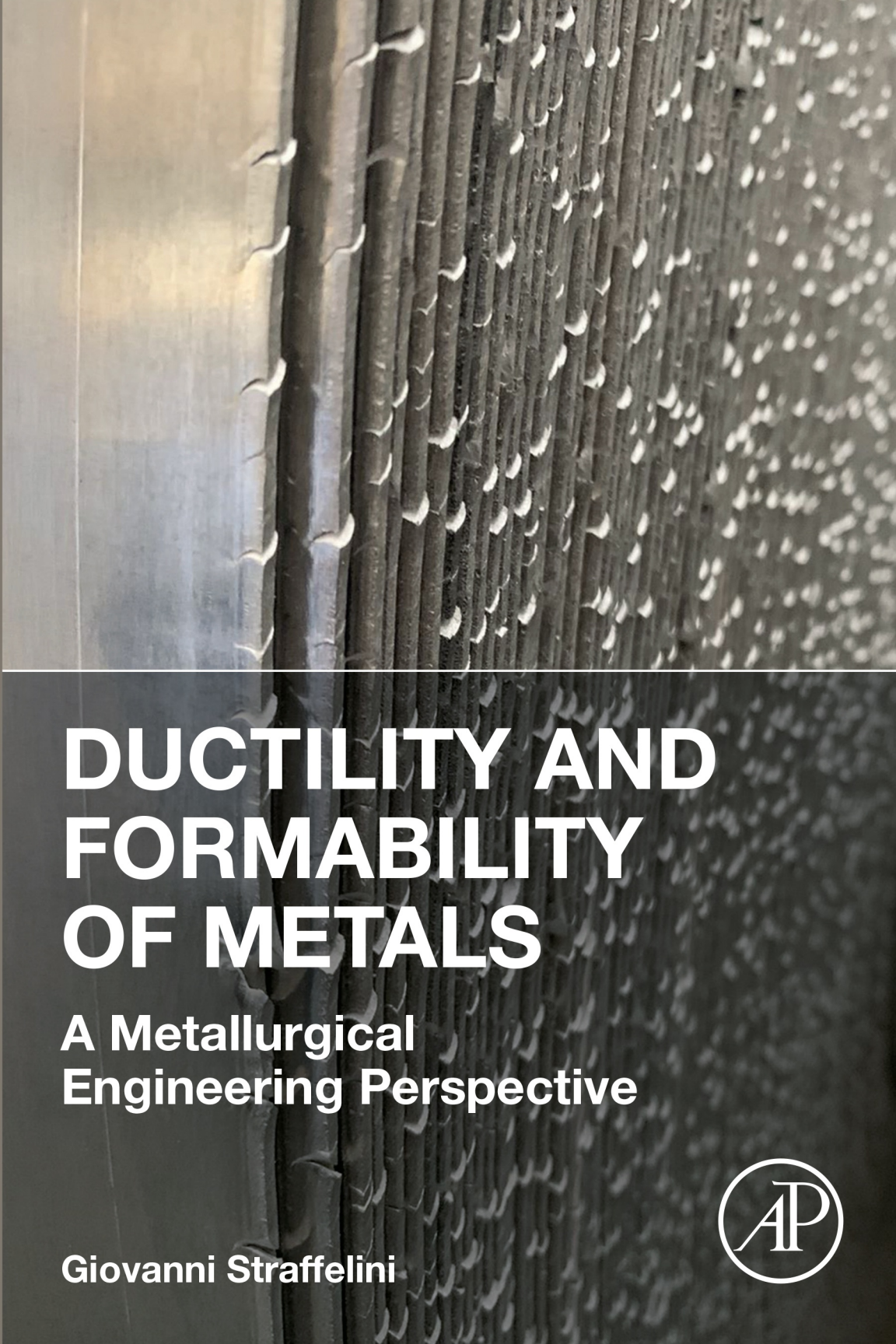




DUCTILITY AND FORMABILITY OF METALS

**A Metallurgical
Engineering Perspective**

Giovanni Straffelini



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A Metallurgical Engineering Perspective

Giovanni Straffelini

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To my wife Fabiana and my daughter Beatrice

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Introduction

This book presents a comprehensive overview of the plastic deformation and ductile fracture behavior of the most important engineering alloys, including steels and Al, Cu, Ti, and Mg alloys, with applications in metal-forming processes. A metallurgical engineering approach is adopted, considering both microstructure and stress/strain conditions.

The first nine chapters introduce the main concepts of the plastic deformation and ductile fracture of metals. The influence of microstructure on the yield behavior is first discussed. The strain-hardening behavior of metals and the influence of strain rate and temperature are then treated. Particular emphasis is given to the meaning and characteristics of the strain-hardening coefficient, n , and the strain rate sensitivity index, m , that play a paramount role in the mechanisms of ductile fracture. A unifying view is presented, and the typical banana-shaped relationship between ductility and yield strength is discussed, considering the positive and negative deviations from the general trend. This is expected to be helpful for the development of new alloys with high ductility without sacrificing strength.

The characteristics of specific phenomena, such as dynamic strain aging and shear banding, are also explained, given their relevance in forming operations. Hot plastic deformation is then discussed, considering the interplay with recrystallization phenomena. Behavior under compression as well as under tension is considered. In fact, most metalworking operations are carried out in compression, but an inhomogeneous plastic deformation may occur (often due to the friction at the tool-workpiece interface), which induces the formation of unexpected secondary tensile stresses that contribute to ductile damage and fracture.

When discussing ductile fracture, a criterion is introduced, selected among the many available, based on the Rice and Tracey (R-T) approach, which is here preferred despite its limitations, since it is based on the physical damage mechanisms. This criterion is a suitable guiding tool to understand and predict the fracture conditions (i.e., the ductility limit) in all the situations where metals are under critical plastic deformation conditions, like those featuring the metal-forming processes. The application of the R-T criterion to metal formability is very important, since the ductile damage of a part under forming is the most important defect to be avoided.

The last seven chapters expose the conventional metal-working operations, i.e., forging, rolling, extrusion, wire drawing, sheet metal forming, and machining by chip removal, following the same pattern, aimed at presenting the formability windows. First, the main processes are described, including mechanical analysis, that are necessary to evaluate the involved deformations, strain rates, and temperature rise, and then the stresses on the tools, the required force, and power are covered. Subsequently,

the main defects that can limit formability are introduced, with particular regard to the formation of ductile cracks, once the ductility limit is reached. The methods to avoid them are presented.

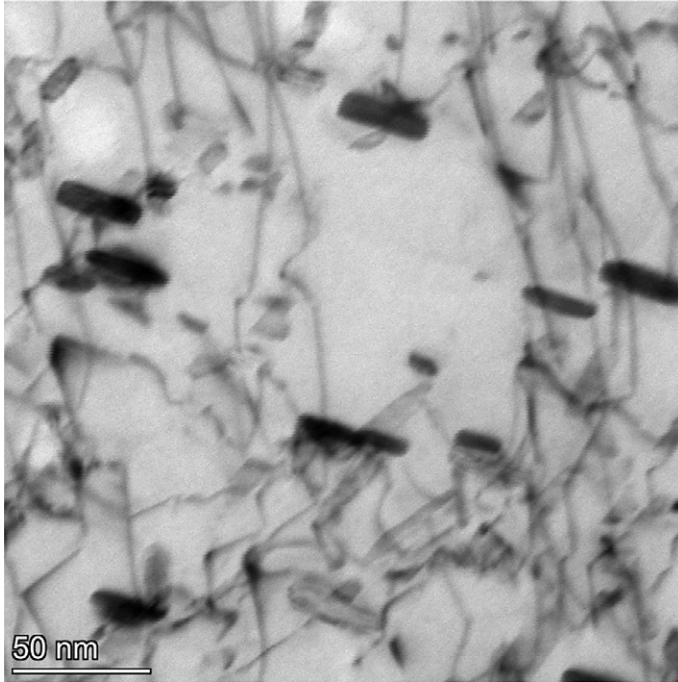
The analysis of the applications and, in particular, the failure mechanisms occurring during specific operations is very useful to better understand the involved phenomena. These aspects, also thanks to the positive interaction with the basic concepts introduced in the first chapters, are meant to provide the reader with a complete set of tools particularly suited to tackling new technological scenarios, such as the traditional working of novel materials, or the plastic shaping of traditional alloys through new methodologies.

The book assumes a basic background in the fundamentals of solid mechanics and metallurgy, and is intended for everyone who has to face problems associated with plastic deformation and ductility in metals, particularly in metalworking processes. In this regard, it might be useful to academicians and students in mechanical, production, and process engineering, metallurgy, and materials science, as well as researchers and professionals in the fields of industrial engineering and physics.

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Onset of yielding in metals

1



Dislocations in a steel with small transition carbides.
Courtesy of Gloria Ischia and Stefano Gialanella.

Chapter outline

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Typically, metals display ductile behavior during mechanical deformation. When subjected to a mechanical stress, they first deform elastically but when the stress reaches a certain level, plastic yielding begins, followed by plastic deformation with large shape changes and energy dissipation. Ductility is then defined as the ability of a metal to undergo large plastic deformation, after the onset of yielding and before fracture.

For the onset of yielding, the applied stress must overcome the metal yield strength. This chapter initially focuses on the measurement of the yield strength and ductility of metals in reference conditions, i.e., with a uniaxial deformation test, at room temperature (RT) and in quasi-static loading. Next, the yield behavior of a single crystal is analyzed, which introduces the basic concepts of dislocations that are required to understand the plastic behavior of metals. Subsequently, the yield behavior of engineering alloys is considered. It is influenced by the main metallurgical features, in particular, the microstructural heterogeneities.

The yield strength of steels and selected non-ferrous alloys is finally presented, together with a short overview of the most used engineering alloys.

1.1 Uniaxial deformation tests

The *yield strength* of a metal is the stress required to initiate plastic deformation. In general, the reference is made to the *uniaxial yield strength*, indicated with σ_Y , which is determined by either uniaxial tension or compression tests. These tests are performed on a universal testing machine (UTM):

- Using unnotched cylindrical or rectangular specimens;
- Under monotonic loading;
- In a displacement-controlled manner, i.e., with a constant travel velocity of the crosshead.

The load-displacement curve is obtained from a load cell, fixed in the crosshead of the machine, which measures the instantaneous applied load, F . Additionally, an extensometer is attached to the specimen, and it measures the displacement $\Delta l = l - l_0$, where l_0 is the original gauge length and l is the instantaneous length. From the recorded curve, the *engineering stress-strain curve* is obtained using the following relationships:

$$s = \frac{F}{A_0} \quad e = \frac{\Delta l}{l_0} \quad (1.1)$$

where:

- s is the engineering stress;
- e is the engineering strain;
- A_0 is the original cross-sectional area of the specimen.

Fig. 1.1A and B shows the typical geometry of a cylindrical and a rectangular specimen for tensile testing with dimensions (in millimeters) complying with the ASTM standard (ASTM E08-91, 1991). A typical representation of a tensile engineering stress-strain curve for a metal is shown in Fig. 1.1C, together with a schematization of the geometry of the specimen during uniaxial deformation.

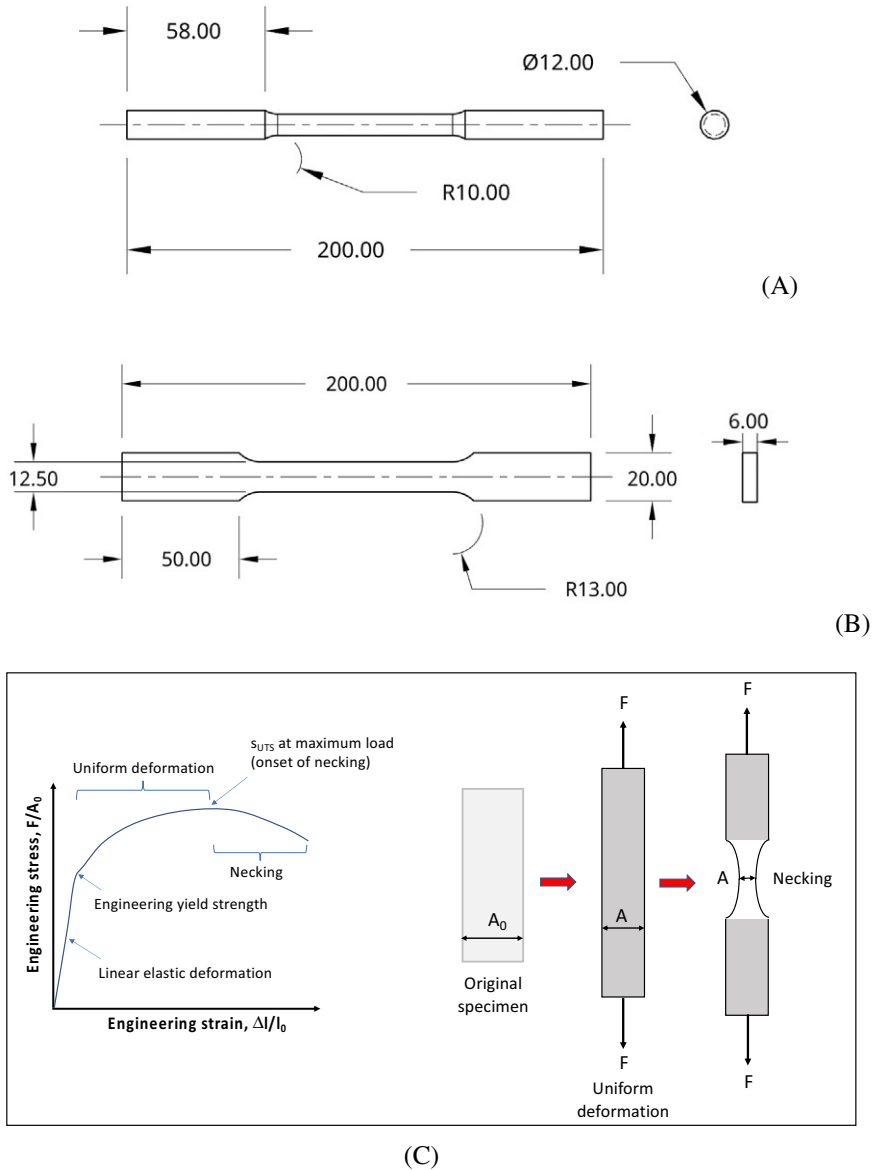


Fig. 1.1 Typical geometry of a cylindrical (A) and a rectangular (B) tensile specimen with dimensions (in millimeters) complying with the ASTM standard (adapted from [ASTM E08-91, 1991](#)). In both cases, the original gauge length, l_0 , is 50 mm. The thickness of rectangular specimens can also be quite small, when testing samples obtained from thin hot/cold flat rolled products, or greater than 6 mm, when testing samples obtained from plates or other parts. (C) Schematic representation of a typical engineering stress-strain curve obtained from a uniaxial tension test and deformation geometry of the tensile specimen.

The *longitudinal elastic modulus*, E , is acquired from the initial linear elastic region of the curve, whereas the *engineering yield strength*, s_Y , is obtained at the onset of plastic deformation, corresponding to a macroscopic deviation from linearity:

$$s_Y = \frac{F_Y}{A_0} \quad (1.2)$$

where F_Y is the load at yielding (further details on the determination of s_Y are given in Sections 1.3 and 1.4). Yielding involves the entire load-bearing section and is also referred to as *general* (or *gross*) *yielding*. After general yielding, plastic deformation is *uniform* until the onset of *necking*, beyond which plastic deformation concentrates on a specific cross section. The engineering tensile stress at the onset of necking is called *ultimate tensile strength*, s_{UTS} , and is given by:

$$s_{UTS} = \frac{F_{\max}}{A_0} \quad (1.3)$$

where $F_{\max}$ is the load at necking. After necking, a decrease in the recorded tensile engineering stress is observed due to the decrease in resisting cross section, which is not compensated by strain hardening. In the case of compression testing, the stress-strain curve is similar to tensile testing but without the occurrence of necking.

In order to better understand the stress and strain behavior in the plastic region, it is important to consider the concepts of *true stress* and *true strain*. The true stress σ is given by:

$$\sigma = \frac{F}{A} \quad (1.4)$$

where A is the instantaneous resisting cross section. The true strain, ϵ , is given by the sum of the strain increments dl/l :

$$\epsilon = \int_{l_0}^l \frac{dl}{l} = \ln \frac{l}{l_0} \quad (1.5)$$

Considering that: $l = l_0 + \Delta l$, Eq. (1.5) can be rewritten in the following way:

$$\epsilon = \ln(1 + e) \quad (1.6)$$

The *true strain rate* at which a metal is deformed is given by the following relationship:

$$\dot{\epsilon} = \frac{\dot{v}}{l} \quad (1.7)$$

where:

- $\dot{v}$ is the crosshead velocity;
- l is the instantaneous gauge length.

In a displacement-controlled test, a constant displacement velocity, $\dot{v}$, is used. In the case of tensile testing, a maximum strain rate is observed at the beginning of the test,

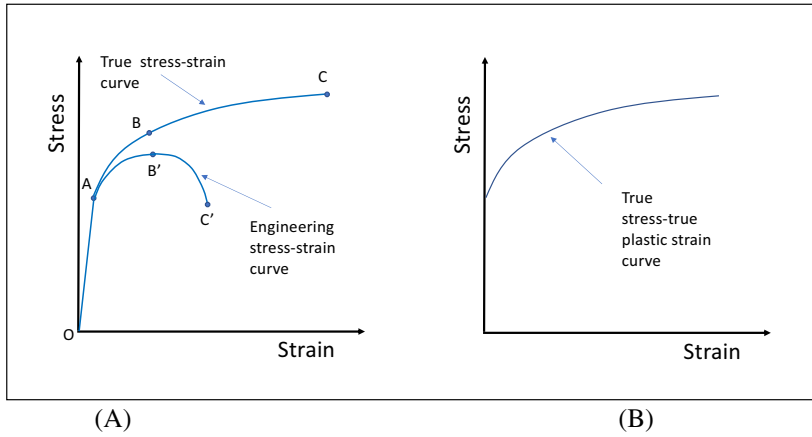


Fig. 1.2 (A) Comparison between an engineering stress-strain curve and the corresponding true stress-true strain curve; section AB (uniform region) is obtained from section AB' , whereas section BC (post-necking region) is obtained from additional measurements of the resisting cross section during tensile testing. (B) Corresponding true stress-true plastic strain curve; note the disappearance of the elastic part (segment OA) of the curve.

which subsides with the testing duration. The opposite is observed in the case of compression testing. The tests carried out in quasi-static conditions display strain rates in the range of 10^{-4} to 10^{-2} s^{-1} (10^{-3} s^{-1} being a typical value). Specific tests with constant strain rate can also be carried out; in these tests, the crosshead velocity must be controlled and it must properly increase during a tensile test.

From the engineering stress-strain curve, it is possible to obtain the *true stress-true strain curve*, where the true stress, σ , is reported as a function of the true strain, ϵ . Fig. 1.2A is a schematization of a tensile true stress-true strain curve. Since in the elastic region $s \cong \sigma$, then we can write: $s_Y \cong \sigma_Y$. The construction of the uniform plastic deformation region (section AB of the curve) derives directly from Eqs. (1.4) and (1.6). The instantaneous cross section, A , can be estimated under the assumption of the constancy of the volume during plastic deformation (only in specific cases, such as in the case of porous materials, is this assumption not satisfied) and neglecting the small volume change associated with elastic strain: $A_0 l_0 = A l$. Therefore:

$$\sigma = \frac{F}{A} = s \cdot \frac{A_0}{A} = s \cdot (1 + e) \quad (1.8)$$

The true stress in the plastic region is also called *flow stress*.

For the construction of the post-necking region (section BC), it is necessary to continuously record the effective reduction of the resisting section, A . This can be done using a proper transverse strain measuring extensometer, or using more sophisticated techniques such as the *digital correlation technique* (Quinta de Fonseca et al., 2005). Using Eq. (1.4), the true stress is then obtained. However, it is an average value since, after necking, the local state of stress becomes multiaxial and the estimated true stress

must be properly corrected (this will be discussed in detail in [Chapter 5](#)). The true strain is given by:

$$\varepsilon = \ln \frac{A_0}{A} \quad (1.9)$$

which is obtained from Eq. (1.5) after setting: $l/l_0 = A_0/A$.

In the analysis of the plastic deformation processes and the ductility of metals, it is necessary to make reference to the concept of *true plastic strain*, ε_p , obtained by subtracting the elastic strain from the true strain:

$$\varepsilon_p = \varepsilon - \frac{\sigma}{E} \quad (1.10)$$

In [Fig. 1.2B](#) a schematization of a true stress-true plastic strain curve is shown. In the plastic deformation of ductile metals and in particular in the case of metalworking operations when plasticity is fully developed, σ/E is low in comparison to ε , and the true plastic strain is considered equal to the true strain, i.e., $\varepsilon_p \cong \varepsilon = \ln(l/l_0)$.

From the tensile test conducted up to fracture, information on the material *ductility* can be gained by measuring the sample cross section at fracture, A_f , on the broken sample, by using Eq. (1.9):

$$\varepsilon_f = \ln \frac{A_0}{A_f} \quad (1.11)$$

Since the elastic strain is recovered after fracture, ε_f actually represents the true plastic strain at fracture. As said, in ductile materials the elastic portion of the true strain is negligible and the true plastic strain almost coincides with the true strain. ε_f is commonly indicated as the *true fracture strain*.

The *ductility* indicates the ability of a metal to undergo large plastic deformation before fracture. Two other engineering parameters are commonly used to express ductility: the *elongation at fracture*, EL, and the *reduction of area at fracture*, RA. They are defined by the following relationships:

$$EL = \frac{l_f - l_0}{l_0} \quad (1.12)$$

$$RA = \frac{A_0 - A_f}{A_0} \quad (1.13)$$

where l_0 is the original gauge length of the specimen determined by two marks at a given distance, and l_f is the distance between the two marks once the fractured ends are fitted together (EL and RA are often expressed as a percentage).

The reduction of area at fracture, RA, is related to ε_f , since: $\varepsilon_f = \ln \left(\frac{1}{1-RA} \right)$. EL is not directly related to ε_f , since it depends on l_0 .

[Table 1.1](#) summarizes the main parameters that can be obtained from uniaxial tensile testing using cylindrical specimens (some parameters will be introduced in the next chapter).

Table 1.1 Main parameters obtained from tensile testing.

Symbol	Parameter	Method	Fields of use
E	Longitudinal modulus of elasticity (Young's modulus)	Engineering curve (alternatively, from loading-unloading cycles or the acoustic resonance method)	Mechanical design for stiffness
σ_Y	Yield strength	Engineering curve	Mechanical design for strength Metalworking operations
K, n	Strength and strain-hardening coefficient (L-H equation; see Chapter 2)	True stress-true plastic strain curve	Metalworking operations
s_{UTS}	Ultimate tensile strength	Engineering tensile curve	In mechanical design and metalworking operations to infer maximum resistance
ϵ_f (EL, RA)	Fracture strain (elongation at fracture and reduction of area at fracture)	Tensile fractured samples	Metalworking operations

Parameters K and n will be introduced in the next chapter. In general, parameters E , σ_Y , K , and n obtained from compression tests are similar to those obtained from tensile tests (important asymmetries will be outlined in the next chapters).

1.2 Onset of yielding in single crystals

The onset of plastic deformation is triggered by the motion of *dislocations* on easy slip planes under the action of a *shear stress*. Dislocations are line defects. There are two basic types of dislocations: *edge* and *screw* dislocations. Additionally, dislocations may also have a mixed character ([Hirth and Lathe, 1992](#)).

[Fig. 1.3A](#) depicts an **edge dislocation**. The figure represents a crystal lattice with a semi-plane of atoms inserted in it. The edge dislocation consists of a line at the end of this semi-plane, around which the atoms are elastically displaced with respect to their ideal position. The motion of the dislocation by *slip* (also called *glide*) is possible when the applied shear stress, τ , exceeds a critical value, τ_c , as shown in [Fig. 1.3B](#), inducing a slight displacement of the atoms at the end of the semi-plane. This allows for the progressive displacement of the dislocation up to the free surface, where it forms a step, as seen in [Fig. 1.3C](#). The slip direction and the length of the step defines the *Burgers vector*. It is perpendicular to the edge dislocation. The dislocation motion is favored on planes with high atom packing (*slip planes*)

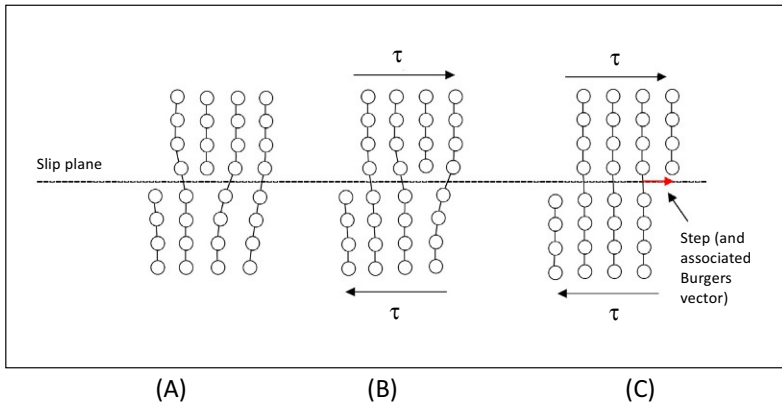


Fig. 1.3 An edge dislocation (depicted in A) slips under the action of shear stress ($\tau > \tau_c$); B. The edge dislocation is perpendicular to the Burgers vector (which also indicates the direction of motion, C). An edge dislocation (i.e., the end of the semi-plane inserted in the crystal lattice) is usually indicated with the symbol: $\perp$.

and in directions with minimum distance between the atom planes (*slip directions*). In these cases, the local distances between the atoms are at a minimum and the atomic displacement is easier. The combination of a slip plane and a slip direction gives a *slip system*.

In addition to slip, edge dislocations can move by *climb*, perpendicularly to the slip plane. This is due to the diffusion of atoms or vacancies located at the end of the semi-plane. When an atom (or a vacancy) exchanges its position with an adjacent vacancy (or atom), a *jog* is formed. If this process involves several atoms or vacancies, a net motion of the dislocation occurs. This process is favored by the application of a shear stress and increases in importance with increasing temperature, which activates the diffusive motion of atoms and vacancies. The important role of climbing at high temperature is better explained in [Chapter 4](#).

[Fig. 1.4](#) depicts a **screw dislocation**. The upper part of the crystal is displaced to the right under the action of a shear stress, τ . The screw dislocation then moves perpendicularly to the direction of the applied stress. The Burgers vector is parallel to the slip direction. However, unlike in an edge dislocation, it is parallel to the dislocation line. This has an important consequence; screw dislocations do not move on a unique slip plane, allowing for the possibility of *cross slip*, which may occur in materials containing two or more slip planes with a common slip direction (such as in metals with fcc structure).

[Fig. 1.5A](#) presents the yield in a single crystal subjected to simple uniaxial tensile loading condition. For convenience, the system is considered to be two-dimensional. In a single crystal, the *yield strength* primarily depends on the orientation of the crystallographic planes with respect to the loading direction. The tangential component of the load F on the plane AA' is: $F \sin \theta$. The corresponding shear stress, τ , is maximum for $\theta = 45^\circ$. However, the slip does not necessarily occur for $\theta = 45^\circ$, but involves a

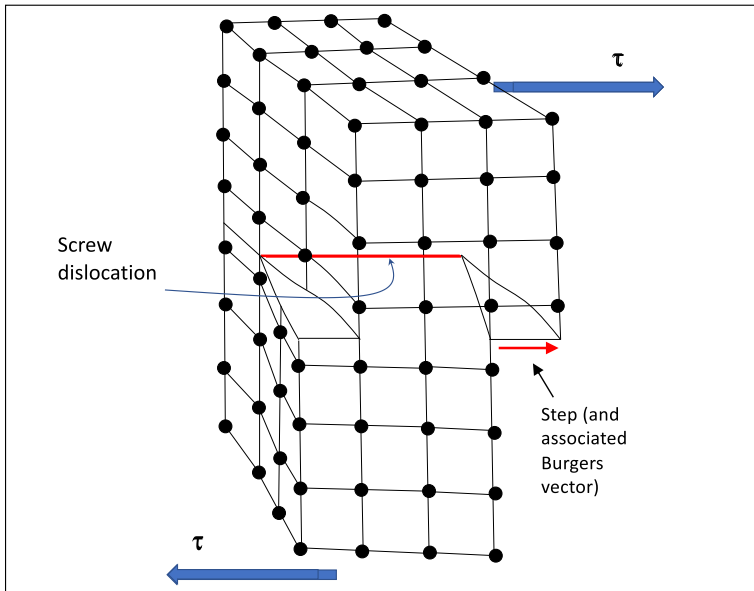


Fig. 1.4 A typical representation of screw dislocation. The screw dislocation is parallel to the Burgers vector (which also indicates the direction of motion).

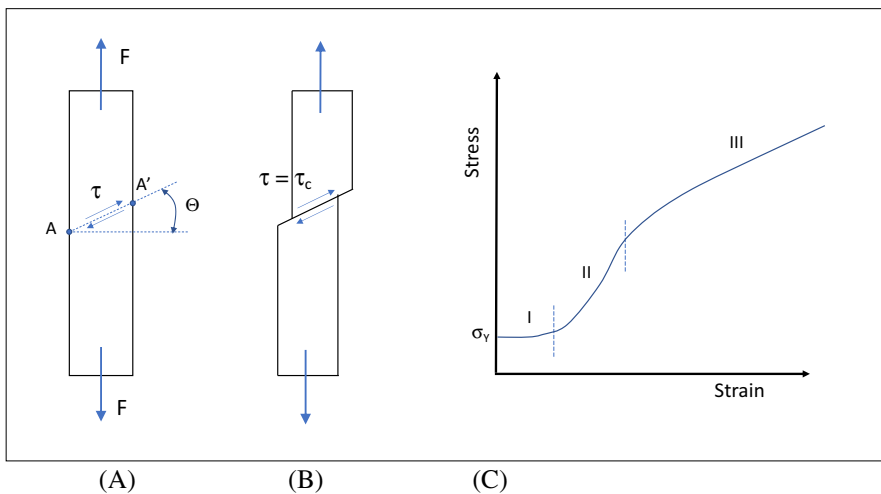


Fig. 1.5 (A) Single crystal subjected to tensile testing; (B) triggering of yield (Stage I) when the shear stress resolved in the easy slip plane reaches the critical value τ_c ; (C) development of plastic deformation.

slip plane at which the *critical resolved shear stress*, τ_c , is reached first. Hence, the yield strength, σ_Y , is given by:

$$\sigma_Y = \frac{\tau_c}{\sin \theta \cdot \cos \theta} \quad (1.14)$$

The critical resolved shear stress is determined by the crystalline structure and slip system; the presence of lattice imperfections, such as vacancies, interstitials, or other dislocations; and temperature. At room temperature (RT), it assumes the following typical values (Dieter, 1988):

- Between 0.48 and 1.3 MPa in metals with a face-centered cubic (**fcc**) crystal lattice (such as Al, Cu, Ag, and γ -Fe);
- Between 0.18 and 0.77 MPa in the basal plane of metals with a hexagonal close-packed (**hcp**) crystal lattice (like Zn and Mg), except in Ti (α -Ti alloys), where it assumes comparatively high values (beyond 10 MPa);
- Comparatively higher values in metals with a body-centered cubic (**bcc**) crystal lattice (such as Mo, α -Fe, β -Ti). For example, 27.5 MPa in α -Fe and 49 MPa in Mo.

The ease of onset of plastic deformation depends on τ_c and on the number of available slip systems. The highest numbers of such systems are present in metals with bcc structure, even if the slip planes are not at maximum packing, leading to high τ_c values. Fcc metals deform more easily, as they are characterized by lower values of τ_c and the presence of 12 easy slip planes. Metals with hcp structures deform easily if properly oriented. Fig. 1.6 shows a schematization of the hcp crystallographic structure. Typically, dislocations move by slip in the three closed-packed *a* directions in the basal systems. Dislocation slip may also occur in the prismatic planes, if specific atoms that

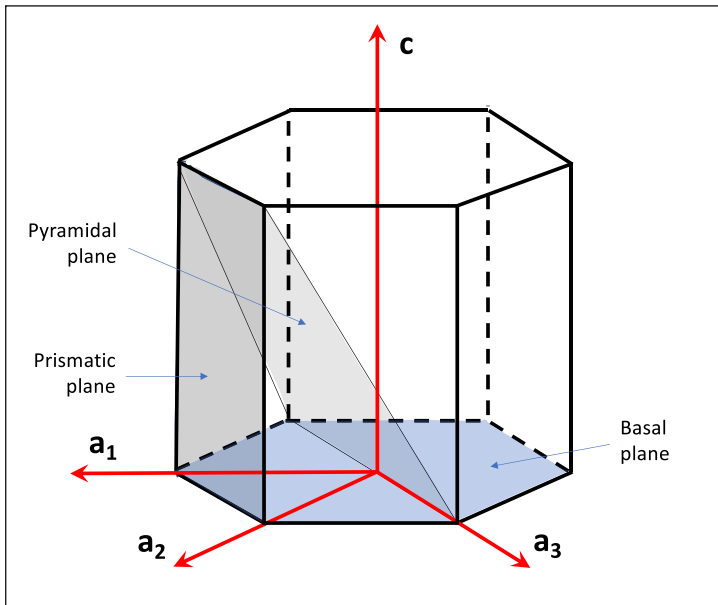


Fig. 1.6 Schematization of the hcp crystallographic structure.

Table 1.2 Plastic deformation characteristics of some single metal crystals (Dieter, 1988; Abbaschian et al., 2009).

Crystal structure	Metals	Possible slip systems	Comments
fcc	γ -Fe Al Cu Ni	Three close-packed directions $\times$ 4 octahedral slip planes = 12 easy slip systems.	Plastic deformation starts at low stress. However, dislocations may easily interact during sliding (because of early activation of multiple slip), producing strain hardening.
bcc	α -Fe β -Ti Mo	Four closed-packed directions. No closed-packed planes, but three (or more) planes with high atomic density.	Plastic deformation is easy but irregular, as no slip planes are well defined (and the τ_c -values are quite high).
hcp	Mg α -Ti Zn Be	Three closed-packed directions on the basal plane only.	Large plastic deformation in single crystals, but only if properly oriented.

increase the packing density are present (such as Al atoms in Ti alloys). However, dislocation slip does not occur in the c direction and the slip in the basal and/or prismatic plane does not produce a strain component parallel to the c -axis (Hosford and Caddell, 1993). As will be seen in the next chapters, this will markedly reduce the plastic deformability of hcp metals.

In Table 1.2, a summary of the main characteristics of slip systems in different metals is given.

If $\tau < \tau_c$, deformation remains elastic. In Stage I of deformation, as seen in Fig. 1.5C, the dislocations slip proceeds easily on the *primary* slip system, causing an increase in the plastic deformation. In properly oriented hcp single crystals, stage I may be particularly extended. However, at a certain stress level, dislocations of other slip systems are activated (multiple slip), which begin to multiply and hinder their own movement, forming dislocation *forests* and *bands*. This leads to the *strain-hardening* phenomenon, which is characterized by the need to significantly increase the applied stress to increase plastic deformation (Stage II).

Successively, the strain hardening may decrease in intensity (Stage III), when sufficiently high temperature favors:

- The movement of edge and screw dislocation by climbing and by cross slip, respectively;
- The *onset of dynamic recovery* phenomena.

As mentioned, edge dislocations may move by climbing. A temperature rise activates the dislocation climb, as it is triggered by the phenomenon of diffusion. Climbing permits the dislocations to surmount microstructural obstacles, facilitating plastic deformation. Screw dislocations may undergo cross slip in materials containing two or more slip planes with a common slip direction. This permits additional slip planes

to be active. Also, cross slip allows the dislocations to surmount microstructural obstacles. Therefore, dislocations climb and cross slip facilitates plastic deformation, leading to a decrease in the strain-hardening rate. Finally, dynamic recovery phenomena, such as dislocations annihilation, reduce the dislocation density during plastic deformation (in particular at large strains and high temperature). Also, this phenomenon facilitates plastic deformation (recovery phenomena and the role of temperature in plastic deformation are discussed in detail in [Chapter 4](#)).

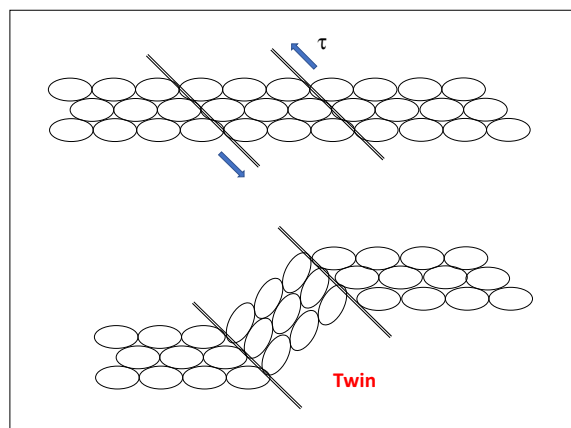
The three stages illustrated in [Fig. 1.5C](#) are not always present. Their presence depends on the crystalline structure, the crystallographic orientation with respect to the loading direction, the purity of the metal, the initial density of the dislocations, and temperature. At high temperatures, Stage III is prevailing.

In metals with few slip systems (such as in hcp structures) or in situations of difficult slip (such as at low temperature and high strain rate, and when τ_c is high, as in bcc metals), the deformation mechanism by *twinning* becomes important, typically occurring together with dislocations slip. This process is shown in [Fig. 1.7](#) ([Abbaschian et al., 2009](#)). Twins may be produced by a shear stress that induces a change in the orientation of a portion of a crystal, forming a mirror image with respect to the parent crystal, with reference plane of symmetry. A twin is a plane defect, whereas a dislocation is a line defect. Correspondingly, the critical shear stress for twinning is larger than that for slip. For example, in monocrystalline Fe and in quasi-static loading, the twinning shear stress is approximately 220 MPa ([Meyers et al., 2002](#)). However, it is almost independent on temperature and strain rate. The twinning shear stress is proportional to the *stacking-fault energy*, *SFE*, which will be introduced in the next chapter.

1.3 Onset of yielding in engineering metals

Engineering metals are typically *polycrystalline* in nature, with a complex microstructure containing different phases and particles, such as non-metallic inclusions and intermetallic precipitates. [Fig. 1.8A](#) provides a typical example of the microstructure of engineering metal. Due to this, the yield process is more complicated than in single crystals.

Fig. 1.7 Representation of deformation by twinning under the action of a shear stress, τ . Modified and redrawn from Abbaschian, R., Abbaschian, L., Reed-Hill, R.E., 2009. *Physical Metallurgy Principles*, fourth ed. Cengage Learning.



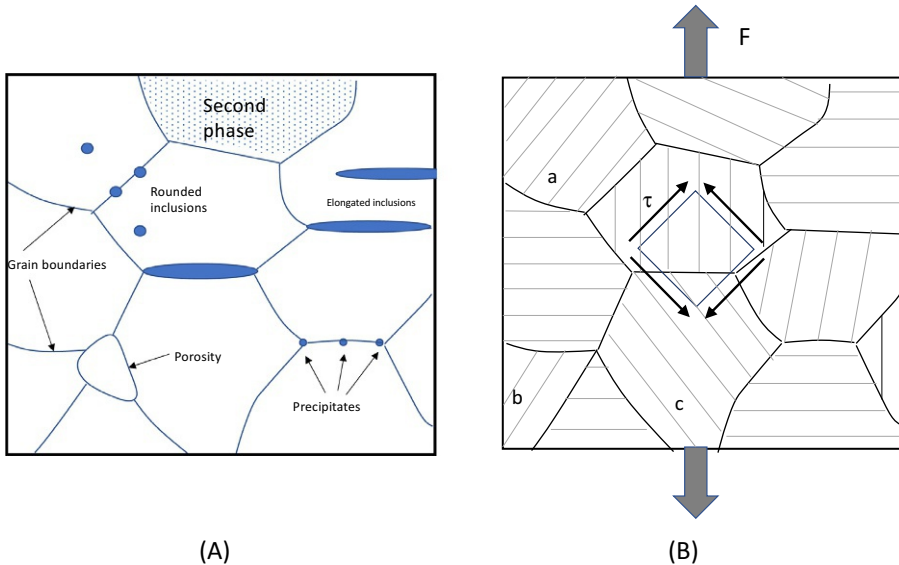


Fig. 1.8 (A) A typical example of the microstructure of an engineering metal. Note that the pores are mainly present in cast alloys and alloys made from the powder metallurgy technique. Elongated inclusions, such as Mn-sulfides, are typically present in hot-rolled steels. (B) Schematization showing the random orientation of most favorable slip planes in polycrystalline metals. Slip initially starts in grains indicated with *a*, *b*, and *c*, where slip planes form an angle close to 45 degrees with the loading direction.

As seen, plastic deformation in single crystals is due to the dislocations slip under the action of a shear stress. Polycrystalline metals contain numerous grains with random orientation, as schematized in Fig. 1.8B. The applied shear stress, τ , is maximum at an angle of 45 degrees to the tensile axis. Therefore the dislocations slip is first initiated in grains that are most favorably oriented (they are indicated with *a*, *b*, and *c* in Fig. 1.8B). Dislocations then pile up at the grain boundaries. Subsequently, different slip systems are activated (multiple slip) and slip initiates also in neighboring less favorably oriented grains. In polycrystalline fcc and bcc metals, this implies the disappearance, or strong reduction, of Stage I (Fig. 1.5C), with the starting of strain hardening almost from the beginning of loading. In hcp metals, plastic deformation is rather difficult and also, for comparatively large average strains, it may involve only the better oriented grains, i.e., the grains with the basal planes favorably oriented with respect to the applied shear stress.

In addition to the dislocations activated and generated to allow the homogeneous plastic deformation of the crystal lattice, *geometrically necessary dislocations* are also generated. They guarantee the local accommodation of the deformations at the heterogeneities (one example is the mentioned dislocation pile-ups at the grain boundaries), leading to the introduction of local strain gradients. This concept will be explained in detail in Chapter 2. In engineering metals, the plastic deformation develops in an inhomogeneous manner, since the microstructure determines

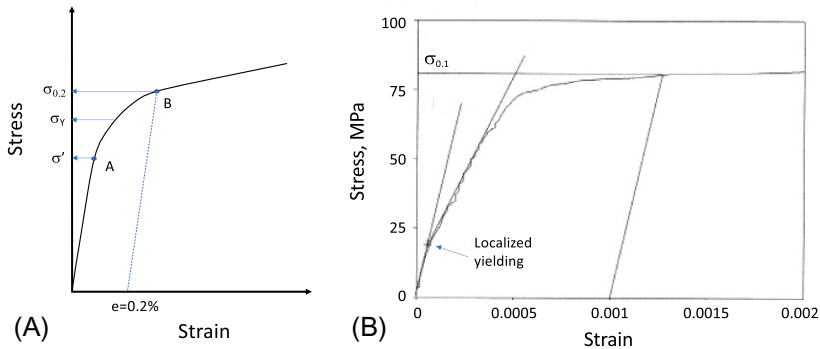


Fig. 1.9 (A) Schematization of stress-strain curve near the yield point with reference to the behavior at generalized yielding and localized yielding. (B) Engineering stress-strain curve of sintered iron with a porosity of 12.7%.

Modified from Straffelini, G., Fontanari, V., Molinari, A., 1999. True and apparent Young's modulus in ferrous porous alloys. *Mater. Sci. Eng. A* 260, 197–202; reproduced with permission of the publisher.

the characteristics (in particular, the density) of the geometrically necessary dislocations.

In engineering metals, the stress-strain curve at the macroscopic *yield point* displays the typical behavior shown in Fig. 1.9A. As mentioned previously, the *yield strength*, indicated by σ_Y , represents the *general yield strength*, corresponding to which the entire cross section of the specimen plasticizes. However, if the region of occurrence is closely inspected, it can be seen that the deviation from linearity starts at σ' , which is lower than σ_Y . In fact, due to the heterogeneous microstructure, the plastic deformations involving the better oriented grains or the more deformable phases occur well before general yielding. Such early deformations occur typically close to pores or hard particles, due to their stress and strain concentrating effects. The general yield point is then reached later, at a point in the *AB* curve, after the spreading of plastic deformation onto the entire resisting section.

The yield strength, σ_Y , is then difficult to determine experimentally. Therefore it is often identified with the stress that produces a permanent engineering plastic deformation of 0.2% after unloading (sometimes 0.1%). It is indicated with $\sigma_{0.2}$, and it is set as: $\sigma_{0.2} = \sigma_Y$. Clearly, at the engineering strain of 0.2%, plasticity is already developed in the material. However, in many cases, the simplification can be accepted. In some cases, σ_Y is obtained using more complex methods. One of them is to consider the stress at which the third-order derivative of the stress with respect to the deformation reaches a minimum (Prangnell et al., 1994).

In metals with *highly heterogeneous microstructure*, a very early deviation from linearity is detected, and the σ' -values are much lower than σ_Y . As an example, Fig. 1.9B shows the engineering stress-strain curve (for nominal strains less than 0.002) for sintered Fe with a porosity of 12.7% (Straffelini et al., 1999). The

occurrence of the anticipated yielding can be clearly detected, which is due to the starting of plasticity at the pore edges. Pores act as microstructural *internal notches* (refer to Section 5.5). A *continuous yielding* is then observed in the stress-strain curve, with a progressive transition from localized plastic deformation to the generalized deformation. The early deviation from linearity, with the corresponding small σ' -value with respect to σ_Y , is called *localized yielding* or *first yielding*. The subsequent determination of the generalized yield point is rather difficult, since the curve shows a continuous bending. In the case of the stress-strain curve in Fig. 1.9B, it is expected that the σ_Y is in the range of 60–80 MPa, and with the simplified approach outlined previously, we obtain: $\sigma_Y = \sigma_{0.1} = 80$ MPa.

All metals with very heterogeneous microstructure show the localized yielding phenomenon and the subsequent continuous yielding. The microstructural feature responsible for plastic localization can be mechanically regarded as an internal notch. Among these metals, we may list:

- Sintered steels with high porosity and/or inhomogeneous matrix microstructure;
- Dual-phase steels (with ferritic-martensitic microstructure);
- Duplex stainless steels (with ferritic-austenitic microstructure);
- Cast irons containing graphite particles (flakes or spheroids) in the microstructure;
- Metal matrix composites reinforced with ceramic particles or fibers;
- Tool steels containing a large fraction of hard carbides;
- As quenched steels (also subjected to low-temperature tempering) containing a comparatively large fraction of retained austenite (its content increases with C content).

In addition, the first yielding phenomenon may also be recorded in metals with hcp structure, as the plastic deformation starts early in grains that are favorably oriented. As seen, dislocations easily move along the a directions in the basal planes, whereas this does not occur in the c direction.

The determination of σ' is strongly dependent on the sensitivity of the extensometer, which typically permits recording strain magnitudes typically on the order of 10^{-4} . A closer look at the curve origin, with an extensometer able to detect strains on the order of 10^{-6} , often provides evidence of an even earlier deviation from linearity. It is due to the development of *micro-plasticizations*, caused by the microstructural heterogeneities and, most of all, residual stresses, which lead to an intensification of the local stress, inducing early motion of a few hundred dislocations (Dieter, 1988). The stress responsible for such a deviation, indicated with σ_e , is defined as the *elastic limit*.

In engineering metals with heterogeneous microstructure, the local values of stress and strain are different from the measured macroscopic *average* values, expressed by the relationships introduced in Section 1.1. To understand this further, consider the two limiting cases of a sample volume made with different material blocks in series, and a sample volume made by different material stripes in parallel. In the first case, the stress is the same everywhere, but the strain differs in the blocks depending on the material deformability (and the plastic strain is zero if the stress is lower than the yield strength of the material of the block). In the second case, the strain is the same everywhere. Hence, for this to be accounted for, the local stress in each stripe is different.

In real engineering materials, the microstructure may contain a mix of different phases in series or in parallel. Therefore both the local stresses and strains may considerably vary from point to point in the microstructural constituents, and this may introduce difficulties and approximations in the modeling in the *continuum* of the stress-strain behavior of metals.

1.4 Upper and lower yield point phenomenon

Annealed metals contain small dislocation density. Many of them, including ferritic and pearlitic-ferritic steels with low solutes, display the *upper and lower yield point phenomenon*, also called the *yield point phenomenon*, shown in Fig. 1.10. After reaching a stress limit value, σ_s (the upper point), the applied stress drops sharply to σ_i (the lower point) and remains approximately constant, as the deformation proceeds at the given strain rate. The explanation of this phenomenon is due to the *Cottrell atmosphere*. Dislocations are proposed to be blocked by solute atoms present as impurities in the crystal lattice, known as dislocations, or Cottrell atmospheres. Examples of such solutes are: C, N, and O atoms in Fe, Mo, and Zn alloys; Zn, Al, As atoms in Cu alloys; and Mg and Cu atoms in Al alloys. Hence, to remove them from anchors, it is necessary to apply a stress, σ_s , able to activate the motion of such dislocations. Subsequently, plastic deformation proceeds with the formation and propagation of the *Lüders bands*, oriented at 45 degrees with respect to the loading axis, i.e., in the direction of the maximum shear stress. The yield is generalized, but it is concentrated in these bands, which grow in size during plastic deformation. The growth is favored by the multiplication of dislocations, which are continuously generated at a small distance in front of the propagating band, until they uniformly spread over the entire specimen gauge length. The length of the *Lüders plateau* (also called *Lüders strain*)

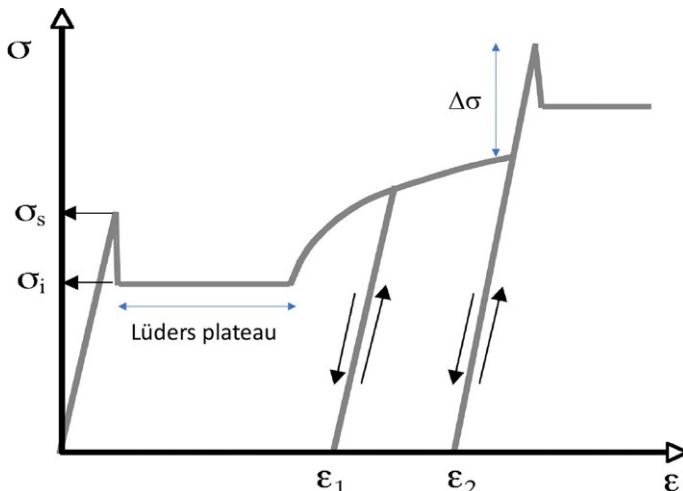


Fig. 1.10 The upper and lower yield point and strain aging phenomenon.

depends on strain rate. As strain rate increases, solutes have less time to dynamically block the moving dislocations and the Lüders strain decreases. Thereafter, the plastic deformation proceeds in the usual way. For practical purposes, in metals that exhibit this behavior, the lower yield stress, σ_l , is often considered as the material yield strength ($\sigma_Y = \sigma_l$).

If after a certain plastic deformation (e.g., ϵ_1 , in Fig. 1.10), a complete unloading is performed rapidly followed by a reloading, the hardened metal does not display the yield point phenomenon, and the yield strength of the deformed metal is determined by its strain hardening. However, if after unloading at a certain deformation (say ϵ_2), sufficient time is provided before reloading, the yield point phenomenon appears again, producing an increase in the yield strength, which is indicated by $\Delta\sigma$ in the figure. Such a phenomenon is called *strain aging* or *strain-age hardening*. It is due to the diffusion of atoms toward the dislocations during the waiting time, and their re-anchoring. In steels, strain aging is slow at RT, since it is necessary to wait a few months before its occurrence. However, strain aging is quicker at higher temperature.

The Lüders bands produce a rough appearance on the metal surfaces. In many cases, such as in sheet metal forming operations, their formation is not acceptable for esthetic reasons. They can be avoided by reducing impurities to low values. In steels, this objective can be pursued by the addition of C and N stabilizing elements such as Ti, Nb, and V. These elements form carbonitrides, which remove C and N atoms from their interstitial positions in the crystal lattice (in this way, the so-called *interstitial free*, IF, steels are produced). Another method consists of making a small preliminary plastic deformation by compression with sufficient intensity on the sheet surface, since no Lüders bands form under compression (this treatment is called *skin-passing*, which is obtained by cold rolling, discussed in Chapter 12). The subsequent tensile deformation of the sheet follows the usual strain-hardening path, without impairing the quality of the surface.

The yield point phenomenon does not appear when the metal contains sufficiently large microstructural heterogeneities, inducing the localized yielding phenomenon.

1.5 Effect of microstructure on the yield strength

The yield strength is given by the stress required to promote the motion of the dislocations on the entire load-bearing section. Therefore σ_Y increases with the difficulty to start such movements; σ_Y is considered to depend on the following parameters:

1. The crystal lattice itself;
2. Alloying elements;
3. Other dislocations;
4. Grain boundaries;
5. Particles or other secondary phases.

In general, it can be stated that:

$$\sigma_Y = \sigma_0 + \sigma_s + \sigma_g + \sigma_d + \sigma_p \quad (1.15)$$

where:

- σ_0 (often called *Peierls-Nabarro* stress) is the stress required to overcome the lattice friction. At RT, it is approximately 54 MPa for α -Fe, 10 MPa for Al, 25 MPa for Cu, 7 MPa for Mg, and 78 MPa for Ti.
- σ_s is the stress required to overcome the lattice elastic distortions exerted by solute atoms (but other factors may also play a role). This effect increases with the difference in the atomic radius between solute atoms and solvent atoms. In α -Fe, P exerts a notable solution hardening effect (also Si, but to a lesser extent). N that is present as interstitial elements also exerts a notable influence on σ_s . In the case of Al alloys, Mg is rather effective in increasing the yield strength and is added up to 6% in some alloys of the class 5000.
- σ_g is the stress required to overcome other dislocations. It is expressed by the *Taylor* equation:

$$\sigma_g = \alpha \cdot G \cdot b \cdot \sqrt{\rho}$$

where:

- α is a material constant (it is approximately 0.8 for steels and Al alloys);
 - G is the shear elastic modulus (approximately 77,000 MPa in α -Fe);
 - b is the modulus of the Burgers vector (e.g., 0.248 nm for α -Fe);
 - ρ is the density of dislocations, expressed as the number of dislocations crossing a unit area.
- For example, in annealed α -Fe, dislocation density is comparatively low (approximately 10^8 cm^{-2}) and σ_g is close to 16 MPa. After heavy plastic deformation (e.g., after cold drawing), ρ may reach values up to 10^{13} cm^{-2} , corresponding to $\sigma_g = 5000 \text{ MPa}$, which is close to the maximum possible yield strength for steels.
- σ_d accounts for the effect exerted by grain boundaries through the well-known *Hall-Petch* relationship:

$$\sigma_d = \frac{k}{\sqrt{d}}$$

where:

- d is the mean grain size;
- k is a material constant.

In the case of ferritic and ferritic-pearlitic steels, k is approximately $17 \text{ MPa mm}^{1/2}$, which is not dependent on the alloying content. Strengthening from grain boundaries is quite limited in the case of Cu and Al, i.e., in metals with an fcc structure, which easily undergo plastic deformation. Different models were proposed to explain the Hall-Petch relationship. One relates σ_d to the stress required to overcome grain boundaries that act as barriers for dislocation motion. Another model relates σ_d to the density of geometrically necessary that is inversely proportional to the grain size, d (refer to [Section 2.1](#)).

- σ_p is the stress required to overcome the fine particles present in the microstructure. *Coherent* particles with atomic lattice that match that of the matrix are deformable, in the sense that they can be crossed by dislocations. In this case, σ_p is proportional to $R^{1/2} f_v^{1/2}$, where R and f_v are the average radius and the volume fraction of the particles, respectively. The higher the radius, the higher is the coherency strain associated to the particle, leading to higher

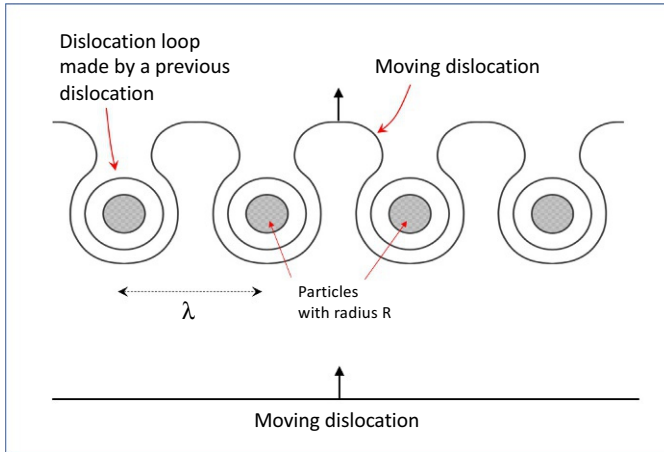


Fig. 1.11 Orowan mechanism showing the resistance offered by incoherent particles to the motion of dislocations. The formation of dislocation loops can also be seen. Modified and redrawn from Abbaschian, R., Abbaschian, L., Reed-Hill, R.E., 2009. Physical Metallurgy Principles, fourth ed. Cengage Learning.

strengthening, i.e., σ_p values. If the particles are *not coherent (incoherent)*, they may be bypassed by dislocations through the *Orowan mechanism*, shown in Fig. 1.11. Note that in this process, each dislocation leaves a dislocation loop around the particle. These dislocations loops are examples of geometrically necessary dislocations. In the case of incoherent particles, a model proposed by Orowan shows that:

$$\sigma_p = \frac{K \cdot b}{\lambda} \cdot \ln \left(\frac{R}{b} \right)$$

where:

- K is a material constant;
- b is the modulus of the Burgers vector;
- λ is the average spacing between the particles;
- R is the average radius of the particles.

1.6 Yield strength of steels

The yield strength of metals depends on microstructure, temperature, and strain rate. Some examples of selected steels and non-ferrous alloys are described in the present and next section. In every case, the yield strength is considered at RT (room temperature) and in quasi-static conditions. In the next chapters, the role of temperature and strain rate will be outlined.

Ferritic steels

In annealed ferritic steels, the yield strength (in megapascals) can be evaluated with Eq. (1.16) (Pickering, 1978):

$$\sigma_Y = 54 + 32.3\%Mn + 83\%Si + 23 \cdot \sqrt{\%N} + \frac{17.4}{\sqrt{d}} \quad (1.16)$$

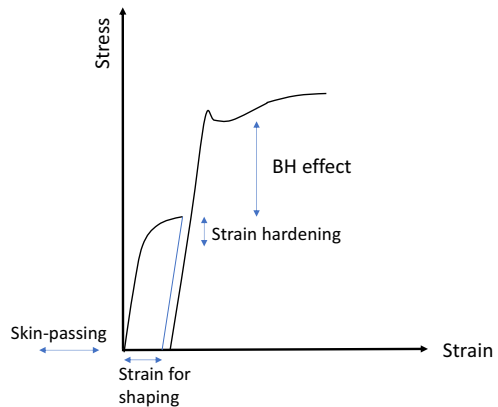
$\uparrow$ σ_0 $\underbrace{\hspace{10em}}_{\sigma_s}$ $\uparrow$ σ_d

where d is the average grain size (in millimeters) and %N indicates the percentage of free N in interstitial solid solution, i.e., not combined as a nitride (all the percentages are in weight). This relationship can also be used for steels with low carbon content, i.e., when C is mainly present as small and isolated carbides at the grain boundaries or in very small pearlite nodules.

The yield strength of ferritic steels is quite low, typically between 150 and 200 MPa. Therefore these steels are called *mild steels*. Since they possess a high ductility, they are mainly used in the form of sheets and subjected to deep-drawing or stretch forming operations. In these steels, σ_Y is considerably increased by increasing the density of dislocations, which is attained by cold working. In the case of heavy cold working, σ_Y values are higher than 350 MPa, coinciding with σ_{UTS} , with rather low ductility. In mild steels, skin-passing is required to avoid formation of the Lüders bands in sheet metal forming operations (Chapter 12).

The so-called *bake hardening (BH) ferritic steels*, also used in sheet metal forming operations, contain very low C content (on the order of 0.02%) and low content of alloying elements. The yield strength varies between 180 and 300 MPa. The C content is not sufficient for the formation of carbides in the microstructure, and then it remains in solid solution. This carbon in the solid solution is responsible for the strain aging effect (Fig. 1.10), which occurs during heating after plastic deformation, generally realized through hot painting. A schematization of this phenomenon, called *BH effect*, is provided in Fig. 1.12. After skin-passing, the steel is subjected to plastic forming with a given strain. This induces a strength increment by strain hardening.

Fig. 1.12 Schematization of the BH effect in a ferritic steel subjected to skin-passing, plastic deformation, and then heating.



Subsequently, the metal is submitted to hot painting, typically at 170°C for 20 min, which induces an additional increase in strength, typically in the range of 50 MPa (Llewellyn and Hudd, 1998). This behavior is very attractive when a high plastic formability must be coupled with a sufficiently high strength of the formed part. A typical example is the dent resistance of automotive body panels.

A special class of ferritic steels is constituted by the *IF steels*, already mentioned in Section 1.4. In these steels, the C and N content in solid solution is very low, generally in the range 25–30 ppm. This is attained through a special process comprising vacuum degassing, followed by the addition of small amounts of Ti and Nb. These elements react with C and N to form carbonitrides, leading to a high reduction in the content of these interstitials. These steels display high plastic deformability and do not display the formation of the Lüders bands during tensile plastic deformation. The strength of IF steels may be increased with the addition of particular elements like small amounts of P, which appreciably increases the strength of ferrite.

Ferritic-pearlitic steels

The yield strength of ferritic-pearlitic steels, with medium C content (on the order of 0.25%), is in the range of 200–400 MPa, and it is influenced by the fraction of pearlite and its interlamellar spacing. These steels are mainly used in structural applications, for the construction of buildings, ship hulls, or pipelines, wherein the fusion welding processes and weldability play an important role. They also contain some Mn, a rather inexpensive element, which provides good solid solution strengthening and grain refinement. Hence, these steels are also called *C-Mn steels*.

The yield strength increases considerably in *fully pearlitic steels*, often used in rail and wire applications. To obtain wires, such steels are subjected to heavy cold deformations in drawing operations. The increase in the dislocation density in the interlamellar regions and the alignment of cementite lamellae in the direction of plastic deformation raises the yield strength to over 3000 MPa (refer to Chapter 14). However, the ductility is drastically reduced, and σ_Y almost coincides with σ_{UTS} .

Micro-alloyed steels

Micro-alloyed steels (also known as high-strength low alloy steels—*HSLA steels*) have varying C content. The lowest values are adopted in ferritic steels for sheet-forming operations, whereas the C content is typically between 0.2% and 0.5% for ferritic-pearlitic steels to be used in civil buildings or in forgings. These steels also contain Mn in the range 0.6% and 1.6%. Above all, they contain low quantities of micro-alloying elements like Nb, V, and Ti (the total amount of such elements is lower than 0.1%). These elements form carbonitrides, which prevent excessive grain growth during austenitization or thermo-mechanical processing (rolling, forging). This allows the formation of fine grain size, in the range of 5–10 μm , after cooling to RT. Carbonitrides also exert precipitation hardening through the Orowan mechanism. In the case of common Nb-based HSLA steels, σ_Y can be estimated with the equations provided previously and by considering the hardening exerted by the Nb-carbonitrides

to be roughly around $3000 \text{ MPa} \times \% \text{Nb}$. By balancing the element content and the grain size, steels with yield strength between 400 and 1000 MPa can be obtained (Gladman, 1997).

Dual phase, TRIP and TWIP steels

These steels were developed by the automotive industry for sheet forming operations, with the aim of obtaining steels with proper plastic deformability and high strength after forming (Llewellyn and Hudd, 1998). Due to this, they are also known as *advanced high-strength steels* (AHSS).

The microstructure of **dual phase** steels comprises ferrite and 5%–20% of martensite. Martensite increases the yield strength to the range of 300–450 MPa, depending on its content. Dual phase steel sheets are generally obtained by hot and cold rolling, followed by intercritical annealing (i.e., annealing in the α - γ region, at a temperature close to 800°C), during which austenitic islands are formed in a ferritic matrix. After annealing, the sheets are rapidly cooled to achieve a duplex microstructure made by hard martensitic islands in a ductile ferritic matrix. Due to the heterogeneous microstructure, these steels do not display the upper and lower yield point phenomenon, but they show a continuous yielding behavior. After plastic deformation, some BH effect is also possibly achieved.

The **TRIP steels** (*TR*ansformation-*IND*uced *PL*asticity steels) contain Mn in the range of 0.2%–2.5% (it can be increased up to 10%), and C in the range of 0.09%–0.4%. They display a multiphase microstructure containing ferrite, bainite without carbides, martensite, and 10%–20% of retained austenite that transforms to martensite during plastic deformation (*TRIP effect*). They are also obtained by hot and cold rolling, followed by intercritical annealing and rapid quenching. However, unlike dual phase steels, the final temperature after quenching is selected to favor the bainitic transformation. The steels contain comparatively high C, Mn, and Si content. Si and Mn inhibit the formation of cementite. The yield strength of TRIP steels is in the range of 400–600 MPa.

Finally, the **TWIP steels** (*TW*inning-*IND*uced *PL*asticity steels) have a C content in the range of 0.15%–0.6%, and high Mn content in the range 15%–25%, favoring the formation of a fully austenitic microstructure (a fully austenitic microstructure is obtained if: $\% \text{Mn} + 13\% \text{C} > 17$; Bouaziz et al., 2011). In these steels, twins form during plastic deformation, if the SFE is sufficiently low to avoid formation of a cellular dislocation structure (this will be explained in the next chapter). The TWIP steel sheets are obtained by hot and cold rolling, followed by recrystallization annealing. The yield strength of TWIP steels typically ranges between 300 and 450 MPa.

Heat-treated steels

High yield strength of 1500 MPa and beyond is obtained in heat-treated (quenched and tempered) alloyed steels with a martensitic microstructure (Krauss, 1980). The high yield strength of steels with untempered martensite or subjected to low-temperature

tempering is due to two main contributions. The first is the interstitial solid solution strengthening exerted by the C atoms trapped in the octahedral sites of the body center tetragonal crystal lattice. The second is the high density of dislocations. For these steels, σ_Y increases with C content, from approximately 1200 MPa at 0.2%C, to 1600 MPa at 0.5%C (Krauss, 2001).

In most cases, heat-treated steels are tempered after quenching. Therefore their microstructure is made of *tempered martensite*. In this case, most of the strengthening effect of C is lost at the expense of increase in ductility. The yield strength is then given by:

- The internal friction (54 MPa);
- The atoms in solid solution (a low contribution);
- The precipitation of carbides;
- The high density of dislocations (up to 10^{12} – 10^{13} cm⁻²);
- The high density of interfaces in martensite.

The morphology of martensite is *lath*-type if C is less than approximately 0.7% wt. It is *plate*-type if C exceeds 1% wt. Finally, it is mixed when C is in intermediate range. Plate martensite is quite irregular. On the other hand, lath martensite is formed in packets, with sizes directly related to the prior austenitic grain size. Yield strength of lath martensite follows a Hall-Petch relationship, with packet size and grain size of the prior austenitic grains (Krauss, 2001). It is observed that the decrease in yield strength with tempering temperature is mainly due to the drop in the dislocation and precipitation strengthening (as the temperature increases, the dislocations are annihilated, making the precipitates larger and more spaced).

Bainitic steels

Bainitic steels display intermediate mechanical properties, between ferritic-pearlitic steels and quenched and tempered steels. Most bainitic steels have C content similar to that of ferritic-pearlitic steels (less than 0.2%, to ensure good weldability) but display better mechanical strength (σ_Y up to 1000 MPa). A typical composition is Fe—0.1C—0.52%Mn—0.11%Si—0.54%Mo—0.0033%B. Note the presence of B, which allows the achievement of a bainitic microstructure by simple normalizing treatment (process involving austenitizing, followed by free air cooling). The microstructure consists of ferrite lamellae, separated by carbides (upper bainite) or ferrite plates containing a fine dispersion of carbides and partly separated by carbides (lower bainite) (Bhadeshia and Christian, 1990). In these steels, the yield strength is due to the internal friction, solid solution strengthening, carbides (if any), and the high dislocation density. Also, bainitic steels, as well as martensitic steels, display a stress-strain curve with an early deviation from linearity. This is due to the high heterogeneity of the microstructure.

Tool steels

Tool steels contain a large amount of C, ranging from 0.35% to 1% or more, which promotes the precipitation of numerous hard carbides (primary and/or secondary). These steels are mainly used in the production of manufacturing tools. Specific steels are developed for cold and hot working and for cutting operations. After a suitable heat treatment, the microstructure of tool steels generally comprises a tempered martensitic matrix with a dispersion of carbides of the added elements. The carbides increase the hardness of the steels, at the expense of ductility and fracture toughness. During mechanical loading, the plastic deformation tends to concentrate early in the metal matrix. If the carbides (intrinsically brittle) are numerous, large, and irregular in shape, their brittle fracture can occur before generalized yielding of the matrix. The fracture may then propagate easily over the entire cross section, following energetically favorable paths offered by the carbides themselves, and is observed to be macroscopically brittle. In this case, σ_Y is possibly evaluated by the means of compression tests (Roberts et al., 1980).

Austenitic and duplex stainless steels

Due to the high Cr content (>15%), austenitic stainless steels are widely used in many engineering applications, owing to their excellent corrosion resistance in different environments. The following experimental relationship is given to evaluate their yield strength (Pickering, 1978):

$$\sigma_Y = 67.8 + 20\%Si + 3.7\%Cr + 14.5\%Mo + 354\%C + 494\%N + \frac{7}{\sqrt{d}} + 2.45\%\delta_{\text{ferrite}} \quad (1.17)$$

In this case, limited solid solution strengthening exerted by the alloy elements (negligible hardening in the case of Ni) and limited strengthening exerted by the grain size refinement can be easily seen. Both effects are due to the fcc structure of the steels, i.e., the presence of a large number of slip systems that allow for an easy dislocations motion. The possibility of having additional hardening due to delta ferrite (δ_{ferrite}) is worth mentioning. Such ferrite can be present, even if in limited quantities, in stainless steels with high Cr content. In general, σ_Y for these steels is comparatively low, in the range of 200–400 MPa. The best way to increase their yield strength is to increase the density of the dislocations through cold working or make use of the martensitic transformation in *metastable* grades (TRIP effect).

The effect of ferrite on the yield strength is amplified in ferritic-austenitic stainless steels (also called *duplex* stainless steels), which have a ferrite content of approximately 50% (Cr content is typically more than 20%, and Ni content less than 5%). The yield strength is in the range of 550–620 MPa. These steels also display excellent corrosion resistance, but are characterized by comparatively poor hot workability.

Table 1.3 summarizes the main kinds of steels along with their typical use and yield strength interval.

Table 1.3 Main kinds of steels, their typical use, and yield strength intervals.

Steel	Main use	σ_Y interval (MPa)
Ferritic steels (annealed)	Sheet forming applications	150–200
Ferritic steels (cold worked)		350 and more
Dual phase, TRIP and TWIP steels	Sheet forming operations	300–450 (up to 600 in TRIP steels)
Ferritic-pearlitic steels	Structural applications such as beams or pipelines	200–400
Pearlitic steels (cold worked)	Wires, rails	Up to 2000
HSLA steels	Sheet forming applications (low-C); structural and mechanical applications (medium-C)	200–800
Heat-treated steels	Mechanical applications	Up to 1400
Bainitic steels	Mechanical applications	200–400
Tool steels	Tools for manufacturing operations	
Austenitic stainless steels	Applications requiring high corrosion resistance; low temperature applications	550–620
Duplex stainless steels	Applications requiring high corrosion resistance	

1.7 Yield strength of selected non-ferrous alloys

Al alloys

The yield strength of Al alloys typically ranges between 100 and 600 MPa. Due to their comparatively low density (approximately 2.7 g/cm^3 with respect to 7.8 g/cm^3 for steels), they display a very attractive strength-to-weight ratio. In general, Al alloys are subdivided into *cast* and *wrought* alloys (Totten and Scott, 2003). Cast alloys typically contain Si or Si-Cu, and in some cases they are heat-treated to increase their strength. Coarse intermetallic compounds that form during solidification have a comparatively small effect on yield strength. Wrought alloys are used in working processes by plastic deformation and may be heat-treatable or non-heat-treatable. In alloys that cannot be heat treated (alloys of classes 1000, 3000, 4000, and 5000), the yield strength depends on the presence of elements in solid solution (in this respect, the addition of Mg in 5000 alloys is very effective) and, greatly, on the density of the dislocations (typically varies between 10^8 cm^{-2} —in annealed alloys—and 10^{10} – 10^{12} cm^{-2} —in cold worked alloys). On the other hand, grain size refining plays a minor role.

In heat-treated alloys undergoing *solution annealing* followed by quenching and *artificial aging* (alloys of the 2000 classes with Cu; 6000 with Mg and Si; and 7000 with Zn and Mg), σ_Y mainly depends on the presence of fine particles, precipitated during aging (typical size is below 0.02 μm). For example, in the case of Al-Cu alloys, with aging, it is possible to induce the formation of dispersed precipitates that are coherent with the crystal lattice (called GP-1 and GP-2 zones) and induction of coherent intermetallic particles (θ' -CuAl₂) with the progression of aging. By appropriately tuning the heat-treatment parameters, i.e., temperature and time, it is possible to optimize the yield strength. An example is shown in Fig. 2.18 of the next chapter, for a 7499 Al alloy. In general, the peak strength is associated with the transition from coherent particles, which can be sheared by dislocations, to incoherent precipitates, which can only be bypassed by dislocations. The use of very high aging temperatures does not allow the precipitation of GP zones, reducing the hardening capability of precipitates, i.e., the peak hardness in aging curves. In some cases, cold working and heat treatment are combined together to achieve the best mechanical properties (T8 treatment).

Table 1.4 shows the typical yield strength values (expressed as $\sigma_{0.2}$) of some Al alloys. The designations are O for annealed alloys, H for strain-hardened alloys, and T for solution-treated alloys. Alloys in the annealed condition, such as the 5052-O alloys, are commonly used in sheet metal forming operations. The 6061 and 6063 are commonly used for producing extrudates and hot forged parts. The T6 treatment (solution quenching followed by artificial aging) is carried out after the shaping process.

Cu alloys

Commercially (polycrystalline) pure Cu with fcc structure is used in applications where high electrical and thermal conductivity is required, such as in electrical wires. The yield strength is quite low, in the range 20–50 MPa, depending on the elements present as impurities. The yield strength of Cu can be increased by increasing the dislocation density, i.e., by cold working, and/or by adding alloying elements. For example, with a cold drawing of 25%, the yield strength of Cu can be increased to 340 MPa. However, in both cases, the electrical and thermal conductivities are decreased (Renwick, 2015).

Of relevance are the Cu-Zn alloys, known as *brasses*: α -brasses contain up to 36% of Zn in solid solution, and retain the fcc structure of pure Cu. With the addition of 10% or 30% of Zn, σ_Y rises to 70 and 75 MPa, respectively, with a slight dependency on the grain size. Cold working is clearly required to achieve higher σ_Y values. The alloys with Zn in the range 36%–45% are known as α - β brasses, as they also contain some β phase, which undergo ordering when temperature is below 470°C. The ordered β' phase, for which the atoms in solid solution are not randomly present but prefer to have solvent atoms as neighbors, has a bcc superlattice. For example, the α - β alloy containing 40% Zn displays a yield strength of 214 MPa in the annealed condition. Due to the limited deformability of β -phase (bcc structure) at low temperature, binary brasses must be hot worked.

Table 1.4 Yield strength of selected Al alloys.

Name (NHT)	State	Main elements	σ_y (MPa)	Name (HT)	State	Main elements	σ_y (MPa)
1100-O	Annealed	0.12Cu	35	2014-T6	Quenching and artificial aging	4.4Cu-0.8Si-0.8Mn-0.5Mg	414
1100-H18	Very strain hardened	0.12Cu	150	6061-T6	Quenching and artificial aging	1Mg-0.28Cu-0.6Si	275
3003-O	Annealed	0.12Cu-1Mn	25	6063-T6	Quenching and artificial aging		215
5052-O	Annealed	2.5Mg-0.25Cr	90	7075-T6	Quenching and artificial aging	2.5Mg-1.6Cu-0.2Mn	500
5052-H38	Strain hardened	2.5Mg-0.25Cr	255				

NHT, non-heat-treatable alloys; *HT*, heat-treatable alloys. The designations are O for annealed alloys, H for strain-hardened alloys, and T for solution-treated alloys. Data from Polmear, I., St John, D., Nie, J-F., Qian, M., 2017. Light Alloys, Metallurgy of the Light Metals, fifth ed. Butterworth-Heinemann.

Further increase in σ_Y can be obtained with the addition of other elements in solid solution, such as Sn, Al, and Ni. Of interest are *Cu-Be alloys*, with Be in the range 2–3 wt%. Very high yield strength values can be achieved through solution quenching and aging treatment, with the precipitation of γ -phase, following a precipitation sequence like that encountered in Al alloys (i.e., coherent GP zones, γ' , and incoherent γ). The yield strength can be increased up to 1200 MPa and beyond.

Ti alloys

Ti alloys have low density (approximately 4.5 g/cm^3) and are characterized by very high strength-to-weight ratio, more than Al alloys and steels (Leyens and Peters, 2003). They are increasingly being used in the aerospace, chemical, energy, and biomedical industries (Gialanella and Malandrucolo, 2020). At RT, pure Ti has an hcp structure (called α -phase), and a yield strength in the range 170–280 MPa, depending on the interstitial impurities, such as O, N, and C. If the O_{eq} , given by: $O+N+C$, is around 1%, σ_Y rises to 600 MPa because of intense solution hardening. The structure changes into a bcc phase, called β -phase, at 883°C .

The yield strength of Ti alloys can be increased by alloying additions and heat treatment. The addition of elements such as C, O, N and, most of all, Al stabilizes the α -phase (they increase the alpha transus, i.e., the temperature below which the alloys have an α structure), and increases the yield strength by solid solution hardening. Al, Sn, and Zr display a strong hardening effect, and in these alloys, σ_Y values up to 900 MPa can be typically attained in annealed conditions. These alloys are classified as α alloys.

The addition of Mo, V, and Cr stabilizes the β phase, since these elements reduce the alpha transus and also the beta transus (i.e., the temperature above which the material has a β structure). By a correct addition of such elements, the $\alpha+\beta$ alloys with a dual phase microstructure can then be obtained. The yield strength of the very common Ti-6 wt%Al-4 wt%V alloy in the annealed state (often code-named *Ti-64*) is typically in the range 850–950 MPa. The yield strength of the α - β alloys can be further increased with quenching and aging treatment. The addition of high amounts of β -stabilizing elements (e.g., Mo, Nb, Ta) promotes the retention of a fully metastable β -structure at RT and it further gives rise to a Ti alloy group, which is of β -alloys.

Mg alloys

The engineering interest in Mg alloys is mainly stimulated by their very low density, which is approximately 1.74 g/cm^3 . Mg has an hcp structure with only three slip systems on the basal plane, leading to a very low inclination to plastic deformation. Pure polycrystalline Mg shows a yield strength of approximately 50–100 MPa. At RT (and below 200°C), dislocation motion mainly occurs on the basal planes because of the low critical shear stress (Table 1.2). However, slip systems are few and the plastic deformation is rather difficult in polycrystalline metal. It involves only the properly

oriented grains. In addition, Mg alloys display an asymmetry in the yield behavior and the yield strength under compression is lower than under tension, due to the incidence of twinning, as explained in the next chapter (Pekguleryuz et al., 2013).

Yield strength may be increased through solid solution strengthening, with the addition of elements such as Al and Zn that strongly increase the critical resolved shear stress and, in some cases, through precipitation hardening. Mg alloys are mainly produced by casting processes, and mechanical properties strongly depend on the (complex) solidification microstructure (porosity, large and variable grain size). Typically, σ_Y for cast alloys ranges between 75 and 150 MPa. In wrought alloys, σ_Y typically varies between 170 and 300 MPa. The most common alloys, also used in wrought conditions, are based on the Mg-Al-Zn system, such as the AZ31 alloy.

Fig. 1.13A summarizes the typical yield strength intervals of selected alloys of particular application interest. Fig. 1.13B shows the typical yield strength interval of the same alloys in the annealed condition, which is the starting microstructural state at which most plastic forming operations are carried out.

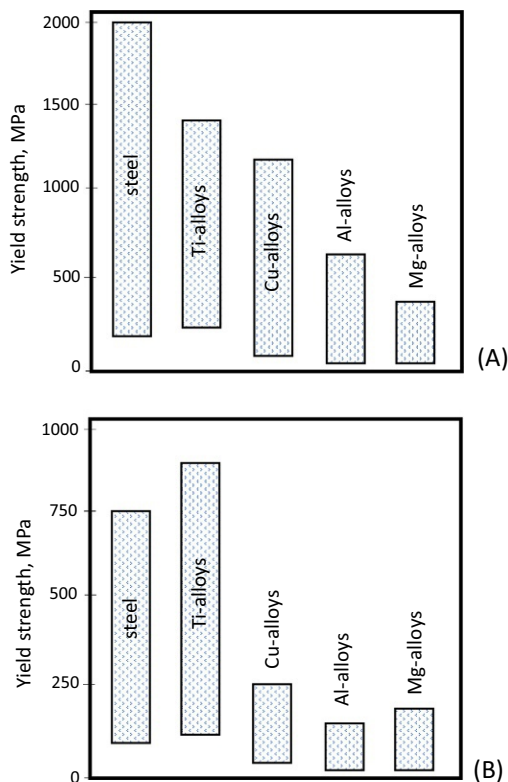


Fig. 1.13 (A) Typical yield strength intervals for selected alloys; (B) typical yield strength intervals of the same alloys in the annealed condition (the data are only indicative).

1.8 Evaluation of yield strength of metals from hardness tests

Hardness tests are widely used to quickly evaluate the yield strength of materials. These tests are employed because they are simple to perform and involve a small part of the material, which makes them a non-destructive procedure. The most widely used indenters have a spherical (Brinell hardness, HB) or pyramidal (Vickers hardness, HV) shape. In both cases, the hardness, H , is given by the ratio of the applied load and the size of the impression left in the metal by the plastic indentation. It is expressed in kg/mm^2 .

The yield strength is then evaluated using the following relationship:

$$\sigma_Y = \frac{H}{A_V} \quad (1.18)$$

where H is the Vickers or Brinell hardness (to be converted into MPa), and A_V is a material constant. Using the slip line method and with reference to materials with perfectly plastic, i.e., non-hardening, behavior, it is seen that A_V is between 2.9 and 3. From experimental tensile tests for Al alloys, it was determined that $A_V = 2.95$ (Hawk et al., 1988). A marked dependence on microstructure was found in steels (Umemoto et al., 2001):

- $A_V = 5.79$ for ferrite;
- $A_V = 4.9$ for pearlite;
- $A_V = 4.27$ for bainite;
- $A_V = 3.17\text{--}4.2$ for tempered martensite.

In fact, the indentation used in Vickers or Brinell hardness tests results in high plastic strains beneath the indenter, typically between 0.25 and 0.35 (Chaudhri, 1998), and then the hardness values are largely affected by material strain hardening behavior.

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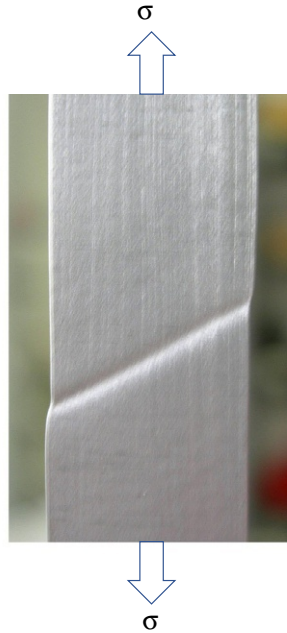
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Strain hardening of metals

2



Localized necking in a 5182 Al alloy sheet deformed under uniaxial tensile loading.

Chapter outline

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With the overcoming of the yield strength, the metal begins to strain harden and the applied stress must be continuously augmented to increase the plastic strain. In the present chapter, the strain-hardening behavior of metals in the reference conditions (room temperature—RT, + quasi-static loading) is presented. The discussion will be focused on the evolution of the dislocation configurations and the contribution of twinning, considering the role of geometrically necessary dislocations (for bcc metals) and stacking faults (for fcc metals). The analysis is based on the Ludwik-Hollomon equation and in particular on the meaning of the strain-hardening coefficient, n , making reference to a simplified approach able to capture the involved microstructural phenomena. The strain-hardening coefficient is a very useful parameter for understanding the ductile behavior of metals, including the ductile fracture mechanism, as will be outlined in Chapter 6.

The role of the different crystal structures is then considered, with particular reference to bcc and fcc steels, and selected non-ferrous alloys. The interplay of the strain evolution with possible microstructural transformations is discussed, showing how optimized transformations can be used for improving the strain-hardening behavior of metals and the corresponding uniform strain, while maintaining a satisfactory yield strength.

The strain-hardening evolution in sheet specimens with rectangular cross section is introduced at the end of the chapter, showing the peculiarities of the plastic deformation in this kind of specimens. This information is very important to characterize the materials for sheet metal forming operations.

2.1 Dislocation configurations and contribution of twinning

During plastic deformation, new dislocations are continuously generated and stored. These dislocations are observed to exert a mutual slowdown as they trap each other, forming dislocation *forests* or *tangles*. Since the interaction among dislocations and their generation are statistical processes, the dislocations that form and accumulate in this way are called *statistically stored dislocations*. However, the total density of dislocations, ρ , is given by the sum of two contributions: the density of the statistically stored dislocations, ρ_s , and the *excess* or *geometrically necessary dislocations*, ρ_G (Fleck et al., 1994):

$$\rho = \rho_G + \rho_s \quad (2.1)$$

Geometrically necessary dislocations are formed to accommodate local plastic deformations at the microstructural heterogeneities, which are typical in engineering metals (Fig. 1.8A), or to accommodate strain gradients due to the geometry of loading, as in bending of a beam. Fig. 2.1 shows the geometrically necessary dislocations around small cementite particles in a low C steel (Gawne and Lewis, 1985).

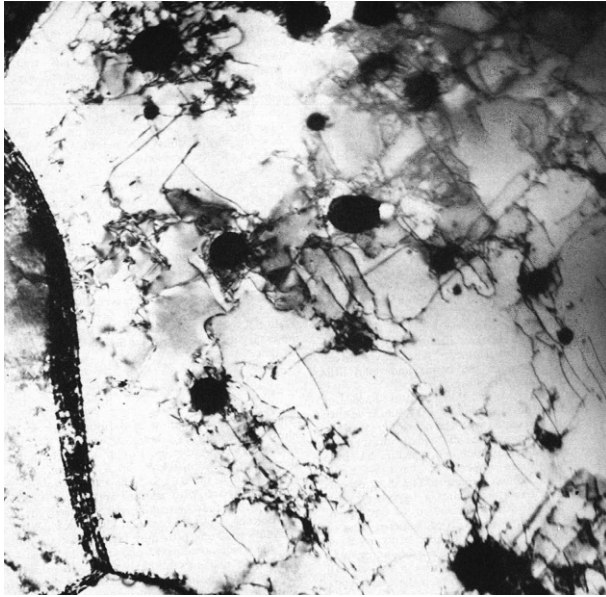


Fig. 2.1 Low carbon steel with geometrically necessary dislocations around cementite particles, obtained using TEM at 20,000 $\times$.

Reproduced from Gawne, D.T., Lewis, G.M.H., 1985. Strain-hardening of high-strength steels. *Mater. Sci. Technol.* 1, 128–135, with permission of the Publisher.

The density of geometrically necessary dislocations is proportional to the strain gradients, i.e., to the ratio between the macroscopic average true plastic strain, ϵ_p , and a length scale, λ_g . λ_g is representative of the size of the region, where the gradient of plastic deformation due to the microstructural heterogeneity is present (Fleck et al., 1994; Courtney, 2000):

$$\rho_G \cong \alpha \cdot \frac{\epsilon_p}{b \cdot \lambda_g} \quad (2.2)$$

where α is a constant and b is the modulus of the Burgers vector. For example, in the case of pure polycrystalline metals, λ_g is proportional to the average grain size, while in a metal containing a dispersion of particles, λ_g is proportional to the average inter-particle spacing (see Section 6.2). In pure metals with a sufficiently large grain size, ρ_G is small and $\rho_s \gg \rho_G$ at each plastic strain. However, engineering metals with a heterogeneous microstructure are characterized by large local plastic strain gradients during plastic deformation, and ρ_G is much greater than ρ_s . Therefore $\rho \approx \rho_G$. In engineering metals, the density and configuration of dislocations depends on plastic strain and on the microstructural characteristics.

In fcc and hcp metals, a dislocation can split during plastic deformation and form two *partial dislocations*, with a modulus of the Burgers vector lower than that of a perfect dislocation. The stress fields of these new dislocations interact, and the two dislocations move away from one other to reach an equilibrium separation. In this way, they produce *faults* in the stacking sequence of closely packed atomic planes (Hull and Bacon, 2011; Kovacs and Zsoldos, 1973). As shown in Fig. 2.2, a *stacking fault* between two partial dislocations can be considered as a lattice region, where a part of the close-packed plane is removed. The elastic energy associated with the stress fields of the two partial dislocations is called *stacking fault energy* (SFE). It is inversely proportional to the length of the fault plane, L , i.e., the distance between the two partial dislocations (Hull and Bacon, 2011):

$$\text{SFE} = \frac{G \cdot b^2}{4\pi \cdot L} \quad (2.3)$$

where G is the shear modulus. This model is also used to evaluate the SFE of a material. Other methods have been developed, based on thermodynamic and physical models (Hirth, 1970; Lee et al., 2010). The experimental data are typically between 1 and 1000 mJ/m² and the SFE of a metal depends on alloying content and increases with temperature.

Depending on the SFE, two distinct situations can be recognized concerning plastic deformation.

Fcc and hcp metals with comparatively **high SFE** form very small stacking faults (low L values, Fig. 2.2), and stacking faults are not expected in metals with a bcc structure. In all these alloys, dislocations are rather free to move. With increasing plastic strain, dislocation density increases and dislocations form forests and tangles. Subsequently dense dislocation walls are formed. They are the dislocation

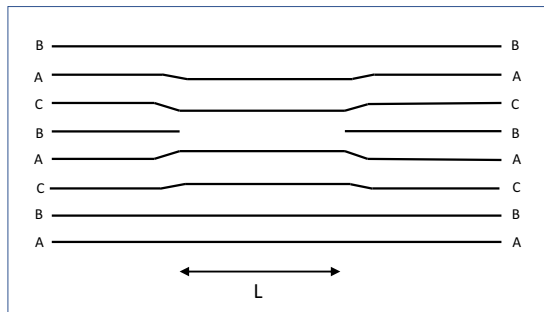


Fig. 2.2 The stacking sequence in an fcc structure, with the presence of an intrinsic stacking fault with length L . A, B, and C indicate the close packed planes with the three different possible positions of the atoms. The fault is surrounded by two partial dislocations, whose modulus of Burgers vector is lower than that of the corresponding perfect dislocation.

Modified and redrawn from Abbaschian, R., Abbaschian, L., Reed-Hill, R.E., 2009. Physical Metallurgy Principles. Cengage Learning.

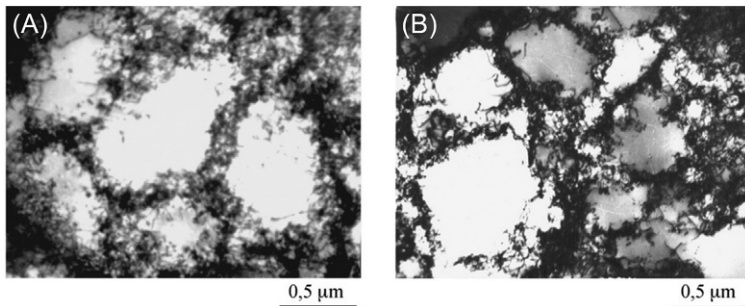


Fig. 2.3 Examples of dislocations with a cellular morphology observed in a polycrystalline Ni_3Fe alloy (A) and in Ni single crystals (B).

Reproduced from Koneva, N.A., Starenchenko, V.A., Lychagin, D.V., Trishkina, L.I., Popova, N.A., Kozlov, E.V., 2008. Formation of dislocation cell substructure in face-centred cubic metallic solid solutions. *Mater. Sci. Eng. A* 483–484, 179–183, with permission of the Publisher.

grain boundaries of *cell substructures* within the original grains under deformation; two examples are shown in Fig. 2.3. The interior of the cells has a very low density of dislocations. The average cell diameter may decrease with plastic strain. However, at a certain strain, *saturation conditions* are attained, with the formation of closely spaced dense dislocation walls, followed by micro-shear localizations. In this way, dislocation *microbands* are formed that are superimposed on the existing dislocation structure and often extend along the grain width. Further straining increases the density of micro-bands. In the micro-bands the dislocations slip can locally accommodate large strains. In addition, the dislocations within the micro-bands can experience annihilation (Thuiller and Rauch, 1994), a phenomenon that is favored by a local temperature rise. This introduces a local softening that results in a decrease in the strain-hardening rate with the progress of plastic deformation (Gawne and Lewis, 1985). It is also argued that the high stored elastic energy, due to the very high dislocation density, induces the dislocations in the walls/micro-bands to move easily, facilitating the overcoming of energy barriers without the necessity of a large increase of the applied shear stress. In this respect, it has been proposed that screw dislocations are easily released from the cell walls into the interior of the cells, breaking the original structure and forming elongated cell structures. This process too is able to accommodate large strains (He et al., 2017). To sum up, after the attainment of saturation conditions, a reduction in the strain-hardening rate (i.e., a *dislocation softening*) takes place.

In fcc metals with a **low SFE** (below approximately 45 mJ/m^2), the size of the stacking faults, L , is rather large. At low strains, slip is then confined on a single slip plane and hence dislocations typically move by *coplanar slip* (Kuhlmann-Wilsdorf, 1989). In addition, the high L values limit cross-slip of screw dislocations, since for cross-slip to occur, the two partial dislocations must recombine into a perfect dislocation.

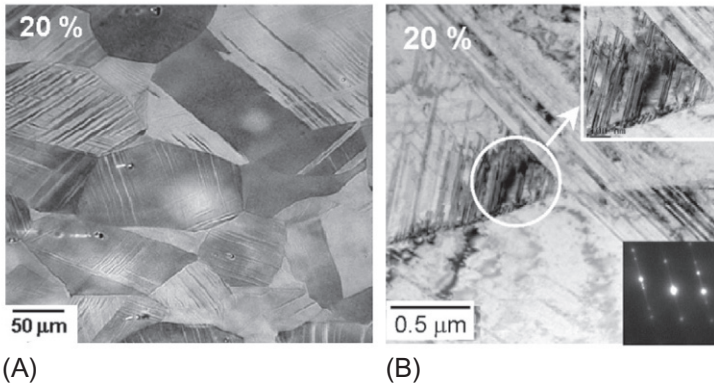


Fig. 2.4 Examples of twins formed in an Fe-0.6%C-22%Mn steel after tensile straining of 20%; (A) is an optical micrograph of the deformed microstructure, whereas the micrograph in (B) was obtained with a transmission electron microscope (TEM).

Reproduced from Park, K.T., Jin, K.G., Han, S.H., Hwang, S.W., Choi, K., Lee, C.H., 2010. Stacking fault energy and plastic deformation of fully austenitic high manganese steels: effect of Al addition. *Mater. Sci. Eng. A* 527, 3651–3661, with permission of the Publisher.

With increasing strain, dislocation arrangement becomes random and the density of stacking faults increases. As a result, the deformation by slip becomes more difficult. Since the critical shear stress for twinning may be comparatively low, as it is generally proportional to the SFE, deformation by twinning gains in importance, occurring together with dislocation slip. In Fig. 2.4, the twins formed in an Fe-0.6%C-22%Mn steel (with an SFE of 21.5 mJ/m²) after tensile straining of 20% are shown (Park et al., 2010). In fcc alloys with a particularly low SFE (below approximately 15 mJ/m²), the strain-induced transformation of austenite into martensite becomes possible during plastic deformation (the transformation mechanism is rather similar to twinning).

In Table 2.1, some examples of metals with fcc structure are reported, along with their SFE values and the typical observed dislocation configurations after straining. In Table 2.2, the role of SFE on the development of plastic deformation of metals is summarized.

2.2 Strain-hardening parameters

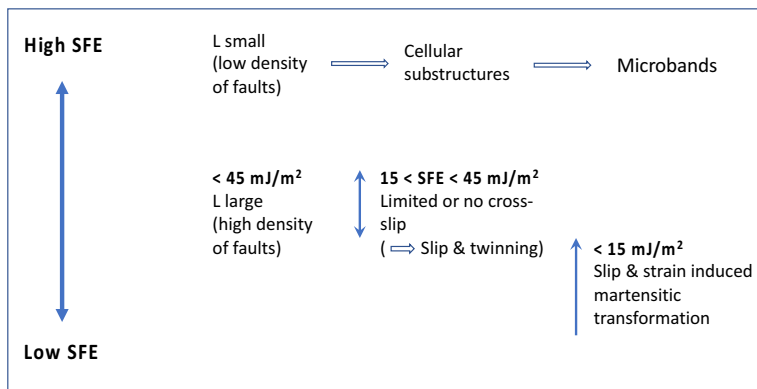
Plastic deformation starts when the applied stress is equal to σ_Y . To increase the plastic deformation, the stress must be continuously increased due to the strain-hardening phenomenon, correlated with the motion, multiplication, and interaction of dislocations. Several experiments have demonstrated that the flow stress, σ , depends on the total density of dislocations through the following relationship:

$$\sigma = \sigma_0 + k_1 \cdot \sqrt{\rho} \quad (2.4)$$

Table 2.1 Typical SFE values and observed dislocation configuration in fcc metals.

Metal	SFE (mJ/m ²)	Dislocation configuration
Al alloys	220–350	Cells
Ni	200	Cells
Ni-40%Co	130	Cells
Cu	80	Cells
Cu-8%Zn	30	Cells
Cu-16%Zn	23	Cells
Cu-2%Al	37	Cells
Cu-24%Zn	16	Coplanar slip
Cu-4.5%Al	7	Coplanar slip
Cu-8%Al	4	Coplanar slip
304 SS	25	Coplanar slip
316 SS	15	Coplanar slip

Data from Rohatgi, A., Vecchio, K.S., Gray, III G.T., 2001. The influence of stacking fault energy on the mechanical behavior of Cu and Cu-Al alloys: deformation twinning, work hardening, and dynamic recovery. *Metall. Mater. Trans. A* 32, 135–145, and other sources.

Table 2.2 Role of SFE on the plastic deformation of metals.

where σ_0 and k_1 are material constants. The influence on the flow stress of microstructural features such as grain boundaries and precipitates can be together accounted for by the concept of geometrically necessary dislocations. As already highlighted, these microstructural features induce formation of geometrically necessary dislocations (refer to Figs. 1.11 and 2.1) that increase ρ and then σ . Therefore, combining Eqs. (2.2) and (2.4) (after setting: $\rho \cong \rho_G$), the contribution of geometrically necessary dislocations to the flow strength is obtained (expressed at the level of continuum

plasticity, i.e., considering the macroscopic average strains; Kubin and Mortensen, 2003):

$$\sigma = \sigma_0 + k_2 \cdot \sqrt{\varepsilon_p} \quad (2.5)$$

where k_2 is a material constant. As mentioned in Section 1.1, when plasticity is fully developed, $\varepsilon_p \cong \varepsilon$. In this case, it is more convenient to use the concept of true strain, ε ; this is done in the following equations.

The dependence of the flow stress on plastic strain is then parabolic. This is a good approximation in case of stable fcc metals, but for bcc metals dislocation motion is strongly influenced by the lattice friction and the picture changes. In addition, during plastic deformation dislocations may form special structures, deformation by twinning gains in importance, and the microstructure may also experience phase transformations. As a result, the exponent of Eq. (2.5) assumes experimental values other than 0.5, which may also change with plastic deformation, as will be outlined in the next sections. Therefore, from generalizing Eq. (2.5), we obtain:

$$\sigma = \sigma_Y + K \cdot \varepsilon^n \quad (2.6)$$

where:

- K is the *strength coefficient*
- n is the *strain-hardening coefficient* (or *strain-hardening exponent*).

Eq. (2.6) is called the *Ludwik equation* and is often simplified in the following manner:

$$\sigma = K \cdot \varepsilon^n \quad (2.7)$$

Eq. (2.7) is known as the *Hollomon equation* or the *Ludwik-Hollomon, L-H, equation*. It clearly underestimates the flow stress at small ε values, since σ tends to 0 when ε tends to 0. However, the equation works well when the curves display a smooth transition from the initial elastic and the plastic part (refer to Section 1.3) and, most of all, it is suitable to model the strain-hardening behavior of metals in the uniform region.

The fitting parameters of Eq. (2.7) are easily obtained from the σ - ε curves, with reference to the region of uniform plastic deformation, and after plotting the curve in a log-log scale. Refer to the schematization in Fig. 2.5A. K is obtained by extrapolating to $\varepsilon = 1$. Clearly, in the case of metals with particularly low ductility, reference to the true plastic strains should be made in Eq. (2.7) (after subtracting the elastic portion of the strain from the true strain), to avoid relevant errors in the evaluation of K and n . The data obtained from tensile tests on cylindrical or rectangular specimens and compressive tests from cylindrical specimens almost coincide. Particular asymmetries will be discussed later in this chapter.

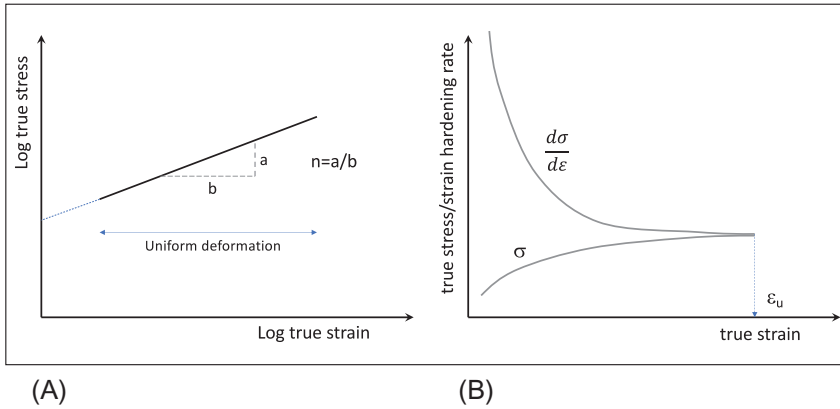


Fig. 2.5 (A) Determination of the average n -value in the uniform deformation region; (B) theoretical condition for necking.

In the following analysis, Eq. (2.7) is utilized with the relevant parameters K and n , to discuss the strain-hardening behavior of different engineering metals. In fact, Eq. (2.7) is simple and allows for some interesting evaluations concerning plastic deformability, referring to the meaning of the strain-hardening coefficient n .

First, taking the first derivative of Eq. (2.7), we obtain:

$$\frac{d\sigma}{d\epsilon} = K \cdot n \cdot \epsilon^{n-1} = n \cdot \frac{\sigma}{\epsilon} \quad (2.8)$$

and thus:

$$n = \frac{\epsilon}{\sigma} \cdot \frac{d\sigma}{d\epsilon} \quad (2.9)$$

which shows that at a given stress and strain level, n is directly proportional to the strain-hardening rate of a metal, expressed by: $d\sigma/d\epsilon$.

Second, a mechanistic model indicates that n coincides with the true strain at the onset of necking during a tensile test (*Considère criterion*). Such a strain is called the uniform strain, ϵ_u , and it is a measure of the ductility of the material. In fact, necking occurs when the reduction in the resisting cross section, following uniform plastic deformation, induces a decrease in the applied force, F (required to maintain the applied deformation velocity) that is not counterbalanced by strain hardening, such that $dF = 0$ (see Fig. 1.1C). Since $F = \sigma A$, at necking: $dF = \sigma dA + A d\sigma = 0$. Considering $d\epsilon = -dA/A$, we get:

$$\frac{d\sigma}{d\epsilon} = \sigma \quad (2.10)$$

and comparing with Eq. (2.8), we finally obtain $n/\epsilon = 1$, and therefore $\epsilon_u = n$. Fig. 2.5B shows the progressive decrease of the strain hardening rate ($d\sigma/d\epsilon$) with the increase in the plastic strain, and the condition for necking based on the L-H equation. It is clearly shown that ϵ_u (and then n) is low when the strain-hardening rate ($d\sigma/d\epsilon$) decreases quickly with plastic strain. Necking limits the plastic tensile deformation of a metal and favors ductile fracture, as will be seen in Chapter 6. Therefore metals with a high value of n are more ductile, as they can delay necking (and fracture) to higher strains.

All these observations imply that coefficient n should be considered as an important measure of the plastic deformability of a metal.

If the deformation mechanism changes *during* plastic deformation, coefficient n is not constant but also changes following the variations in $d\sigma/d\epsilon$. In this case, reference to the average n value should be made, and the concept of *instantaneous* n value is very useful. Eq. (2.9) can be used to estimate it at a given level of strain (or stress; say: ϵ' and σ'), after calculation of $d\sigma/d\epsilon$ from the true stress-true strain curve. The procedure is outlined in Fig. 2.6.

A useful parameter that expresses the ability of a material to undergo strain hardening after the onset of yielding is the ratio: s_{UTS}/σ_Y . Considering that at necking $\epsilon = n$, we can write: $\sigma_{UTS} = K \cdot n^n$ (where σ_{UTS} is the true stress at necking), and then:

$$s_{UTS} = \frac{\sigma_{UTS}}{\epsilon^n} = \frac{K \cdot n^n}{\epsilon^n} \quad (2.11)$$

Setting $\sigma_Y = \sigma_{0.2}$ (Section 1.2): $\sigma_Y = K \cdot 0.002^n$, and then:

$$\frac{s_{UTS}}{\sigma_Y} = \left(\frac{n}{0.0054} \right)^n \quad (2.12)$$

Hence, the ratio of s_{UTS}/σ_Y is directly correlated with n , and increases with it. The dependency of n with the ratio of s_{UTS}/σ_Y is graphically shown in Fig. 2.7. This

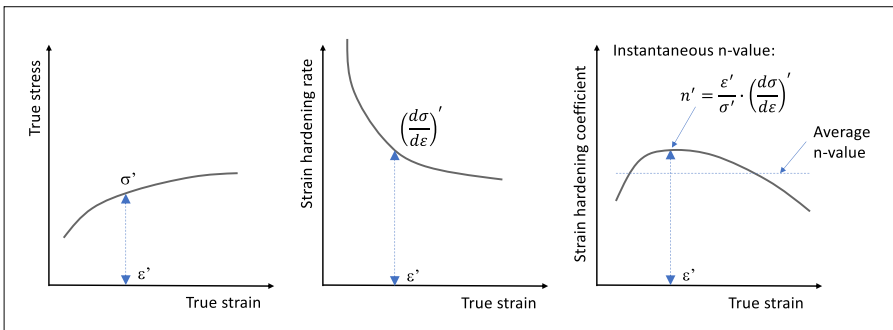


Fig. 2.6 Procedure for the determination of the instantaneous n value during plastic deformation. The $d\sigma/d\epsilon$ curve must be obtained from the σ - ϵ curve (Fig. 2.5B).

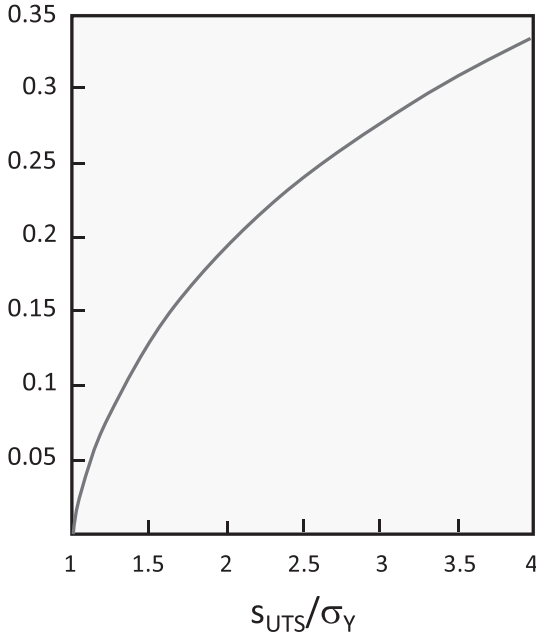


Fig. 2.7 Variation of the strain-hardening coefficient, n , with the yield ratio (a simple fit is: $n = 0.025 + 0.53 \cdot \log(s_{UTS}/\sigma_Y)$).

diagram is a simple way to obtain n from s_{UTS} and σ_Y , when the stress-strain curves approximately follow the L-H equation (Rajendran et al., 2014). Knowing n and σ_Y , from Eq. (2.11), it is also possible to estimate K .

2.3 Strain hardening of bcc steels

To understand the role of microstructural features on the experimental strain-hardening behavior (and then on n) of bcc steels, the interplay between the strain hardening realized by the geometrically necessary dislocations and the phenomenon of dislocation softening described in Section 2.1 must be considered.

As the intensity of dislocation softening increases, the intensity of the strain-hardening rate, $d\sigma/de$, decreases, and this is accompanied by a decrease in n . The role of dislocation softening can be rationalized using a very simple model that considers the rate of accumulation of geometrically necessary dislocations, which is a function of the microstructural parameter λ_g (Gawne and Lewis, 1985). In fact, a decrease in λ_g induces a marked increase in the dislocation density as strain increases, and this makes the local attainment of *dislocation softening* easier, also increasing its intensity and favoring achievement of comparatively low n values.

Let us consider first the case of *ferritic steels* (in the annealed condition). As seen in Section 1.4, their tensile stress-strain behavior is typically characterized by the upper and lower yield point phenomenon and the subsequent appearance of the Lüders

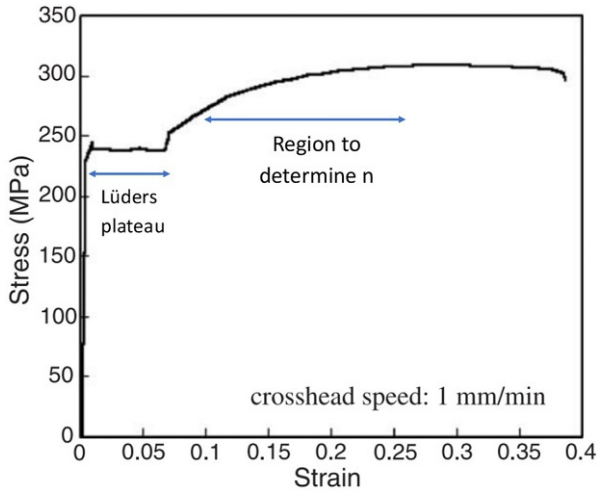


Fig. 2.8 Engineering stress-strain curve for a mild ferritic steel, containing 0.036%C. The typical region used for the determination of n is indicated. Reproduced from Sun, H.B., Yoshida, F., Ohmori, M., Ma, X., 2003. Effect of strain rate on Lüders band propagating velocity and Lüders strain for annealed mild steel under uniaxial tension. *Mater. Lett.* 57, 4535–4539, with permission of the Publisher.

bands. As indicated in Fig. 2.8, the strain-hardening coefficient, n , must then be determined (following the procedure described in Fig. 2.5A), making reference to the uniform deformation region, i.e., between the end of the Lüders plateau and the necking point. The average n values for this kind of steel are typically in the range: 0.18–0.25. They are mainly influenced by the average grain size, d , and the following experimental relationship has been proposed (Morrison, 1966):

$$n = \frac{5}{10 + d^{-1/2}} \quad (2.13)$$

where d is the average grain size in millimeters (Eq. 2.13 is also called the *Morrison equation*). For example, for $d = 15 \mu\text{m}$, $n = 0.27$, and for $d = 10 \mu\text{m}$, $n = 0.25$. In ferritic **IF steels** (that do not display the Lüders plateaus) the n values are higher, typically in the range 0.25–0.34. Experimental values in the range 0.29–0.37 were also obtained for some IF steels stabilized with Ti (Antoine et al., 2006). These steels had average grain sizes in the range $38.1\text{--}26.7 \mu\text{m}$, and the n values predicted from the Morrison equation are in good agreement with the experimental data, as shown in Fig. 2.9 (the experimental data are indicated with open circles, whereas the predicted values are shown with filled circles).

The influence of the grain size, d , on n can be explained considering the proposed simplified model. In fact, parameter λ_g is proportional to d , and therefore it decreases as d is decreased. Consequently, the average n is expected to decrease with decreasing

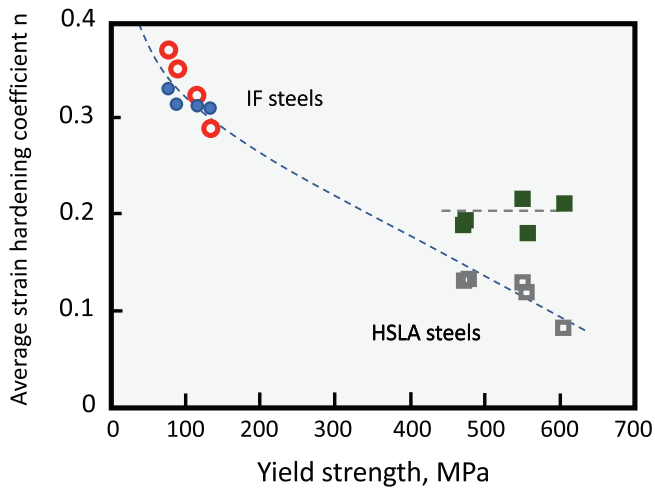


Fig. 2.9 Experimental dependence of n vs yield strength for two ferritic steels: IF steels stabilized with Ti (data from Antoine, P., Vandeputte, S., Vogt, J.-B., 2006. Empirical model predicting the value of the strain-hardening exponent of a Ti-IF steel grade. *Mater. Sci. Eng. A* 433, 55–63), and low-C HSLA steels (data from Silva, R.A., Pinto, A.L., Kuznetsov, A., Bott, I.S., 2018. Precipitation and grain size effects on the tensile strain-hardening exponents of an API X80 steel pipe after high-frequency hot-induction bending. *Metals* 8, met8030168). The experimental data are represented by *open symbols*, whereas the *filled symbols* are the corresponding data obtained using the Morrison equation.

d , as experimentally observed, since the saturation conditions are realized earlier and with higher intensity (during plastic deformation) as the grain size is decreased.

Fig. 2.9 also shows the results for ferritic low-C HSLA steels, containing Nb and Ti and then hardened by the precipitation of fine carbonitrides (Silva et al., 2018). The experimental n values are reported with open squares. They are seen to decrease with increasing the yield strength. The corresponding n values calculated using the Morrison equation are shown with filled squares (the grain size is in the range 3.2–5.7 μm). In this case, the predicted n values are always greater than the corresponding experimental values (and are not proportional to the yield strength). This behavior can be explained considering that, as seen in Section 1.5, fine incoherent particles in the microstructure induce an increase, during plastic deformation, in the density of geometrically necessary dislocations, due to the Orowan mechanism (Fig. 1.11). Therefore the precipitates realize a decrease in λ_g , and hence induce an increase in the yield strength and, at the same time, a decrease in n . In other words, the strain-hardening behavior of these steels, and also the experimental n values, are determined by the fine precipitates, characterized by an interparticle spacing that is much lower than the grain size.

This result is rather general: all finely dispersed incoherent precipitates display a prevailing influence on n , inducing a decrease in n . In this regard it should be noted

that limited effects are exerted by microstructural particles that are large and widely dispersed in the microstructure. The corresponding λ_g value is in fact too big to influence the strain hardening rate and then n (one example is Mn-sulfides, refer to [Section 7.1](#), with mean spacings typically in excess of 80 μm).

In *pearlitic and ferritic-pearlitic steels*, the average n values decrease with decreasing grain size, extension of the pearlite nodules, and pearlite interlamellar spacing, as all these parameters induce a decrease in λ_g .

Very low values of n are displayed by *cold-worked ferritic steels*. In these steels, the density of dislocations is very high, and the dislocations are arranged in cells with dense dislocation walls, depending on the degree of cold working. During tensile deformation, intense dislocation softening is attained very early (from the beginning of the test), and the strain hardening is correspondingly low. Therefore these metals are characterized by very high values of the yield strength and very low values of n . For example, low-C steels strain hardened to approximately 300 MPa show n values less than 0.05. This behavior is actually displayed by all metals in the cold worked condition. Typically, the n values are below 0.1, depending on the degree of cold working.

A similar behavior is displayed by *martensitic* (quenched and tempered) *steels*. In these steels, the density of dislocations is very high and many submicrometric precipitates are present. Hence, intense dislocation softening occurs from the beginning of plastic straining and the n values are typically below 0.1. Different behavior is displayed by *as quenched* or *low temperature tempered martensitic steels* (tempering $T < 200^\circ\text{C}$). These steels with medium-high C content display a continuous yielding behavior due to the presence of retained austenite that begins to deform at low stress. Consequently, these steels are characterized by n values that are approximately 0.2–0.25 at the beginning of the deformation, since deformation localizes in the austenitic areas, and then continuously decreases to very low values as plastic deformation spreads into the martensitic regions.

A distinct behavior is displayed by porous metals. [Fig. 2.10](#) shows the dependence of n on volume fraction of porosity for *porous iron* obtained by powder metallurgy ([Straffelini, 2005](#)). Coefficient n decreases with increasing porosity fraction and with increasing the angularity of pores, expressed by the K_p coefficient (K_p increases with increasing the angularity of pores). This behavior is partly due to a decrease in λ_g as porosity is increased, but mainly to a decrease in the load-bearing section during straining, due to the large increase in volume of the pores (this phenomenon is strongly influenced by K_p). This last effect is called the *geometrical softening effect*. The role of this contribution is confirmed by the results of compression tests, which shows an increase in n in the tests for materials with the same porosity, since compression promotes pore closure during deformation ([Kuhn and Ferguson, 1990](#)).

This last observation allows for obtaining a general concept. In engineering metals, it is rather common that n measured under compression is higher than under tension. This is because engineering metals may contain pores and/or soft phases that introduce a geometrical softening effect during tensile deformation (that induces a decrease in n) and a hardening effect during compression (that induces an increase in n).

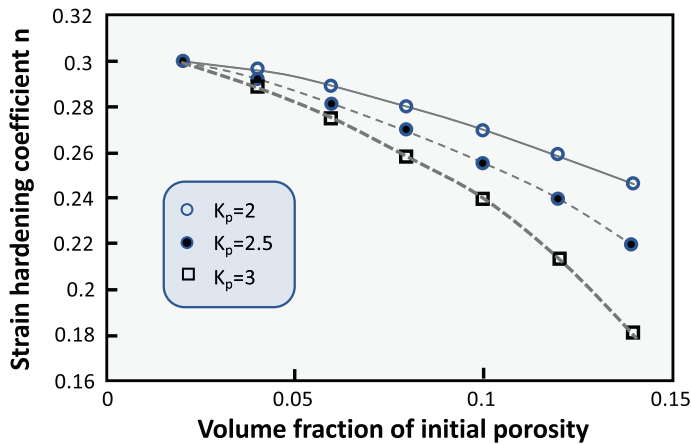


Fig. 2.10 Dependence of n on volume fraction of porosity for porous iron and for three different K_p values. The curves were obtained using experimental data and a strain-hardening model. Redrawn from Straffelini, G., 2005. Strain hardening behaviour of powder metallurgy alloys. Powder Metall. 48, 189–192, with permission of the Publisher.

Dual phase steels have a duplex ferrite-martensite microstructure. They also display a continuous yielding behavior, since deformation starts early in ferrite, and then spreads in the surrounding martensitic islands. Different parameters play an important role in the plastic deformation, including fraction, size, shape, composition, and distribution of the phases. In general, the strain hardening of dual phase steels markedly changes with plastic strain, depending on the partitioning of the plastic strain among the different phases (Paruz and Edmonds, 1989). Since the recorded flow stress, σ , and true strain, ϵ , are macroscopic average values (as discussed at the end of Section 1.3), the calculated n values assume intermediate values between that of ferrite, approximately 0.3, and martensite, approximately 0.1, that change with plastic deformation.

A particular situation is encountered in metals that display special phenomena during plastic deformation. An example is **TRIP steels**. Here, the presence of a complex and heterogeneous microstructure decreases λ_g and this would induce a reduction of n . However, the unstable austenitic regions in the microstructure, characterized by a low SFE, transform to martensite during plastic deformation. This process is called *strain-induced martensitic transformation* and plays a major role. It is accompanied by an increased plasticity (the so-called *Transformation Induced Plasticity*—TRIP), due to the additional strain induced by the volume difference between austenite and martensite, and—most importantly—by an additional increase in the flow strength due to the formation of martensite. As an example, Fig. 2.11A displays the tensile true stress-true strain behavior of a TRIP steel with chemical composition: Fe-15%Mn-3%Al-3%Si (Frommayer et al., 2003). For small strains, a moderate strain hardening occurs and n is close to 0.2. However, for plastic strains above 0.15, the flow stress increases noticeably with strain. This is due to the transformation of austenite into martensite,

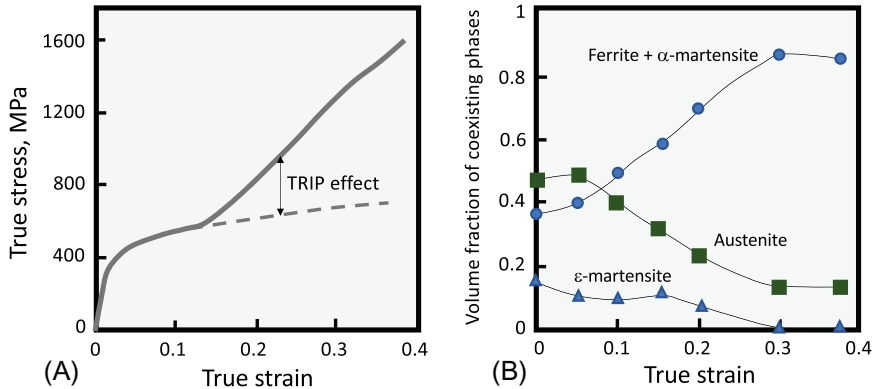


Fig. 2.11 (A) True stress-true strain curves of a TRIP steel at a strain rate of 10^{-4} s^{-1} ; (B) volume fractions of coexisting phases in the TRIP steel.

Adapted and redrawn from Frommayer, G., Brux, U., Neumann, P., 2003. Supra-ductile and high-strength manganese-TRIP/TWI steels for high energy absorption purposes. *ISIJ Int.* 43, 438–446.

as demonstrated by the X-ray diffraction data displayed in Fig. 2.11B. The strength increase due to the phase transformation induces a marked increase in the strain-hardening rate, $d\sigma/d\varepsilon$. This, in turn, induces a rise in n , which increases with plastic deformation to a peak value (close to 0.9 for the material of Fig. 2.11). Subsequently, when plastic deformation spreads into the martensitic regions the strain-hardening rate (and also n) decreases. Finally, necking occurs when $d\sigma/d\varepsilon$ reaches σ (or, similarly, when the instantaneous n value is equal to ε).

The evolution of the strain-hardening rate and of the instantaneous strain-hardening coefficient for TRIP steels are schematically shown in Fig. 2.12. Further details on the TRIP effect will be provided in the next section when dealing with the strain-hardening behavior of austenitic stainless steels.

To summarize:

1. In mild (low C) annealed ferritic steels, the n values (obtained from the uniform region) are in the range 0.18–0.25; higher values are recorded in IF steels;
2. The n values decrease with decreasing grain size and increasing the fraction of small precipitates or other microstructural features that decrease λ_g ;
3. Cold worked steels contain a large density of dislocations, and tempered martensitic steels also contain a large density of precipitates and interfaces; in both cases the corresponding n values are low;
4. The microstructural features that increase strength also reduce λ_g and therefore the strain-hardening coefficient, as shown graphically in Fig. 2.13 (the reported intervals are only indicative);
5. One exception is represented by TRIP steels because of the hardening phase transformation (retained austenite into martensite) during plastic deformation. This indicates an attractive strategy to improve the strain-hardening behavior of metals while obtaining a satisfactory strength.

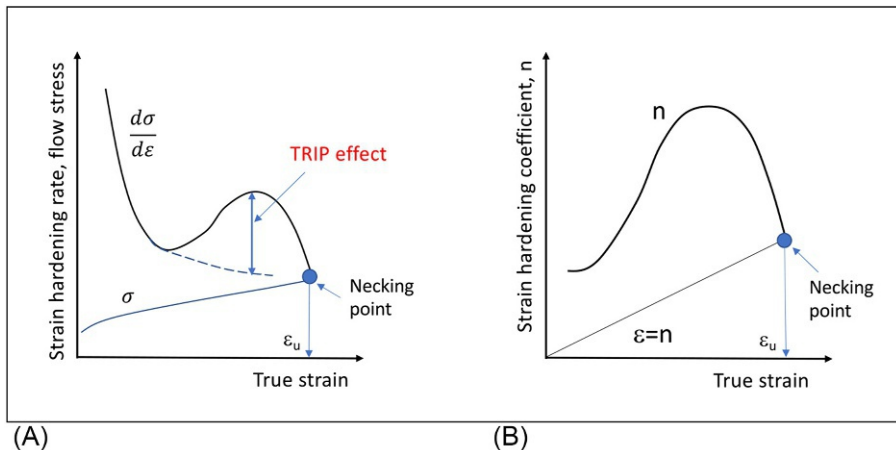


Fig. 2.12 Influence of the TRIP effect on the evolution with the true strain of the strain-hardening rate (A) and the instantaneous strain-hardening coefficient (B), with identification of the necking point.

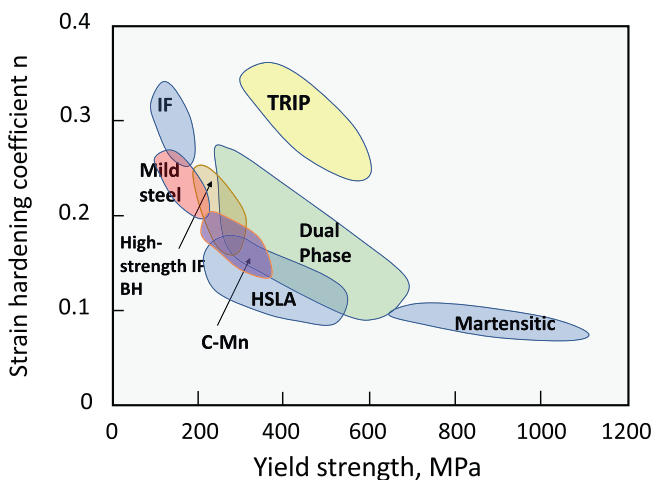


Fig. 2.13 Experimental dependence of the average strain-hardening coefficient vs yield strength for some types of steels (the reported intervals are only indicative). It is recalled that yield strength increases markedly with cold working and, correspondingly, n decreases to values typically below 0.1.

2.4 Strain hardening of fcc steels

Austenitic stainless steels in the annealed condition display a strain-hardening behavior that generally develops in two stages, as in Fig. 1.5C. During the first stage, at rather low plastic strains, dislocation motion is quite easy (by coplanar slip) and the coefficient n is low, around 0.14 or less. Subsequently, the dislocation density

increases with straining. Many slip planes are activated simultaneously, and different grains are involved. As a result, the strain-hardening rate increases and coefficient n rises, assuming values typically between 0.35 and 0.5 in stable steels (i.e., in steels that do not change their microstructure during plastic deformation), in agreement with the prediction ($n = 0.5$) introduced in Section 2.2.

As an example, Fig. 2.14 shows the true stress-true plastic strain curve of a 316L austenitic stainless steel in a double logarithmic scale (Soussan et al., 1991). As mentioned, the slope of the curve is related to n (Fig. 2.5A). In the first stage, the slope of the σ - ϵ curve is small and then increases in the second stage. The following equation was proposed to fit the experimental data:

$$\sigma = K \cdot \epsilon^n + \Delta$$

where:

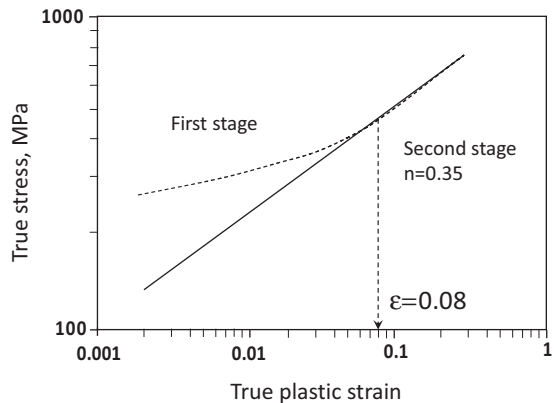
$$\Delta = \exp(K_L + n_L \cdot \epsilon) \quad (2.14)$$

K_L and n_L are two material constants (Eq. 2.14 is known as the *Ludwigson equation*). For the steel in Fig. 2.14, $K_L = 4.947$ and $n_L = -56.986$, and the transitional strain is 0.08. In the second stage, the recorded strain-hardening coefficient is 0.35.

Austenitic stainless steels of type 316 are also called *18-10 steels*, due to the typical content of 18%Cr and 10%Ni. Type 304 steels, called 18-8 steels (18%Cr, 8%Ni), are characterized by a lower SFE mainly because of the lower Ni content (Rhodes and Thompson, 1977; Allain et al., 2004). Fig. 2.15 shows the true stress-true strain curves for a 316 and a 304 stainless steel at RT (Byun et al., 2004). The 304 steel displays a strain-hardening behavior that is similar to that depicted in Fig. 2.12A. This behavior is due to the formation of twins and martensite regions (TRIP/TWIP effect) during plastic deformation. As expected, for the 316 steel, this effect is much less marked.

Type 301 stainless steel has an SFE that is lower than that of 304 (it is below 15 mJ/m², Lee et al., 2010). This kind of steel is then expected to be rather *metastable*, and to display a pronounced TRIP effect during plastic deformation (Table 2.2). As a matter of fact, it was observed that in this steel the transformation of austenite into

Fig. 2.14 True stress-true plastic strain curve for an austenitic stainless steel in log-log scale. Modified and redrawn from Soussan, A., Degallaix, S., Magnin, T., 1991. Work-hardening behaviour of nitrogen alloyed austenitic stainless steel. Mater. Sci. Eng. A 142, 169–176.



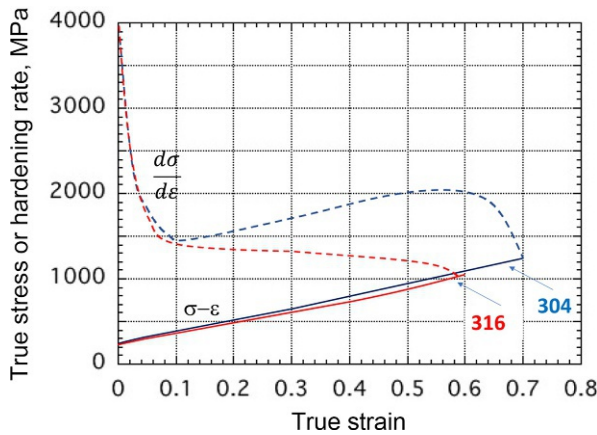


Fig. 2.15 True stress-true strain curves and corresponding strain-hardening curves for 304 and 316 austenitic stainless steels. Redrawn from Byun, T.S., Hashimoto, N., Farrell, K., 2004. Temperature dependence of strain hardening and plastic instability behaviors in austenitic stainless steels. *Acta Mater.* 52, 3889–3899.

martensite during tensile testing started at a very low true strain, approximately 0.05, and the martensite fraction increased up to 80% near the end of the test (Talyan et al., 1998). The corresponding peak value for n (near failure) was rather high: 1.07. The authors also tested a 304L stainless steel. In this case, the martensitic transformation started later ($\epsilon \sim 0.22$) and the fraction of martensite near the end of the test was close to 50%. The measured n value near to failure was much lower: 0.52.

A distinct behavior is displayed by high-Mn austenitic steels, such as the **Hadfield steels** with high C-content (approximately 1.15%C and 12%Mn, to achieve high hardness), and the **TWIP steels**. In these steels, very high strain-hardening rate is attained due to an intense contribution of twinning (Bouaziz et al., 2011). The formation of twins during plastic deformation increases strain hardening. This is mainly because twins act as sub-grains and thus as obstacles for dislocation slip, inducing formation of high density of geometrically necessary dislocations. Of course, a sufficiently high Mn content is required to have a fully austenitic micro-structure, but its content, together with the content of other elements such as C, Si, and Al, must be balanced to obtain the TWIP effect, i.e., a sufficiently low SFE to avoid formation of a cellular dislocation structure. As mentioned, SFE should be in the range of 15–45 mJ/m².

Fig. 2.16 shows the true stress-true strain curves and the corresponding strain-hardening rates ($d\sigma/d\epsilon$) of three fully austenitic high-Mn steels containing 0.6%C, 22%Mn, and three different amounts of Al: 0%, 3%, and 6% (Park et al., 2010). The SFE of the steels increases with increasing the Al content. The following values are reported: 21.5 mJ/m² (0% Al); 36.5 mJ/m² (3% Al); and 50.7 mJ/m² (6% Al). A clear TWIP effect is displayed by alloys containing 0% and 3% Al, for which the SFE falls in the reported experimental interval for twinning. Therefore the resulting uniform strain, ϵ_u , is maximum for the steel with 3%Al, which observed an abundant twin formation, in particular at high strains, and lower for the steel with 6%Al, characterized by an insignificant contribution of twinning to plastic deformation.

To summarize:

1. In fcc austenitic steels, plastic flow is typically easy at the beginning of straining ($\epsilon < 0.05$ – 0.08 , first stage of deformation), where the strain-hardening coefficient is low;

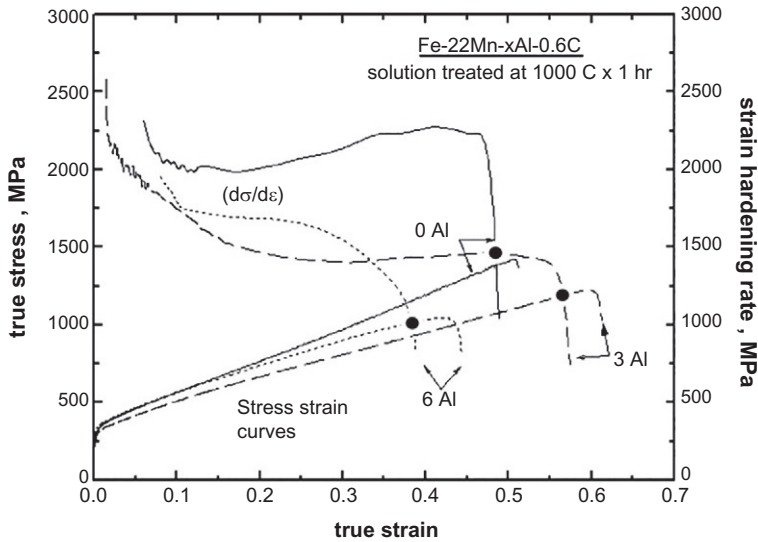


Fig. 2.16 True stress-true strain curves and the corresponding strain rate evolutions of three steels containing 0.6%C, 22%Mn, and three different amounts of Al: 0%, 3%, and 6%. The tests were carried out at RT and at 10^{-3} s^{-1} . The *circular dots* indicate the necking points. Modified from Park, K.T., Jin, K.G., Han, S.H., Hwang, S.W., Choi, K., Lee, C.H., 2010. Stacking fault energy and plastic deformation of fully austenitic high manganese steels: effect of Al addition. *Mater. Sci. Eng. A* 527, 3651–3661; reproduced with permission of the Publisher.

2. With increasing of plastic strain (second stage), strain hardening occurs due to the increase in the dislocation density, and n increases to values of around 0.35–0.5;
3. If the SFE is very low, less than approximately 15 mJ/m^2 , part of the austenite will transform into martensite during straining, and this will increase strain hardening, reaching instantaneous n values of 0.8 or more (TRIP effect);
4. A similar strain hardening is observed when SFE is between 15 and 45 mJ/m^2 , i.e., when twinning exerts an important contribution to plastic deformation (TWIP effect);
5. Austenitic steels have a yield strength (in the annealed condition) comparable to that of ferritic-pearlitic steels (Table 1.3); however, they exhibit a much higher strain-hardening ability; this behavior is particularly marked in the metastable grades, characterized by the TRIP or TWIP effects during plastic deformation.

2.5 Strain hardening of selected non-ferrous alloys

Al alloys

It is first important to note that Al alloys, with an fcc structure, have a rather low melting point, in the range $460\text{--}670^\circ\text{C}$, and therefore RT is comparatively high, close to $0.4T_m$ (where T_m is the melting temperature in Kelvin). This has two main consequences with regards to the strain-hardening behavior of these alloys:

- As already mentioned, SFE of metals increases with temperature. At RT Al alloys are hence characterized by high SFE values (Table 2.1). Therefore they do not display any TRIP or

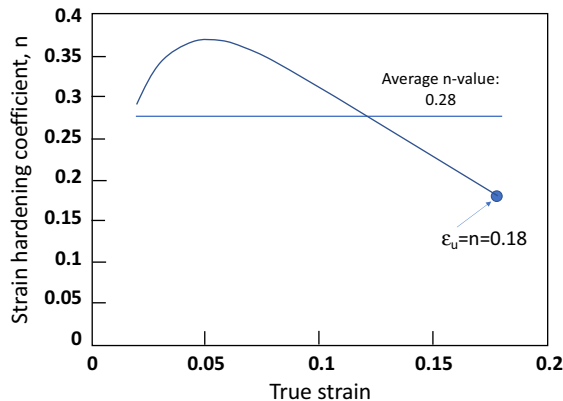


Fig. 2.17 Evolution of the instantaneous n value with plastic strain for an Al-3%Mg alloy. Data obtained from Duan, X., Jain, M., Metzger, D., Kang, J., Wilkinson, D.S., Enbury, J.D., 2005. Prediction of shear localization during large deformation of a continuous cast Al-Mg sheet. *Mater. Sci. Eng. A* 394, 192–203.

TWIP effect during plastic deformation at RT (and below RT). On the contrary, plastic deformation typically proceeds with the formation of dislocation cells and, therefore, n is strongly dependent on the microstructural heterogeneities, i.e., on parameter λ_g ;

- Bulk thermal softening phenomena due to dislocation recovery, explained in more detail in [Chapter 4](#), are also very important at RT ([Kocks and Mecking, 2003](#)). They significantly reduce the strain-hardening rate during plastic deformation, inducing a decrease in the instantaneous n value with strain. An example is shown in [Fig. 2.17](#), which depicts the evolution of n , calculated with Eq. (2.9), for an Al-3%Mg alloy (AA5754-O). For small strains, n is seen to slightly increase, but then continuously decreases because of the prevalence of thermal *softening*. The average n value is 0.28, but the uniform elongation at the onset of necking, ϵ_u , is only 0.18 and it coincides with n at necking.

The average strain-hardening coefficient, n , is close to 0.3 for pure Al. It is quite high in annealed alloys (close to 0.32–0.34, mainly because of dynamic strain aging, DSA, that will be discussed in the next chapter), whereas it typically decreases with alloying, when microstructural features such as precipitates are present in the microstructure.

The role of precipitates is shown in [Fig. 2.18](#), which shows the evolution of the yield strength and ultimate tensile strength for an AA 7449 alloy (8.3% Zn, 2.2% Mg, and 1.9% Cu), subjected to solution annealing and aging for different durations, at temperatures in the range of 120–160°C ([Fribourg et al., 2011](#)). For aging durations shorter than 12 h, under-aging conditions were achieved, whereas for durations longer than 12 h over-aging conditions were achieved. In the case of the over-aged samples, the σ_{UTS}/σ_Y ratio decreases from approximately 1.04 (peak aged alloy) to 1.61 for the alloy aged for 10,000 h. From [Fig. 2.7](#), the average strain-hardening coefficient, n , is estimated to increase from approximately 0.03 to approximately 0.13. This behavior can be explained by considering that hardening is induced by incoherent precipitates, whose spacing increases from approximately 8 nm to approximately 60 nm with the aging duration. Correspondingly, λ_g increases and the density of geometrical dislocations decreases during plastic deformation, due to a lower accumulation of Orowan loops around the precipitates. As said, an increase in λ_g is reflected in an increase in the average strain-hardening coefficient n .

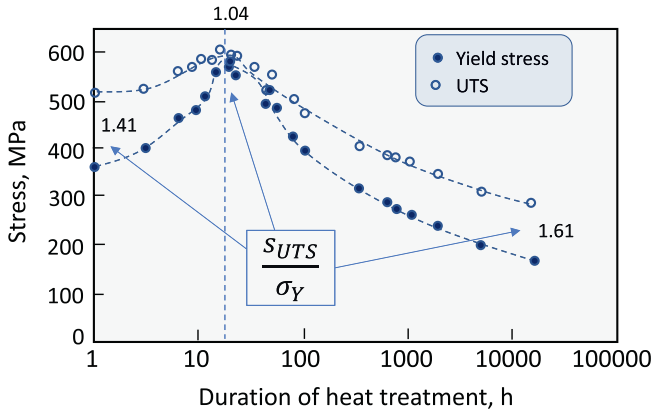


Fig. 2.18 Evolution of the yield strength and of the ultimate tensile strength as a function of the aging duration in the case of a 7499 Al alloy.

Redrawn from Fribourg, G., Brechet, Y., Deschamps, A., Simar, A., 2011. Microstructure-based modelling of isotropic and kinematic strain hardening in a precipitation-hardened aluminium alloy. *Acta Mater.* 59, 3621–3635.

In the case of the under-aged samples, the picture is different. In this case, the precipitates are very small in size and they behave as particles that can be crossed by dislocations (Section 1.5). Therefore they can be considered as a second phase in the matrix microstructure, with a much lower intrinsic deformability than the matrix. Therefore, with increasing the aging duration, the volume fraction of these precipitates increases, from approximately 0.025 to approximately 0.055 in the peak aged condition, and correspondingly the estimated average n value decreases, from approximately 0.1 to approximately 0.03.

Cu alloys

Cu alloys with an fcc structure display high strain hardening rates during plastic deformation. As a result, they generally display comparatively high average n values, in the range 0.2–0.5, like austenitic stainless steel. As an example, Fig. 2.19 shows the engineering stress-strain curves of a commercial copper and a brass. The average n values are also indicated.

Also for these alloys, strain hardening is particularly marked when the TWIP effect is present. Fig. 2.20A shows the compressive true stress-true strain curves for Cu and some Cu-Al alloys (Rohatgi et al., 2001). In the case of Cu, the strain-hardening coefficient is high, approximately 0.54. With increasing Al content, the strain-hardening increases appreciably. The curves in Fig. 2.20B show a large increase in the strain-hardening rate at intermediate strains, especially for materials containing 4% and 6% Al. This was attributed to the contribution of twinning, as already evidenced in TWIP steels. At a strain of 0.2, n is approximately equal to 0.84 for the alloy with 6%Al, and the average n value of this alloy is approximately 0.6.

As a matter of fact, the authors observed the formation of twins in these alloys, as shown by the arrows in the micrograph of Fig. 2.20C. The transition in the plastic

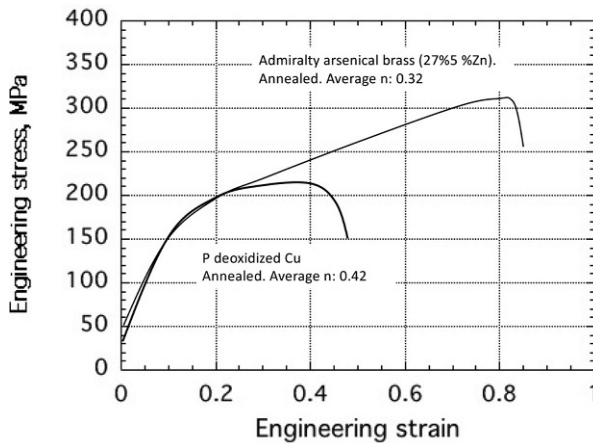


Fig. 2.19 Engineering stress-strain curves for a commercial P copper and a brass.

Redrawn from Copper Development Association Inc., 2020. Application Data Sheet: Mechanical Properties of Copper and Copper Alloys at Low Temperatures, 144/8.

deformation process is due to a change in the SFE determined by the alloying elements. As seen in Table 2.1, the SFE in pure Cu is comparatively high, and plastic deformation proceeds with the formation of dislocation cells. However, as the alloying content is increased, the SFE decreases, and for the alloys with 4% and 6% Al, it is sufficiently low to favor the contribution of twinning that is particularly important under compression.

In this respect, it must be observed that pure Cu is also reported to twin readily (also when strained in tension) if the deformation rate is sufficiently high and/or the temperature is sufficiently low. As already mentioned, twinning gains in importance when dislocation slip becomes more difficult.

Mg and Ti alloys

Plastic deformation in hcp Mg and Ti alloys is rather complex, as very easy dislocation slip at RT is possible on the basal planes, but not on the other planes for which the critical resolved yield stress is too high. In polycrystalline hcp metals, plastic deformation is restricted only to the properly oriented grains, while the other grains may remain elastically deformed up to large nominal strains. Therefore the first yielding phenomenon may be observed during tensile testing. Because of slip difficulties, twinning may play a role in plastic deformation. Several twinning modes are reported, and the contribution of twinning to plastic deformation is particularly effective under compression.

Fig. 2.21A shows the tensile true stress-true strain curves for an AZ31B Mg alloy (containing 3% Al; Wu and Lin, 2010) and at two strain rates. The yield strength is approximately 235 MPa and the authors report an average n value of 0.17, irrespective of the strain rate. This value is rather similar to the uniform strain, ϵ_u , which is in the range of approximately 0.18–0.2. The low n values are rather typical in Mg alloys. They originate from the initial easy in plastic deformation in the properly oriented grains and the subsequent very reduced strain hardening, due to the difficulties in activating additional slip systems (as it occurs, e.g., in fcc steels). One way to increase n is to produce alloys with a small grain size or to add rare-earth elements able to

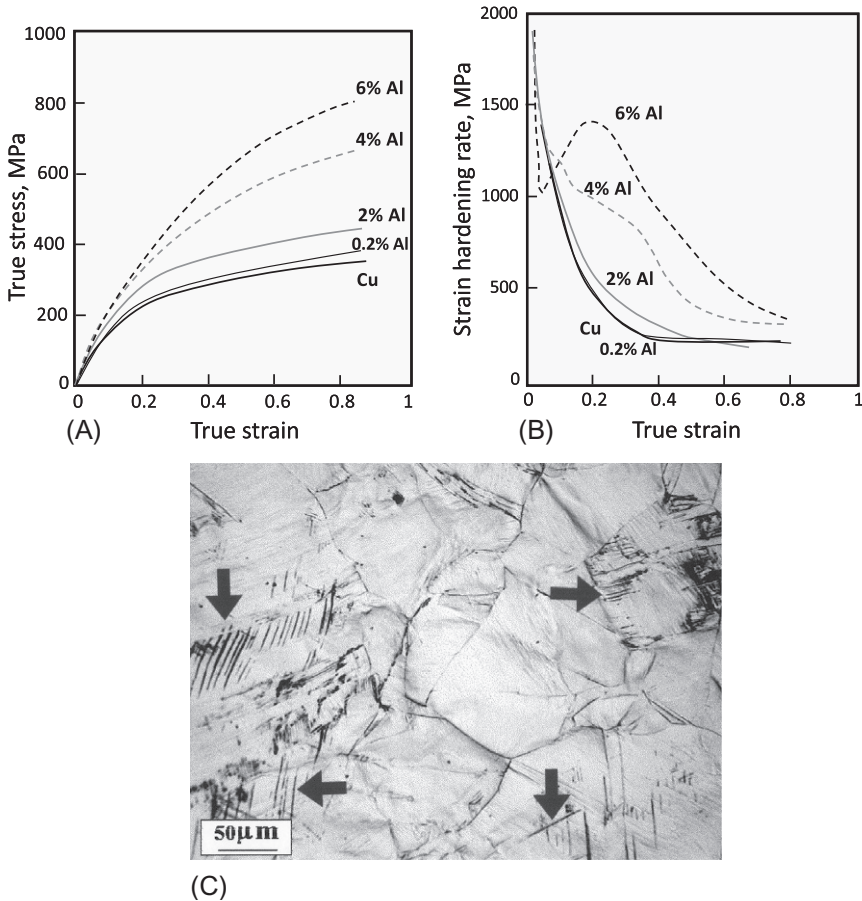


Fig. 2.20 (A) Flow curves of Cu and some Cu-Al alloys at RT and at $5 \cdot 10^{-4} \text{ s}^{-1}$ and (B) corresponding strain-hardening rate curves, (C) optical micrograph of the alloy with 4% Al after a true strain of 0.3 showing deformation twins (marked with *arrows*). Redrawn (A, B) and reproduced (C) from Rohatgi, A., Vecchio, K.S., Gray, III G.T., 2001. The influence of stacking fault energy on the mechanical behavior of Cu and Cu-Al alloys: deformation twinning, work hardening, and dynamic recovery. *Metall. Mater. Trans. A* 32, 135–145, with permission of the Publisher.

randomize the grains' orientation. This means that more properly oriented grains are available to increase the intensity of strain hardening (Yamashita et al., 2001).

Another way to increase the ductility of Mg alloys is to increase the deformation temperature. In fact, as the temperature is increased, the critical resolved shear stress for the dislocation slip along prismatic and pyramidal plans significantly decreases, as shown in Fig. 2.21B for an extruded AZ31 Mg alloy (Boehlert et al., 2012; Barnett et al., 2004).

Fig. 2.22 shows the engineering stress-strain curves for an extruded Mg-Sn-Zn-Al-Na alloy, tested in tension (in both the extrusion and transverse directions) and in compression (in the extrusion direction only). In Mg alloys, twinning is easier in

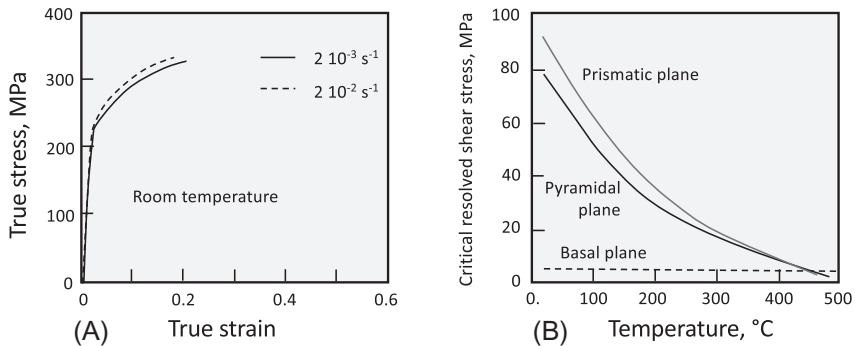


Fig. 2.21 (A) Tensile true stress-true strain curve at three different strain rates for a AZ31B Mg alloy sheet (containing 3% Al; adapted and redrawn from Wu, H.Y., Lin, F.-Z., 2010. Mechanical properties and strain hardening behavior of Mg alloy AZ31B-H24 thin sheet. *Mater. Sci. Eng. A* 527, 1194–1199); (B) dependence of the critical resolved shear stress with temperature and for different slip systems in an extruded AZ31 Mg alloy (redrawn from Boehlert, C.J., Chen, Z., Gutierrez-Urrutia, I., Llorca, J., Perez-Prado, M.T., 2012. In situ analysis of the tensile and tensile-creep deformation mechanisms in rolled AZ 31. *Acta Mater.* 60, 1889–1904).

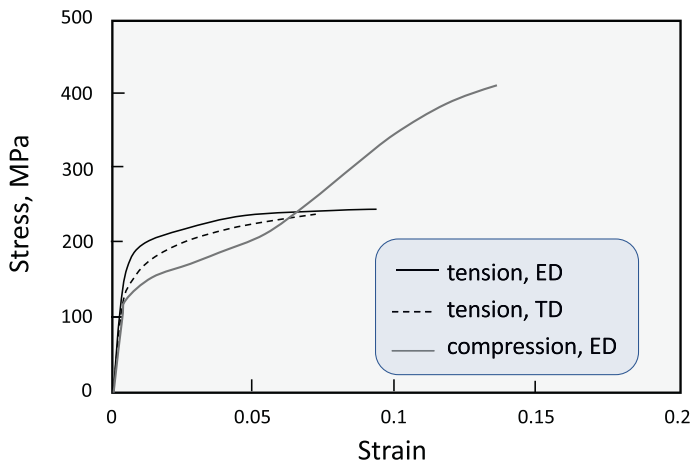


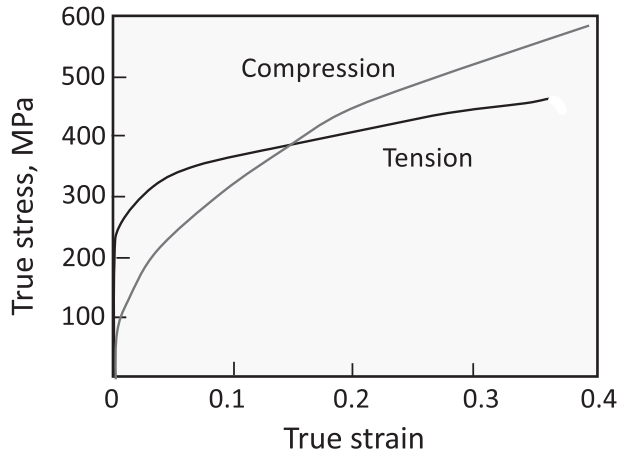
Fig. 2.22 Tensile and compressive stress-strain curves for an extruded Mg alloy in the extruded (ED) and transverse (TD) directions.

Redrawn from Davis, A.E., Robson, J.D., Turski, M., 2019. Reducing yield asymmetry and anisotropy in wrought magnesium alloys—a comparative study. *Mater. Sci. Eng. A* 744, 525–537.

compression than under tension. Consequently, the yield strength under compression is lower than under tension, due to the contribution of twinning to the onset of plastic deformation. However, the subsequent strain hardening is much more pronounced than in the case of tensile testing, because of the intense formation of twins during plastic deformations (Davis et al., 2019). The stress-strain curves of Mg alloys under compression are then similar to those of the TWIP steels.

Fig. 2.23 True stress-true strain curves obtained under tension (RD-T) and compression (RD-C) for CP Titanium.

Redrawn from Lin, P., Hao, Y., Zhang, B., Zhang, S., Chi, C., Shen, J., 2017. Tension-compression asymmetry in yielding and strain hardening behavior of CP-Ti at room temperature. *Mater. Sci. Eng. A* 707, 172–180.



In general, Ti alloys exhibit low strain-hardening ability and low uniform strains. Pure Ti and most Ti α -alloys display a deformation behavior that is rather like that of Mg alloys. Fig. 2.23 shows the results of mechanical tests carried out on commercially pure Ti (CP Ti) under tension and compression (Lin et al., 2017). In the case of pure compression testing, the yield stress is lower than under tension, but strain hardening increases noticeably during plastic straining. The average strain-hardening coefficient is close to 0.15 under tension and close to 0.6 under compression. This behavior is attributed to the occurrence of twinning under compression, which anticipates yield onset and increases strain hardening (Salem et al., 2003). In general, in Ti α -alloys, twinning plays an important role in compressive plastic deformation, and these alloys display a tensile and compressive *asymmetry* in the flow strength.

For the α - β Ti-64, n is less than 0.1 (approximately 0.05) and the asymmetry in the tension-compression yield behavior is moderate (Gilles et al., 2011).

Metastable β alloys (bcc), may display twinning or martensitic transformation during tensile plastic deformation, depending on the β phase stability, and therefore on the chemical composition (Banerjee and Williams, 2013). In these cases, the strain-hardening behavior is similar to that displayed by TWIP and TRIP steels, even if of lower intensity. The average strain-hardening coefficients in these alloys are typically in the range 0.1–0.25 in tensile testing, but the development of metastable β alloys with properly balanced TWIP and TRIP effects is a promising strategy to obtain Ti alloys with enhanced plastic properties (including ductility). An example is shown in Fig. 2.24, referring to a novel Ti-4Mo-3Cr-1Fe alloy (Ren et al., 2020). The comparatively intense strain hardening was attributed to a profuse formation of twins during plastic deformation. The uniform strain ($\epsilon_u = 0.27$) is comparatively high.

Summary

Table 2.3 reports a summary of the typical n values of some alloys in the annealed condition. As seen, the strain-hardening coefficient, n , is influenced by several factors, and may vary by changing the microstructure of the alloy and during plastic

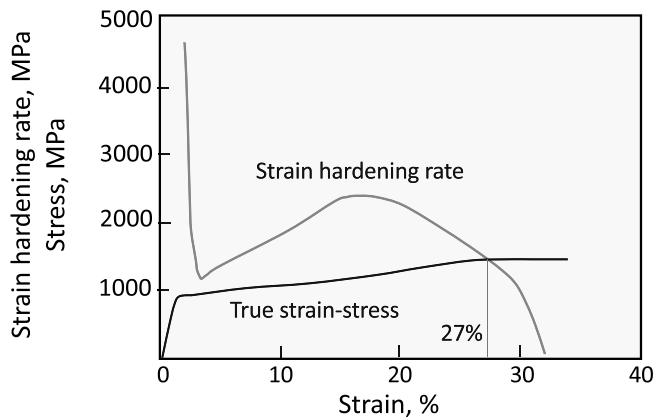


Fig. 2.24 Tensile true stress-true strain and strain-hardening curve for a β solution treated Ti-4Mo-3Cr-1Fe alloy.

Redrawn from Ren, L., Xiao, W., Kent, D., Wan, M., Ma, C., Zhou, L., 2020. Simultaneous enhanced strength and ductility in metastable β -Ti alloy by stress-induced hierarchical twin structure. *Scr. Mater.* 184, 6–11.

Table 2.3 Typical average n values for some annealed engineering alloys (in the reference conditions).

Alloy	Average n -values	Alloy	Average n -values
Pure ferrite and IF steels	0.25–0.34	TWIP steels	0.3–0.45 (Markedly changing with strain if SFE is low, reaching values up to 0.8 or more)
(Mild) ferritic steels	0.2–0.3	Al alloys	0.1–0.34 (Decreasing with strain because of thermal softening)
TRIP steels	0.2–0.35 (Markedly changing with strain, reaching values up to 0.6 or more)	Cu alloys	0.2–0.5 (Markedly changing with strain if SFE is low)
Dual phase steels	0.1–0.2	Ti alloys	0.05–0.15 (Higher values under compression; higher values in metastable β alloys)
Austenitic SS	0.25–0.4 (Markedly changing with strain if SFE is low, reaching values up to 0.8 or more)	Mg alloys	0.05–0.15 (Higher values under compression)

deformation. The data in the table should be considered as a quick reference for simple evaluations and comparisons. If yield strength is increased by heat treatment of cold working, n correspondingly decreases, typically below 0.1.

2.6 Strain-hardening coefficient and uniform elongation

Uniform elongation, indicated here with ϵ_u , is an important indication of metal ductility since it represents the maximum plastic strain attainable in uniform tensile deformation before necking. The necking condition is given by Eq. (2.10):

$$\frac{d\sigma}{d\epsilon} = \sigma$$

and it is then delayed when $d\sigma/d\epsilon$ remains high with increasing σ or ϵ . Using a simple model introduced in Section 2.2 and based on the L-H equation, we have: $\epsilon_u = n$.

Fig. 2.25 collects representative experimental data for different metals considered in this chapter. The average n values are considered. The figure shows a general tendency of increasing ϵ_u with increasing n , showing that a proportionality between ϵ_u and n is identified.

In many cases, the onset of necking is reached for ϵ_u values that are slightly lower than the average n value. This may occur in materials characterized by instantaneous n values that strongly change during deformation, such as metals that display TRIP and TWIP effects, and is rather typical in metals that experience thermal softening during plastic deformation (such as Al alloys). In some cases, $\epsilon_u < n$ because of the geometrical softening exerted by an anticipated microstructural damage (such as voids formation) at microstructural heterogeneities (see Chapter 6). A discrepancy between the ϵ_u values

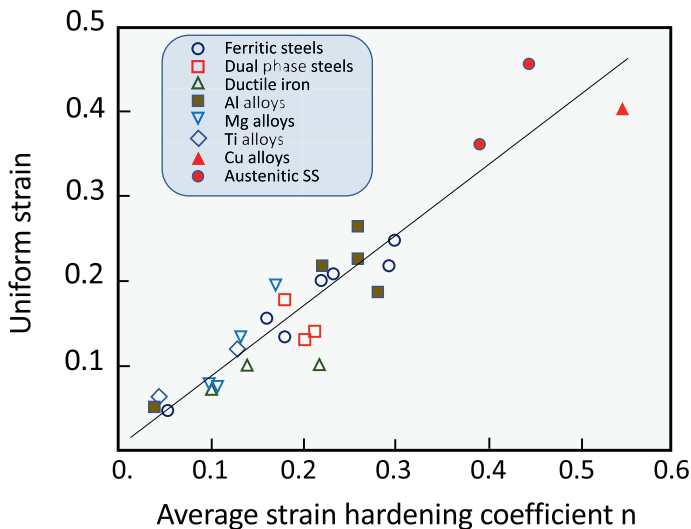


Fig. 2.25 ϵ_u values vs average n values for different alloys considered in this chapter.

and the average n values is also observed in metals with a heterogeneous microstructure, in which the local stresses and strains are much different than the average σ and ϵ values.

2.7 Plastic anisotropy in sheet specimens

Uniaxial tensile tests can also be carried out using samples with a *rectangular geometry* (Fig. 1.1B). A case in point is of samples cut from metal sheets to be used in sheet working operations like deep drawing or stretch forming (Chapter 15). In general, such sheets have small thickness, in the range of 1 mm or less, and are produced by hot rolling, followed by cold rolling and batch or continuous annealing. Thus, the mechanical properties are often *anisotropic*, since they vary with the orientation of the specimen with respect to the rolling direction. As schematized in Fig. 2.26, the specimens for tensile testing are obtained from rolled sheets with reference to the rolling direction. The three directions are often indicated with 0, 45, 90, i.e., forming angles of 0, 45, 90 degrees with the rolling direction.

In general, we can distinguish between two kinds of material anisotropy or texture: *fiber texture* and *crystallographic texture* (Dieter, 1980). Fiber texture is produced by the orientation along the metal flow of microstructural features, such as precipitates, inclusions, and voids. Mechanical properties such as yield strength and ductility are dependent on fiber texture.

The crystallographic texture in sheets is determined by the preferred orientation of certain crystallographic planes. As seen, plastic deformation involves crystallographic planes that allow easy dislocation slip. This implies that in polycrystalline alloys, during plastic working, the grains tend to rotate and align the most favorable slip planes along the direction of maximum plastic deformation. This orientation increases with increase in the intensity of deformation. When cold rolled alloys are subjected to recrystallization annealing, new grains nucleate and grow in the preferred orientations. Consequently, the attitude toward plastic deformation of a recrystallized sheet

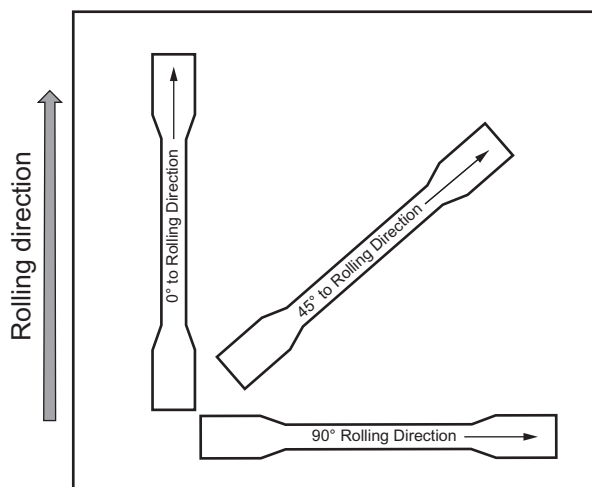


Fig. 2.26 Schematic showing the orientation of sheet specimens for uniaxial tensile testing with respect to the rolling direction.

depends on preferred orientations and changes with respect to the longitudinal, width, and thickness direction.

The plastic anisotropy of metal sheets with crystallographic texture is typically quantified by the so-called *Lankford parameter*, R , determined with reference to the uniform deformation region of the stress-strain curve. Indicating with ε_t and ε_w the plastic strains in the thickness and width (or transversal) direction, R is given by (Hosford and Caddell, 1993):

$$R = \frac{\varepsilon_w}{\varepsilon_t} \quad (2.15)$$

where the width and thickness strains are given by:

$$\varepsilon_w = \ln \left(\frac{w}{w_0} \right) \quad \varepsilon_t = \ln \left(\frac{t}{t_0} \right) \quad (2.16)$$

where:

- w_0 and t_0 are the initial width and thickness of the specimen;
- w and t are the width and thickness after unloading.

Since ε_t is difficult to measure with adequate precision, the longitudinal strain, ε_l , is measured with an extensometer and then ε_t is calculated from the following relationship: $\varepsilon_t = -(\varepsilon_l + \varepsilon_w)$, with the assumption of the constancy of the volume under plastic deformation. The unloading is typically done after an engineering strain of 15%–20%, and in any case, well within the uniform region. For some alloys, such as Mg and Ti alloys, which are characterized by low uniform strains, the unloading must be carried out after engineering strains of 10% or less. R is determined in the three directions: 0, 45, 90, and an average value, R_m , is calculated using:

$$R_m = \frac{R_0 + 2R_{45} + R_{90}}{4} \quad (2.17)$$

For an isotropic material, plastic deformation is the same in all directions; hence, $R = 1$. In general, R values above unity are preferred. In fact, with $R > 1$, we have $\varepsilon_t < \varepsilon_w$. Therefore the thickness reduction during plastic straining is less pronounced. Reducing sheet thinning during plastic forming allows increasing the intensity of the forming operation. Of course, the situation changes if large plastic deformability in the thickness direction is required.

In *ferritic (mild) steels* with bcc structure, many (111) planes are usually oriented parallel to the sheet plane during plastic deformation, such as cold rolling. After recrystallization, some anisotropy remains, and the corresponding R_m values are in the range of 1.4–1.8. Besides the intensity of plastic deformation, other factors play a role. One is the presence of interstitials, such as C and N. Both induce a decrease in R_m , because they decrease the fraction of favorably oriented (111) planes after recrystallization (Lewellyn and Hudd, 1998). Correspondingly, high R_m values, more than

1.8, can be achieved in *IF steels* with very low total C and N contents (below 0.004%). Another factor that allows achievement of high R_m values is the presence of a *pancake grain morphology*, with equiaxed grains that are quite large in the sheet plane and smaller in the thickness direction. This microstructure is obtained with a proper precipitation of intermetallics, such as Al-(carbo)nitriles, during annealing or cooling after hot rolling. During recrystallization, grain boundary Al-(carbo)nitriles reduce the grain growth in the thickness direction, allowing the formation of a pancake microstructure. The increased density of grain boundaries in the thickness direction reduces the plastic deformability in this direction, thus contributing to increasing R_m .

Al alloys with an fcc structure also display crystallographic texture. However, the resulting Lankford parameter is in the range of 0.6–0.8, i.e., below unity. Depending on the composition and thermomechanical processing conditions, i.e., on the hot and cold rolling parameters, huge texture modifications are observed due to the strong selective recrystallization mechanisms occurring during nucleation and growth of new grains (Hirsch and Al-Samman, 2013). For example, in Al, Al-Mg, and Al-Mg-Si alloys (classes: 5000 and 6000), the nucleation and growth of the highly textured cube component on particles or small shear bands is observed. Other complex textured phases may nucleate in these alloys and, as a result, an anisotropic behavior is recorded. As an example, Fig. 2.27 shows the experimental R values for the three 0, 45, 90 orientations for a 6010 Al-Mg-Si alloy subjected to cold rolling and annealing, after different ingot pre-annealing processes (indicated with C1, C2, and C3). Apart from C3, R_0 and R_{90} show the lowest values.

A characteristic behavior is displayed by *Ti* and *Mg alloys* with an hcp structure. For these alloys, slip is easy only on the basal plane. Therefore, during sheet rolling processes, the grains are markedly rotated to align the basal planes with that of the sheet. Strong crystallographic texture develops after annealing, and this is revealed by high R_m values; in the range of 3–5 for α -Ti alloys (Hosford and Caddell, 1993), and higher than 4 for Mg alloys (Boehlert et al., 2012). As shown in Fig. 2.27 for a Ti-64 alloy, the R values vary significantly with the specimen orientation. However, if the temperature is increased, dislocation slip also occurs on non-basal planes, because of the decrease in the relevant critical resolved shear stress, as seen in Fig. 2.21B for a Mg alloy. Consequently, the Lankford parameter decreases, tending to unity, as seen in Fig. 2.28 for a AZ31 Mg alloy (Boehlert et al., 2012). Note that the high R values for the Mg alloys are mainly determined by the very low plastic strains in the thickness direction, ϵ_t , and are then not satisfactory in this case. Research efforts are made to obtain R values close to unity also at RT with increased ϵ_t values, by a proper modification of the chemical composition and microstructure.

In Table 2.4, typical R_m values for some alloys are listed.

2.8 Localized necking in sheet specimens

In the case of tensile testing of cylindrical specimens, necking occurs after a uniform plastic region, which is followed by ductile fracture. The same occurs when testing rectangular specimens with sufficiently low width-to-thickness ratio (Cabezas and Celentano, 2004). However, during testing sheet specimens with small thickness

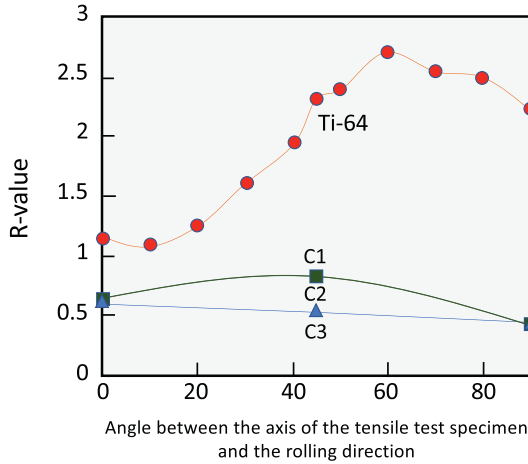


Fig. 2.27 Experimental R values for different orientations, for a 6010 Al-Mg-Si alloy subjected to cold rolling and annealing after different ingot pre-annealing processes, indicated with C1, C2, and C3 (data from Hirsch, J., Al-Samman, T., 2013. Superior light metals by texture engineering: optimized aluminium and magnesium alloys for automotive applications. *Acta Mater.* 61, 818–843), and for a Ti-64 alloy (data from Gilles, G., Hammami, W., Libertiaux, V., Cazagu, O., Yoon, J.H., Kuwabara, T., Habraken, A.M., Duchene, L., 2011. Experimental characterization and elasto-plastic modelling of quasi-static mechanical response of TA-6 V at room temperature. *Int. J. Solids Struct.* 48, 1277–1289).

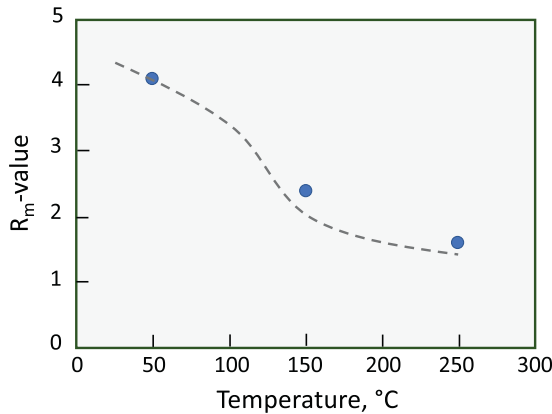
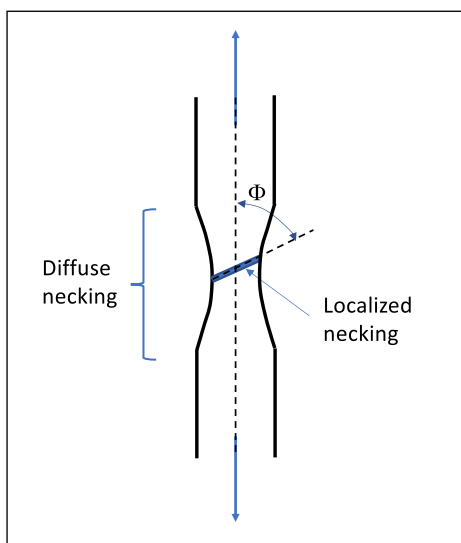


Fig. 2.28 Experimental R_m values obtained at different temperatures for an AZ31 Mg alloy (Boehlert et al., 2012). The dotted curve in the picture is a fit of the results of previous studies reported for comparison by the authors.

(and width-to-thickness ratio more than 7–8), after the onset of necking (also named *diffuse necking* in this case), which involves the specimen gauge, and its length is approximately between one to two times the specimen width, a second kind of necking may be observed before fracture. It is called *localized necking* and is characterized by a strong localization of plastic deformation in a straight band, forming an angle Φ with

Table 2.4 Typical R_m values for some alloys in the annealed condition.

Alloy	R_m range	Alloy	R_m range
(Mild) ferritic steels	1.4–1.8	TRIP steels	~0.9
IF steels	1.8–2.8	Al-alloys	0.6–0.8
HSLA steels	0.9–1.3	Cu-alloys	0.6–0.9
Dual phase steels	0.9–1.2	Ti-alloys	2–5
Austenitic stainless steels	0.7–1.1	Mg-alloys	4–5

**Fig. 2.29** Diffuse and localized necking in thin sheet specimens subjected to uniaxial tensile testing.

the loading direction, as shown in Fig. 2.29. Diffuse necking does not produce a worsening of the surface appearance of the sheet, whereas localized necking introduces a generally unacceptable surface appearance. As noted, diffuse necking may occur when the longitudinal plastic strain is equal to the strain-hardening exponent n . A simplified model predicts that localized necking occurs when:

$$\frac{d\sigma}{d\varepsilon} = \frac{\sigma}{2} \quad (2.18)$$

i.e., when the longitudinal strain increases to $2n$, and the corresponding angle Φ is $\arctg \left(\sqrt{R_m / (1 + R_m)} \right)$ (Hill, 1950). In isotropic materials, $R_m = 1$ and then $\Phi = 54.74^\circ$.

However, the situation in practice is much more complex. Quite often, the experimental values of Φ are greater than 54.74 degrees, and localized necking occurs at a

longitudinal strain, which is generally lower than $2n$. This behavior is due to the presence of microstructural and geometrical *imperfections* that trigger anticipated localization. Typical imperfections are:

- Microstructural regions with larger plastic deformability (often due to different textures) than the average matrix deformability;
- Microstructural regions with the presence of voids (also nucleated at particles following the applied strain) in a higher density than in the neighboring regions;
- Geometrical surface defects, such as surface roughness or scratches;
- Geometrical variations of the specimen thickness along its gage, with the presence of regions with smaller thickness than the average.

These imperfections induce the development of plastic deformation by shear in small regions, and its subsequent spreading along the specimen width. The strain for this plastic shear localization and Φ , in the testing of sheet specimens, depend on the size and distribution of the imperfections and on the sheet aspect ratio (width-to-thickness ratio). It is generally reported that strain localization easily occurs when the imperfection size is close to half the specimen thickness (Tvergaard, 1993).

For example, in a sheet specimen made with a dual-phase low-C steel and submitted to tensile testing, diffuse necking was observed at an engineering strain of 0.182, corresponding to $\epsilon_u = \ln(1 + 0.182) = 0.17$, which is rather close to the experimental strain-hardening exponent of $n = 0.16$ (Li et al., 2016). Subsequently, two intersecting bands were observed to form in the neck region. This strain localization started at approximately $\epsilon = \ln(1 + 0.255) = 0.23$, in a band forming an angle Φ close to 65 degrees. Hence, ϵ at localized necking was lower than $2n$, and Φ was greater than 54.74 degrees. As tensile loading continued, plastic deformation localized in one of the two bands (leading to the nucleation of a ductile crack in the center of the specimen and finally to ductile fracture).

An important case is that of Al alloys and in particular the Al-Mg alloys (class 5000). In these alloys, thermal softening induces a decrease in n with plastic deformation and this anticipates diffuse necking. For example, in the case of the alloy in Fig. 2.17, diffuse necking started for $\epsilon = 0.18$, which is lower than the average n value of 0.28 (Duan et al., 2005). In addition, the presence in the microstructure of regions with special textures that act as imperfections, such as regions with cube texture, and the presence of negative values of the strain rate sensitivity factor (which will be introduced in the next chapter) favored a pronounced shear localization that significantly anticipates localized necking (and fracture). Fig. 2.30 shows diffuse and localized necking in a 5182 Al alloy.

To summarize:

- Sheet specimens display diffuse necking and localized necking during tensile plastic deformation;
- As usual, diffuse necking occurs when $\epsilon \sim n$; in Al alloys that undergo softening during plastic deformation, ϵ at necking is lower than the average n value;
- Typically, localized necking occurs for $\epsilon < 2n$ (Hill model), since it is anticipated by microstructural and geometrical imperfections.

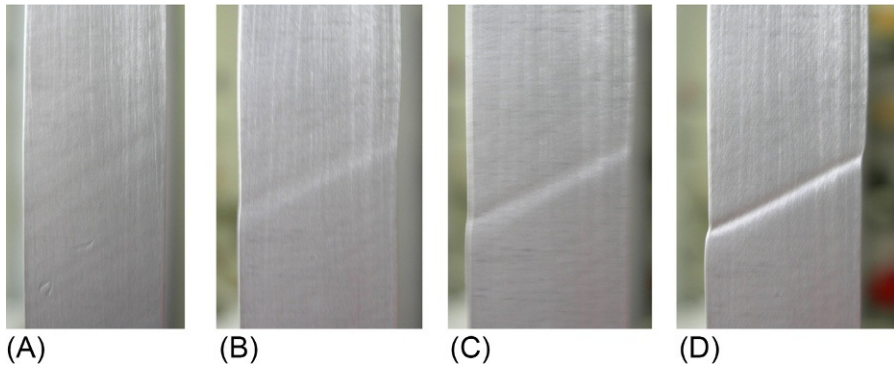


Fig. 2.30 Development of localized necking in an AA5182 Al alloy sheet specimen under uniaxial tensile loading (width: 12.5 mm, thickness: 1.15 mm). Diffuse necking (A) started at $\epsilon_u = 0.22$, which is lower than the average strain-hardening coefficient, $n = 0.3$ (Buzzatti, 2005). In (B) and (C), the development of localized necking is shown.

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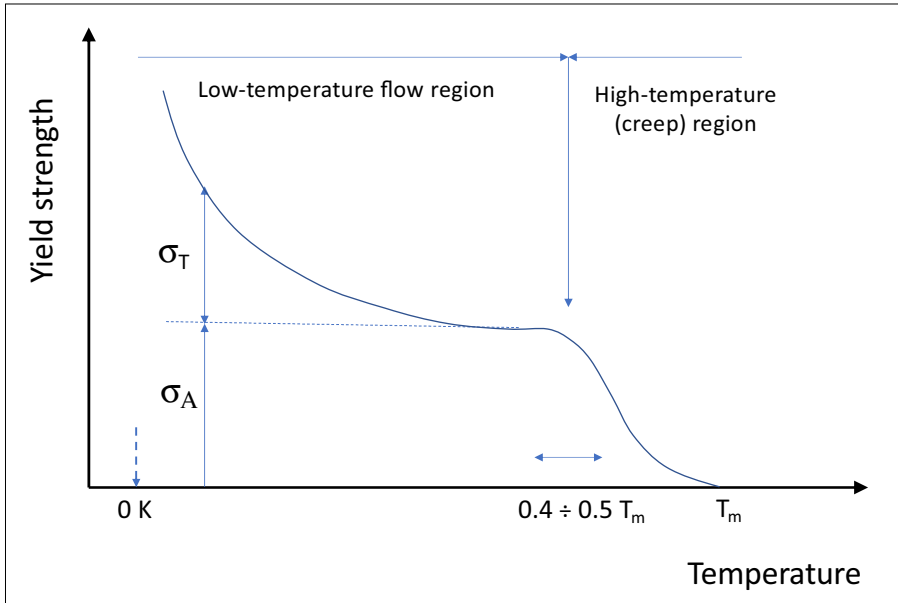
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Plastic deformation: Influence of strain rate and temperature (for $T < 0.4\text{--}0.5T_m$)

3



Schematization showing the two main flow regions of metals.

Chapter outline

- 3.1 Thermally activated slip of dislocations 72
- 3.2 Influence of temperature and strain rate on strain hardening 77
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The plastic behavior of metals is highly influenced by temperature, and two main regions can be identified. For temperatures above $0.4\text{--}0.5T_m$ (where T_m is the absolute melting temperature of the alloy), high-temperature creep deformation occurs; this will be discussed in the next chapter. The present chapter focuses on the deformation behavior of metals for temperatures below $0.4\text{--}0.5T_m$, addressing the role of temperature and strain rate. This is done by introducing first the classical model of thermally activated slip of dislocations. It is the basic frame to understand the role of temperature and strain rate on the plastic deformation of metals in this temperature region, including their influence on the strain-hardening coefficient, n . Subsequently, the particular behavior of Al and Cu alloys, with moderately high SFE, and that of metastable austenitic stainless steels, with low SFE, is outlined.

Next, the dependence of the metal flow stress on strain rate is explained by considering the concept of strain rate sensitivity index, m , which has an important physical meaning for understanding the ductility of metals (as will be seen in Chapters 6–8). In this respect, the phenomenon of dynamic strain aging (DSA) is described, and its role in determining negative m values is identified.

Finally, the metal temperature rise during plastic deformation is analyzed. The important phenomenon of shear banding (including adiabatic shear banding, ABS) is then introduced.

3.1 Thermally activated slip of dislocations

Fig. 3.1A shows the dependence of yield strength on temperature for a 316 austenitic stainless steel (fcc structure), and for pure Fe (bcc structure). In Fig. 3.1B this dependence for 5052 Al alloys (fcc structure) is also shown. All the tests were carried out in

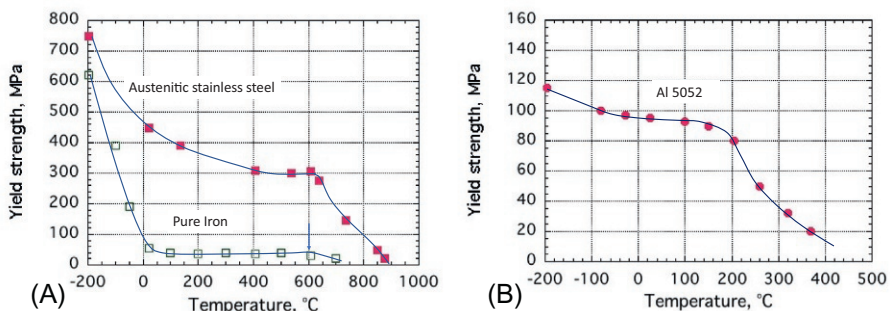


Fig. 3.1 Yield strength dependence ($\sigma_Y = 0.2\%$ offset yield strength) on temperature for (A) an ultrafine-grained type 316 austenitic stainless steel, and for pure iron and (B) for a 5052 Al alloy. A: Redrawn from Varin, R.A., 1987. The effects of temperature and strain rate on the plastic flow and ductility of ultrafine-grained type 316 austenitic stainless steel. *Mater. Sci. Eng.* 94, 93–107; Clarke, D.K., Comstock, R.J., Jr, Mataya, M.C., Von Tyne, C.J., Matlock, D.K., 2008. Effect of strain rate on the yield stress of ferritic stainless steel. *Metall. Mater. Trans. A* 39, 752–762; B: Data from Brandes, E.A., Brook, G.B. (Eds), 1998. *Smithells Light Metals Handbook*, Butterworth Heinemann.

Table 3.1 Melting temperatures for metals with approximate boundaries with the high-temperature region.

Metal	Melting temperature, T_m (°C)	$0.4\text{--}0.5T_m$, in °C (K)	Homologous temperature at RT
Low-C steel	1425–1550	405–635 (680–900)	0.17–0.16
Austenitic stainless steel	1400–1450	400–585 (670–860)	0.18–0.17
Ti alloys	1600–1670	480–700 (750–970)	0.16–1.15
Al alloy	460–670	20–200 (295–470)	0.4–0.32
Cu alloys	930–1000	210–360 (480–635)	0.25–0.23
Mg alloy	350–650	–25–190 (250–460)	0.45–0.32

The homologous temperature at RT is also listed (it is given by RT/T_m , where temperatures are expressed in Kelvin).

quasi-static conditions. Two regions can be recognized. For temperatures beyond $\sim 600^\circ\text{C}$ for steels (approximately corresponding to $0.48T_m$, where T_m is the absolute melting temperature of the metal) and beyond $\sim 200^\circ\text{C}$ for the Al alloy (approximately corresponding to $0.4T_m$), the yield strength decreases with temperature, tending to zero. This is the high-temperature creep region, typically occurring in metals for temperatures above $0.4\text{--}0.5T_m$, as discussed in the next chapter. Table 3.1 lists the melting temperatures for some metals, along with the approximated boundaries with the high-temperature region.

For temperatures below $0.4\text{--}0.5T_m$, the yield strength decreases with increasing temperature tending to a plateau, i.e., to a stress value independent of temperature (and also on strain rate, as will be seen later). Such a decrease is very noticeable in bcc Fe, and less marked in the austenitic stainless steel and particularly in the Al alloy, both with an fcc structure. In this temperature region, the yield strength, σ_Y , can then be expressed by the following relationship:

$$\sigma_Y = \sigma_A + \sigma_T(\dot{\epsilon}, T) \quad (3.1)$$

where

- σ_A is the plateau value, i.e., the so-called *athermal* component of the yield stress;
- σ_T is the *thermal* component of the yield stress, which depends on temperature and strain rate.

In this temperature region, the yield strength can be rationalized within the frame of the *thermally activated slip* of dislocations.

As far as strain rate is concerned, Fig. 3.2 shows that in practice it spans from 10^{-8} s^{-1} to 10^6 s^{-1} . Very low strain rates are encountered in low stress creep deformation at high temperature. Very high strain rates are encountered in high-velocity metal forming, like explosive forming. The strain rate interval between moderate (10^{-4} s^{-1}) to high (10^3 s^{-1}) strain rates is of interest for the thermally activated slip

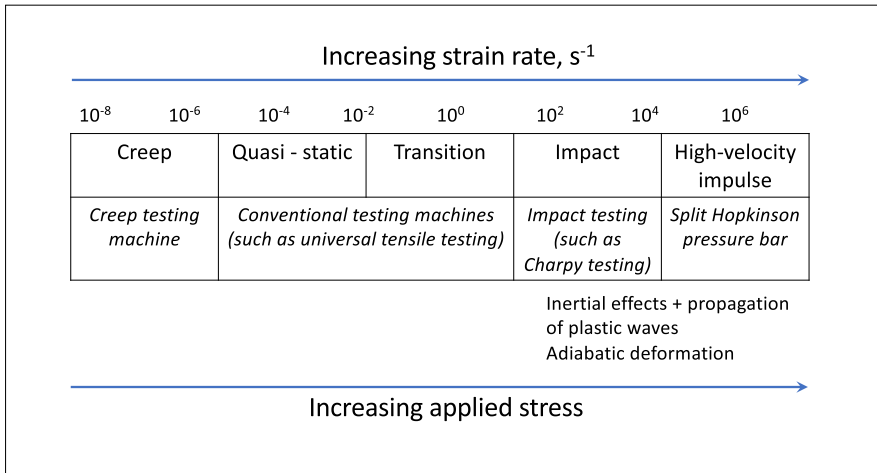


Fig. 3.2 Schematics showing the typical strain rate ranges and corresponding tests.

of dislocations. Note that with conventional universal tensile testing machines, strain rates up to approximately $10^2 s^{-1}$ are attained. For higher strain rates, up to $10^3 s^{-1}$ or more, special equipment must be used, such as the *split Hopkinson pressure bar* (Chen and Song, 2011). In this technique, a specimen is placed between two bars, the *incident* and the *transmission* bar. The incident bar is impacted by a striker bar, and this produces an elastic wave. This wave is propagated in the incident bar and then transmitted into the sample. With proper strain gauges in both the incident and transmitted bar, the strain, strain rate, and applied stress can be obtained as a function of time.

In the thermally activated slip region, a decrease in temperature and an increase in strain rate induce an increase in the yield strength since plastic deformation is made more difficult. As known, the slip of dislocations is slowed down by the presence of microstructural obstacles (Section 1.5). In the model of thermally activated slip of dislocations, it is assumed that thermal fluctuations assist dislocations in overcoming *short-range stress-field obstacles* (such as solute atoms, internal lattice friction due to elastic distortions at the dislocation lines, and coherent precipitates) during their motion (see Fig. 3.3), whereas they are not able to assist dislocations in overcoming *long-range stress-field obstacles* (formed from grain boundaries, second phases, incoherent precipitates, and other dislocations in some cases). Hence, the stress σ_A is determined by the resistance to dislocations motion exerted by the long-range stress-field obstacles, whereas σ_T is determined by the ability of dislocations to overcome short-range stress-field obstacles. Within this frame, it is assumed that short-range stress-field obstacles provide a *potential-energy barrier*, ΔG , which must be surmounted by each dislocation during its motion. The probability, p , that a dislocation is able to jump over the obstacle, i.e., to have an energy equal to or greater than ΔG , is given by the following relation (Meyers et al., 2002):

$$p = A \cdot e^{\frac{-\Delta G}{kT}} \quad (3.2)$$

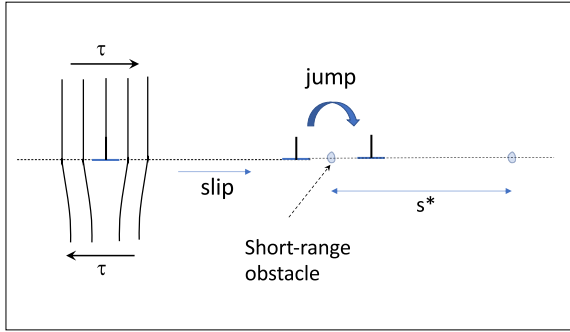


Fig. 3.3 Representation of the jump of an edge dislocation over a short-range obstacle (very small in size) assisted by a thermal fluctuation. s^* indicates the mean spacing between two obstacles.

where

- A is a material constant;
- k is the Boltzmann constant (1.38×10^{-23} J/K);
- T is the absolute temperature.

The probability p is related to the frequency, ν_1 , of successful jumps over the barrier, i.e., $p = \nu_1/\nu_0$, where ν_0 is the vibrational frequency of a dislocation. The transit time of a slipping dislocation is given by the travel time between two neighboring short-range obstacles at a distance s^* (indicated with t_t), and the time spent at an obstacle to wait for a successful jump (indicated with t_w). In general, t_t is negligible in comparison to t_w . Hence, the average dislocation velocity is $v = s^*/t_w$. Time t_w is the inverse of ν_1 , and then: $v = s^* p \nu_0$. In general, the average dislocations velocity is related to the average of the strain rate, $\dot{\epsilon}$, through the Orowan relation:

$$\dot{\epsilon} = \alpha \cdot \rho \cdot b \cdot v \quad (3.3)$$

where

- α is the orientation factor;
- ρ is the dislocation density;
- b is the modulus of the Burgers vector.

Therefore, from Eq. (3.3) and with the relevant substitutions, the following is obtained:

$$\dot{\epsilon} = \dot{\epsilon}_s \cdot e^{-\frac{\Delta G(\sigma, T)}{kT}} \quad (3.4)$$

where

$$\dot{\epsilon}_s = \alpha \cdot A \cdot \rho \cdot b \cdot \nu_0 \cdot s^* = \alpha \cdot \rho \cdot b \cdot v_s \quad (3.5)$$

where $v_s = A s^* \nu_0$ is a material property.

As the temperature is increased, the stress required to move the dislocations decreases, as temperature favors the overcoming of the energy barrier due to the thermal fluctuations. On the contrary, as strain rate is increased, the stress required to

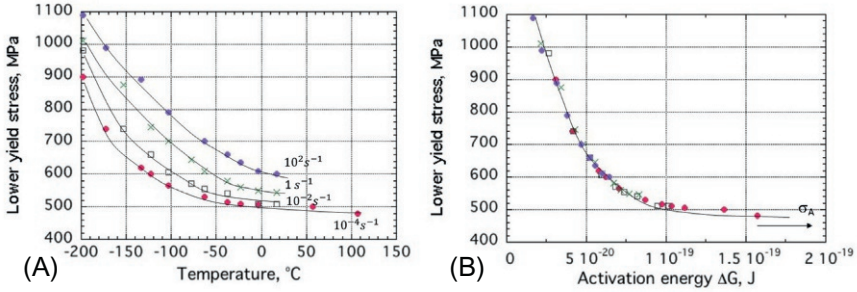


Fig. 3.4 (A) Yield strength as a function of temperature and strain rate for a normalized steel (composition: Fe-0.16%, C-0.3%, Si-1.36%, Mn-0.63%, Ni-0.16%V); (B) yield strength as a function of the free energy for dislocation motion.

Redrawn from Dahl, W., 1992. Mechanical properties, in: *Steel, A Handbook for Materials Research and Engineering*, Vol. 1, Verein Deutscher Eisenhüttenleute.

move the dislocations increases, as less time is allowed at dislocations to overcome the energy barrier. As seen in Fig. 3.1A, the influence of temperature (and also strain rate) is particularly pronounced in bcc metals for which dislocations must overcome large friction stresses in their motion.

To check the validity of the proposed model, the data in Fig. 3.4A are considered. They show the yield strength of a normalized C-Mn steel (with ferritic-pearlitic microstructure) as a function of temperature and strain rate (Dahl, 1992). The experimental data are then replotted in Fig. 3.4B using parameter ΔG in the x-axis. ΔG is obtained from Eq. (3.4):

$$\Delta G(T, \dot{\epsilon}) = -kT \cdot \ln \frac{\dot{\epsilon}}{\dot{\epsilon}_s} \quad (3.6)$$

setting $\dot{\epsilon}_s = 10^8 \text{ s}^{-1}$ ($\dot{\epsilon}_s$ is considered as a fitting parameter; note that $\dot{\epsilon}_s \gg \dot{\epsilon}$). Therefore $\ln(\dot{\epsilon}/\dot{\epsilon}_s) < 1$). It can be observed that all the experimental points fall on the same interpolating line. In addition, they asymptotically tend to a stress value that is independent of ΔG . With reference to Eq. (3.1) this asymptotic value is the athermal component of σ_Y , indicated with σ_A . To fit the data in Fig. 3.4B, a power law was proposed (Dahl, 1992):

$$\sigma_Y = \sigma_A + \sigma_T = \sigma_A + \sigma_T^* \cdot \left(1 - \frac{\Delta G}{\Delta G_0}\right)^M \quad (3.7)$$

where

- σ_T^* is σ_T for $T=0\text{K}$;
- ΔG_0 is ΔG for $\sigma_T=0$;
- M is a material constant.

For the steel in Fig. 3.4A, we obtain: $\sigma_A = 425$ MPa, $\sigma_T^* = 1025$ MPa, $\Delta G_0 = 2 \times 10^{-19}$ J, and $M = 3.65$.

3.2 Influence of temperature and strain rate on strain hardening

The analysis carried out in the previous section considers the yield strength, σ_Y , of metals. During the subsequent plastic deformation, strain hardening occurs, due to the increase in the dislocation density, ρ . In some cases, plastic deformation is not determined by dislocation slip only, but particular phenomena and microstructural transformations may play an important role. This complicates the picture greatly. The aim of the present section is to present and discuss the dependence of the strain-hardening coefficient, n , on temperature and strain rate for some selected alloys. It is a difficult task, since many factors are involved and some of them are also inter-related. The analysis is then simplified.

We initially focus attention on **bcc Fe** and **ferritic steels**. These metals prevalently deform by dislocation slip in this temperature interval. The contribution of twinning is significant only at very low temperatures and very high strain rates (Meyers et al., 2002), and thus is neglected in the present analysis.

The sensitivity of bcc metals to strain rate and temperature is rather marked, as shown by the noticeable influence of temperature and strain rate on the yield strength (see Figs. 3.1A and 3.4A). Fig. 3.5A shows the experimental dependence of n on temperature for Fe and different Fe-Ni alloys below RT (Chen et al., 1981). The tensile tests were carried out in quasi-static conditions, in the range 8.3×10^{-4} to $8.3 \times 10^{-3} \text{ s}^{-1}$. The coefficient n is seen to decrease as temperature is decreased.

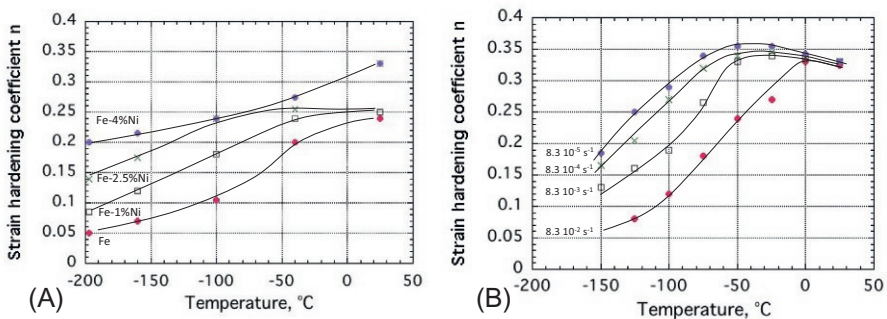


Fig. 3.5 (A) Experimental dependence of n on temperature for pure Fe and Fe-Ni alloys; (B) Experimental dependence of n on temperature for a C10 steel (0.1%C) at different strain rates.

A: Data from Chen, Y.T., Atteridge, D.G., Gerberich, W.W., 1981. Plastic flow of Fe-binary alloys - I. A description at low temperatures. *Acta Metall.* 29, 1171–1185; B: Redrawn from Dahl, W., 1992. Mechanical properties, in: *Steel, A Handbook for Materials Research and Engineering*, Vol. 1, Verein Deutscher Eisenhüttenleute.

To explain this, we start considering that a decrease in n is associated with the attainment of more intense saturation conditions (i.e., dislocation softening) during plastic deformation, as seen in the previous chapter. This could be realized by an increase in the dislocation density. This phenomenon was actually observed, and it was typically ascribed to the formation of more dense dislocation walls (Lee and Lin, 2001; Park and Niewczas, 2008). It can be explained within the model of thermally activated slip of dislocations, which assumes that a lower temperature and an increased strain rate make dislocation slip more difficult; therefore an increased dislocation density (obtained by the multiplication of dislocations during deformation) is required to achieve, in these conditions, a given plastic strain.

Fig. 3.5B shows the dependence of n on temperature and strain rate (in the range of 8.3×10^{-5} to $8.3 \times 10^{-2} \text{ s}^{-1}$) for a Fe-0.1%C steel below RT (Dahl, 1992). As for the Fe alloys in Fig. 3.5A, n is seen to decrease with decreasing temperature. In addition, n is seen to increase with increasing strain rate, also in agreement with the theory of thermally activated slip of dislocations.

Two other examples are shown in Fig. 3.6. The first refers to some mild ferritic steels (Hosford and Caddell, 1993). The second refers to a martensitic steel with 0.35%C (Lee and Lam, 1996). In both cases, the tests were conducted at RT. For both steels, the coefficient n decreases with strain rate. As expected, n values for the martensitic steel are much lower than those of the ferritic steels. The behavior described for Fe is generally observed in other bcc metals as well.

Approaching $0.4\text{--}0.5T_m$, the dependence of the flow stress on temperature and strain rate decreases in importance, since the flow stress becomes mainly dependent on its athermal component. As a result, the coefficient n is expected not to be very dependent on temperature and strain rate. The picture changes if special phenomena take place during plastic deformation. An important phenomenon is dynamic strain aging (DSA), discussed in Section 3.4. In the temperature interval characterized by

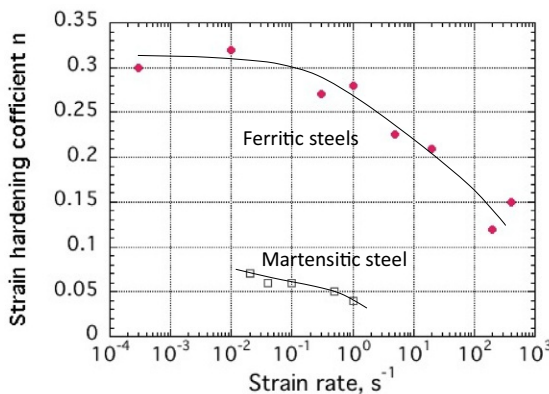


Fig. 3.6 The n values plotted as a function of strain rate for some ferritic steels (data from Hosford and Caddell, 1993; Rana et al., 2009) and a high-strength martensitic steel (data from Lee and Lam, 1996).

the DSA phenomenon ($0.15\text{--}0.4 T_m$), n may increase (Mishra et al., 1989). In addition, for temperatures close to the boundary with the creep region, coefficient n generally decreases, since incipient creep phenomena introduce a material softening.

A particular behavior is displayed by **fcc metals** with a high-moderate SFE. The plastic behavior of these metals is greatly influenced by the mechanisms that characterize the creep temperature region. Therefore coefficient n shows a particular experimental dependence on temperature and strain rate.

The creep mechanisms are outlined in Chapter 4. Here it is simply anticipated that *dynamic recovery* (DRV) plays a very important role at high temperature. DRV is easier in metals with a high SFE. It induces *thermal softening*, i.e., a decrease in the strain rate, $d\sigma/de$ with strain, and then in n . Therefore, when DRV comes into play, an increase in temperature induces a decrease in n (DRV increases in intensity), whereas an increase in the strain rate induces an increase in n (DRV has less time to occur).

One important case is that of Al alloys. As seen in Section 2.5, these metals are characterized by a high SFE and then DRV phenomena are very intense during plastic deformation. Fig. 2.17 shows that at RT, which corresponds to $\sim 0.4T_m$ and is then very close to the boundary with the high-temperature creep region, the instantaneous n value decreases with plastic strain because of thermal softening, and n at necking is lower than the average n value.

Consider now Fig. 3.7. It shows the true stress-true strain curves for an AA 6061 Al alloy (in the T4 state) at four temperatures from RT to -196°C (77 K), and at a strain rate of $6.6 \times 10^{-2} \text{ s}^{-1}$ (Schneider et al., 2015). A decrease in temperature induced an increase in the yield and flow strength and an increase in the uniform elongation, ϵ_u , and then in n . At RT the authors report an average n value close to 0.32, whereas ϵ_u (corresponding to n at necking) amounts to approximately 0.24, confirming the importance of thermal softening. The figure shows that the experimental ϵ_u values increase

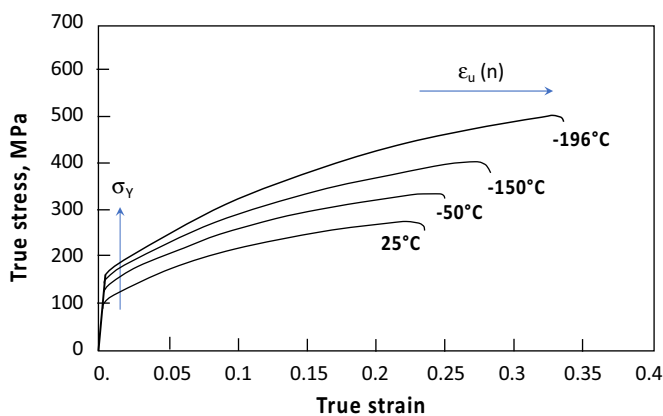


Fig. 3.7 True stress-true strain curves for an AA 6014 (T4) Al alloy at different temperatures, from 20°C to -196°C .

Redrawn from Schneider, R., Heine, B., Grant, R.J., Zouaoui, Z., 2015. Aluminium sheet metal forming at low temperatures, IOP Conf. Ser.: Mater. Sci. Eng. 74, 012014.

with decreasing temperature. At 77 K, ϵ_u and n are close to 0.35. This trend of n with temperature below RT was also observed for pure Al, for which n was found to increase from approximately 0.26 at RT to approximately 0.65 at -196°C (Hosford and Caddell, 1993). Similar trends were also observed in other Al alloys (Park et al., 2015).

As said, this special behavior can be explained considering the high SFE values of (fcc) Al alloys that strongly favor DRV phenomena. With decreasing temperature, the SFE decreases, and the intensity of such phenomena also decreases; this in turn induces an increase in ϵ_u and in the average n values.

A similar behavior is also displayed by pure Cu and Cu alloys characterized by intermediate SFE values, as shown in Table 2.1. Fig. 3.8 shows the engineering stress-strain curves of annealed Cu (with 0.028% P) and an average grain size of $51\text{ }\mu\text{m}$ (Copper Development Association Inc, 2020). Yield strength and elongation at necking (and also at fracture) increase with decreasing temperature. At RT (295 K) the average n value (estimated using the approach described in Fig. 2.27) is approximately 0.4 whereas ϵ_u is approximately 0.33, highlighting the role of thermal softening. As temperature is decreased, DRV phenomena are reduced in intensity, and then n and ϵ_u are seen to increase. It is noted that some twinning is generally observed (especially at low temperatures) and may contribute to further increasing the strain-hardening rate (Wang and Ma, 2003).

Also for fcc metals, approaching $0.4\text{--}0.5T_m$ the coefficient n becomes not very dependent on temperature and strain rate.

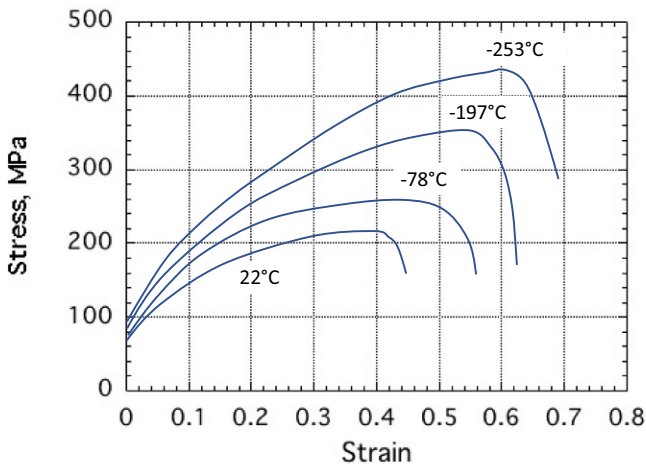


Fig. 3.8 Engineering stress-strain curves for annealed copper at different temperatures, from 22°C to -253°C .

Redrawn from Copper Development Association Inc., 2020. Application Data Sheet: Mechanical properties of Copper and Copper Alloys at Low Temperatures.

In a few cases, the scenario is more complicated due to specific phenomena that may occur during plastic deformation, like:

1. Dynamic strain aging (DSA), already mentioned and specifically addressed in [Section 3.4](#);
2. The transformation of austenite into martensite (TRIP effect): it occurs in fcc metals with a very low SFE ([Table 2.2](#));
3. Plastic deformation by twinning (TWIP effect): it is particularly important in metals characterized by difficult slip, such as fcc metals with moderate-low SFE and hcp metals, and is generally favored by high strain rates and low temperatures.

We first consider the important case of **metastable austenitic stainless steel**, which may display the TRIP effect during plastic deformation because of the comparatively low SFE. As seen in [Section 2.4](#), important metastable grades are AISI 304 and AISI 301 stainless steels.

The top part of [Fig. 3.9](#) shows the true stress-true strain curves of a 304 stainless steel at $5 \times 10^{-4} \text{ s}^{-1}$ and at different temperatures below RT ([Cios et al., 2017](#)). In the bottom part of the figure, the calculated strain-hardening rates, given by $d\sigma/d\varepsilon$, are also shown as a function of the true strain. The comparison of these curves with [Fig. 2.11A](#) shows that at RT (23°C) the TRIP effect is almost absent (or not sufficiently intense to produce modifications in the strain-hardening rate evolution; see also [Fig. 2.15](#)). However, its intensity increases markedly with decreasing temperature, because of corresponding decrease in the SFE that favors the TRIP effect. The TRIP effect induces a marked increase in the strain-hardening rate at the

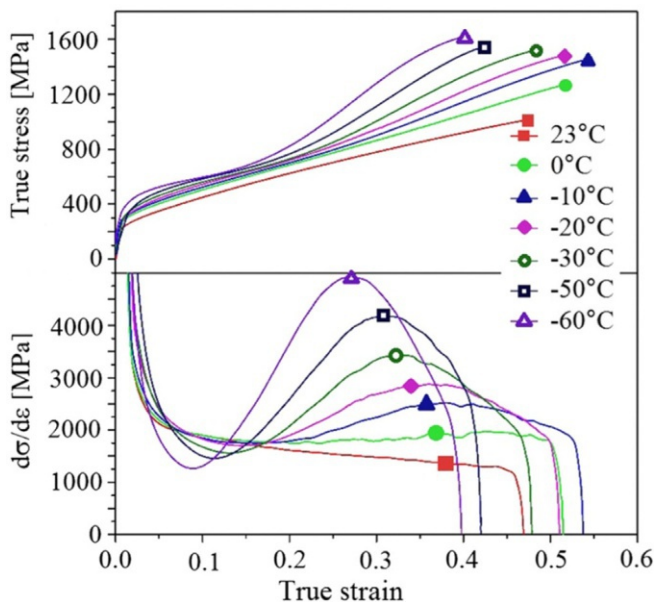


Fig. 3.9 True stress-true strain curves and corresponding strain-hardening rates at different temperatures for a 304 stainless steel ([Cios et al., 2017](#)).

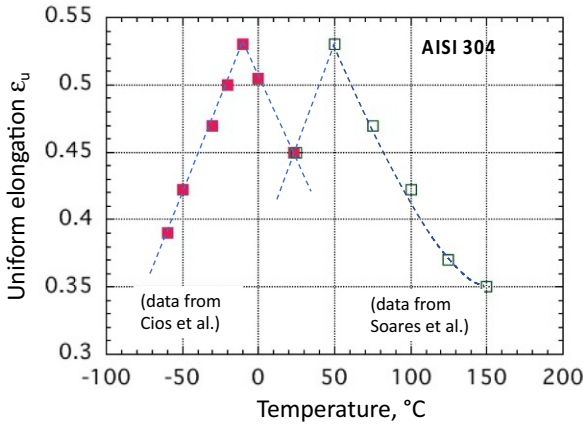


Fig. 3.10 Evolution of the uniform elongation ϵ_u with temperature for two AISI 304 austenitic stainless steels (data from Cios et al., 2017; Soares et al., 2017). The steel at the right side has a lower SFE.

intermediate strains. Subsequently, the strain-hardening rate drops following the formation of a large fraction of martensite, whose strain hardening rate is very low, as outlined in the previous chapter. At -60°C , for example, the instantaneous n value at the maximum strain rate (corresponding to $\epsilon \sim 0.27$) is greater than unity. Subsequently, n continuously decreases down to ~ 0.37 , at which necking conditions are attained.

A consequence of this behavior is the particular dependence of the uniform strain, ϵ_u , (corresponding to the n value at the onset of necking) on temperature. As depicted in Fig. 3.10 (left side for the steel of Fig. 3.9), a maximum is attained at -10°C . Below this temperature ϵ_u decreases with decreasing temperature since, as already mentioned, martensitic transformation gains in importance and this anticipates the drop in the strain-hardening rate. Above the maximum, ϵ_u decreases with temperature, because of the increase in the SFE and the progressive reduction in the martensitic transformation. The martensitic transformation is reported to almost disappear for temperatures in excess of 200°C . In Fig. 3.10, the data for another 304 steel are included (right side). The whole trend is similar even if the maximum of ϵ_u is shifted to the right, because of a lower SFE, due to the lower Mn content (Soares et al., 2017). These results show that the ϵ_u values are maximum in a temperature interval where the contribution of the martensitic transformation is intermediate.

Information on the influence of the strain rate on TRIP effect and on the corresponding strain-hardening behavior for type 301 steel (another important metastable austenitic stainless steel) is shown in Fig. 3.11 (Talyan et al., 1998). In the figure, the evolution of the martensite content due to the transformation of austenite is shown. At the strain rate of 10^{-3} s^{-1} , the strain-hardening rate increases remarkably during straining. This is due to the TRIP effect. However, as the strain rate is increased, the fraction of austenite transformation decreases, along with a decrease in strain hardening. This is attributed to the temperature rise during plastic

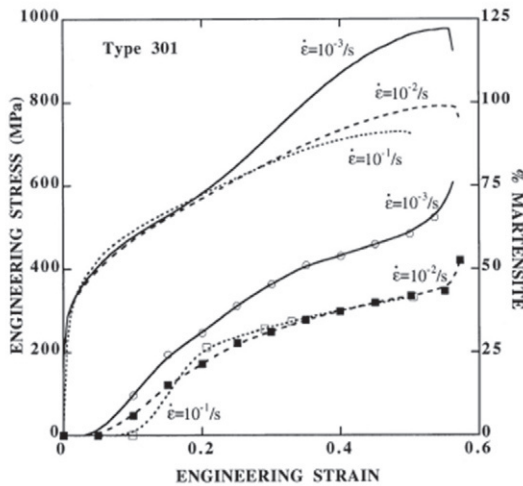


Fig. 3.11 Engineering stress-strain curves and (at the bottom of the graph) volume fraction martensite transformed vs. engineering strain for an austenitic 301 stainless steel. From Talyan, V., Wagoner, R.H., Lee, J.K., 1998. Formability of stainless steel. *Metall. Mater. Trans. A* 29, 2161–2172, reproduced with permission of the publisher.

deformation, induced by an increase in the strain rate (see [Section 3.5](#)). It increases the SFE, and then decreases the transformation rate to martensite.

The very common type 316 austenitic stainless steel is thermodynamically stable in a wide range of temperature, because of the comparatively high SFE ([Section 2.4](#)). However, the behavior may change in special conditions such as, for example, at rather low temperature. [Fig. 3.12](#) shows the true stress-true strain curves with the corresponding strain-hardening rates (dotted lines) for a 316LN stainless steel in the annealed condition ([Byun et al., 2004](#)). It can be seen that at -150°C the typical strain-hardening behavior of the TRIP/TWIP effect is present. As a matter of fact, the reduction in the SFE in passing to -150°C produced the formation of many twins from the beginning of plastic deformation (TWIP effect), along with some martensite regions (TRIP effect) when plastic deformation was sufficiently large.

Hadfield and TWIP steels (with high Mn content) undergo an intense TWIP effect during plastic deformation, since the chemical composition is balanced to have the right SFE for twinning. [Fig. 3.13](#) shows the true stress-true strain curves for a 22% Mn-0.6% C TWIP steel at different temperatures, from 77 to 673 K, i.e., -196°C to 400°C ([Bouaziz et al., 2011](#)). It is first observed that at temperatures close to RT, the curves display a serrated flow that is typical of the DSA phenomenon, considered in [Section 3.4](#). This phenomenon typically occurs in this kind of steel, together with the TWIP (and, sometimes, the TRIP) effect. It is also seen that the uniform strain, ϵ_u , decreases with increasing temperature, as indicated by the arrows. This is because the TWIP effect decreases in intensity with increasing the SFE. On the contrary, the TWIP effect generally increases in intensity with increasing strain rate, facilitating plastic deformation ([Liu et al., 2012](#)).

With increasing the strain rate above approximately 10^3 s^{-1} , a marked rise in the yield strength and in the strain-hardening coefficient, n , is observed ([Blazynski, 1987](#)).

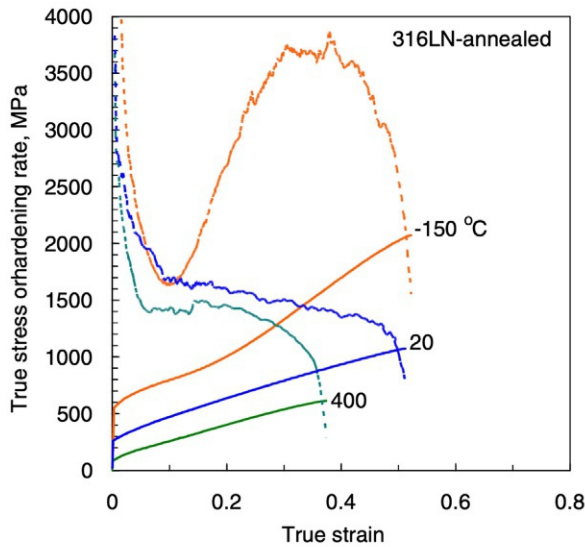


Fig. 3.12 True stress-true strain curves and corresponding strain-hardening rates (*dotted lines*) for a 316LN stainless steel in the annealed condition.

From Byun, T.S., Hashimoto, N., Farrell, K., 2004. Temperature dependence of strain hardening and plastic instability behaviors in austenitic stainless steels, *Acta Mater.* 52, 3889–3899, reproduced with permission of the publisher.

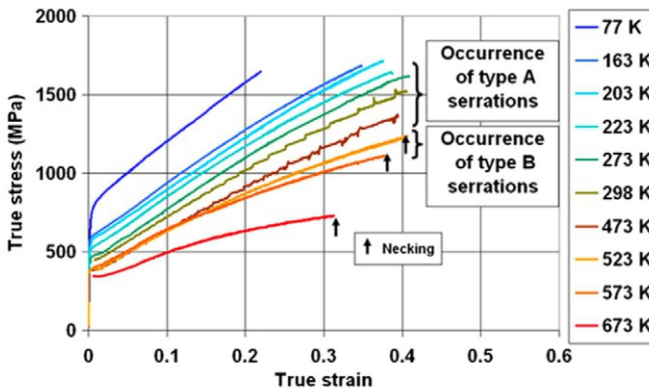


Fig. 3.13 True stress-true strain curves for a 22%Mn-0.6%C TWIP steel at different temperatures.

From Bouaziz, O., Allain, S., Scott, C.P., Cugy, P., Barbier, D., 2011. High manganese austenitic twinning induced plasticity steels: a review of the microstructure properties relationships. *Curr. Opin. Solid State Mater. Sci.* 15, 141–168, reproduced with permission of the publisher.

Coefficient n increases by about 50% or more, depending on the strain rate. For example, for the martensitic steel in Fig. 3.7, the coefficient n was found to increase from 0.04 at 1 s^{-1} , to 0.19 at 1000 s^{-1} (Lee and Lam, 1996). Different explanations have been proposed to justify this behavior. It is reported that for very high strain rates, no activation is required for dislocation motion, which is limited by the energy dissipated during the lattice distortions (damping effect). In addition, stresses and the strains are not uniform along the sample length, as the process has a short duration. Therefore the *inertial effects* due to the acceleration of any volume element of the sample cannot be neglected. These effects are claimed to diffuse plastic deformation over the sample gauge length. Finally, thermal softening effects due to material heating (see Section 3.5) may come into the picture and complicate the analysis a great deal (Regazzoni et al., 1987).

To summarize:

1. Bcc alloys, like ferritic steels, display a pronounced sensitivity to temperature and strain rate; correspondingly, the strain-hardening coefficient, n , decreases with decreasing temperature and increasing strain rate; however, when temperature increases and approaches $0.4\text{--}0.5T_m$, n becomes rather insensitive to temperature and strain rate (provided that special phenomena, such as DSA, do not occur); this behavior is indicated with (A) in Fig. 3.14;
2. Fcc alloys display a limited sensitivity to temperature and strain rate; in some alloys with high-intermediate SFE, including Al and Cu alloys, n is seen to increase with decreasing temperature and increasing strain rate, since incipient DRV phenomena (typical of the creep region) are progressively suppressed; this behavior is indicated with (B) in Fig. 3.14;
3. Fcc alloys with low SFE (such as the metastable stainless steels) display TRIP/TWIP phenomena during plastic deformation; ϵ_u and n display maximum values when such phenomena have a balanced intensity; this behavior is indicated with (C) in Fig. 3.14;
4. In passing to very high strain rates (beyond 10^3 s^{-1}), the strain-hardening coefficient is seen to significantly increase.

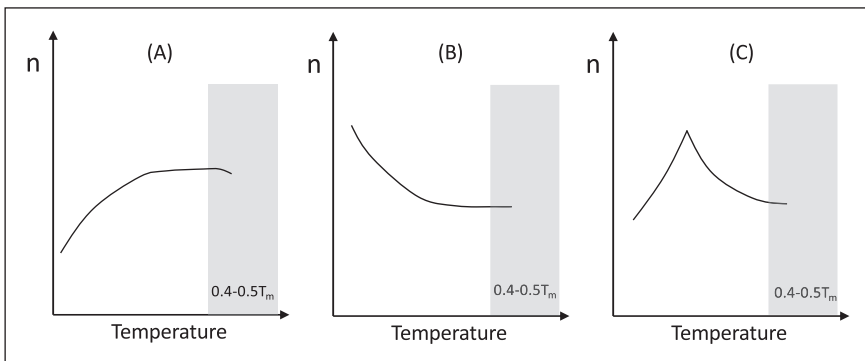


Fig. 3.14 Schematizations showing the three typical trends of the average strain-hardening coefficient, n , with temperature in the region below $0.4\text{--}0.5T_m$.

3.3 Strain rate sensitivity

The flow stress of metals typically increases with the strain rate $\dot{\epsilon}$. This is shown as an example in Fig. 3.4A. In general, the experimental data can be simply fitted using a power equation:

$$\sigma \propto \dot{\epsilon}^m \quad (3.8)$$

where exponent m is also called the *strain rate sensitivity (SRS) coefficient or index*.

Coefficient m depends on the material, temperature, and strain rate interval. In addition, it may also vary with the plastic strain, and then reference to an *average value* is often made. In general, coefficient m assumes values lower than 0.1 for temperatures below $0.4\text{--}0.5T_m$. From a physical viewpoint, m is related to the difficulty of dislocations to surmount short-range stress field obstacles. This difficulty increases with strain rate as dislocations have less time for successful jumps: high m values hence signify a high difficulty for dislocations in overcoming such obstacles (with the corresponding increase in the applied stress required for slip).

In bcc alloys, such as in ferritic steels, coefficient m assumes comparatively high values, because of the pronounced effect of the friction stress that makes slip difficult. A typical value at RT is 0.07. However, m is seen to decrease with increasing yield strength. In Fig. 3.15A, the experimental average m values for a set of ferritic steels are plotted as a function of the yield strength. The comparison with the n values discussed in Section 2.3 (see in particular Fig. 2.13) shows that, to a first approximation, the m and n values in bcc steels are proportional. This is shown in Fig. 3.15B for some typical examples.

Coefficient m assumes comparatively low values in fcc alloys because of the lower energy barriers in dislocation slip. For Al, m is close to 0.005, and for a 6082 Al alloy (in T6 state), it is 0.008 at RT. For Cu, m is also below 0.01 at RT. Commercially pure Ti (with hcp structure) displays an intermediate behavior. Coefficient m is 0.032 at RT; it decreases to 0.005 at -196°C and increases to 0.034 at 200°C (Sun et al., 2016). In fcc metals the relationship between m and n is not clear, since both these parameters may appreciably change with strain.

From an experimental point of view, m is often simply determined by a tensile test and by increasing the strain rate in a stepwise manner, in the so-called *jump test*, schematized in Fig. 3.16. The coefficient m is then given by:

$$m = \frac{\log\left(\frac{\sigma_1}{\sigma_2}\right)}{\log\left(\frac{\dot{\epsilon}_1}{\dot{\epsilon}_2}\right)}, \text{ where } \sigma_1 \text{ and } \sigma_2 \text{ are}$$

the flow stresses at the strain rates $\dot{\epsilon}_1$ and $\dot{\epsilon}_2$, respectively. In this way, it is also possible to obtain the dependence of m on the plastic strain.

As already observed in the previous section, a particular strain-hardening behavior is displayed by metals that exhibit special transformations during plastic deformation, in specific temperature/strain rate intervals. In these cases, the corresponding m values strongly depend on strain, strain rate, and temperature. Let us consider the important case of metastable austenitic stainless steels. They

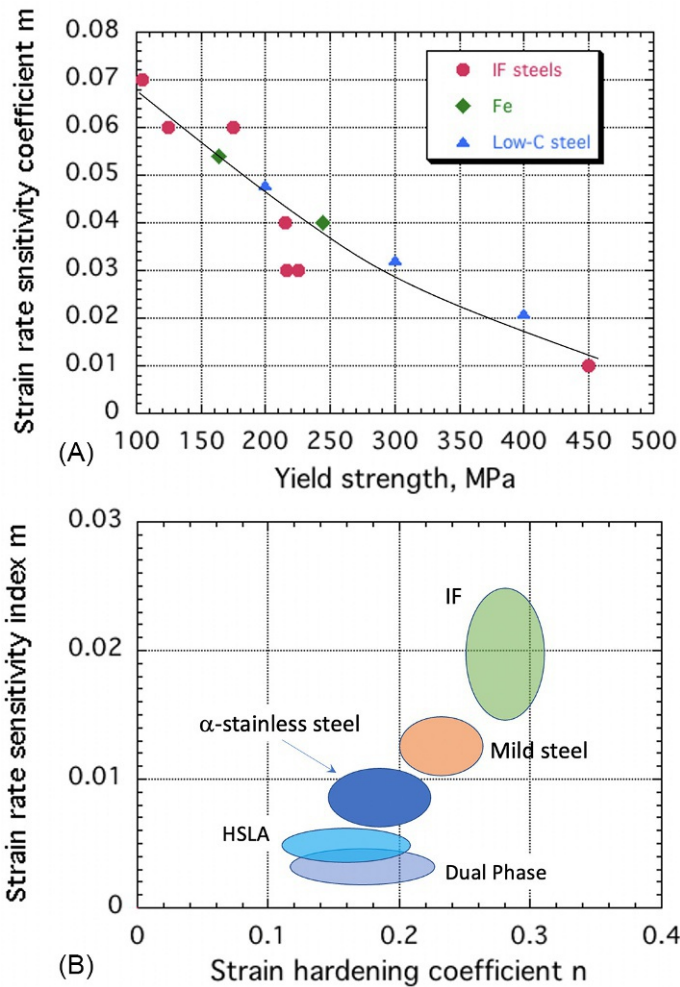


Fig. 3.15 (A) Average values of m vs. yield strength for a set of special IF steels, pure Fe, and low-C steels (data from [Rana et al., 2009](#); [Li et al., 2017](#); [Hosford and Caddell, 1993](#)); (B) schematic relationship between n and m for some industrial bcc steels.

have an fcc structure and then typically display low m values, in the range of 0.015–0.02. However, the m values may further decrease and become negative during plastic deformation, if the TRIP effect comes into play. The curves in [Fig. 3.11](#) show that at large strains (after $\epsilon \approx 0.2$), the stress decreases with increasing strain rate, and hence m is negative in this strain interval. A specific example is shown in [Fig. 3.17](#), which presents the dependence of m on plastic strain at different temperatures for an AISI 316L stainless steel ([Park et al., 2011](#)). At RT m becomes negative at an engineering strain of $\epsilon \approx 0.25$. The figure also shows that

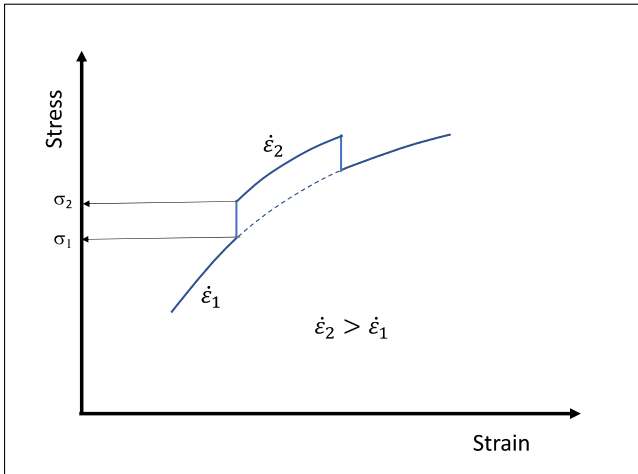
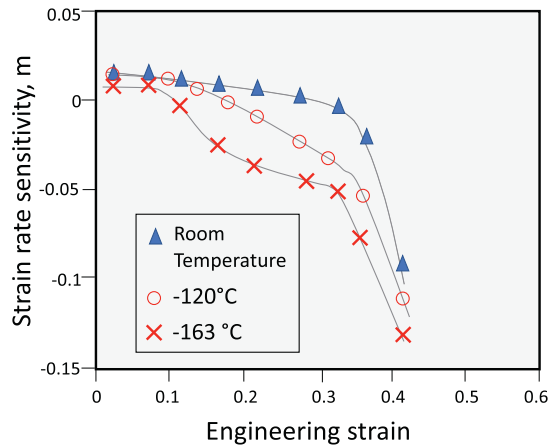


Fig. 3.16 Tensile jump-test to estimate the strain rate sensitivity coefficient.

Fig. 3.17 Experimental dependence of the strain rate sensitivity index, m , on plastic strain for an AISI 316L stainless steel tested at different temperatures.

Data from Park, W.S., Chun, M.S., Han, M.S., Kim, M.H., Lee, J.M., 2011. Comparative study on mechanical behavior of low temperature application materials for ships and offshore structures: part I-experimental investigations. Mater. Sci. Eng. A 528, 5790–5803.



the engineering strains at which m becomes negative decrease with decreasing temperature.

A very important situation occurs in metals that undergo dynamic strain aging (DSA) during plastic deformation. As seen in the next section, this phenomenon can also be responsible for negative m values.

For strain rates above approximately 10^3 s^{-1} , the flow stress markedly increases with strain rate. Therefore the coefficient m significantly increases with strain rate (El-Magd et al., 2006). This means that for very high strain rates, the flow stress increases with strain (because of the increase in n) and with strain rate (because of the increase in m). As said, however, the analysis is quite complicated due to the concurrent presence of inertial as well as thermal effects.

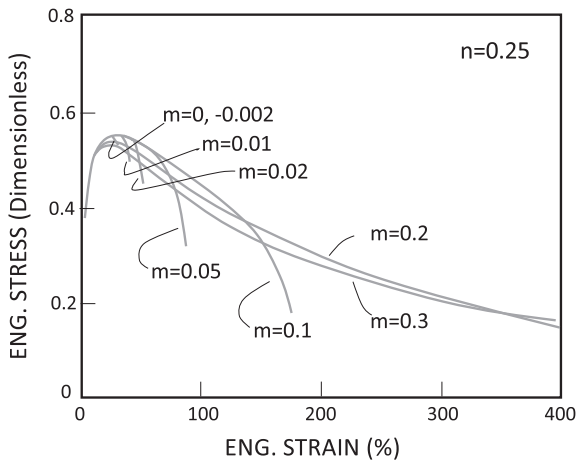


Fig. 3.18 Engineering stress-strain curves from FEM for $n = 0.25$ and different m values. Redrawn from Chung, K., Wagoner, R.H., 1988. Effects of work-hardening and rate sensitivity on the sheet tensile test. Metall. Mater. Trans. A 19, 293–300.

When a single m value adequately characterizes the influence of strain rate on the flow stress, at least in specific temperature and strain rate intervals, the use of Eq. (3.8) is very useful, and m has an important physical meaning. The simulated curves in Fig. 3.18 were obtained using a finite element model (FEM) of a tensile specimen, with a strain-hardening coefficient of 0.25 and different m values, from -0.002 up to 0.3 (Chung and Wagoner, 1988). In all cases, necking is seen to occur for a true strain close to 0.25 (in agreement with the Considère criterion, as the influence of m on necking is comparatively small). However, the flow behavior after necking is seen to markedly depend on m . For low m values, the load drops quickly after necking, implying that the strain localization in the neck region is very pronounced. But, *as m is increased, the strain localization becomes increasingly difficult*. This can be explained with reference to the schematization in Fig. 3.19. As the strain rate increases

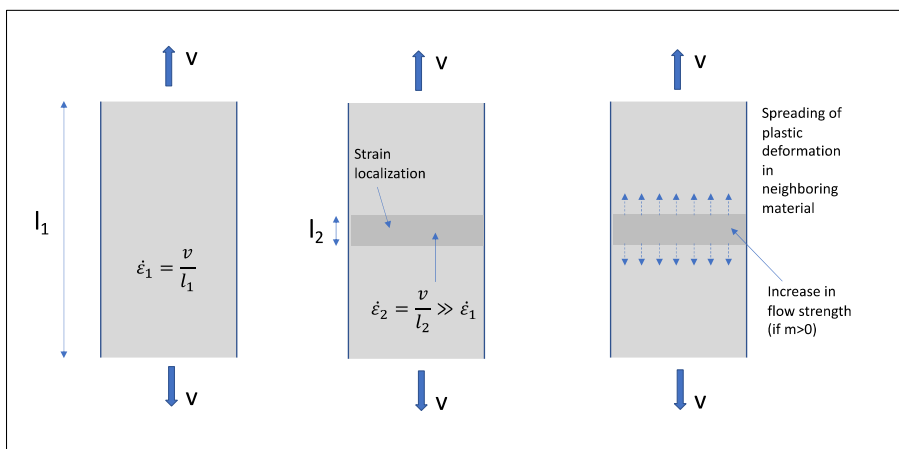


Fig. 3.19 Schematization showing the role of positive m values in retarding plastic strain localization.

in the region of incipient localization, the stress required for the local increase in plastic deformation also increases *depending on m* . If m is high, such an increase is also high. Therefore localization (with the corresponding strain intensification) is reduced, as neighboring material is involved in the deformation process.

For very high m values ($m > 0.2$), an upward concavity in the stress-strain behavior is obtained (up to a certain strain), resulting in the slow development of strain localization. Such high m values can be attained at high temperatures (close to or beyond approximately $0.4\text{--}0.5T_m$) and at very high strain rates.

To summarize:

- For temperatures below $0.4\text{--}0.5T_m$, the strain rate sensitivity coefficient, m , assumes values lower than 0.1;
- Bcc metals display m values greater than fcc metals; in bcc steels, the m and n values are approximately proportional;
- Coefficient m may assume negative values when metals display phase transformations during plastic deformation (such as the TRIP or TWIP effects);
- Coefficient m assumes high values in the case of very high strain rates, above approximately 10^3 s^{-1} ;
- With increasing m , strain localization is made more difficult.

3.4 Dynamic strain aging (DSA)

A *serrated pattern* in the tensile stress-strain curve in specific intervals of temperature and strain rate is typically observed in a number of metals, including steels and Al alloys. Fig. 3.13 shows some examples for a TWIP steel (refer to the curves close to RT); Fig. 3.20 shows another example of this behavior for an Al-4.5%Mg alloy (a non-heat treatable Al-5182 alloy), subjected to tension test at RT (Zadra et al., 2008). This phenomenon is called the *Portevin-Le Chatelier (PLC) effect* or *dynamic strain aging (DSA)* and is analogous to the upper and lower yield-point phenomenon, described in Section 1.4. Often DSA is associated with this phenomenon, as can be appreciated by a closer look at the stress-strain curve of Fig. 3.20 at small strains.

As for the Lüders bands, the DSA effect induces a surface quality degradation of the metal. In fact, DSA leads to the formation of numerous band-type markings, which are aesthetically unacceptable, especially in sheet formed parts. Fig. 3.21 shows the development of such markings with time in an Al sheet sample subjected to tensile testing (Franklin et al., 2000). In this case, the test was conducted with a displacement velocity on the order of 0.1 mm/s, whereas the markings front was observed to travel with a velocity on the order of 1 cm/s. In general, it is experimentally observed that markings front velocity, v_f , is related to the sample displacement velocity, v , as:

$$v_f = \frac{v}{\delta\epsilon} \quad (3.9)$$

where $\delta\epsilon$ is the strain difference across the front.

For the onset of the yield point and DSA phenomena with the corresponding formation of bands, a critical plastic strain must be reached. Successively, the

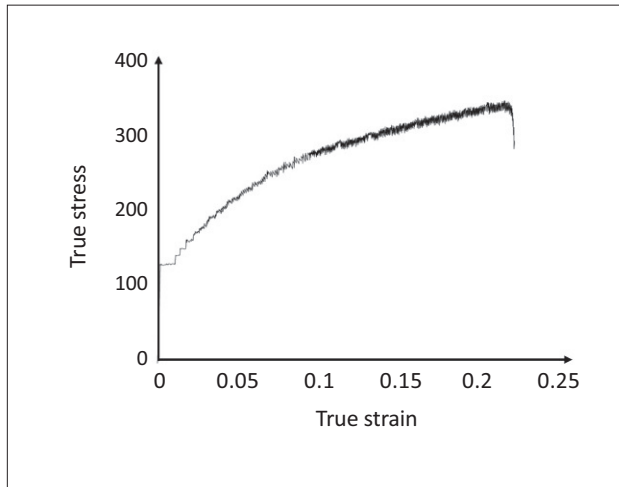


Fig. 3.20 True stress-true strain curve for an Al-4.5%Mg alloy, with the serrated flow. From Zadra, M., Straffelini, G., Molinari, A., 2008. Tensile ductility of AA5182: effect of impact loading. *Mater. Sci. Technol.* 24, 1259–1264, reproduced with permission of the publisher.

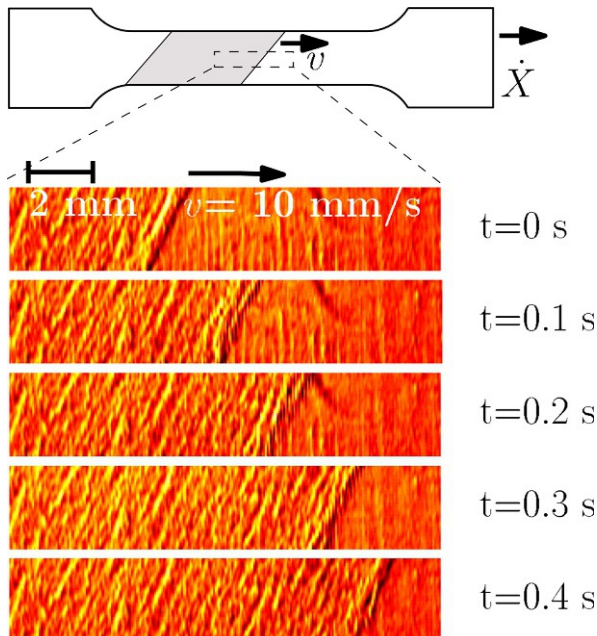


Fig. 3.21 Propagation of band markings during tensile testing of an Al alloy. From Franklin, S.V., Mertens, F., Marder, M., 2000. Portevin-Le Chatelier effect. *Phys. Rev. E* 62, 8195–8206, reproduced with permission of the publisher.

macroscopic strain during tensile deformation increases due to the spreading of these bands along the sample length. Correspondingly, an instantaneous stress drop occurs during the formation of a new band. However, different from the yield-point phenomenon, in DSA the applied stress must be continuously increased to propagate the surface bands. Some researchers argue that the two phenomena are similar, and the difference is due to the combination of displacement velocity and bands front velocity, which depends on the material, temperature, and strain rate.

The DSA phenomenon is usually explained by considering the role of solute atoms. Examples are N and C atoms in ferritic steels, and Mg atoms in Al alloys. If their diffusion rate is higher than the velocity of the dislocations, they can exert pinning of mobile dislocations. Therefore the serrated pattern is the result of the continuous anchoring and release of such dislocations by the solute atoms present in the crystal lattice. Hence, the DSA phenomenon is influenced by chemical composition and microstructure. For instance, in steels, Mn forms Mn-C and Mn-N pairs that reduce the mobility of interstitial atoms, which are responsible for DSA through dislocation pinning (Akghar and Serajzadeh, 2012). In addition, the presence of microstructural obstacles that increase the density of geometrically necessary dislocations may also cause difficulty in dislocation pinning. For example, DSA intensity was observed to be reduced in steels with high content of pearlite (Hall, 1970). However, the DSA phenomenon is also observed in pearlitic steels, being mainly due to the dislocation locking exerted by C atoms dissolved in the ferrite lamellae by the decomposition of cementite (Gonzales et al., 2003).

As mentioned, DSA occurs at specific intervals of temperature and strain rate. For the TWIP steel of Fig. 3.13, such an interval is approximately RT–200°C

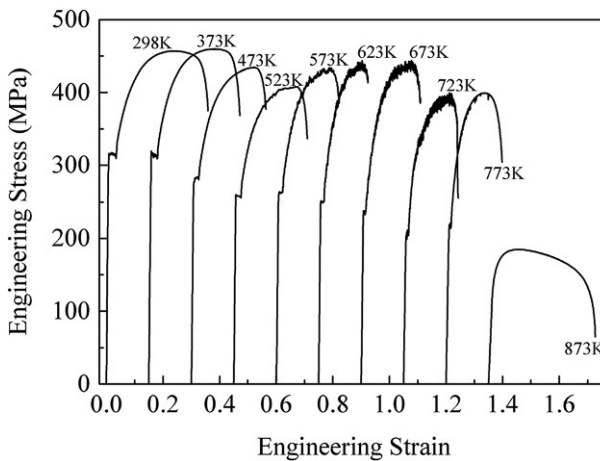


Fig. 3.22 Engineering stress-strain curves recorded from 298 K (20°C, i.e., RT) up to 873 K (600°C) at 10^3 s^{-1} for a ferritic stainless steel.

From Soares, G.C., Quiroz, R.R.U., Santos, L.A., 2020. Effects of dynamic strain aging on strain hardening behavior, dislocation substructure, and fracture morphology in a ferritic stainless steel. *Metall. Mater. Trans. A* 51, 725–739, reproduced with permission of the publisher.

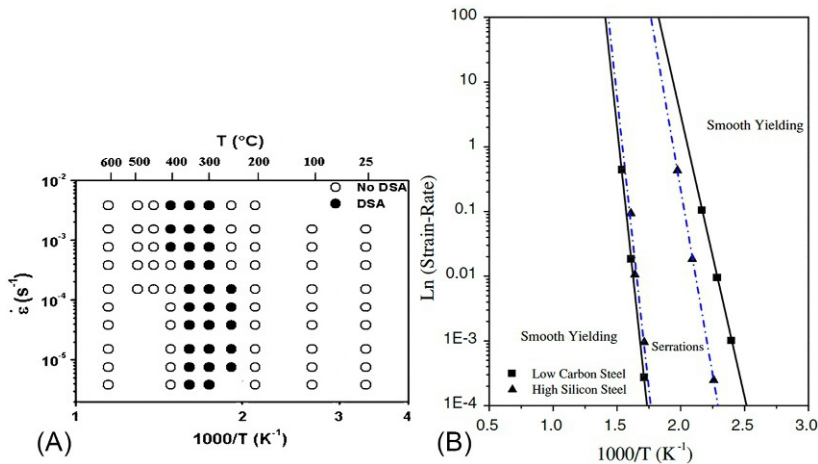


Fig. 3.23 Maps showing the temperature and strain rate domain, where DSA is occurring for (A) a ferritic Cr-Mo-V steel, and (B) two ferritic low carbon and high silicon steels. From Gupta, C., Chakraborty, J.K., Banarjee, S., 2010. Dynamic strain aging in new generation Cr-Mo-V steel for reactor pressure vessel applications. *Metall. Mater. Trans. A* 41, 3326–3339; Akhgar, J.M., Serajzadeh, S., 2012. Flow behavior and mechanical properties of a high silicon steel associated with dynamic strain aging. *J. Mater. Eng. Perform.* 21, 1919–1923, reproduced with permission of the publisher.

(0.17 – $0.27T_m$). Fig. 3.22 shows the engineering stress-strain curves carried out from RT up to 600°C at 10^{-3} s^{-1} for a ferritic stainless steel (grade 430; Soares et al., 2020). In this case, the serrated flow occurs in the temperature interval between 250°C and 500°C (0.28 – $0.4T_m$). At sufficiently low temperatures, the dislocation pinning does not occur, since, after the upper and lower yield point, the interstitial atoms do not have enough diffusion rate to pin the moving dislocations. With increasing temperature, the solute diffusivity increases, and at a critical temperature, DSA begins. However, if the temperature becomes too high, plastic deformation becomes very easy. This means that the dislocations slip with high velocity, and solutes do not exert a pinning effect due to their insufficient diffusion rate.

In the two maps of Fig. 3.23, the conditions for DSA are shown for a ferritic Cr-Mo-V steel, and two ferritic high silicon (HS) and low carbon (LC) steels. In general, with an increase in the strain rate, the temperature interval for DSA is shifted towards higher temperatures, since atoms have less time to diffuse to dislocations during plastic deformation. This induces a less effective pinning action. The transition to the DSA interval (and its exit at a higher temperature) may be expressed by an Arrhenius-type equation:

$$\dot{\epsilon} = A \cdot \exp\left(-\frac{Q}{k \cdot T}\right) \quad (3.10)$$

Table 3.2 DSA appearance temperature at $\sim 10^{-3} \text{ s}^{-1}$ for some metals considered in the present chapter.

Metal	T for DSA at 10^{-3} s^{-1}	Average m in the critical interval	Reference
5182-O Al-alloy	$\sim -45^\circ\text{C}$	-0.015	Picu et al. (2005)
TWIP steel	$\sim 25^\circ\text{C}$	-0.007	Chung et al. (2011)
Low carbon (LC) steel	$\sim 130^\circ\text{C}$	-0.01	Akghar and Serajzadeh, 2012
Ferritic high silicon (HS) steel	$\sim 170^\circ\text{C}$	-0.01	Akghar and Serajzadeh, 2012
Pearlitic steel	$\sim 175^\circ\text{C}$	-0.015	Gonzales et al. (2003)
Ferritic Cr-Mo-V steel	$\sim 280^\circ\text{C}$	-0.01	Gupta et al. (2010)

where Q is the activation energy for the DSA process and A is a material constant (further details on A are given in Section 14.3). For the ferritic low-C steel, Q is 72.2 kJ/mol and very close to the activation energies for N and C diffusion in ferrite (74 and 85 kJ/mol). This demonstrates the very likely role of both interstitials in the process. It is also seen that Q increases with alloying content, thus justifying the increase in temperature for the onset of the process. For the Al-Mg alloys (class 5000), the DSA phenomenon is generally attributed to the diffusion of Mg, and it occurs typically in the temperature range -50 to 100°C (0.25 to $0.43T_m$). Note that in the Al-Mg-Si alloys (class 6000) the DSA phenomenon is not detected, since Mg forms intermetallic compounds with Si, and then it is not available in solid solution.

In Table 3.2, the DSA appearance temperature at 10^{-3} s^{-1} is reported for different alloys. In general, this phenomenon is observed to occur for temperatures in the broad interval 0.15 – $0.4T_m$.

With the presence of serrations, the metal flow strength increases, as the dislocation slip is made more difficult. This is shown in Fig. 3.24. Here, the flow stress (at a true strain of 0.1) of a Mn-C steel (with 0.14%C, 1.37%Mn) is plotted as a function of temperature and for three different strain rates. It is seen that the flow stress does not follow the reference trend with the temperature displayed in Fig. 3.1, but the stress is seen to increase in the DSA region (in the range of 500–1000 K). In addition, with increasing strain rate, the critical interval for DSA is shifted to higher temperatures and the corresponding peak stress comparatively decreases (because of the increased temperature). This means that within the critical temperature interval, the flow stress decreases as the strain rate is increased, resulting in a negative value of m .

Fig. 3.25 shows the average experimental m values as a function of temperature for the two steels from Fig. 3.23B and for a 5182-O Al alloy (Picu et al., 2005). In the temperature intervals where DSA occurs, m values as low as -0.04 and -0.08 are attained for the HS and LC steels, and as low as -0.02 for the Al alloy. For the steel

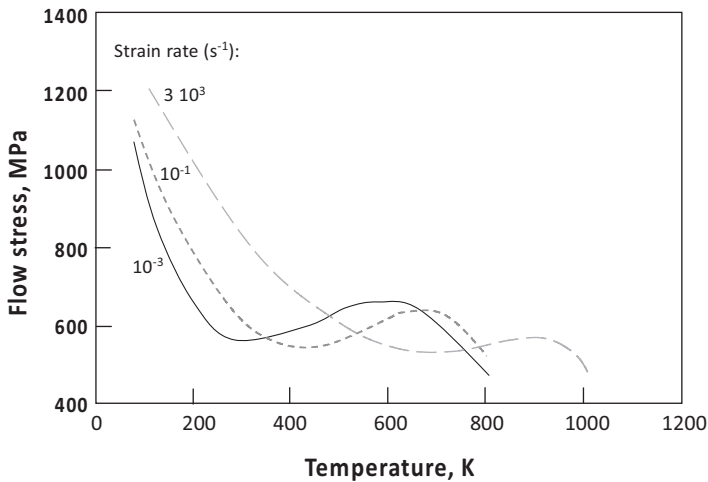


Fig. 3.24 Effect of strain rate and temperature on the flow stress (at a true strain of 0.1) for a C-Mn steel (0.14%C, 1.37%Mn).

Redrawn from Nemat-Nasser, S., Guo, W.G., 2003. Thermomechanical response of DH-36 structural steel over a wide range of strain rates and temperatures. *Mech. Mater.* 35, 1023–1047.

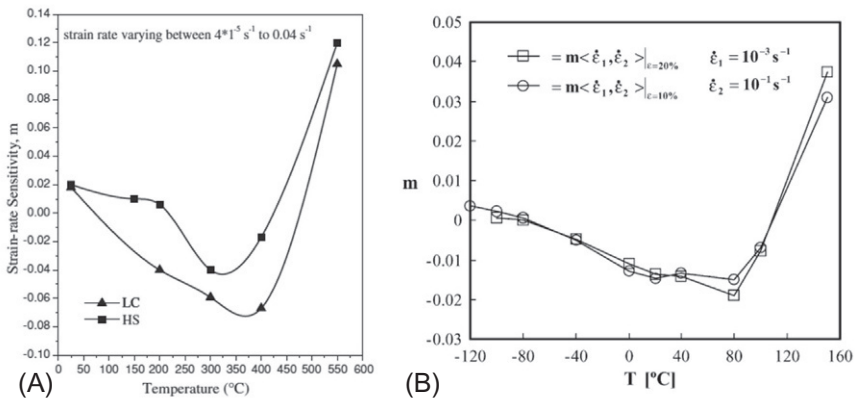


Fig. 3.25 Dependence of SRS index on temperature for (A) the two steels from Fig. 3.23B and for (B) a 5182-O Al alloy.

A: From Akhgar J.M., Serajzadeh S., 2012. Flow behavior and mechanical properties of a high silicon steel associated with dynamic strain aging. *J. Mater. Eng. Perform.* 21, 1919–1923, reproduced with permission of the publisher; B: From Picu, R.C., Vincze, G., Ozturk, F., Gracio, J.J., Barlat, F., Maniatty, A.M., 2005. Strain rate sensitivity of the commercial aluminium alloy AA5182-O. *Mater. Sci. Eng. A* 390, 334–343, reproduced with permission of the publisher.

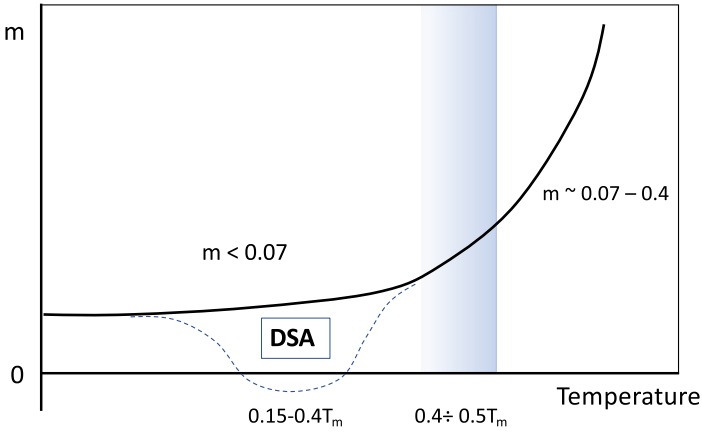


Fig. 3.26 Simple representation of typical m dependence on temperature and the role of DSA.

of Fig. 3.23A, m values as low as -0.01 are reported in the corresponding temperature interval. Some typical mean m values in the critical intervals are included in Table 3.2. The presence of negative m values plays an important role in the plastic deformation and ductile fracture of a metal, as outlined in Chapters 6 and 7.

To summarize:

- DSA is observed in different alloys, including some steels and Al-Mg alloys (class 5000); it is characterized by serrated evolution of the flow stress in a temperature interval $0.15-0.4T_m$;
- As schematized in Fig. 3.26, DSA induces a decrease in the strain rate sensitivity coefficient, m , which can also assume negative values.

3.5 Plastic deformation and temperature rise

Plastic deformation is an irreversible process and most of the plastic work is converted into heat. This occurs at every temperature, i.e., above and below $0.4-0.5T_m$. Upon increasing the strain rate, the deformation process occurs faster, and heat has a shorter time to diffuse away from the region where it is generated. If the heat remains concentrated in the generation region, the process is *adiabatic*. Consequently, the local temperature rise can be rather high.

A useful parameter to evaluate the adiabaticity of a process is the *thermal diffusion distance*, d , defined by the following relationship:

$$d = \sqrt{a \cdot t} \quad (3.11)$$

where

- a is the thermal diffusivity of the material (given by: $k/\rho c$, where k is the thermal conductivity, ρ is density, and c is the specific heat);
- t is the deformation time.

For example, if a steel sample ($a = 17.5 \times 10^{-6} \text{ m}^2/\text{s}$) with a diameter of 10 mm is plastically strained to $\varepsilon = 0.5$ at a strain rate $\dot{\varepsilon} = 10^{-3} \text{ s}^{-1}$ (quasi-static loading), Eq. (3.11) gives $d = 93.5 \text{ mm}$. Therefore d is greater than the sample dimension, and it can be understood that the heat has enough time to diffuse to a longer distance. However, if $\dot{\varepsilon} = 10^3 \text{ s}^{-1}$ (impact loading), d becomes 0.09 mm, which is much less than the sample dimension. Hence, a large heat quantity does not diffuse that far away from the point where it is generated. In Table 3.3, representative thermal properties of some alloys are listed. It can be clearly seen that the thermal diffusivity is comparatively low in Ti alloys and in austenitic stainless steels, and for these materials, an adiabatic process occurrence is more likely. This is also shown in Fig. 3.27 where the calculated thermal diffusion distance, d , is plotted as a function of strain rate for a plastic strain $\varepsilon = 0.5$.

In deformation processes close to adiabatic conditions, the temperature rise can be high and this, in turn, influences the plastic deformation. In fact, in thermally activated processes (involving dislocation slip and also dislocation climb, as seen in the next

Table 3.3 Density and representative thermal properties of some alloys.

Material	Density (kg/m^3)	Specific heat (J/kg K)	Thermal conductivity (W/mK)	ρc (MJ/Km^3)	$a \times 10^6$ (m^2/s)
Low-C steel	7800	440	60	3.43	17.5
316L SS	8000	490	14	3.92	3.6
Al alloys	2700	860	165	2.32	71
Cu alloys	8700	380	80	3.3	24.2
Ti alloys	4500	560	8	2.5	3.2

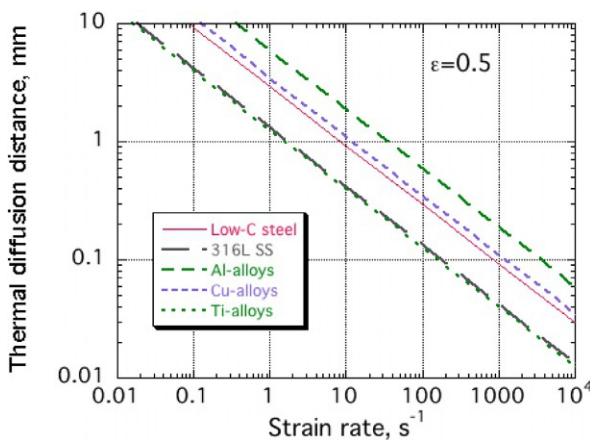


Fig. 3.27 Calculated thermal diffusion distance for $\varepsilon = 0.5$ and for the alloys listed in Table 3.3.

chapter), a temperature rise can reduce the flow stress, and then induce the so-called *thermal softening* of the metal.

In adiabatic conditions, the temperature rise, ΔT , is given by:

$$\Delta T = \beta \frac{u}{\rho c} \quad (3.12)$$

where

- β is the fraction of plastic work that is effectively dissipated into heat (in general, β is between 0.95 and 1; the rest of the energy is used in producing new defects such as dislocations, or to promote phase transformations);
- ρ and c are the density and heat capacity of the metal;
- u is the specific dissipated energy (energy for unit volume).

The specific energy u can be calculated using the following relationship:

$$u = \int_0^{\epsilon} \sigma d\epsilon = \bar{\sigma} \epsilon \quad (3.13)$$

where

- ϵ is the total true plastic strain;
- $\bar{\sigma}$ is the mean stress, between $\epsilon=0$ and ϵ .

When using the L-H equation, we obtain:

$$\bar{\sigma} = \frac{K \epsilon^n}{n+1} \quad (3.14)$$

For the calculation of the mean stress, it must be considered that the flow stress also depends on temperature, and generally decreases with it. It should then be evaluated using one of the constitutive equations reported in the [Section 3.7](#).

In the case of laboratory testing, temperature rise due to plastic deformation should be appropriately considered in the measurement of specific properties at a given temperature, even at comparatively low strain rates. In the case of temperatures at or below RT, a cooling system is recommended. One way is to cool the specimen using forced convection, such as flowing water. In the case of high temperature testing (refer to the next chapter), the sample temperature is continuously monitored with a contact thermocouple, and a retrofitting system is used to calibrate the heating power. This is done with the aim of maintaining a constant sample temperature during the test.

3.6 Formation of shear bands

The curves in [Fig. 3.18](#) were numerically obtained by considering a model material with a homogeneous microstructure. As seen, the curves show a sudden drop of load after necking, particularly when m is close to 0 or negative. The combination of

negative m values with a heterogeneous microstructure may induce important modifications in the deformation behavior of metals. In these cases, plastic deformation tends to localize in the imperfections, due to the action of local *shear* stress that is maximum at 45 degrees from the loading direction (see Fig. 1.8B). Correspondingly, the strain rate increases in these regions and becomes much higher than the average value (refer to Fig. 3.19). In the case of negative m values, the flow resistance of the material decreases, and this increases the strain localization, inducing an avalanche effect. This phenomenon is further exacerbated by local temperature rise due to localized plastic deformation, and the associated thermal softening. As a result, plastic deformation localizes in the so-called *shear bands*.

In addition to coefficient m , the strain-hardening coefficient, n , also plays an important role in favoring or impeding the localization of plastic deformation in shear bands. In fact, if n is high, the material can locally strain harden and this diffuses plastic deformation in the neighboring regions. This reduces or strongly limits any further development of strain localization. On the contrary, if n is low, the local strain hardening may not be sufficient to avoid strain localization (even if the m value is positive). Therefore, coefficients m and n play a similar role in strain localization (n expresses the flow stress increase as the strain is increased; m expresses the flow stress increase as the strain rate is increased), as schematized in Fig. 3.28.

An important example is that of Al-Mg alloys (class 5000). Fig. 3.29 shows the results of a finite element modeling of the evolution of plastic strain during the tensile test of an Al 5182 aluminum alloy sheet with $m = -0.0172$ and average n value close to 0.26, decreasing with strain because of thermal softening (Straffelini et al., 2013). In the model, the presence of soft phases in the microstructure (the imperfections) was considered, and the development of heavy shear localization in these phases is clearly predicted. In Fig. 3.29C, localized necking is shown (compare with Fig. 2.30) and it occurs for a true strain, ϵ , lower than $2n$. The shear

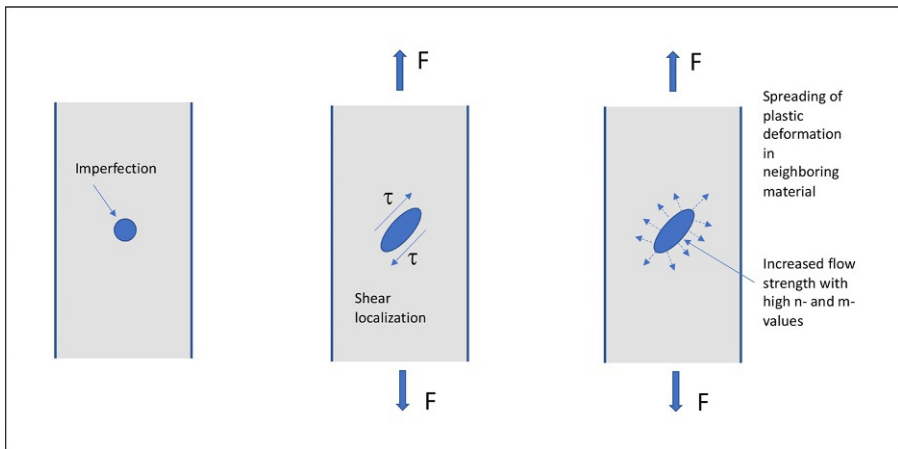


Fig. 3.28 Schematization showing the role of high n and m values in reducing the formation of shear bands.

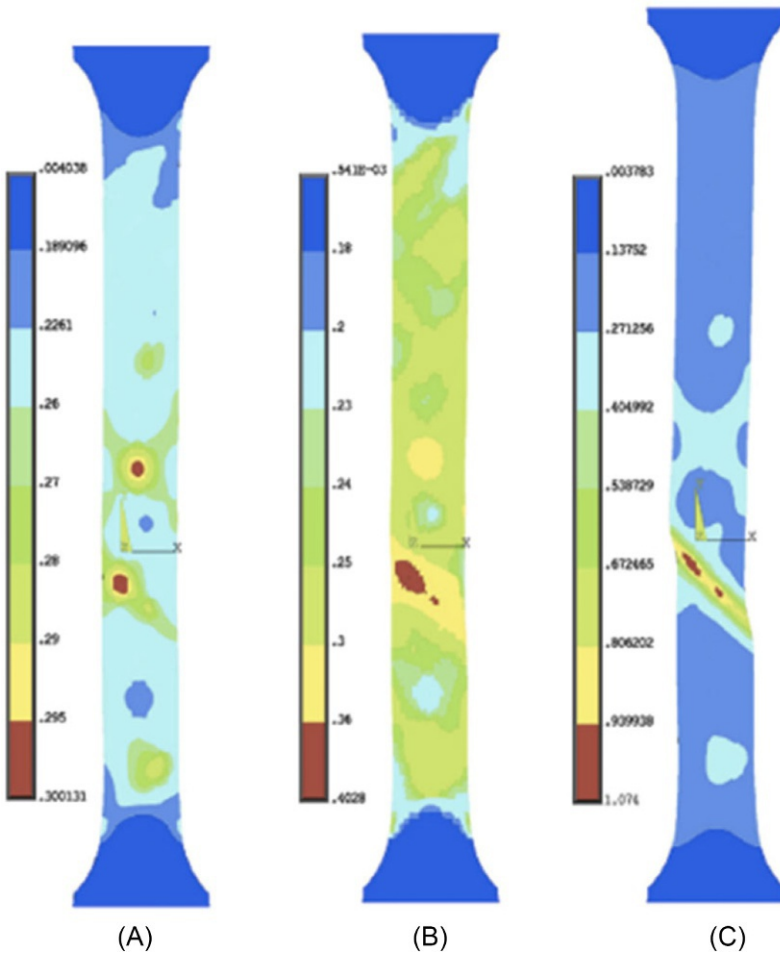


Fig. 3.29 Predicted (by FEM) evolution of the plastic strain during tensile testing of a sheet specimen made with 5182 Al alloy at nominal elongations of (A) 21.82%, (B) 22.5%, (C) 23.1%.

From Straffellini, G., Fontanari, V., Zadra, M., 2013. Influence of specimen width on the deformation and fracture behaviour of AA5182 sheets. *Eng. Fract. Mech.* 109, 262–272, reproduced with permission of the publisher.

localization leads to high strains within the band and then quickly to ductile fracture (Fig. 3.30).

An interesting example is shown in Fig. 3.31. It shows two deformed and fractured samples of an HSLA steel and a Ti-64 alloy (Bourcer et al., 1984). The tensile tests were conducted using rectangular specimens (thickness: 1.52 mm) that contained a hole in their center (diameter: 0.51 mm). The hole provided a large geometrical imperfection to promote a shear strain localization. This phenomenon was observed in the Ti-64 alloy, featuring a very low strain-hardening coefficient

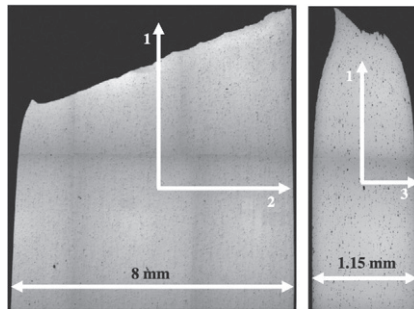


Fig. 3.30 Morphology of a sheet specimen made with 5182 Al alloy at fracture. From Straffellini, G., Fontanari, V., Zadra, M., 2013. Influence of specimen width on the deformation and fracture behaviour of AA5182 sheets. *Eng. Fract. Mech.* 109, 262–272, reproduced with permission of the publisher.

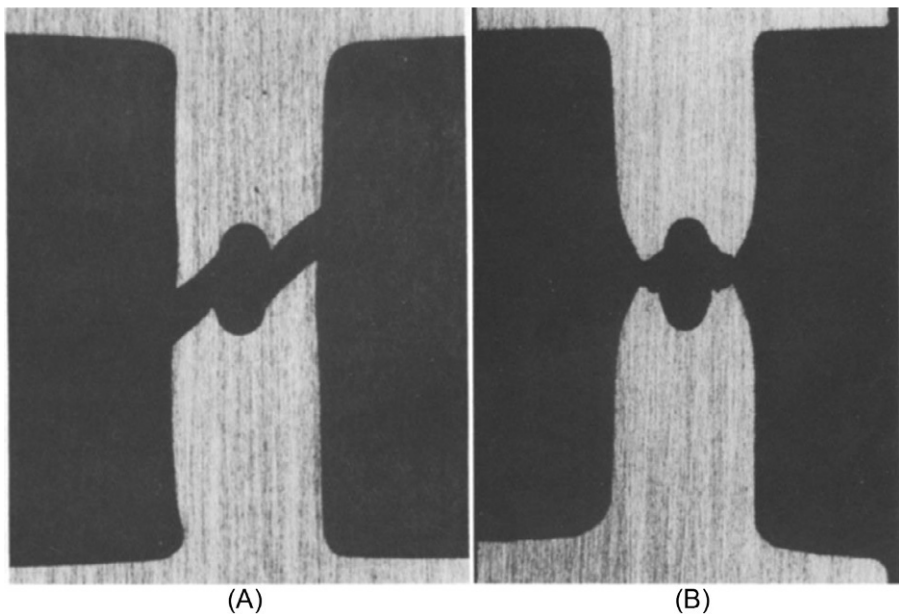


Fig. 3.31 Deformation and fracture appearance of Ti-64 (A) and HSLA steel (B) specimens containing a large geometrical imperfection. From Bourcer, R.J., Smelser, R.E., Richmond, O., Koss, D.A., 1984. Deformation and fracture at isolated holes in plane-strain tension. *Int. J. Fract.* 24, 289–297, reproduced with permission of the publisher.

(see [Section 2.5](#)). For this material, the local strain hardening was not sufficient to counterbalance the shear localization tendency induced by the hole. On the contrary, the local strain hardening in the steel (characterized by a comparatively high strain-hardening coefficient) was sufficient to avoid this localization, and the local plasticity developed intensively, involving different plane orientations. As a result, fracture

ultimately occurred perpendicularly to the loading direction (and after a large longitudinal strain).

Spontaneous shear localization in metals is then favored by the presence of:

- Microstructural and geometrical imperfections;
- Negative m values and low n values;
- A marked tendency to decrease the flow strength with increasing temperature (thermal softening).

Several Al alloys are particularly prone to shear localization at RT because:

1. They contain soft regions in their microstructure (formed by special textures, as shown in [Section 2.8](#));
2. RT falls in the range of approximately $0.3\text{--}0.4T_m$, i.e., in a temperature interval in which DSA may occur, providing negative m values and the n values are comparatively low due to thermal softening;
3. The flow strength is largely sensitive to a temperature rise.

As a result, shear localization followed by shear fracture is often observed in sheets (where shear localization is easier for geometrical reasons, related to the high width-to-thickness ratio), and also in cylindrical specimens (see [Fig. 8.3](#)).

As seen in the previous section, with increasing strain rate, thermal softening increases in importance, since the process approaches adiabatic conditions (typically, adiabatic conditions are attained at strain rates greater than 10^3 s^{-1}). As a result, the formation of shear bands is facilitated (even if coefficients n and m increase with increasing strain rate, and this may render spontaneous shear localization more difficult). If the intense material softening, following the adiabatic deformation, prevails, an avalanche process is initiated and the local plastic deformation can become very high. The shear bands forming in this way are denominated *adiabatic shear bands* (ASBs).

A particular situation occurs when very large shear deformations are *forced* in a narrow region by external loading. This may occur, for example, in metalworking processes. The simplest situation regards cutting, by punching or chip removal, which induces heavy shear strain localization in the cutting planes. Also, in the case of forging or rolling, narrow regions with high shear deformation may form, following the phenomenon of barreling induced by the workpiece-die friction (explained in detail in [Chapter 11](#)). Very often, forced shearing takes place at high strain rates, i.e., in adiabatic conditions. In these cases, ASBs may form rather easily. As an example, [Fig. 3.32](#) shows an ASB formed in an annealed austenitic stainless steel.

The local temperature rise within the ASB is very high. As an example, [Fig. 3.33](#) shows the temperature evolution, recorded using infrared thermal imaging, in a flat-hat specimen made with a 6061-T6 Al alloy subjected to forced localized shear deformation with very high strain rates, i.e., in adiabatic conditions ([Nie et al., 2020](#)). It is seen that the temperature rise in the shear band is close to 500°C . The local maximum temperature (770 K) is then close to $0.83T_m$. The evolution of the recorded applied force is also shown. Shortly after the attainment of the maximum temperature rise, the fracture is initiated and propagated, as shown by the subsequent force drop. In a Ti-64 alloy,

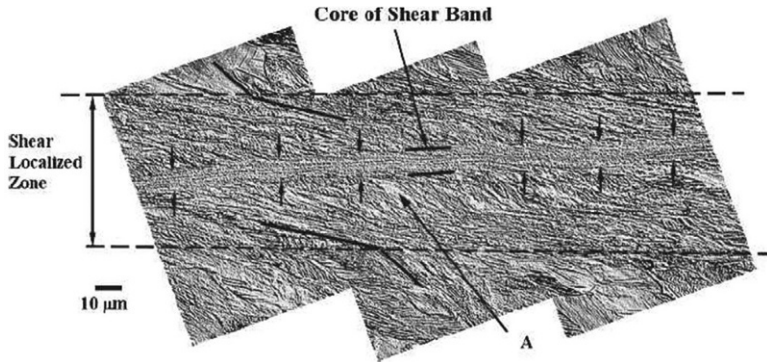


Fig. 3.32 ASB in an annealed austenitic stainless steel.

From Xue, Q., Gray III, G.T., 2006. Development of adiabatic shear bands in annealed 316L stainless steel: part I. Correlation between evolving microstructure and mechanical behavior. *Metall. Mater. Trans. A* 37, 2435–2446, reproduced with permission of the publisher.

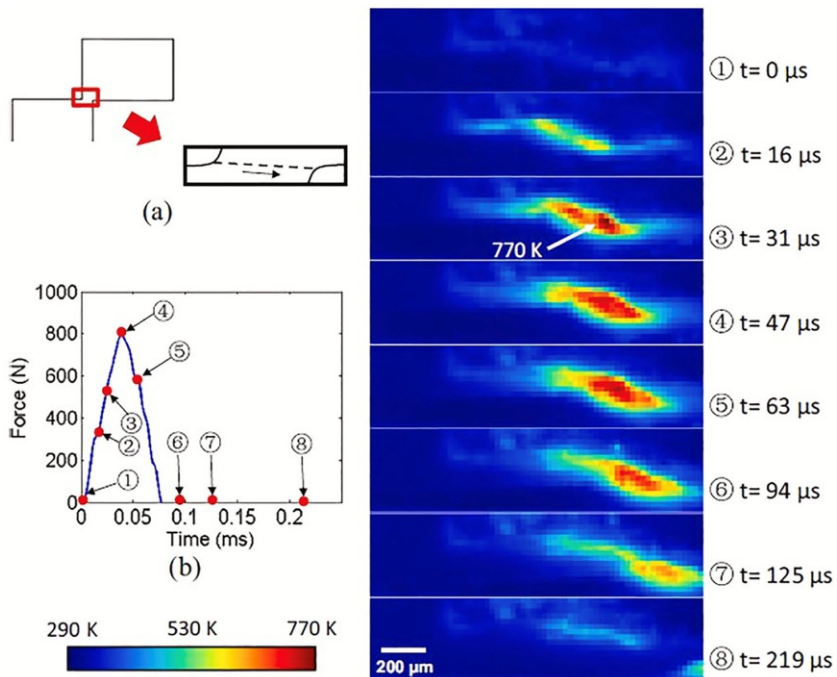


Fig. 3.33 Temperature evolution, recorded using infrared thermal imaging, in a flat-hat specimen of a 6061-T6 Al alloy. The evolution of the applied force is also shown.

From Nie, Y., Claus, B., Gao, J., Zhai, X., Kedir, N., Chu, J., Sun, T., Fezzaa, K., Chen, W.W., 2020. In situ observation of adiabatic shear band formation in aluminum alloys. *Exp. Mech.* 60, 153–163, reproduced with permission of the publisher.

the initiation of ASB was found to occur at an average sample temperature of the range 100–300°C, but at a very high local temperature of 1100°C, corresponding to approximately $0.75T_m$ (Ranc et al., 2008). In steels, temperatures of 1400°C were measured within the ASB, corresponding to $\sim 0.9T_m$ (Zhou et al., 1996). In some cases, the temperature rise can be so high as to induce local melting. In general, the temperature in the shear bands spans the range $0.5\text{--}1T_m$ (Yan et al., 2021).

Metallographic investigations have shown that, in addition to *deformed shear bands*, determined by the large plastic deformations only (typically formed in Al alloys at RT and at low strain rates), *transformed shear bands* may also form, characterized by local microstructural transformations. The high local temperature (in the range of high-temperature creep) combined with high plastic strain may induce dynamic recrystallization (DRX, a phenomenon considered in the next chapter), which greatly increases the local softening of the metal and significantly contributes to the shear localization (Yan et al., 2021). In steels, there may be the formation of martensite, following austenitization due to large local temperature rise, and then fast cooling. In austenitic stainless steels with low SFE, the strain-induced formation of martensite may also occur. The transformed regions are also called *white layers*.

Forced ASBs form easier with increasing:

- The process adiabaticity (and then with decreasing the thermal conductivity and increasing the strain rate);
- The local heat generation (and then with increasing u , i.e., the metal strength; see Eq. 3.13).

Some key parameters play a conflicting role, and most material parameters highly depend on temperature. Therefore it is very difficult to obtain a reliable model for the determination of the occurrence of ASBs, even if some approaches are presented in the literature (Staker, 1981; Meyers, 1994; Komanduri and Hou, 2002). Considering the thermal properties of metals (and then the evaluations of Fig. 3.27) and their typical strength, it can be expected that forced ASBs are comparatively easy to form in Ti alloys and in high-alloyed steels, such as stainless steels; on the contrary, they are expected to form with more difficulty in Al alloys. The formation of ASBs is expected to be more difficult at high temperatures, because of the decrease in the flow stress (see the next chapter) that decreases u , and then the local temperature rise.

The formation of spontaneous or forced adiabatic shear bands may have different consequences. The most important one is the attainment of very large local plastic strains that may favor the damage mechanisms, leading to anticipated ductile fracture. This can be beneficial, such as in cutting operations, or detrimental, such as in forging or rolling operations, or in dynamic loading of engineering structures, for which ductile crack formation can ultimately cause catastrophic failure.

To summarize:

- *Spontaneous shear localization* in narrow bands at 45 degrees with the loading direction is induced by the maximum shear stress and the presence of microstructural and geometrical

imperfections; it is favored by negative m values and low n values, and by a noticeable metal tendency to thermal softening;

- Al alloys of class 5000 fulfill these requirements and are particularly prone to spontaneous shear localization;
- Plastic strain localization may be accentuated under adiabatic conditions, i.e., with sufficiently high local strain rates; in this case, a special form of shear bands develop, the so-called adiabatic shear bands (ASBs);
- In ASBs, the temperature rise is rather high, and this may promote phase transformations such as dynamic recrystallization that greatly soften the metal;
- Formation of ASBs is greatly favored when shear deformation is forced by the external loading at comparatively low temperature; forced ASBs form easily in alloys with low thermal conductivity and high flow strength (such as Ti alloys and high alloy and high strength steels).

3.7 Constitutive equations

A *constitutive equation* describes and predicts the flow stress of a material as a function of strain, strain rate, and temperature:

$$\sigma = f(\varepsilon, \dot{\varepsilon}, T) \quad (3.15)$$

The Ludwik (2.4), Ludwik-Hollomon (2.5), and Ludwigson (2.11) equations are examples of constitutive equations. However, their practical use is restricted to specific temperature and strain rate intervals. Their wider application would require the knowledge of the dependency of the involved parameters (such as K and n in the L-H equation) with temperature and strain rate.

In the present section, some important constitutive equations, which can account for the influence of temperature ($T < 0.4\text{--}0.5T_m$) and strain rate (below approximately 10^3 s^{-1}) will be presented. It is recalled that plastic deformation is really a complex phenomenon, and it is quite impossible to get a constitutive equation able to correctly predict the flow stress in every condition. Additionally, the deformation path plays an important role, and its consideration would require a more comprehensive analysis (some aspects on this point are introduced in Chapter 5). In general, the selected equation should be sufficiently accurate for the type of computation in which it is intended to be used.

The first class of constitutive equations derives directly from the theory of thermally activated slip of dislocations, introduced in Section 3.1. Following this approach, two equations were obtained by Zerilli and Armstrong (1987) for bcc and fcc metals (in the case of fcc metals, the dependency of the thermal component of the flow strength, σ_T , on plastic strain is considered). Subsequently, the authors merged the two equations into one in the following form:

$$\sigma(T, \varepsilon, \dot{\varepsilon}) = \sigma' + \frac{B}{e^{\beta \sigma' T}} + \frac{B_0}{e^{\alpha \sigma' T}} \cdot \sqrt{\dot{\varepsilon}} \quad (3.16)$$

where

- $\beta = \beta_0 - \beta_1 \cdot \ln \dot{\epsilon}$
- $\alpha = \alpha_0 - \alpha_1 \cdot \ln \dot{\epsilon}$
- $B, B_0, \beta, \beta_0, \alpha$, and α_0 are material constants (the strain-hardening coefficient, n , is here set to 0.5, but in principle, a different value could be chosen).

In Table 3.4, materials constants to be used in Eq. (3.17) and for some metal alloys are listed.

Eq. (3.16) is *additive*, meaning that the different contributions are summed up together. The second class of constitutive models is made up of *multiplicative* equations, meaning that the role of strain rate and temperature is accounted for by multiplying the reference equation (valid for RT and in quasi-static conditions) with proper factors.

These models can then be introduced starting from the basic L-H equation: $\sigma = K\epsilon^n$. Let σ_0 be the flow stress at RT and under quasi-static conditions (say $\dot{\epsilon}_0 = 10^{-3} \text{ s}^{-1}$). If σ is indicated as the flow stress at $\dot{\epsilon}$ (with $\dot{\epsilon} > \dot{\epsilon}_0$), Eq. (3.8) can be used to obtain:

$$\frac{\sigma}{\sigma_0} = \left(\frac{\dot{\epsilon}}{\dot{\epsilon}_0} \right)^m$$

Using the L-H equation for σ_0 :

$$\sigma = K \cdot \epsilon^n \cdot \left(\frac{\dot{\epsilon}}{\dot{\epsilon}_0} \right)^m \quad (3.17)$$

where K , n , and m are materials properties that depend on temperature. Eq. (3.17) is often rewritten in the following way:

$$\sigma = C \cdot \epsilon^n \cdot \dot{\epsilon}^m \quad (3.18)$$

Table 3.4 Experimental materials constants to be used in Eq. (3.18).

	304L stainless steel (Lee and Lin, 2001)	HY-100 high-strength steel (Zerilli and Armstrong, 1995)	Ti-6Al (Zerilli and Armstrong, 1995)	AA 5086 (H32) (Wagenhofer et al., 1999)
σ_a (MPa)	96.12	400	690	200
B (MPa)	16.05	1155	1485	0
β_0 (K^{-1})	5.63×10^{-3}	3.56×10^{-3}	3.09×10^{-3}	0
β_1	1.84×10^{-3}	9.67×10^{-5}	7.47×10^{-5}	0
B_0 (MPa)	1127.58	1255	1135	1283
α_0 (K^{-1})	3.88×10^{-12}	2.02×10^{-3}	1.4×10^{-4}	3.71×10^{-3}
α_1 (K^{-1})	1.59×10^{-12}	1.75×10^{-4}	0	1.43×10^{-4}

Table 3.5 Experimental values for M to be used in Eq. (3.19) (data from Meyers, 1994, and other sources).

Metal	M	Metal	M	Metal	M
Pure Fe	0.55	CMnAl TRIP	0.62	Al-2024 T3	1
Fe-0.08%C	1	409 stainless steel	0.23	Al-5457	0.78
Fe-0.12%C-1.37%Mn	0.32	CP copper	1.09	Al-6061 T6	1.34
Fe-0.4%C-1.65%Ni	1.03	Cu-Zn	1.68	Al-7075 T6	1.13
S7 tool steel	1	Ti-6Al-4V	1.1		

and it is also called the *extended L-H equation*. By comparing Eqs. (3.17) and (3.18): $C = K/\dot{\epsilon}_0^m$.

To account for the role of temperature, an additional multiplying factor can be used in Eq. (3.17). The following equation is then often used:

$$\sigma = K \cdot \epsilon^n \cdot \left(\frac{\dot{\epsilon}_1}{\dot{\epsilon}_0} \right)^m \cdot \left(1 - \left(\frac{T - T_0}{T_m - T_0} \right)^M \right) \quad (3.19)$$

where

- T_0 is the absolute RT;
- T_m is the absolute melting temperature of the alloy;
- M is an experimental constant.

Some experimental values for M are listed in Table 3.5. The well-known *Johnson-Cook (J-C) equation* is like Eq. (3.19). In the J-C equation, the strain hardening at RT is modeled using the Ludwik equation (Eq. 2.4), and the effect of the strain rate is accounted for with the following factor: $(1 + C \cdot \ln \dot{\epsilon}/\dot{\epsilon}_0)$, where C is an experimental parameter (for example, $C = 0.022$ for a low-C steel, and $C = 0.015$ for the Al-2024 T3 alloy). Eq. (3.19) is simple and suitable for several applications. Specific limitations are discussed in Liang and Khan, 1999, where other models are also presented.

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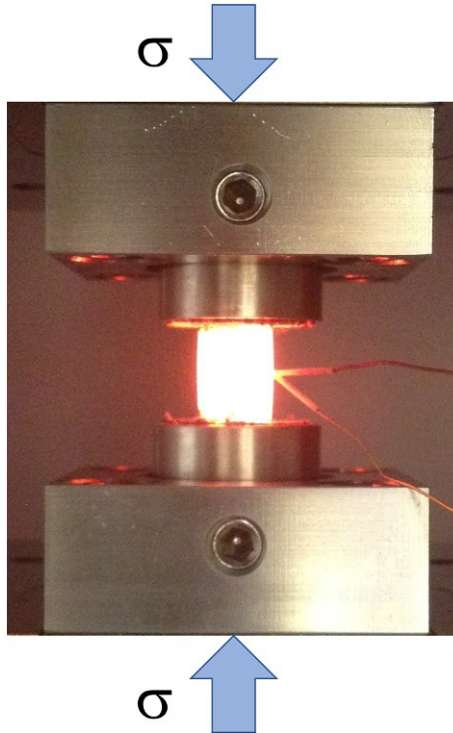
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Plastic deformation at high temperature ($T > 0.4\text{--}0.5T_m$)

4



Hot compression test. Specimen heating is made through electric current and associated Joule effect. Temperature is controlled with a thermocouple.
Courtesy of Dynamic Systems Inc.

Chapter outline

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For temperatures above $0.4\text{--}0.5T_m$, plastic deformation occurs rather easily. Additional mechanisms facilitate plastic deformation:

- Dynamic recovery with dislocations cross-slip and dislocations climb (at comparatively high stresses);
- Diffusion creep and grain boundary sliding (at comparatively low stresses).

In any case, the flow stress decreases with temperature and assumes very low values when the temperature is high. This is very attractive for plastic deformation processes, as will be outlined in the next chapters.

In this chapter, the high-temperature deformation of metals under compression and at comparatively high stresses is extensively analyzed, since most high-temperature deformation processes are carried out in this manner. The role of SFE and the interplay with recrystallization phenomena is elucidated. Specific constitutive equations, useful in the process design, are also introduced.

The tensile behavior of metals at high temperature and under high stresses is also introduced since it is a reference framework for analyzing the hot ductility of metals (Chapter 8). At the end of this chapter, the high-temperature tensile behavior of metals under low stresses is considered; it is the reference framework for the phenomenon of superplasticity (also outlined in Chapter 8).

4.1 Creep of metals: An overview

As shown in Fig. 3.1, the yield strength of metals decreases with temperature, tending to minute values for $T > 0.4\text{--}0.5T_m$. For temperatures above $0.4\text{--}0.5T_m$, plastic deformation is quite easy and time dependent. Therefore reference to *viscous* or *creep* deformation is made.

In structural parts employed at high ambient temperature, creep plays a negative role. The applied stress, σ , is often a tensile stress and it is comparatively low, approximately below $10^{-4} \sigma/E$, where E is the Young's modulus (which decreases with temperature as well). This stress induces an increase in the plastic deformation over time. Usually, the plastic deformation is required to be low, and less than an admissible value after a certain period of time. To obtain design relevant data, high-temperature tensile tests are carried out. The specimen is heated and held at a constant temperature during the test, and it is loaded with a constant stress. Hence, the material is subjected to two external loads: tensile stress (σ) and temperature (T). It reacts with a time-dependent plastic deformation, as shown in Fig. 4.1A. After an almost instantaneous elastic deformation (given by σ_y/E), there is an initial stage during which the hardening rate is greater than the softening rate. This stage is called *primary creep*. Subsequently, a second stage is entered, during which the material deforms at a constant strain rate, $\dot{\epsilon}$, under a dynamic equilibrium between hardening and softening. This is the main deformation stage that covers the prevalent life span of a mechanical component. It is called *secondary creep* or *steady-state flow*. The corresponding strain rate is very low, typically between 10^{-10} and 10^{-6} s^{-1} . Finally, at the end of the secondary creep stage, particular microstructural damage

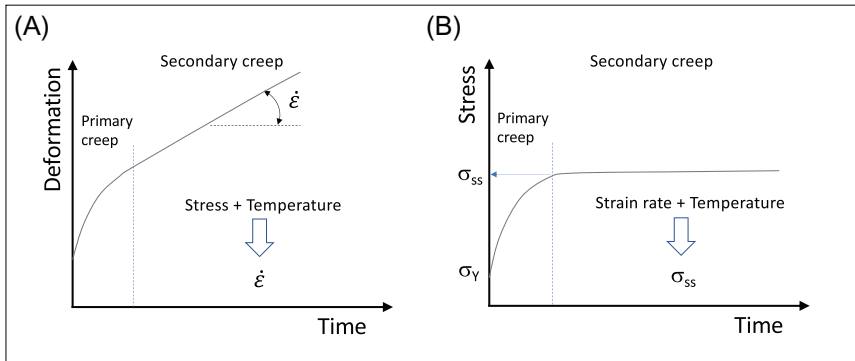


Fig. 4.1 Creep curves obtained at high temperature with (A) constant stress, and with (B) constant deformation rate.

phenomena may prevail, which lead to an increase in the strain rate (*tertiary creep*, not shown in the figure), resulting in specimen fracture (Frost and Ashby, 1982).

On the contrary, creep plays a positive role in plastic forming processes. Processes such as rolling, forging, and extrusion are typically conducted at a temperature beyond $0.4\text{--}0.5T_m$ to exploit the positive influence of temperature on the plastic deformability of metals. In these cases, plastic deformation is by compression and a specific (usually large) strain is applied to the workpiece at a given temperature, by means of a tool, typically at strain rates in the range of 5×10^{-4} and 10^2 s^{-1} . The corresponding stress is comparatively high, in the range of $10^{-4}\text{--}10^{-2} \sigma/E$. To determine this stress, and subsequently to calculate the process load and power, laboratory compression tests are carried out at fixed temperatures and strain rates. The evolution of the applied stress with time is presented in Fig. 4.1B. The presence of the two stages of primary and secondary creep is indicated. Note that in compression tests, the maximum possible strains are on the order of 2, due to the barreling phenomenon, caused by the friction between the workpiece and the die (explained further in Chapter 11). Barrelling may be limited by adopting proper lubrication at the tool-workpiece interface. To avoid this shortcoming, hot torsion tests can be carried out (Section 5.7).

In this chapter, the compression behavior of metals at comparatively high strain rate and large strains is first considered. This is the so-called *high-stress creep regime*. Subsequently, the tensile behavior in the high-stress creep regime is introduced. It is discussed further in Chapter 8 when reporting the hot ductility of metals (this topic is important since during metal forming processes by compression, unexpected tensile stresses may appear, and these may lead to ductile fracture).

In the last section of this chapter, the tensile behavior of metals at comparatively low strain rate conditions is discussed (in the so-called *low-stress creep regime*). The methods for structural design are not presented, but the attention is focused on a special phenomenon, *superplasticity*, that is responsible for high ductility, as is also seen in Chapter 8.

Table 4.1 summarizes the general characteristics of high-stress and low-stress creep regimes, and the main fields where these phenomena are encountered.

Table 4.1 General characteristics of high-stress and low-stress creep and the main fields where these phenomena are encountered.

	Typical strain rate range (s^{-1})	Compression deformation	Tensile deformation
High-stress creep $\sigma/E > 10^{-4}$	5×10^{-4} – 10^2	<i>Metal working processes (creep is positive)</i>	<i>Metal working processes</i> Unexpected tensile secondary stresses <i>Structural parts (creep is negative)</i>
Low-stress creep $\sigma/E < 10^{-4}$	10^{-10} – 10^{-6} (Up to 10^{-3} in the case of superplasticity)		<i>Structural parts (creep is negative)</i> <i>Metal working processes</i> Special behavior: <i>superplasticity (creep is positive)</i>

4.2 Primary and secondary high-stress creep

Fig. 4.2 shows the creep curves obtained at 1050°C by compression for a steel with composition 1.1% C, 6% Cr (Bernsthein et al., 1985). The tests were conducted at three strain rates and the recorded true stress is plotted as a function of the applied strain.^a At this temperature ($\sim 0.77T_m$), the elastic modulus is rather low, and the tests were carried out at σ/E values in the range of $2\text{--}8 \times 10^{-3}$, i.e., in the typical conditions of high-stress creep.

The testing temperature (1050°C) is high and the steel has an austenitic fcc structure during plastic deformation. At the end of the tests, the samples were rapidly cooled to RT to freeze the microstructure and allow the observation of the dislocation structures. These are reported in the figure (note that the martensite-start temperature for this steel is below 0°C and then no martensite transformation occurred upon cooling). The first analysis is conducted on the ascending part of the curve, i.e., the primary creep region. For small strains (typically below 0.1), the dislocation density increases with plastic strain, and dislocations form tangles and cells, typical of the strain hardening in cold conditions, as seen in the previous chapter. As a matter of fact, by fitting this part of the curve with the L-H equation, a strain-hardening coefficient of $n = 0.5$ is obtained, which is in broad agreement with the trends for austenitic fcc steels introduced in the previous chapter.

Subsequently, plastic deformation becomes increasingly easier since *thermal softening* is superimposed on strain hardening. Softening is determined by different thermally activated phenomena, promoted by the easy diffusion of atoms and vacancies at high temperatures. These main phenomena are:

^a Please note that the strain is considered to be positive, even if the deformation is by compression; this is also done throughout this chapter and often in the next chapters for the sake of simplicity.

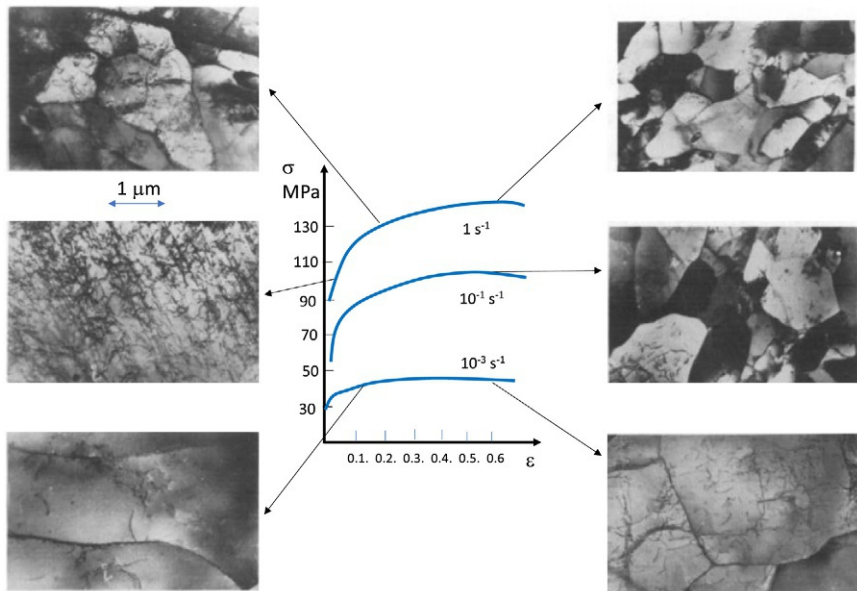


Fig. 4.2 Flow curves for a 1.1%C–6%Cr steel at 1050°C obtained by compression and related microstructures at specific strains.

From Bernstein, M.L., Kaputkina, L.M., Prokoshkrin, S.D., Dobatkin, S.V., 1985. Structural changes during hot deformation of austenite in alloy steels. *Acta Metall.* 33, 247–254, reproduced with permission of the publisher.

- *Climbing* of edge dislocations;
- *Cross-slip* of screw dislocations;
- *Annihilation* of dislocations;
- *Formation* of sub-grains by dislocations *polygonization*.

All these mechanisms are together referred to as *dynamic recovery* (usually indicated with *DRV*).

As seen, for $T < 0.4-0.5T_m$, thermal fluctuations help dislocations in overcoming short-range obstacles (Fig. 3.3). For $T > 0.4-0.5T_m$, temperature helps dislocations in overcoming long-range obstacles as well, by climbing and cross slip. In particular, the climbing of edge dislocations plays a very important role. It is schematized in Fig. 4.3A. As seen in the figure, in this process, a dislocation can move perpendicularly to the slip plane and then surmount a long-range obstacle. This may happen because the atoms of the edge dislocation (i.e., at the end of the semi-plane inserted in the crystal lattice, as seen in Fig. 1.3) diffuse in neighboring vacancies. Temperature increases the diffusivities of atoms and the concentration of vacancies in the lattice, hence favoring the climbing of dislocations. It has been reported that the velocity of climbing greatly depends on the SFE. In metals with high SFE, climbing is much easier and this helps DRV.

The same happens to the cross-slip of screw dislocations. As already discussed in Section 1.2, cross slip involves perfect dislocations and permits the overcoming of

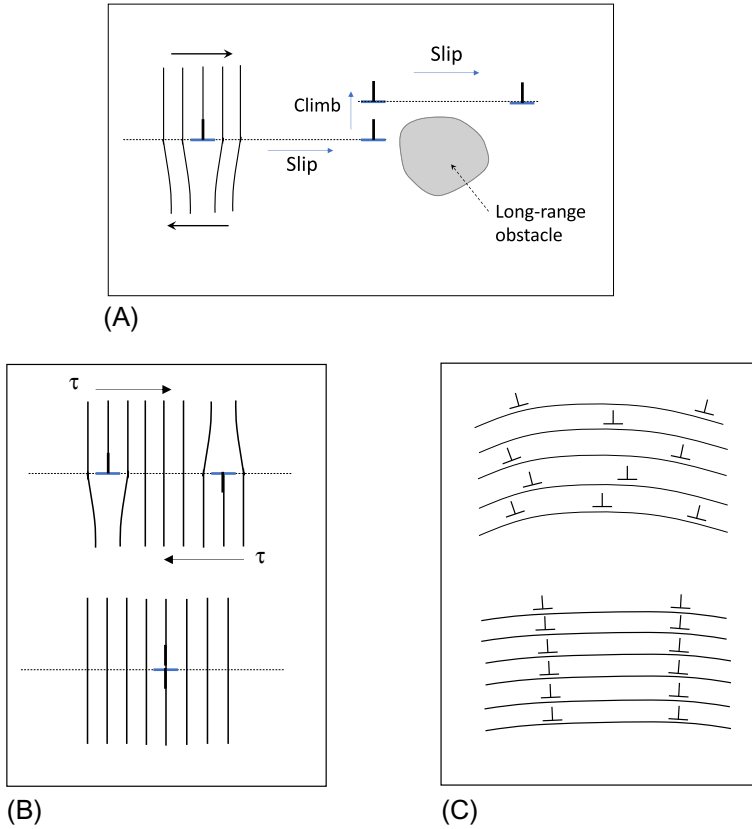


Fig. 4.3 (A) Schematization showing the dislocation climb to surmount a long-range obstacle; (B) schematization of annihilation of two dislocations and (C) polygonization of several dislocations (with the formation of low-angle sub-grain boundaries).

microstructural obstacles. The presence of partial dislocations, which is typical in metals with low SFE, reduces the possibility of cross slip and also the dislocation mobility.

The mechanisms of annihilation and polygonization of dislocations are schematized in Fig. 4.3B and C. The annihilation process involves dislocations moving on the same plane but with opposite orientation. The two dislocations disappear when they meet. The polygonization process (characterized by the formation of low-angle grain boundaries) is due to the tendency of the moving dislocations to reach a low energy configuration, minimizing the mutual interactions. These two processes induce a reduction in the dislocations density and favor the formation of substructures within the grains that provide low resistance to the dislocation motion. In other words, they favor DRV. Metals with a high SFE contain a very low density of stacking faults. Dislocation motion necessary for annihilation and polygonization is rather easy in these metals, and then DRV is intense.

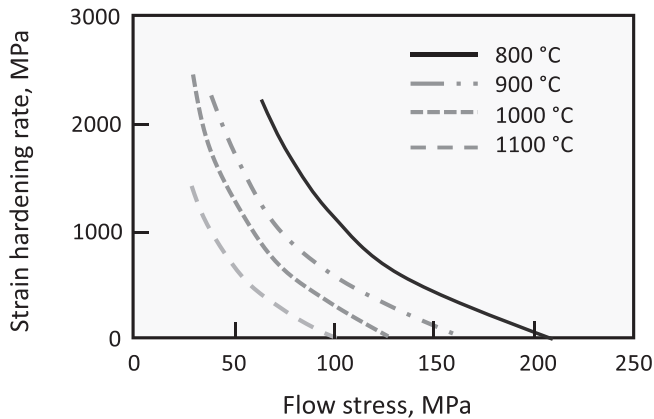


Fig. 4.4 Experimental dependence of the strain-hardening rate on flow stress in the primary creep regime and for a micro-alloyed steel at different temperatures.

Redrawn from Laasraoui, A., Jonas, J.J., 1991. Prediction of steel flow stresses at high temperatures and strain rates. *Metall. Trans. A* 22A, 1545–1558.

To summarize, we can say that the DRV phenomena are all easier in metals with high SFE. DRV phenomena are also thermally activated: with increasing temperature, the intensity of the DRV phenomena increases, whereas as the strain rate increases, the intensity decreases, as less time is given for their occurrence.

The resulting metal softening is characterized by a decrease in the slope of the σ - ϵ curve of the primary creep, i.e., a continuous decrease in the strain-hardening rate, $d\sigma/d\epsilon$ (with a corresponding decrease in the instantaneous strain-hardening coefficient, n), as is clearly shown in the curves of Fig. 4.2. As a further example, Fig. 4.4 shows the experimental dependence of the strain-hardening rate on the flow stress at different temperatures for a micro-alloyed steel (Laasraoui and Jonas, 1991). As expected, at each flow stress the strain-hardening rate, $d\sigma/d\epsilon$, is seen to decrease with temperature. The deformed microstructure is made of typically elongated grains in the direction of plastic deformation.

As mentioned, when the material softening is enough to counterbalance strain hardening, the strain-hardening rate becomes zero and the secondary (steady-state) creep is entered. The transition from primary to secondary creep occurs at a critical strain. In metals with a *high SFE*, DRV phenomena are rather intense. Hence, DRV is able to dynamically counteract strain hardening. The transition to the secondary creep regime is gentle, with the true stress-true strain curve becoming horizontal, as shown in Fig. 4.5A. The transition strain is indicated with ϵ_m and the *equilibrium stress at steady state*, i.e., in the secondary creep, is indicated with σ_{ss} .

In metals with comparatively *low SFE*, DRV is slow. As a consequence, metal softening is difficult and the stored (elastic) energy increases with strain (due to the high density of dislocations). Therefore it may become so high as to induce *dynamic recrystallization (DRX)* at a certain plastic strain. DRX is characterized by the nucleation and growth of new (dislocation-free) grains and takes place *during* plastic deformation. The strain at the onset of DRX is indicated with ϵ_c in Fig. 4.5B. As long as the

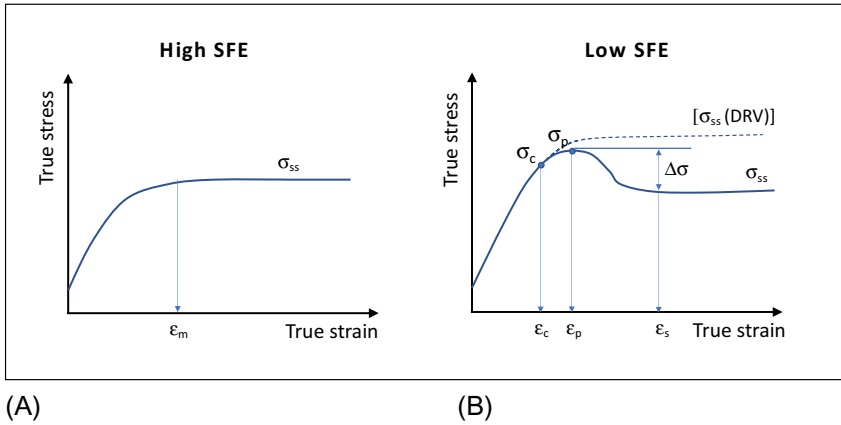


Fig. 4.5 Schematization of the stress-strain curves during hot deformation for metals with high SFE (A), and metals with low SFE (B). In this second case, the steady-state region starts at ϵ_s and may be characterized by periodic oscillations of the equilibrium stress when the strain rate is comparatively low.

fraction of recrystallized grains is low, the flow stress continues to increase with strain. But at a certain strain, ϵ_p , a peak stress, σ_p , is reached and subsequently the flow stress decreases with increase in the strain. When recrystallization involves the whole microstructure, steady-state conditions are reached and the corresponding stress is indicated with σ_{ss} . The decrease in the flow (true) stress from σ_p to σ_{ss} ($\Delta\sigma$ in the figure) is due to the increased deformability of the recrystallized grains. It depends on temperature and strain rate. As the temperature increases, the SFE increases and $\Delta\sigma$ decreases since DRV gains in importance. As strain rate is increased, $\Delta\sigma$ is also decreased, since the shortened deformation time may become insufficient for the growth of the recrystallized nuclei.

In the case of deformations at high temperature and low strain rate (typically below 0.01 s^{-1}), the steady-state region is characterized by periodic oscillations of the equilibrium stress (Medina and Hernandez, 1996).

Table 4.2 shows a list of alloys characterized by easy DRV (and then difficult DRX), and alloys characterized, on the contrary, by difficult DRV (and then easy DRX). The first group of alloys (with high SFE) displays a compressive stress-strain curve similar to that shown in Fig. 4.5A; the second group (with low SFE) displays a stress-strain behavior similar to that shown in Fig. 4.5B. The interesting case of Ti alloys is noted. Differently from steels, Ti alloys undergo easy DRX at low temperatures, when they have an α (hcp) structure; on the other hand, they undergo DRX with many more difficulties at high temperature, when they assume a β (bcc) structure.

As an example, Fig. 4.6 shows experimental peak strains, ϵ_p , and the corresponding peak stresses, σ_p , for different steels, tested at high temperatures and at a strain rate of 1 s^{-1} . In general, the peak strains decrease with increasing the intensity of softening; hence they decrease with increasing temperature (as shown in Fig. 4.6A) and decreasing strain rate. The peak strains in Fig. 4.6A are typically in the range 0.3/0.7, corresponding

Table 4.2 Alloys that typically display the stress-strain behavior schematized in Fig. 4.5A (high SFE) and Fig. 4.5B (low SFE).

Easy DRV, difficult DRX (High SFE alloys)	Difficult DRV, easy DRX (Low SFE alloys)
α -Fe (bcc) alloys, such as ferritic stainless steels Most Al alloys β -Ti (bcc) alloys	Fe and steels that have an austenitic (fcc) structure at high temperature Austenitic stainless steels Some 7000 Al alloys (7050, 7075) α -Ti (hcp); α - β Ti-64 Cu alloys (fcc) Mg alloys (hcp)

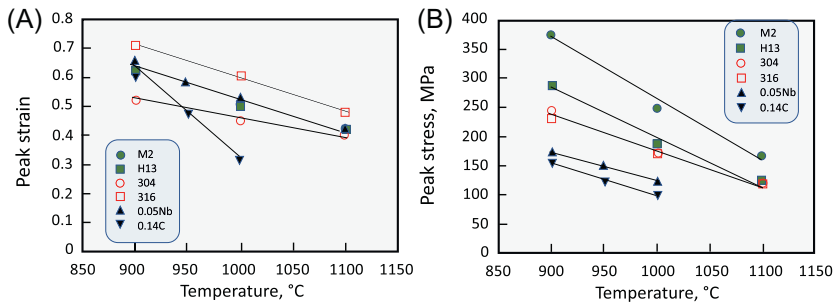


Fig. 4.6 (A) Experimental peak strains, ϵ_p , and (B) peak stresses, σ_p , for different steels, tested at high temperatures and at a strain rate of 1 s^{-1} . Data from Ryan, N.D., McQueen, H.J., 1986. Effects of alloying upon the hot workability of carbon, microalloyed, tool, and austenitic stainless steels. J. Mech. Work. Technol. 12, 279–296.

to reductions of 25%–50%; they broadly increase with increasing the steel strength. The peak stresses are typically two times greater than the corresponding yield stresses. The values reported in Fig. 4.6B are higher for the M2 tool steel. For this steel, the carbides are not completely dissolved in the microstructure, even at the highest temperatures here considered. For the H13 grade, the carbides dissolve in the microstructure at high temperature and the peak stresses become similar to those of the stainless steels. The low-C steels (including the HSLA steel with 0.12%C) display the lowest stress values.

The critical strains, ϵ_m and ϵ_p , are generally expressed by the following relationship (Sellars, 1990):

$$\epsilon_{m,p} = A \cdot d_0^p \cdot Z^q \tag{4.1}$$

where

- A , p , and q are material constants to be determined experimentally;
- d_o is the grain size of the metal at the testing temperature, T (in K), and before plastic deformation (typically in μm);

- Z is the Zener-Hollomon (Z - H) parameter or temperature-compensated strain rate, defined by:

$$Z = \dot{\epsilon} \cdot \exp\left(\frac{Q}{R \cdot T}\right) \quad (4.2)$$

where Q (in kJ/mol) is the activation energy for the plastic deformation at the steady state (Mc Queen and Ryan, 2002), and R is the gas constant ($R = 8.31$ J/mol K).

A similar equation is proposed for the peak stress. In some cases, the role of the grain size is not considered (in most cases, it is difficult to determine, or it plays a minor role), and the analysis is simplified using the following relationships (Mirzadeh et al., 2012):

$$\epsilon_{m,p} = A' \cdot Z^q, \quad \sigma_{m,p} = B' \cdot Z^t \quad (4.3)$$

All the involved parameters, including the Q values, are obtained by the experimental data, such as those reported in Fig. 4.6. Some examples are reported in Table 4.3. The listed metals have a low SFE except for commercially pure Al. Therefore the reported data refer to the calculation of ϵ_p or σ_p (Fig. 4.5B), whereas only the data for pure Al should be used to calculate ϵ_m (Fig. 4.5A).

Table 4.3 Experimental values for Eqs. (4.2) and (4.3).

Alloy	A'	q	B'	t	Q (kJ/mol)	
C-Mn steels	$6.97 \cdot 10^{-4} \cdot d_0^{0.3}$	0.17			312	Sellars (1990) Mirzadeh et al. (2012)
HSLA steel (0.34C, 0.72Si, 1.52Mn, 0.018Ti, 0.083V)	0.0079	0.1026	0.567	0.147	394	
API X65 (HSLA) steel (0.072C, 0.2Si, 1.45Mn, 0.015Ti, 0.047Nb, 0.05V)	0.0045	0.153	0.37	0.160	330	Rakhshkhorshid and Hashemi (2013)
316 SS	$6.03 \cdot 10^{-2} \cdot d_0^{0.75}$	0.053			402	Mc Queen and Ryan (2002) Garcia et al. (2001) Sellars (1990)
99.9% purity Cu	0.00776	0.185			197	
Commercial purity Al	0.002	0.2			158	

The listed metals have a low SFE except for commercially pure Al.

Q mainly depends on the alloy composition and is related to the activation energy for self-diffusion that is responsible for the recrystallization processes and the dislocation motion (in particular the climbing of dislocations). In pure austenite it is 276 kJ/mol, and this value can be considered as a lower bound for Q in steels. In fact, Q increases with the content of solutes and also with the presence of microstructural obstacles that slow down the atoms/vacancy diffusivity (Menapace et al., 2018).

The stress-strain behavior in the primary creep regime is modeled by different methods. One is based on the *Voce equation* that is able to account for the thermal softening contribution (Kocks and Mecking, 2003):

$$\sigma = \sigma_{ss} + (\sigma_Y - \sigma_{ss}) \cdot \exp\left(-\frac{\varepsilon}{\varepsilon^*}\right) \quad (4.4)$$

where

- σ_{ss} is the equilibrium stress to which the flow stress tends asymptotically;
- σ_Y is the stress for $\varepsilon \sim 0$ and represents the yield strength;
- ε^* is a characteristic strain that defines the rate with which the flow strength tends to the equilibrium stress (it is proportional to Z).

The strain-hardening rate is thus given by:

$$\frac{d\sigma}{d\varepsilon} = \frac{\sigma_{ss}}{\varepsilon^*} - \frac{\sigma}{\varepsilon^*} \quad (4.5)$$

This equation shows that the strain-hardening rate decreases as the flow stress increases, in agreement with the data in Fig. 4.4. By comparing Eq. (4.5) and Eq. (2.9), the instantaneous strain-hardening coefficient, n , can be estimated as:

$$n = \frac{\varepsilon}{\varepsilon^*} \cdot \left(\frac{\sigma_{ss}}{\sigma} - 1\right) \quad (4.6)$$

As ε increases, σ tends to σ_{ss} and $(\sigma_{ss}/\sigma - 1)$ tends to 0 (when approaching the secondary creep). Therefore n typically decreases with an increase in the true strain. This behavior is shown in Fig. 2.17 for an Al alloy. The general trend is schematized in Fig. 4.7, where the influence of T and $\dot{\varepsilon}$ is also indicated.

It is worth noticing at the end of this section that sometimes the stress-strain behavior in the primary creep regime is simply estimated using Eq. (3.18) (the extended L-H equation), for which some experimental data are available. For example, in the case of Mn-C steels, the modified Misaka's equation is often used (Siciliano et al., 1996):

$$\sigma = 9.81 \cdot (0.78 + 0.137[\text{Mn}]) \cdot \exp\left(0.126 - 1.75[\text{C}] + 0.595[\text{C}]^2 + \frac{2851 + 2968[\text{C}] - 1120[\text{C}]^2}{T}\right) \cdot \varepsilon^{0.21} \cdot \dot{\varepsilon}^{0.13}$$

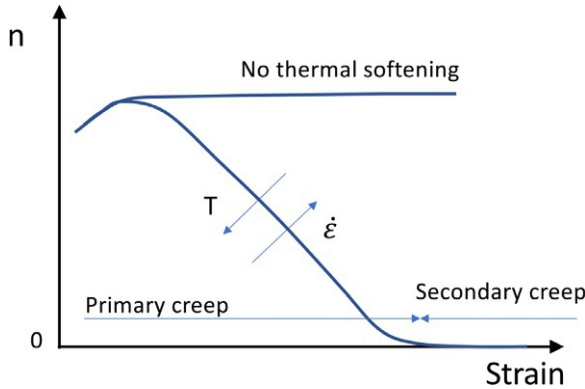


Fig. 4.7 General evolution of the strain-hardening coefficient, n , with true strain in the absence (see also Fig. 2.5) and presence of thermal softening. In the second case, the role of temperature and strain rate is indicated.

where

- σ is given in MPa;
- $[C]$ and $[Mn]$ are the C and Mn concentrations in wt%;
- T is the absolute temperature.

This approach is rather approximate, since the extended L-H equation is not able to account for the role exerted by thermal softening.

The analysis carried out in this section considers the creep at rather high stress, in the range of 10^{-4} – $10^{-2} \sigma/E$ (and corresponding strain rates in the range of 5×10^{-4} – 10^2 s^{-1}). As seen, the creep mechanisms are determined by the dislocation motion (slip and climb) and its interplay with dynamic recovery (and, possibly, DRX). Therefore the high-stress creep is also called *dislocation creep*.

To summarize:

- For $T > 0.4$ – $0.5T_m$, thermal softening (induced by DRV phenomena) decreases the slope of the σ - ϵ curve, and the instantaneous strain-hardening coefficient, n , decreases tending to 0;
- DRV is easier in metals with high SFE; these metals show a gentle transition to the secondary creep (characterized by the absence of strain hardening);
- In metals with low SFE, DRV is slow, and DRX may then start at a critical strain, ϵ_c .

4.3 High-stress creep and dynamic recrystallization

Alloys with low SFE have low efficiency for recovery processes and easily undergo dynamic recrystallization (DRX) during plastic deformation at high temperatures. As an example, Fig. 4.8 shows the microstructure of an AZ31 Mg alloy, extruded at 250°C , and at different extrusion strains (Chen et al., 2007). In Fig. 4.8A, the microstructure of the alloy in the as-cast condition is shown. The other two figures show that the hot extrusion was characterized by DRX with a grain refinement that increases with the extrusion strain. For $\epsilon = 1.94$, DRX is partial, but for $\epsilon = 3.18$, the average grain size is $10 \mu\text{m}$ (see Chapter 13 for details on the extrusion process).

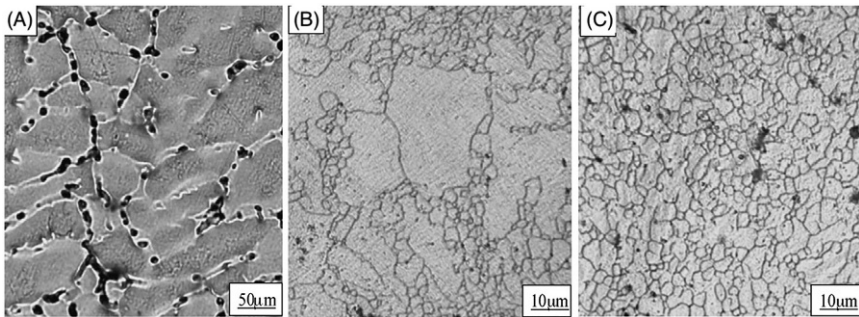


Fig. 4.8 Microstructure of an AZ31 Mg alloy in the as-cast condition (A), after extrusion at 250°C with $\epsilon = 1.94$ (B), and $\epsilon = 3.18$ (C).

From Chen, Y., Wang, Q., Peng, J., Zhai, C., Ding, W., 2007. Effects of extrusion ratio on the microstructure and mechanical properties of AZ 31 Mg alloy. *J. Mater. Process. Technol.* 182, 281–285, reproduced with permission of the publisher.

As seen, DRX starts at a critical strain, ϵ_c , which is a fraction of the peak strain, ϵ_p , i.e., $\epsilon_c = b \epsilon_p$, where b is a constant lower than unity (Poliak and Jonas, 2003). Parameter b was found to be close to 0.8 for C-Mn steels, with C content from 0.06% to 0.25% and Mn content from 0.3% to 1.7% (Sellars, 1990), and in the range of 0.55–0.7 for some austenitic stainless steels and tool steels (Mc Queen and Ryan, 2002). For practical purposes, it can also be set: $\epsilon_c \sim \epsilon_p$.

Parameter ϵ_c can be used to establish the ease with which DRX may occur during high-temperature plastic deformation. For example, for a C-Mn steel subjected to $\dot{\epsilon} = 5 \text{ s}^{-1}$, and with $d_0 = 100 \text{ μm}$, using the data of Table 4.3 ϵ_c is found to be 0.2 at 1300°C, and 1 at 870°C (Sellars, 1990). In these conditions, at 870°C is much more difficult to reach the onset of DRX.

The kinetics of DRX can be represented by the Avrami equation:

$$X = 1 - \exp \left[-0.693 \left(\frac{t}{t_{0.5}} \right)^k \right] \quad (4.7)$$

where

- X is the fraction of recrystallized material;
- k is a material constant;
- t is the time (with t being 0, when $\epsilon = \epsilon_c$, i.e., at the start of DRX);
- $t_{0.5}$ is the time for 50% recrystallization.

For steels, k is typically in the range 1–2.5, and it decreases with an increase in both the initial grain size (d_0) and Z . As an example, Fig. 4.9 shows the experimental evolution of X for some tool steels (Mc Queen and Imbert, 2004). In the graphs, the X values (from 0.5 to 0.99) are displayed between ϵ_c (the critical strain for the onset of DRX) and ϵ_s (the onset of steady state).

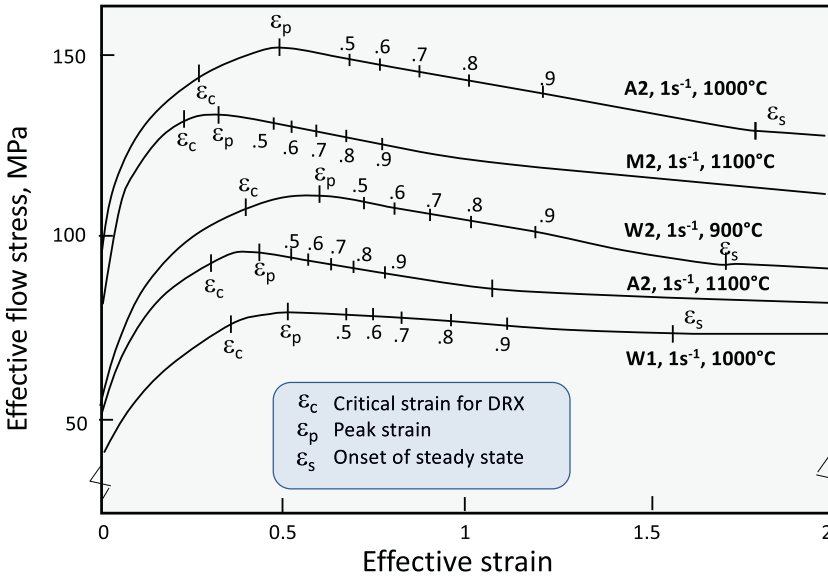


Fig. 4.9 Hot stress-strain curves for some tool steels (AISI A2, M2, and W1) showing the X values of DRX, and derived from the Avrami equation (and confirmed by microscopy) Redrawn from Mc Queen, H.J., Imbert, C.A.C., 2004. Dynamic recrystallization: plasticity enhancing structural development. *J. Alloys Compd.* 378, 35–43.

It has been found that the recrystallized grain size, d (in μm), can be expressed by the following relationship when DRX is complete before the end of the deformation process:

$$d = \frac{C}{Z^c} \quad (4.8)$$

where C and c are two material constants. For example, for Mn-C steels: $C = 16,000$ and $c = 0.23$ (Hodgson and Gibbs, 1992). In general, d is lower than d_0 , and recrystallization induces a refinement of the grain size. In the case of steels, the microstructure changes upon cooling to RT, following the transformation of austenite. In the case of the formation of ferritic-pearlitic microstructures, the grain size after cooling is proportional to the grain size, d , of austenite after DRX. In other alloys, such as Al and Mg alloys, no phase transformation takes place upon heating to the deformation temperature and also upon cooling to RT. Therefore the recrystallized grain does not change its size upon cooling. However, grain growth may subsequently occur after DRX, particularly when cooling does not immediately follow the end of the deformation process.

Often at the end of the deformation process, DRX is incomplete, i.e., $X < 1$. In these cases, recrystallization continues afterward, and the process is called *meta-dynamic* (MDRX) or *post-dynamic recrystallization* (Maccagno et al., 1996). In fact, even

the formation of a small fraction of DRX grains is sufficient to achieve complete MDRX, since the recrystallized grains act as nuclei for subsequent grain growth. The grain size after MDRX can be expressed with a relationship similar to Eq. (4.8). For a Mn-C steel, $C = 26,000$ and $c = 0.23$. Also, after completion of MDRX, grain growth can occur.

If $\varepsilon < \varepsilon_c$, no DRX occurs and at the end of the process, static recrystallization (SRX) may take place, possibly followed by grain growth.

To summarize:

DRX starts during plastic deformation when $\varepsilon = \varepsilon_c$. Therefore:

- DRX becomes easier as the applied strain, ε , increases and/or as ε_c decreases;
- ε_c decreases as Z decreases (Eq. 4.1), i.e., as strain rate decreases and temperature increases;
- ε_c decreases as d_0 decreases (i.e., as pre-heating temperature/time decrease);
- ε_c decreases as Q decreases (see Table 4.3).

4.4 High-stress creep and strain rate sensitivity index

Using a simplified model, it has been observed that the constitutive equation for the steady-state high-stress (dislocation) creep is a *power-law relationship* (Frost and Ashby, 1982):

$$\dot{\varepsilon} = \frac{A' \cdot D_v \cdot E \cdot b}{RT} \cdot \left(\frac{\sigma_{ss}}{E} \right)^{m'} \quad (4.9)$$

where

- A' and m' are material constants;
- E is the elastic modulus;
- b is the modulus of the Burgers vector;
- D_v is the bulk self-diffusion coefficient, which is given by:

$$D_v = D_0 \cdot e^{-\frac{Q}{RT}} \quad (4.10)$$

where D_0 is a constant and Q represents the activation energy for self-diffusion. As said, the effect of temperature in favoring plastic deformation and strain rate is due to its strong influence on atomic and vacancy diffusion. Eqs. (4.9) and (4.10) can be merged to obtain the following relationship (Dieter, 1988):

$$\dot{\varepsilon} = A \cdot \exp \left(-\frac{Q}{RT} \right) \cdot \sigma_{ss}^n \quad (4.11)$$

where A is a constant.

As already mentioned, in metalworking processes, when the metal is shaped at a given strain rate and at a certain temperature, the flow stress must be obtained in order to calculate the process load. Therefore Eq. (4.11) is usually rewritten as:

$$\sigma_{ss} = C \cdot \dot{\varepsilon}^m \quad (4.12)$$

where $C = 1/A \cdot \exp(Q/RT)$ and $m = 1/m'$, where m is the strain rate sensitivity index, introduced in Section 3.3.

As an example, Fig. 4.10A shows the stress-strain curves for a 304L austenitic stainless steel obtained by the means of compression tests at four temperatures and four strain rates (El Wahabi et al., 2003). In agreement with Eq. (4.12), the steady-state

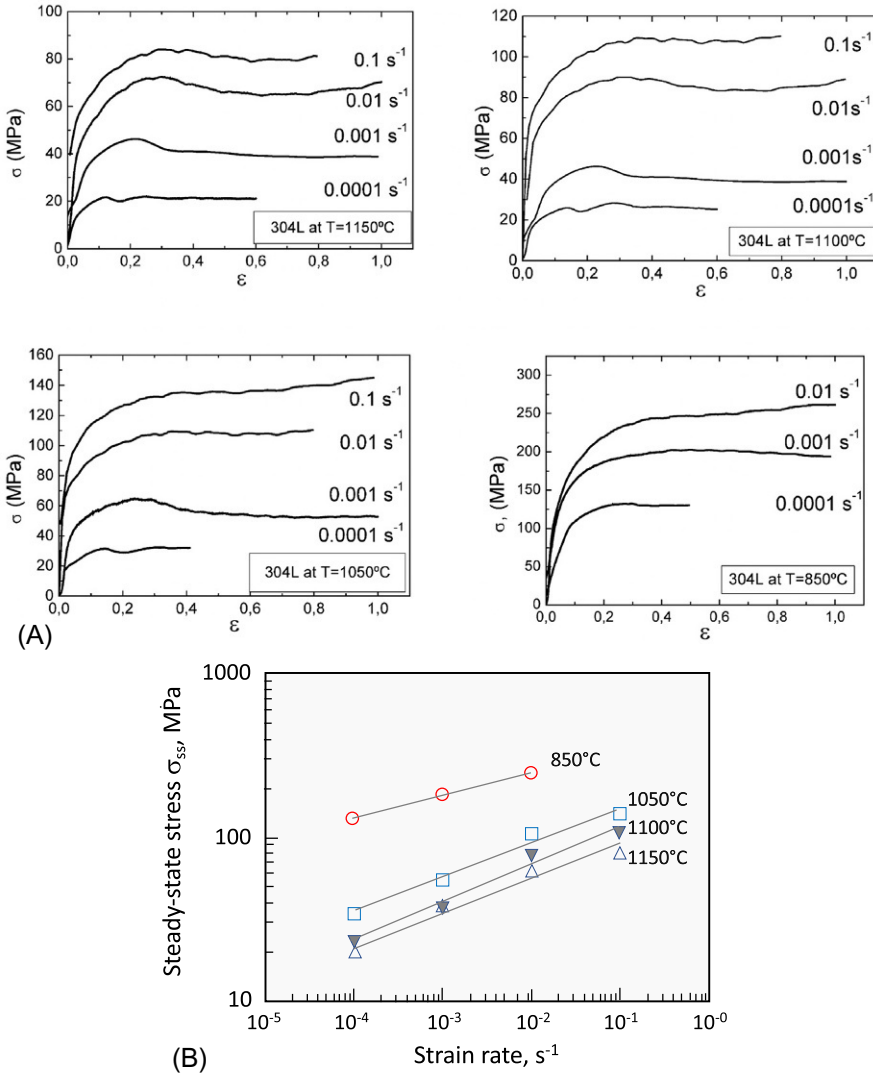


Fig. 4.10 (A) Creep curves for a 304L austenitic stainless steel obtained by the means of compression tests at two temperatures and two strain rates. (B) Plots of the steady-state stresses vs. strain rate.

A: From El Wahabi, M., Cabrera, J.M., Prado, J.M., 2003. Hot working of two AISI 304 steels: a comparative study. *Mat. Sci. Eng. A* 343, 116–125, reproduced with permission of the publisher.

Table 4.4 Experimental values for the parameters of Eq. (4.12).

Metal	$0.45T_m$ (°C)	Temperature range (°C)	C (MPa)	m
Al 1100	~140	200–500	76–15	0.07–0.22
Al 5083	~100	240–480	290–60	0–0.2
Cu	~335	300–900	200–40	0.02–0.16
Cu alloys	~300	200–800	400–20	0.027–0.23
Low-C steel	~550	900–1200	160–58	0.09–0.15
Medium-C steel	~550	900–1200	150–50	0.15–0.19
316 Austenitic SS	~550	600–1200	200–50	0.18–0.23
Ti	~600	600–900	190–30	0.1–0.32
Ti-64	~570	600–1000	620–65	0.075–0.13

Data for annealed metals, adapted from Altan, T., Oh, S., Gegel, H., 1983. Metal Forming, Fundamentals and Applications. American Society for Metals.

stress (σ_{ss}) is seen to increase with strain rate and decrease with testing temperature. Most curves display DRX, with the presence of the typical peak stress, because of the low SFE of the steel. In Fig. 4.10B, the experimental σ_{ss} values are plotted as a function of strain rate (in logarithmic scales) for the investigated testing temperatures. The experimental points can be properly fitted with straight lines, showing that σ_{ss} and $\dot{\epsilon}$ are modeled with a power-law relation. In particular, m values were 0.14 at 850°C, and approximately 0.22 at the other three temperatures.

In Table 4.4, typical experimental data for C and m are listed for different alloys (adapted from Altan et al., 1983). The C and m values are reported for specific temperature intervals. Within each temperature interval, C is seen to decrease with temperature, whereas m is seen to significantly increase. We have seen in Section 3.3 that m is typically below 0.07 for $T < 0.4-0.5T_m$. For $T > 0.4-0.5T_m$, m is higher and falls in the range of 0.07–0.4. In Fig. 4.11, the experimental dependence of m with the homologous temperature for different alloys is displayed. This behavior can be explained considering that m increases with the increase in the difficulty with which dislocations overcome microstructural barriers. At high temperatures, dislocations are mainly required to overcome (by climb) long-range obstacles and this is comparatively more difficult than the overcoming of short-range obstacles that is the prevailing phenomenon controlling dislocation motion at low temperatures. As a consequence, an increase in strain rate at a given temperature is accompanied by a marked increase in the flow stress as less time is allowed at dislocations to overcome the microstructural barriers. This difficulty increases with increasing temperature, because of the corresponding decrease in the dislocation density and, hence, in the number of the available mobile dislocations. In general, the strain rate sensitivity coefficient, m , is also found to depend on the plastic strain, in particular in passing from the primary to the secondary creep. However, this dependence is rather limited and no clear trends can be recognized (Altan et al., 1983).

It is worth noticing that Eq. (4.12) is not always satisfactory in modeling the flow stress of metals at high temperature (Lin and Chen, 2011). An improvement is possible

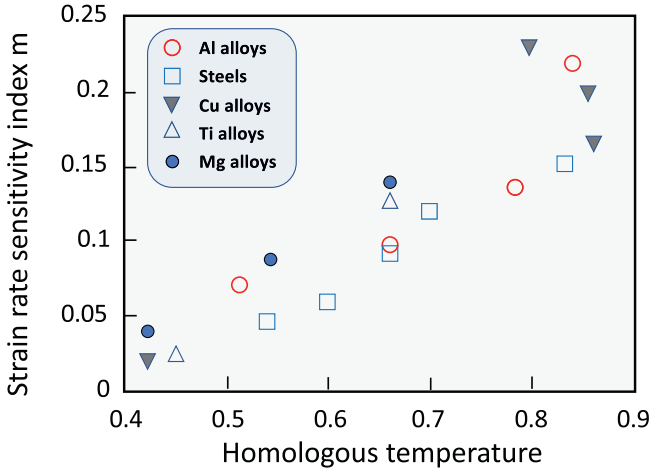


Fig. 4.11 Dependence of strain rate sensitivity coefficient m on homologous temperature for different alloys.

by substituting σ_{ss}^n in Eq. (4.11) with $\sinh(\alpha\sigma_{ss})^n$, where α is a material constant. Considering the definitions of the hyperbolic law and the Z - H parameter, the following Arrhenius-type equation can then be obtained:

$$\sigma_{ss} = \frac{1}{\alpha} \cdot \ln \left\{ \left(\frac{Z}{A} \right)^{\frac{1}{n}} + \left[\left(\frac{Z}{A} \right)^{\frac{2}{n}} + 1 \right]^{\frac{1}{2}} \right\} \quad (4.13)$$

where α , A , and n are materials constants. Specific experimental data can be found in [Mc Queen and Ryan \(2002\)](#) and [Sheppard and Jackson \(1997\)](#).

To summarize:

- For $T > 0.4-0.5T_m$, the strain rate sensitivity index, m , is greater than at lower temperatures and falls in the range 0.07–0.4;
- Coefficient m is observed to increase with the homologous temperature.

4.5 High-stress creep under tensile deformation

As mentioned in the introduction to this chapter, in specific metalworking processes, materials may undergo hot tensile deformation. This may occur in the case of compressive metal forming processes, such as forging and rolling, because of inhomogeneous deformations, generally caused by the metal-tool friction. Therefore knowledge of the hot tensile deformation of metals (including their fracture behavior) is crucial.

During compression tests, necking does not occur. However, necking is very easy under hot tensile testing. Assuming, to a first approximation, a symmetric tensile-compressive strain-strain behavior, the occurrence of necking can be simply inferred from the compression curves. In fact, from these curves the evolution of n with plastic

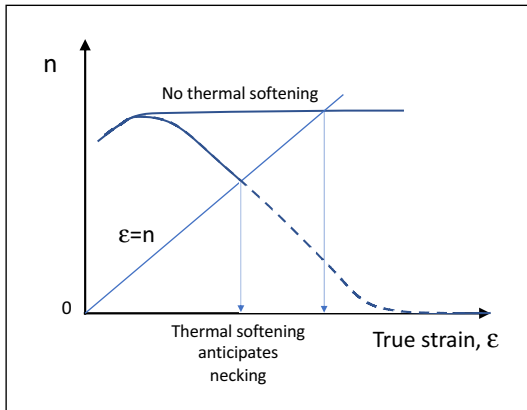


Fig. 4.12 Schematization showing the influence of thermal softening in the primary creep region to the onset of necking.

strain can be obtained. As shown in Fig. 4.7, n decreases during plastic deformation due to thermal softening. During tensile deformation, necking occurs when $d\sigma/d\varepsilon = \sigma$, i.e., using the L-H equation, when $\varepsilon = n$ (Section 2.2). Therefore Fig. 4.12 simply schematizes the conditions of necking during hot tensile testing, which takes place during primary creep (neglecting here the role of m). Using Eq. (4.6) (obtained from the Voce equation) and setting $\varepsilon = n$, the flow stress for necking (σ') can be simply obtained as:

$$\sigma' = \frac{\sigma_{ss}}{1 + \epsilon^*} \quad (4.14)$$

In Eq. (4.14), the peak stress, σ_p , should be used in place of the equilibrium stress, σ_{ss} , if the stress-strain behavior is similar to that shown in Fig. 4.5B. In any case, during hot tensile deformation, necking is expected to occur before the attainment of steady-state conditions, since $\sigma' < \sigma_{ss}$ (or σ_p).

Figs. 4.7 and 4.12 allow us to understand the role of temperature and strain rate on necking. It is seen that an increase in temperature and a decrease in strain rate favor thermal softening, thus anticipating the occurrence of necking. As an example, Fig. 4.13 shows the experimental dependence of the uniform strain at necking (ϵ_u) on temperature in the case of a 9Cr-1Mo ferritic steel (Toloczko et al., 2003), and on temperature and strain rate for an AZ31 Mg alloy (Rodriguez et al., 2016).

As seen in Section 3.3, the strain rate coefficient, m , also plays a role in necking. However, for high m values (typical of high-temperature deformation, including the primary creep region), necking is further shifted to smallest strains (Fig. 3.18) and then its role can be neglected, at least to a first approximation. The role of m cannot be neglected when considering the post-necking deformation. High m values make strain localization very difficult. This means that plastic deformation is *very diffuse* after necking during high-temperature deformation.

Fig. 1.2A shows a schematization of the engineering stress-strain curve at RT, and the corresponding true stress-true strain curve. After necking plastic deformation

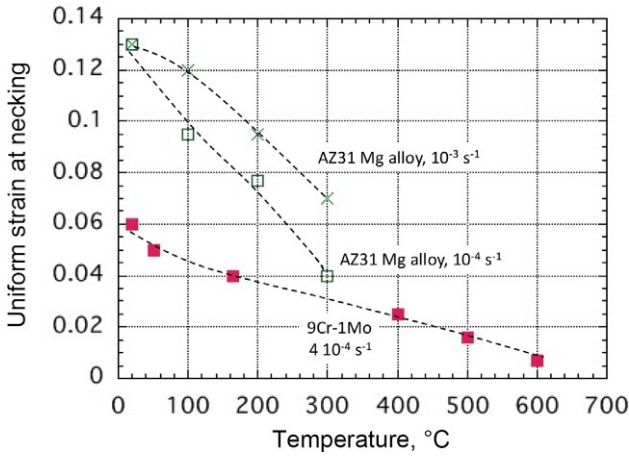


Fig. 4.13 Experimental dependence of the uniform strain at necking (ϵ_u) on temperature in the case of a 9Cr-1Mo ferritic steel (data from Toloczko et al., 2003), and on temperature and strain rate in case of an AZ 31 Mg alloy (data from Rodriguez et al., 2016).

concentrates on a specific cross section and the engineering stress decreases, whereas the true stress increases, because of strain hardening. In the engineering stress-strain (s - e) curve at high temperature, the stress also drops after necking. However, in the true stress-true strain (σ - ϵ) curve, a steady-state true stress is reached after necking, since the strain-hardening rate is counterbalanced by the strain-softening rate, similar to the hot compression tests. As an example, Fig. 4.14 shows the tensile true stress-true strain curves at $900^{\circ}C$ ($\sim 0.64T_m$; $m \sim 0.18$) and at four strain rates for a high-strength steel (Zhang et al., 2020).

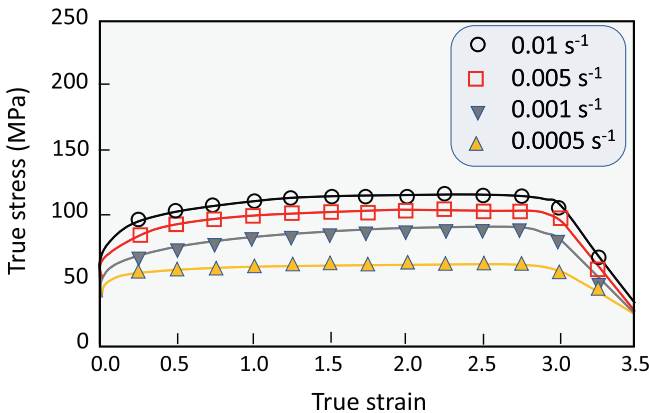


Fig. 4.14 True stress-true strain tensile curves for a Cr-Mn-Si-Ni alloyed steel at $900^{\circ}C$ and for strain rates from $5 \times 10^{-4} s^{-1}$ to $10^{-2} s^{-1}$.

Redrawn from Zhang, J.-Y., Jiang, P., Zhu, Z.-I., Chen, Q., Zhou, J., Meng, Y., 2020. Tensile properties and strain hardening mechanism of Cr-Mn-Si-Ni alloyed ultra-strength steel at different temperatures and strain rates. J. Alloys Compd. 842, 155856.

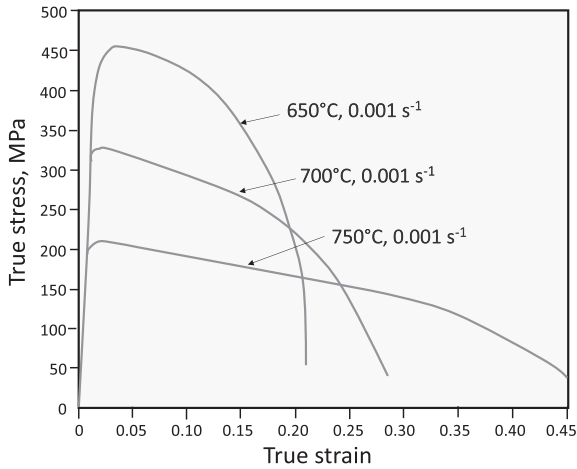


Fig. 4.15 True stress-true strain curves obtained under tension for a Ti alloy tested at three temperatures (from 650°C to 750°C) and at 10^{-3} s^{-1} .

Redrawn from Zhang, C., Li, D., Li, X., Li, Y., 2021. An experimental study of the tension-compression asymmetry of extruded Ti-6.5Al-2Zr-1Mo-1V under quasi-static conditions at high temperature. *Metals* 11 (8), 1299, met11080000.

The picture changes a bit when thermal softening becomes very intense. This may happen when the local strain reaches the conditions for the onset of DRX, i.e., when $\varepsilon = \varepsilon_c$. As seen in Fig. 4.5B for compression testing, DRX induces a drop in the true stress. The same occurs in tensile testing. An example is shown in Fig. 4.15, which shows tensile tests carried out for a near- α Ti alloy (that is rather prone to DRX; see Table 4.2), at a strain rate of 10^{-3} s^{-1} and at three different temperatures from 650°C ($\sim 0.48T_m$) to 750°C ($\sim 0.53T_m$) (Zhang et al., 2021). The flow curves display a development of DRX, leading to a drop in the flow stress (true stress) after the peak. Subsequently, at a certain strain the flow stress starts to significantly decrease and the curve displays a clear downwards curvature. This is induced by the additional geometrical softening exerted by the nucleation and growth of voids in the metal microstructure under tensile testing, leading quickly to ductile fracture, as is discussed in detail in Chapter 8.

The typical tensile stress-strain behavior under high-stress creep is shown in Fig. 4.16. The main involved phenomena are also indicated.

To summarize:

- For $T > 0.4-0.5T_m$, necking is rather easy during tensile testing because of thermal softening;
- The uniform strain at necking decreases with increasing temperature and decreasing strain rate;
- After necking, plastic deformation is very diffuse because of the corresponding high m values;
- In the case of DRX during plastic deformation, the true stress (flow stress) observes a drop during plastic deformation (as in compression tests);
- A large drop in the flow stress is also observed before fracture, because of the nucleation and growth of damaging voids (Chapter 8).

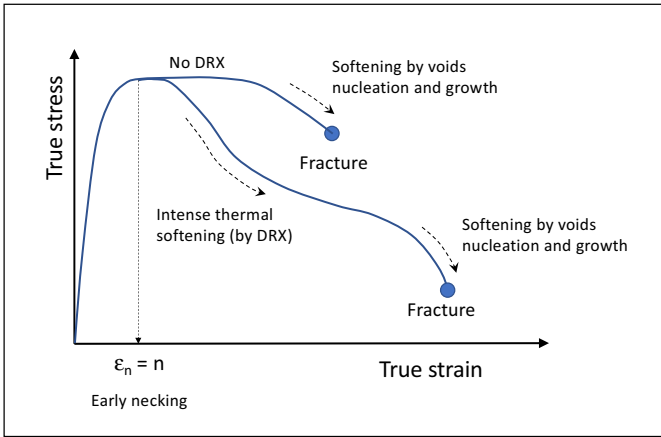


Fig. 4.16 Typical tensile stress-strain behavior of metals under high-stress creep.

4.6 Low-stress creep

Fig. 4.17 displays the typical deformation mechanisms of metals as a function of the normalized stress, σ/E , and of the homologous temperature, T/T_m (with T in K). If $\sigma/E > \sim 10^{-2}$ plastic deformation occurs by dislocation slip. If σ/E is lower and $T/T_m < 0.4-0.5$, deformation remains elastic. However, if $T/T_m > 0.4-0.5$, plastic deformation by *dislocation creep* (slip/climbing/cross-slip) occurs, as outlined in the previous sections. If σ/E is not sufficient to induce any massive dislocation motion (typically $\sigma/E < 10^{-3} - 10^{-4}$), plastic deformation may still occur by *diffusion creep* and *grain boundary sliding* (GBS). The situation is complex, and the two mechanisms may act together or one after the other (first diffusion creep and later GBS).

The mechanism of diffusion creep is outlined in Fig. 4.18. In the figure, a polycrystalline microstructure is represented with an applied stress. The applied stress induces

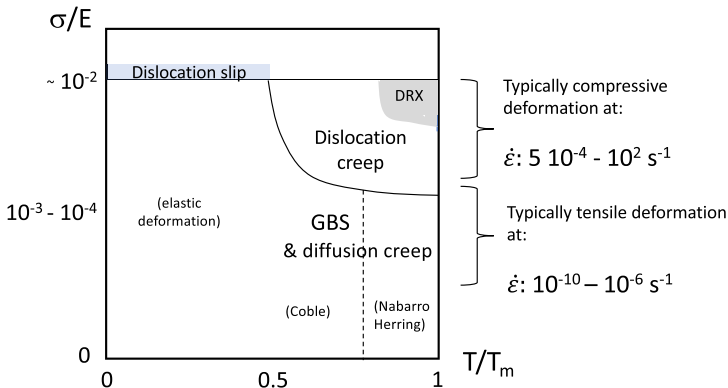


Fig. 4.17 Simple map for the deformation mechanism.

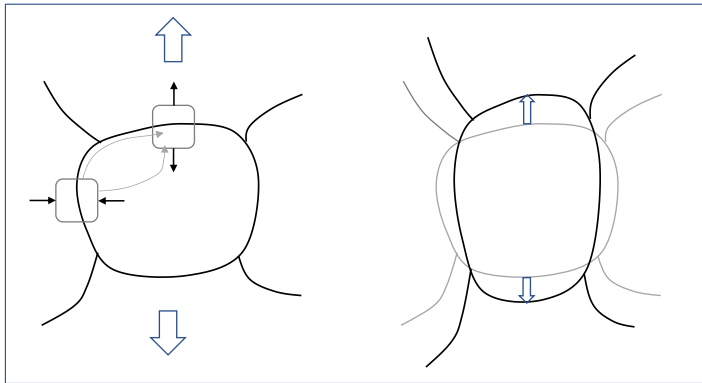


Fig. 4.18 Representation of deformation by diffusion creep.

a tensile/opening action at the grain boundaries (gbs) that are perpendicular to the loading direction. At the same time, a compression/closing action is applied at the gbs that are parallel to the loading direction. Consequently, a driving force is established that forces the atoms to diffuse from the gbs under compression towards those under tension, where more free space is available. A diffusion of vacancies in the opposite direction also takes place. As a result of this material transfer, the grains elongate in the loading direction (Frost and Ashby, 1982).

The creep mechanism of GBS is outlined in Fig. 4.19. In Fig. 4.19A, a region where three grains meet is shown. Some parallel scratches are drawn corresponding to grains 2 and 3, and across their boundary. Because of the applied stress, a shear stress is active at the boundary between grains 2 and 3. This may then induce the sliding of the two grains, with a rotation of the grains that produces a geometric displacement in the direction of loading, as revealed by the displacement of the scratches

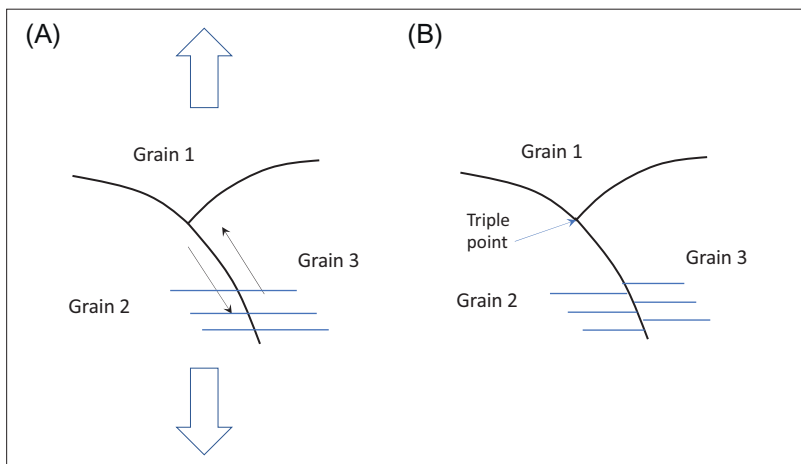


Fig. 4.19 Deformation by grain boundary sliding. At the triple point, a cavity (pore) may form.

(Fig. 4.19B). The sliding is possibly favored by the weakening of the material at the gbs. Here, segregations may be present. They reduce cohesion and also the local melting point of the alloy. This allows formation of a thin liquid that facilitates the sliding. In general, to allow the grain sliding, the local stress and strains must be accommodated. Without this accommodation, GBS would result in the formation of cavities (pores), mainly located at the triple points, which may lead to the early material fracture (refer to Section 8.6). Accommodation may occur through diffusion phenomena and dislocations motion. Diffusion phenomena impede the formation of voids by filling them with diffusing atoms. This mechanism is typically active together with diffusion creep.

From theoretical models and experimental data, it has been shown that the deformation rate for low-stress creep can be described by the following relationship (Langdon, 1994):

$$\dot{\epsilon} = \frac{A \cdot D \cdot E \cdot b}{RT} \cdot \left(\frac{b}{d}\right)^p \cdot \left(\frac{\sigma}{E}\right)^n \quad (4.15)$$

where

- A , p , and n are materials constants;
- E is the elastic modulus;
- b is the modulus of the Burgers vector;
- d is the average grain size;
- D is the diffusion coefficient, given by:

$$D = D_0 \cdot \exp\left(-\frac{Q}{RT}\right) \quad (4.16)$$

where D_0 is a constant and Q is the activation energy of the relevant diffusion process.

Eq. (4.15) is similar to Eq. (4.9). An important difference is the presence of the grain size, d , in Eq. (4.15). As a matter of fact, an increase in d induces a decrease in the strain rate for both diffusion creep and GBS. In fact, an increase in d increases the diffusion distance between the gbs under compression and those under tension (Fig. 4.18). In addition, an increase in d reduces the effectiveness of the accommodation mechanisms that allow GBS. Therefore, one way to decrease the strain rate is to increase the average grain size. For example, in the case of turbine blades, the strain rate should be sufficiently low to increase the component life. Hence, the blades are often made of single crystals.

When creep is controlled by bulk diffusion (*Nabarro-Herring creep*), $p = 2$ and $n = 1$, whereas when it is controlled by grain boundary diffusion (*Coble creep*), $p = 3$ and $n = 1$. For GBS, $p = 2$ and $n = 2$. Since the strain rate sensitivity index, m , is given by $m = 1/n$, m is equal to 1 when the main deformation mechanism is diffusion creep, and equal to 0.5 when the main deformation mechanism is GBS. The expected m values are then greater than those encountered in the case of dislocation creep.

To summarize:

- For $T > 0.4-0.5T_m$ and $\sigma/E < 10^{-3} - 10^{-4}$ (low stress creep), plastic deformation occurs by diffusion creep and grain boundary sliding (GBS);
- The corresponding m values are rather high (theoretically in the range of 0.5 and 1).

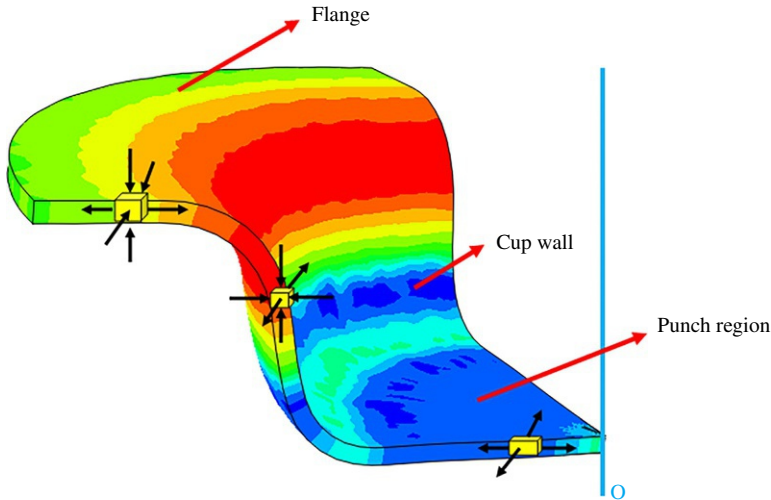
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Plastic deformation in a triaxial state of stress

5



Schematic section of a drawn cup, showing the state of stress on different locations of the part. From Dewang, Y., Sharma, V., Batham, Y., 2020. Influence of punch velocity on deformation behavior in deep drawing of aluminum alloy. *J. Failure Anal. Prev.* 21, 472–487. <https://doi.org/10.1007/s11668-020-01084-5>, reproduced with permission of the publisher.

Chapter outline

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The previous chapters examined the onset of yielding and the subsequent plastic evolution during simple uniaxial tensile or compressive loading. However, in many applications, including most metalworking processes, plastic deformation develops in a triaxial (also called multiaxial) state of stress. Therefore, two questions now arise. Which conditions result in the onset of yielding in a triaxial state of stress? How will the subsequent plastic deformation develop?

In this chapter, the von Mises criterion is used to model the onset and development of plastic deformation. Two simplifying assumptions are adopted. First, plastic deformation paths are considered proportional, i.e., the ratio among the different strain increments is considered constant during plastic deformation. Second, the metal is considered isotropic, i.e., with the same properties in every direction. Only a brief discussion of the anisotropic behavior of metal sheets is included.

The plastic behavior of notched samples is then introduced. This is of paramount importance to understand the post-necking behavior of metals during tensile testing, which plays a pivotal role in determining the ductile fracture process and the measurement of the ductility of metals, and also in understanding the impact behavior of metals investigated using notched specimens (Chapter 9).

5.1 Onset of yielding in a triaxial state of stress

Let us indicate the main loading direction, such as the tensile direction in a tensile test, as “1” and the other two directions perpendicular to 1 (and perpendicular to each other) as “2” and “3.” Consider that such directions are also *principal directions*. Under uniaxial conditions, σ_1 is positive (tensile loading) or negative (compressive loading), and $\sigma_2 = \sigma_3 = 0$, where σ_2 and σ_3 are the stresses along directions 2 and 3. We have seen that yielding in these conditions occurs when $\sigma_1 = \sigma_Y$, where σ_Y is the material uniaxial yield strength. This section answers the question on the conditions required for the onset of yielding in a triaxial state of stress, i.e., when σ_2 and/or σ_3 are different from 0.

We need to develop a *yield model* (or *yield criterion*) to answer this. In the case of *isotropic* materials, i.e., for materials that display the same properties in all directions, two important models for yielding are available: the *Tresca yield criterion* and the *von Mises yield criterion*. Both models are justified by the physical evidence, outlined in Chapter 1, that plastic deformation occurs by slip (or twinning) through a shear mechanism, i.e., upon the action of the shearing component of the stress tensor.

It is convenient to present and use the yield criteria with the concept of *effective* (or *equivalent*) stress, σ_{eff} , and with the related assumption that yielding in a triaxial state of stress begins when:

$$\sigma_{\text{eff}} = \sigma_Y \quad (5.1)$$

Under the assumption that σ_1 , σ_2 , and σ_3 are the principal stresses and σ_1 indicates the *maximum principal stress*, whereas σ_3 indicates the *minimum one* (so that: $\sigma_1 > \sigma_2 > \sigma_3$), in the case of the Tresca criterion we have:

$$\sigma_{\text{eff}} = \sigma_1 - \sigma_3 \quad (5.2)$$

This criterion can be deduced by the assumption that yielding begins when the *maximum shear stress* reaches a critical value. In fact, each plane containing any two principal directions also contains a maximum shear stress at an angle of 45 degrees with the principal directions. For example, in the plane containing the principal stresses σ_1 and σ_2 the maximum shear stress is: $\tau_1 = |\sigma_2 - \sigma_3|/2$. Therefore the greatest of the three maximum shear stresses is: $\tau_{max} = (\sigma_1 - \sigma_3)/2$. If the critical value of τ_{max} at yielding is denoted with C , then under tensile loading: $C = \sigma_Y/2$, implying that yielding occurs when $\tau_{max} = \sigma_Y/2$, and then: $\sigma_1 - \sigma_3 (= \sigma_{eff}) = \sigma_Y$. Though the criterion is very simple, it is typically a lower bound of the experimental data.

In the von Mises criterion, the effective stress is given by the following relationship:

$$\sigma_{eff} = \frac{1}{\sqrt{2}} \cdot \sqrt{(\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2} \quad (5.3)$$

This relationship is a better fit for the experimental data in the case of isotropic ductile metals. Compared to the Tresca criterion, it could be argued that it works better since it considers all the three maximum shear stresses, and not only the maximum one. The criterion can also be deduced from the hypothesis that yielding occurs when the octahedral shear stress reaches a critical value or when the distortion energy reaches a critical value (Dieter, 1988).

5.2 Effective strain and generalized constitutive equations

After the onset of yielding, how does the plastic deformation develop in a triaxial state of stress? To answer this question, we consider the area below the true stress-true plastic strain curve obtained in tensile loading. It defines the *specific plastic energy*, u , i.e., the energy required to plastically deform a unit volume of metal. The specific plastic energy is given by the integral of $du = \sigma_1 d\epsilon_1$. In a triaxial state of stress and for isotropic materials, du can be written in terms of the principal components:

$$du = \sigma_1 \cdot d\epsilon_1 + \sigma_2 \cdot d\epsilon_2 + \sigma_3 \cdot d\epsilon_3 \quad (5.4)$$

where $d\epsilon_1$, $d\epsilon_2$, and $d\epsilon_3$ are the elemental true plastic strains in the principal directions 1, 2, and 3, respectively (the strains ϵ_1 , ϵ_2 , and ϵ_3 are true plastic strains, indicated with ϵ_p in Section 1.1; as mentioned, we can consider them as true strains, ϵ , if the elastic portion of the strain is negligible). Therefore the *effective strain*, ϵ_{eff} , can be defined with the following relationship (Hosford and Caddell, 1993):

$$du = \sigma_{eff} \cdot d\epsilon_{eff} \quad (5.5)$$

On these bases, the *von Mises effective strain* is:

$$d\epsilon_{eff} = \frac{\sqrt{2}}{3} \cdot \sqrt{(d\epsilon_1 - d\epsilon_2)^2 + (d\epsilon_2 - d\epsilon_3)^2 + (d\epsilon_3 - d\epsilon_1)^2} \quad (5.6)$$

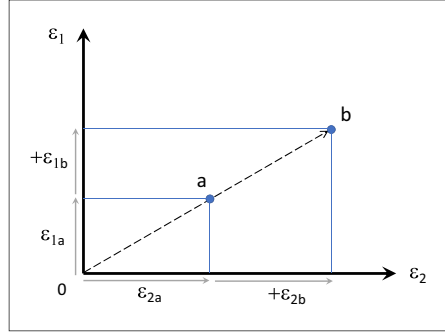


Fig. 5.1 Two-dimensional representation of a deformation path, 0-*a*-*b*, with constant $\Delta\epsilon_1/\Delta\epsilon_2$ ratio.

Under the assumption that the plastic strains at a given point of the *deformation path* depend only on the corresponding stresses, Eq. (5.6) can be rewritten in the following way (Dowling, 1999; Ragab and Bayoumi, 1998):

$$\epsilon_{eff} = \frac{\sqrt{2}}{3} \cdot \sqrt{(\epsilon_1 - \epsilon_2)^2 + (\epsilon_2 - \epsilon_3)^2 + (\epsilon_3 - \epsilon_1)^2} \quad (5.7)$$

However, in most cases (and in particular in metalworking operations), the stress and strain condition depends on the *deformation path*. Eq. (5.7) is then valid when the ratios of the principal strain increments, $d\epsilon_1/d\epsilon_2$ or $d\epsilon_1/d\epsilon_3$, are constant during the entire plastic deformation process, i.e., the deformations (and the stresses) are proportional. As an example, Fig. 5.1 represents a two-dimensional deformation path that is experienced by a material volume. In the beginning, when no plastic deformation is exerted on the material, $\epsilon_1 = \epsilon_2 = 0$. The material is then deformed to point *a*, with the application of ϵ_{1a} and ϵ_{2a} . Subsequently, the material is deformed to point *b*, with the addition of ϵ_{1b} and ϵ_{2b} . The whole deformation path is then given by the arrow: 0-*a*-*b*, starting from the origin. The ratio $\epsilon_{1a}/\epsilon_{2a}$ is equal to the ratio $\epsilon_{1b}/\epsilon_{2b}$. Therefore $\Delta\epsilon_1/\Delta\epsilon_2$ is constant and the deformation path is proportional.

In the case of a tensile (or compressive) uniaxial deformation, $\sigma_1 = \sigma_{eff}$ and $\epsilon_1 = \epsilon_{eff}$, implying that a tensile stress-strain curve, σ - ϵ , is equivalent to a $\sigma_{eff} = \epsilon_{eff}$ curve. Consider, for example, the RT stress-strain behavior of a metal, whose uniform region is properly fitted with the L-H equation (Eq. 2.7). Thus, the L-H equation can be generalized to a triaxial state of stress:

$$\sigma_{eff} = K \cdot \epsilon_{eff}^n \quad (5.8)$$

where the constants K and n are obtained from a uniaxial stress-strain test. In other words, the uniaxial tensile test can be used to obtain the material constants (such as K and n in the L-H equation), which can be used in more general (triaxial) deformation conditions.

This approach can be applied to any of the constitutive equations proposed in the previous chapters. As an example, Eq. (3.19) can be rewritten in its generalized form:

$$\sigma_{eff} = K \cdot \varepsilon_{eff}^n \cdot \left(\frac{\dot{\varepsilon}_{eff}}{\dot{\varepsilon}_0} \right)^m \cdot \left(1 - \left(\frac{T - T_0}{T_m - T_0} \right)^M \right) \quad (5.9)$$

As a last example, we may examine the conditions for the onset of dynamic recrystallization, setting that DRX starts when: $\varepsilon_{eff} = \varepsilon_{eff,c}$, where $\varepsilon_{eff,c}$ must be calculated using the given relations in a generalized form.

In the following chapters, the generalized equations will be utilized to study ductile fracture and the metalworking operations, which occur in triaxial conditions, under the assumption of proportional deformation path (and setting, as usual: $\varepsilon_p \cong \varepsilon$). This approach has some shortcomings:

- In many cases, plastic deformation does not develop along proportional deformation paths (one example concerns the barreling effect in the upsetting of a cylindrical specimen);
- The analysis is valid for materials with an isotropic behavior. Some metals do not display an isotropic behavior and others are isotropic at only small plastic strains and they turn into displaying an anisotropic behavior at large plastic strains, when special crystallographic textures develop;
- The possible presence of the *Bauschinger effect* induces an asymmetry in the tensile-compressive behavior. This effect softens the material under reverse deformation, for example, in compressive deformation after tensile plastic deformation. This is due to the formation of new dislocations under compressive loading, which will annihilate the pre-existing dislocations formed under tension and with an opposite sign. As a result, the yield strength under compression recorded after tensile deformation, is lower in magnitude than that recorded under tension. In general, deformation paths may comprise positive and negative strains but the effective stress and the effective strain are positive quantities that do not account for possible asymmetries.

A more detailed analysis is required for avoiding all these shortcomings, which is out of the scope of the present text.

The last topic that must be considered in this section regards the relationship between the stress and the strain components. In the case of elastic deformation, this relationship is based on Hooke's equation. Analogously, similar equations were proposed for the plastic deformation. Such equations are called *flow rules* or Nadai and Hencky equations. In the case of a proportional deformation path, they can be written in the following way:

$$\begin{aligned} \varepsilon_1 &= \frac{1}{E_p} \cdot [\sigma_1 - 0.5 \cdot (\sigma_2 + \sigma_3)] \\ \varepsilon_2 &= \frac{1}{E_p} \cdot [\sigma_2 - 0.5 \cdot (\sigma_3 + \sigma_1)] \\ \varepsilon_3 &= \frac{1}{E_p} \cdot [\sigma_3 - 0.5 \cdot (\sigma_1 + \sigma_2)] \end{aligned} \quad (5.10)$$

In Hooke's equations, the Poisson's ratio, ν , is present in place of 0.5. The Poisson's ratio is defined by: $\nu = -\varepsilon_2/\varepsilon_1$, referring to uniaxial deformation. In the case of plastic deformation: $\varepsilon_1 + \varepsilon_2 + \varepsilon_3 = 0$, because of the constancy of the volume. For symmetry reasons, $\varepsilon_2 = \varepsilon_3$ for uniaxial deformation and hence $\nu = 0.5$.

The parameter, E_p , in Eq. (5.10) is given by $\sigma_{eff}/\varepsilon_{eff}$ (Brozzo, 1979). This parameter generally varies during plastic deformation due to strain hardening.

5.3 Relevant stress and strain states

In this section, some relevant stress-strain states often encountered in practice are presented. In any case, reference is made to the principal directions (indicated with: 1, 2, and 3), considering that direction 1 is the main one, such as the loading direction in the case of tensile or compressive tests. In addition, reference is made to the von Mises criterion for plastic deformation.

Uniaxial state of stress

A uniaxial state of stress is encountered in tensile or compressive testing, during the uniform deformation. The relevant stress and strains are listed in Table 5.1. In the table, two additional parameters are included: the *mean of principal stresses*, σ_m , and the *triaxiality factor*, T . They are defined as:

$$\sigma_m = \frac{\sigma_1 + \sigma_2 + \sigma_3}{3} \quad (5.11)$$

$$T = \frac{\sigma_m}{\sigma_{eff}} \quad (5.12)$$

Table 5.1 Stress and strain parameters for the uniaxial state of stress.

Stress	Strain
σ_1 positive or negative	ε_1 positive or negative
$\sigma_2 = \sigma_3 = 0$ because of free deformation	$\varepsilon_2 = \varepsilon_3 = -\varepsilon_1/2$ from symmetry and
in directions 2 and 3	volume constancy
Effective stress (von Mises)	Effective strain (von Mises)
$\sigma_{eff} = \sigma_1 $	$\varepsilon_{eff} = \varepsilon_1 $
Mean of principal stresses	Triaxiality factor
$\sigma_m = \frac{\sigma_1}{3}$	$T = \frac{1}{3}$ if σ_1 is positive
	$T = -\frac{1}{3}$ if σ_1 is negative

Plane strain

Plane strain (Table 5.2) is observed when the strain component in one direction, say direction 2, is impeded, i.e., $\epsilon_2 = 0$. For example, in specific conditions, plane strain is present in flat rolling and in close-die forging operations.

Fig. 5.2 shows the geometry of a specimen for tensile testing (direction 1) in plane strain conditions. Plastic deformation occurs in the thin section at the notch. The material surrounding the notch remains elastic during loading, impeding the plastic

Table 5.2 Stress and strain parameters for plane strain conditions.

Stress σ_1 positive or negative $\sigma_3 = 0$ because of free deformation in direction 3 $\sigma_2 = \sigma_1/2$ obtained from the second flow rule	Strain ϵ_1 positive or negative $\epsilon_2 = 0$ $\epsilon_3 = -\epsilon_1$ from volume constancy
Effective stress (von Mises) $\sigma_{eff} = \frac{\sqrt{3}}{2} \cdot \sigma_1 $	Effective strain (von Mises) $\epsilon_{eff} = \frac{2}{\sqrt{3}} \cdot \epsilon_1 $
Mean of principal stresses $\sigma_m = \frac{\sigma_1}{2}$	Triaxiality factor $T = \frac{\sqrt{3}}{3}$ if σ_1 is positive $T = -\frac{\sqrt{3}}{3}$ if σ_1 is negative

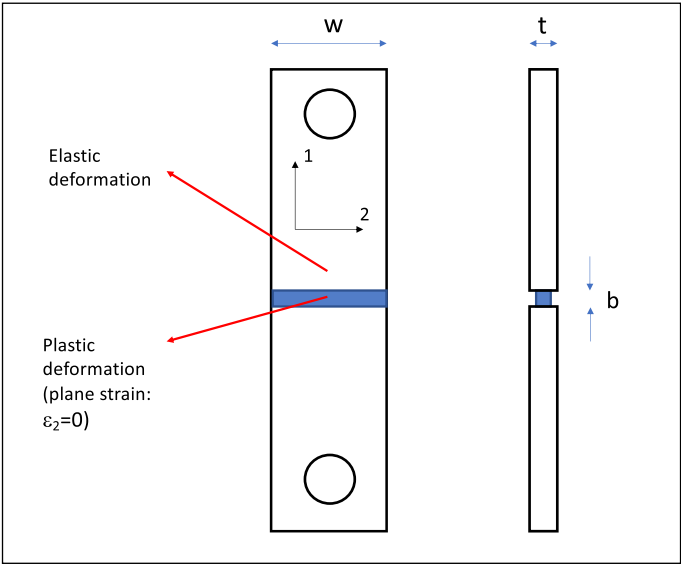


Fig. 5.2 Example of a specimen for tensile testing under plane strain conditions. When w is sufficiently larger than b and t , conditions close to plane strain (with $\epsilon_2 \sim 0$) are reached in the plastic region, i.e., in the notch region with the smallest cross section of the specimen.

deformation in direction 2. Since the plasticized material would contract in direction 2, if it were free to move (because of the constancy of the volume), a tensile stress in direction 2 (σ_2) is required to avoid this.

Plane stress

Plane stress (Table 5.3) occurs when the stress component in one direction, say direction 3, is zero, i.e., $\sigma_3 = 0$, and therefore the other two components lie on the same plane. Plane stress is typically present in sheet metal working operations. As shown in Fig. 5.3, the surfaces of the thin sheet are free, and then σ_3 (i.e., the stress perpendicular to them), is 0. In general, the ratio σ_2/σ_1 is indicated with α , and the ratio ϵ_2/ϵ_1 with β . A particular situation occurs when $\sigma_1 = \sigma_2$, i.e., $\alpha = 1$ (and $\beta = 1$). This state of stress is called *equi-biaxial plane stress*.

In the case of proportional plastic deformation, Eq. (5.10) can be used to relate the constants α and β :

$$\alpha = \frac{2\beta + 1}{2 + \beta} \quad (5.13)$$

Table 5.3 Stress and strain parameters for plane stress conditions.

Stress	Strain
σ_1, σ_2 positive or negative	ϵ_1, ϵ_2 positive or negative
$\sigma_3 = 0$ because of free deformation in direction 3	$\epsilon_3 = -(1 + \beta)$ from volume constancy
Effective stress (von Mises) $\sigma_{eff} = \sigma_1 \cdot \sqrt{1 - \alpha + \alpha^2}$	Effective strain (von Mises) $\epsilon_{eff} = \epsilon_1 \cdot \frac{2}{\sqrt{3}} \cdot \sqrt{1 + \beta + \beta^2}$
Mean of principal stresses $\sigma_m = \frac{\sigma_1 + \sigma_2}{3}$	Triaxiality factor (equi-biaxial) $T = \frac{2}{3}$ if σ_1 is positive $T = -\frac{2}{3}$ if σ_1 is negative

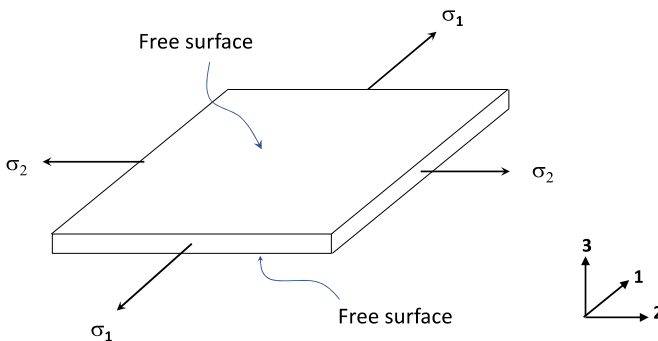


Fig. 5.3 A schematic drawing showing the plane state of stress present in a thin sheet under plastic deformation with the two surfaces free from any external loading.

Pure shear

Fig. 5.4A shows a material element subjected to a shear stress, τ . In Fig. 5.4B, the corresponding Mohr's circle is shown. In Fig. 5.4C, the element is rotated clockwise by 45 degrees to depict the corresponding stresses in the two principal directions, 1 and 2 (note that $\sigma_3 = 0$; pure shear is then a special case of plane stress). The conditions close to pure shear (Table 5.4) are encountered in shearing and machining operations, as schematized in Fig. 5.5 for punch shearing. The deformation induced by the shear stress, τ , can also be evaluated regarding the corresponding shear strain, γ , defined in Fig. 5.6. In case of pure shear, $\gamma = 2\varepsilon_1$, and $\varepsilon_{eff} = \gamma/\sqrt{3}$.

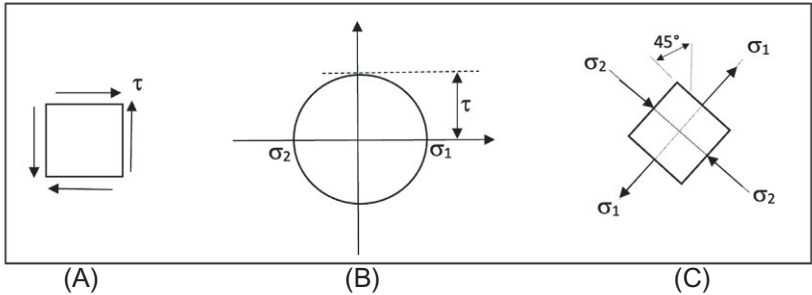


Fig. 5.4 Mohr's circle for pure shear.

Table 5.4 Stress and strain parameters for pure shear.

Stress	Strain
σ_1 positive or negative	ε_1 positive or negative
$\sigma_2 = -\sigma_1$	$\varepsilon_3 = 0$
$\sigma_3 = 0$	$\varepsilon_2 = -\varepsilon_1$ from volume constancy
Effective stress (von Mises) $\sigma_{eff} = \sqrt{3} \cdot \sigma_1 = \sqrt{3} \cdot \tau$	Effective strain (von Mises) $\varepsilon_{eff} = \frac{2}{\sqrt{3}} \cdot \varepsilon_1 = \frac{\gamma}{\sqrt{3}}$
Mean stress $\sigma_m = 0$	Triaxiality factor $T = 0$

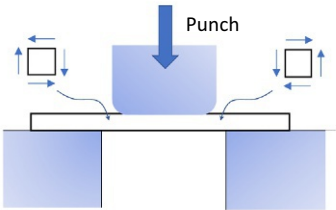


Fig. 5.5 Sheet shearing process with the indication of the regions of the sheets undergoing shear deformation.

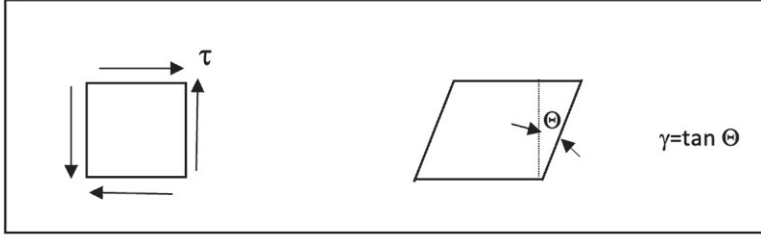


Fig. 5.6 Definition of shear strain, γ .

5.4 Anisotropy in plastic deformation

In general, it is assumed that metals display an *isotropic* behavior under deformation. However, in some cases this assumption is not valid.

In [Section 2.7](#), we have seen the important case of sheet metal forming processes and the common plastic anisotropy of sheet specimens due to the presence of crystallographic textures. The parameter R_m was also introduced. As mentioned, it can be assumed that in sheet metal forming processes, plane stress conditions are attained in the sheets, with $\sigma_3 = 0$, as direction 3 is the sheet thickness direction ([Fig. 5.3](#)). The analysis is rather complex. [Hill \(1950\)](#) formulated the most widely used approach to this topic with a modification of the von Mises yield equations. The resulting equations for σ_{eff} and ε_{eff} are:

$$\sigma_{eff} = |\sigma_1| \cdot \sqrt{1 - \frac{2 \cdot R_m}{R_m + 1} \cdot \alpha + \alpha^2} \quad (5.14)$$

$$\varepsilon_{eff} = |\varepsilon_1| \cdot \frac{1 + R_m}{\sqrt{1 + 2 \cdot R_m}} \cdot \sqrt{1 + \frac{2 \cdot R_m}{1 + R_m} \cdot \beta + \beta^2} \quad (5.15)$$

and:

$$\beta = \frac{(1 + R_m) \cdot \alpha - R_m}{(1 + R_m) - R_m \cdot \alpha} \quad (5.16)$$

Other criteria are proposed in the literature that provide a better fit of the experimental data in the case of metals with a specific behavior, such as Al alloys. These alloys display R_m values that are lower than unity (typical values are in the range 0.6–0.8; refer to [Table 2.4](#)), and the Hill criterion must be modified to model this behavior. In 1990, Hill proposed an improvement of his criterion. However, new criteria have been continuously developed and validated ([Bruschi et al., 2014](#)).

5.5 Plastic deformation at a notch root

In this section, the development of plastic deformation at a notch root is analyzed. This is an important aspect of yielding and plastic deformation in a triaxial state of stress. Additionally, notched specimens are often used to investigate the ductility of metals (also under impact conditions).

Fig. 5.7 represents the tensile loading of a rectangular specimen with two symmetrical notches at both sides. The coordinate axes are also principal directions and therefore will be denoted in the usual way. The upper crosshead travels with a constant velocity, v , and F is the applied load. The initial length and width of the specimen, indicated with L_0 and w_0 , respectively, are also shown in the figure. The notch angle is indicated with Θ and the notch depth with a . The root radius and the specimen thickness (not shown in the figure) are denoted by ρ and t , respectively. It is assumed that t is bigger than a , in such a way that at the notch root, a plane strain condition is present. In fact, the metal under elastic deformation that is close to the notch impedes the metal plastic contraction along its front, i.e., in direction 3 ($\epsilon_3 = 0$). The situation is similar to the testing sample in Fig. 5.2.

The typical load-displacement (F - ΔL) curve is shown in Fig. 5.8A. From the origin up to the point A, the deformation in the specimen is elastic everywhere, with an

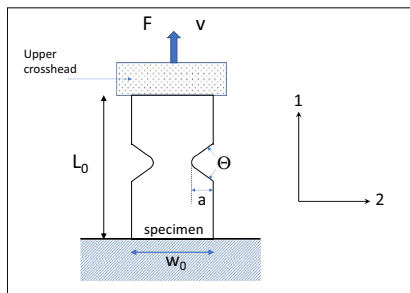


Fig. 5.7 A notch rectangular specimen subject to tensile loading.

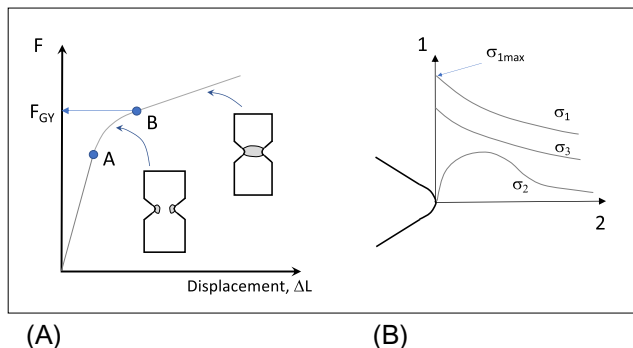


Fig. 5.8 (A) Force-displacement curve of a notched specimen. (B) Elastic stress distribution beneath a notch in plane strain.

elastic stress concentration at the notch root. A schematic of the stress distribution along the three directions (1, 2, and 3) in the notch is shown in Fig. 5.8B. The stress σ_1 reaches a maximum value at the notch root. It is indicated with σ_{1max} in the figure. The ratio between σ_{1max} and the nominal stress, σ_n (load divided by resisting section, i.e., $F/(w_0 - 2 \cdot a) \cdot t$) is the elastic stress concentration factor, K_t .

At a critical value of F , plastic deformation starts at the notch root, when $\sigma_{eff} = \sigma_Y$, where σ_Y is the uniaxial yield strength of the material. For the sake of simplicity, we use here the Tresca criterion. Then: $\sigma_{eff} = \sigma_{1max} - 0 = K_t \cdot \sigma_n = \sigma_Y$. Localized yielding at the notch root starts when $\sigma_n = \sigma_Y/K_t$.

Now, if load F is increased, plastic deformation spreads over the resisting section. Moving from the notch root along direction 2, the stress σ_2 increases from its initial value of 0 (refer to Fig. 5.8B), indicating that the stress σ_1 must also increase to satisfy the Tresca criterion ($\sigma_1 = \sigma_Y + \sigma_2$). In particular, using the Tresca criterion, the slip-line theory, and neglecting the material strain hardening in the plastic region, we obtain (Hill, 1950):

$$\sigma_1 = \sigma_Y \cdot \left[1 + \ln \left(1 + \frac{x_2}{\rho} \right) \right] \quad (5.17)$$

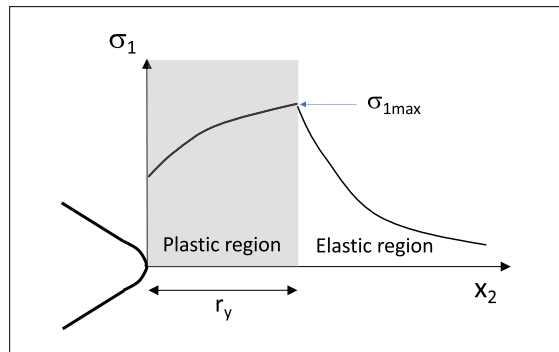
where x_2 is the distance from the notch root along direction 2. This behavior is shown in Fig. 5.9. The maximum value of σ_1 is attained at the interface between the plastic and the elastic region. It is indicated with σ_{1max} in the figure, and can be obtained from Eq. (5.17) by substituting x with r_y , where r_y is the *plastic radius*. The ratio σ_{1max}/σ_Y is indicated with K_p and it is called the *plastic stress intensification factor*.

With further increase in the load F , r_y and σ_{1max} increase up to a limiting value, reached when $r_y = r_0$, where (Brozzo, 1979):

$$r_0 = \rho \cdot \left(\exp \frac{\pi - \Theta}{2} - 1 \right) \quad (5.18)$$

Beyond this point, the plastic radius r_y increases without any additional increase in σ_1 , up to the complete plasticization of the resisting section. This condition is indicated with point B in Fig. 5.8A and is called *generalized (or gross) yielding*. At this point:

Fig. 5.9 Representation of the plastic evolution at the notch root when the load increases along the line AB in Fig. 5.6A (at the notch root $\sigma_1 = \sigma_Y$, following the Tresca criterion).



$K_p = 1 + (\pi - \Theta)/2$. Its maximum value, attained when $\Theta = 0$, is 2.57. The maximum K_p value is also indicated with: $K_{p,max}$.

Note that if the specimen thickness, t , is comparatively smaller than a , plane stress conditions are attained at the notch root (since plastic deformation in direction 3 is free), and $K_p = 1$, i.e., no stress intensification occurs in the plastic region.

Fig. 5.8 can be compared with Fig. 1.9A, and the analysis conducted in Section 1.3. In engineering materials, porosity and other microstructural heterogeneities may then be regarded as *internal notches*. During mechanical loading, plastic deformation first starts close to such heterogeneities because of their stress-concentrating effect. This phenomenon is termed localized yielding. Subsequently, plastic deformation propagates, involving the whole load-bearing section. We may then have a continuous yielding up to generalized yielding.

In the plastic region, σ_1 is given by Eq. (5.17), whereas σ_2 can be estimated from the Tresca criterion: $\sigma_2 = \sigma_1 - \sigma_Y$. Stress σ_3 can be estimated from the flow rules (the third one), setting $\epsilon_3 = 0$: $\sigma_3 = \sigma_1 - \sigma_Y/2$. We can then evaluate the triaxiality factor, T , in the plastic region at the notch root:

$$T = \frac{\sigma_m}{\sigma_Y} = \frac{1}{2} + \ln \left(1 + \frac{x_2}{\rho} \right) \quad (5.19)$$

At the notch root, i.e., for $x_2 = 0$, $T = 1/2$ is obtained for plane strain conditions using the Tresca criterion. Moving from the notch root along direction 2, T increases.

As mentioned, in point B of the force-displacement curve, generalized yielding is attained. The corresponding load is indicated with F_{GY} . It can be calculated by integrating σ_1 on the resisting section. Using Eq. (5.17), the following relationship is obtained (Knott, 1973):

$$F_{GY} = 2 \cdot \sigma_Y \cdot (w_0 - a) \cdot B \left(1 + \frac{a}{w_0 - a} \right) \cdot \ln \left(1 + \frac{w_0 - a}{a} \right) \quad (5.20)$$

The ratio, L , between F_{GY} and the force required to yield a specimen without notches and with the same load-bearing section is given by:

$$L = \frac{F_{GY}}{2 \cdot \sigma_Y \cdot (w_0 - a) \cdot B} = \left(1 + \frac{a}{w_0 - a} \right) \cdot \ln \left(1 + \frac{w_0 - a}{a} \right) \quad (5.21)$$

Parameter L is called the *constraint factor*.

5.6 Tensile testing and post-necking behavior

Cylindrical specimens

In Section 1.1, the uniaxial tension test using cylindrical smooth (i.e., un-notched) specimens were described in detail. Fig. 1.2 represents the engineering stress-strain curve and the corresponding true stress-true strain curve. In point B (or B'), necking

is attained and the geometry of the specimen changes. Due to this, the state of stress in the resisting section also changes. It becomes triaxial and this must be properly taken into account in obtaining the material true stress-true strain curve (and also in evaluating the metal ductility).

Fig. 5.10A shows a cylindrical tensile specimen after necking. Because of necking, a circumferential notch is formed. The radius of the load-bearing section (A) is indicated with a , whereas the curvature radius of the necking profile is indicated with R . The evolution of σ_1 on the load-bearing section for a non-hardening material has been obtained by Bridgman (1952):

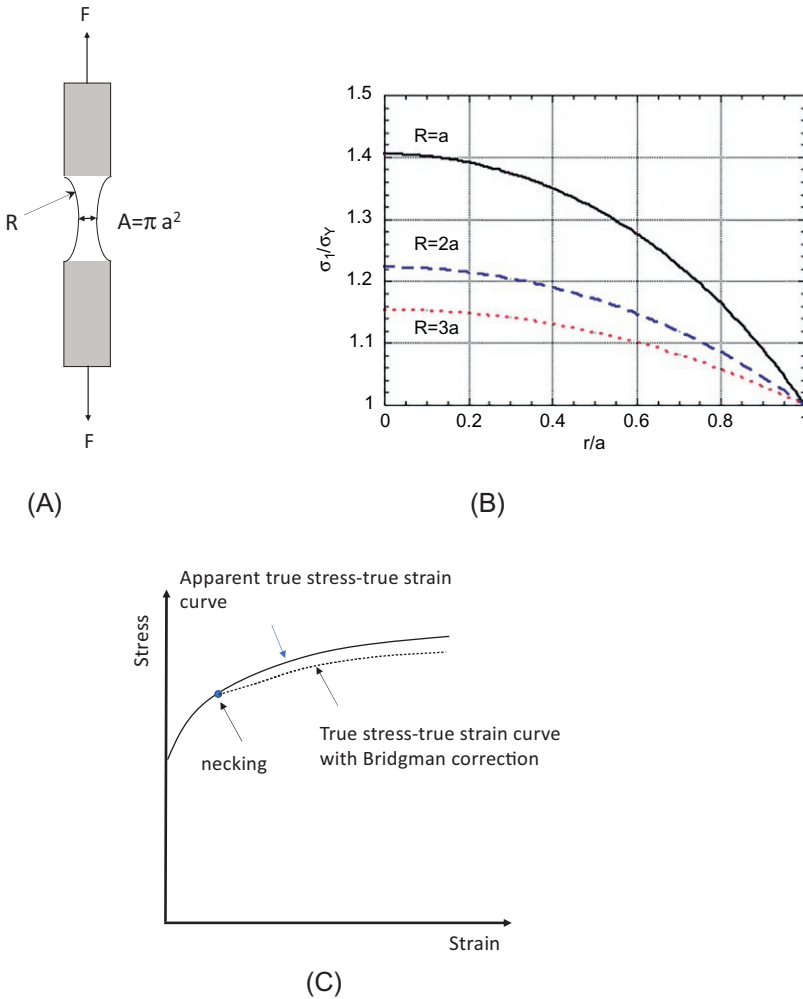


Fig. 5.10 (A) Representation of a cylindrical tensile specimen after necking; (B) evolution of σ_1 on the load-bearing section for different R/a values; (C) effect of the Bridgman correction to the true stress-true strain curve.

$$\sigma_1 = \sigma_Y \cdot \left(1 + \ln \frac{a^2 + 2aR - r^2}{2aR} \right) \quad (5.22)$$

where r is the radial coordinate ($r=0$ refers to the center of the specimen and $r=a$ refers to a point on the outer surface). Eq. (5.22) is schematized in Fig. 5.10B for different R/a values.

The recorded force, F , during tensile testing is then influenced by the presence of the circumferential notch. Therefore the recorded true stress in direction 1, simply given by: $\sigma_1 = F/\pi a^2$, is an average value that must be corrected considering the constraint effect exerted by the notch. Considering the proposed model, F can be expressed by the integral of σ_1 on the resisting section, as done in the previous section to obtain Eq. (5.20) for the notched specimen shown in Fig. 5.7. Therefore to obtain the real stress acting in the material in direction 1, it is necessary to divide the recorded true stress (σ_1) by the corresponding constraint factor (L). For the geometry of Fig. 5.10A, L is given by the following relationship (Bridgman, 1952):

$$L = \left(1 + \frac{2R}{a} \right) \cdot \ln \left(1 + \frac{a}{2R} \right) \quad (5.23)$$

This is called the *Bridgman correction*.

As already mentioned in Section 1.1, the true strain in the neck region is given by:

$$\varepsilon_1 = \ln \frac{A_o}{A} \quad (5.24)$$

Experimental observations have shown that it is almost uniform across the resisting section (of course it changes along the loading direction; it is worth noticing here that $\varepsilon_2 = \varepsilon_3 = -\varepsilon_1/2$, as in uniaxial loading). In Fig. 5.10C, a depiction of the true stress-true strain curve after the Bridgman correction is shown.

The triaxiality factor is maximum at the center of the minimum cross section and it is given by:

$$T = \frac{\sigma_m}{\sigma_Y} = \frac{1}{3} + \ln \left(1 + \frac{a}{2R} \right) \quad (5.25)$$

where σ_Y may be replaced with σ_{eff} to account for strain hardening. Of course, $T = 1/3$ in the uniform region, before necking. T then increases with the ratio a/R . As mentioned in Section 1.1, ε_1 can be obtained by continuously measuring the load-bearing section with a proper transverse strain measuring extensometer. The estimation of R and its variation with straining is rather difficult and more sophisticated methods are required. An experimental relationship for estimating R as a function of ε_1 has been proposed by Ragab (2000):

$$\frac{a}{R} = 3.2 \cdot \left[1 - \exp \left(-\frac{\varepsilon_1 - n}{3} \right) \right] \quad (5.26)$$

Quite often, the plastic deformation and fracture behavior of metals is investigated using cylindrical and *circumferentially notched specimens*. By changing the geometry of the notch, it is possible to change the (initial) triaxiality factor. This is useful, for example, in investigating the damage evolution and fracture behavior of metals and in particular the influence of triaxiality on the ductility of metals (outlined in the next chapter). As an example, Fig. 5.11 shows the experimental true stress-true strain curves for a steel with 0.2% of C, obtained using two types of specimens. The first is a usual smooth cylindrical specimen with a diameter of 6 mm. The second is a circumferentially notched cylindrical specimen. Two notch radii were adopted: $R = 1$ mm and $R = 2$ mm, with a radius of 3 mm of the notch cross section in both cases. From Eq. (5.23), it is seen that L is equal to 1.53 and 1.3, and from Eq. (5.25), it is seen that T is equal to 1.25 and 0.89, respectively (of course, these are initial values, and during tensile testing, the ratio a/R changes due to plastic deformation). The experimental curves in Fig. 5.11 show that at each strain level, the corresponding flow stress increases with the stress triaxiality (this is the so-called *notch strengthening effect*; see the vertical arrow on the left of the figure). In passing, it is also noted that with increasing triaxiality, the true strain ϵ_1 at fracture (i.e., the ductility) decreases (see the horizontal arrow on the right of the figure).

In the figure, the true stress-true strain curve is also reported after the Bridgman correction, obtained using Eqs. (5.23) and (5.26). The true stress-true strain curve before necking (for true strains up to approximately 0.2) and the Bridgman corrected curve after necking (with true strains up to 1.3) can be fitted together by the following L-H equation: $\sigma = 900\epsilon^{0.2}$. This suggests that this method, although approximate, can be used to assess the true stress-true strain behavior of metals at large strains. This is

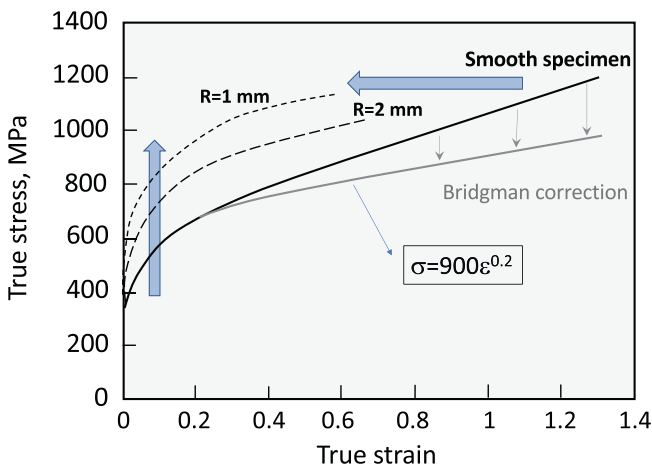


Fig. 5.11 Experimental tensile true strain-true stress curves for an unnotched and two notched specimens made with a steel with 0.2%C (redrawn from Dos Santos and Ruggeri, 2003), on the basis of data reported by Yasuda et al. (2000). With increasing notch severity (i.e., the triaxiality), the flow stress increases (see vertical arrow on the left) and the strain at fracture decreases (see horizontal arrow on the right).

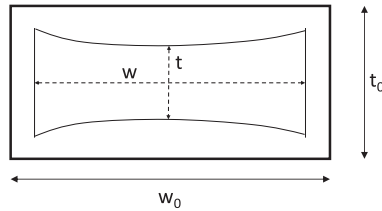


Fig. 5.12 Cushion-like shape of a rectangular tensile specimen after necking.

very useful in applications where large plastic strains are involved, such as in metal-working applications, or to estimate the large strains encountered at the notch/crack tip in fracture mechanical tests (Kamaya and Kawakubo, 2011).

When the a/R ratio is very high, the strain localization and the triaxiality parameter can become very high at the notch root. In this case, plastic damage (and then fracture) may occur at the notch tip before plastic deformation is able to spread on the entire load-bearing section (in this case the notch mechanically behaves like a crack).

Rectangular specimens

Tensile testing using rectangular specimens (refer to Fig. 1.1) allows obtaining stress-strain curves that are similar to those obtained using conventional cylindrical specimens. This is true if the width to thickness ratio of the rectangular specimen is sufficiently low, i.e., the strain gradients in the resisting cross section are not too prominent (Zhang et al., 1999). Also, localized deformation (in narrow bands after diffuse necking; see Section 2.8) is generally not observed in rectangular specimens with a sufficiently low width to thickness ratio.

However, the post-yielding behavior in rectangular specimens is more complex and the Bridgman correction cannot be easily used to obtain the real true stress-true strain curve. In particular, after necking, the cross section develops a particular *cushion-like shape*, shown in Fig. 5.12. For materials with isotropic behavior, the approach proposed by Scheider can be used to evaluate proper factors to correct the true-stress strain curve (Scheider et al., 2004).

5.7 Hydraulic bulge test and torsion test

In the case of tensile testing of a sheet specimen, diffuse necking occurs for $\epsilon_1 = n$, and localized necking for $\epsilon_1 = 2n$, or before, if imperfections are present (and m is negative). Considering the difficulty in using the Bridgman correction outlined in the previous section, it is very difficult to obtain the stress-strain curve for large strains, exceeding the uniform strain. Similarly, the barreling phenomenon limits the maximum possible strains in the case of compression testing of cylindrical specimens. However, both sheet metal working operations and bulk working operations (such as extrusion, rolling, or forging) generally involve plastic strains that are much larger than those encountered in the case of sheet tensile testing and in compression

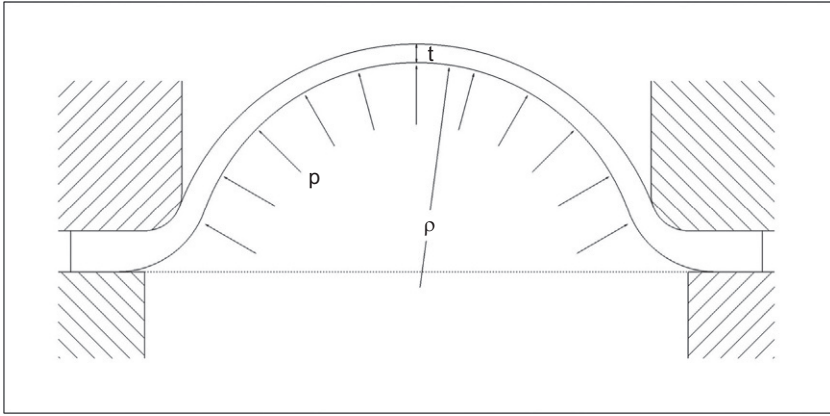


Fig. 5.13 A typical representation of the hydraulic bulge test (p : oil pressure; t : dome apex thickness; ρ : curvature radius at the apex).

testing. Therefore different special tests, such as the *hydraulic bulge test* (for sheet specimens) and the *torsion test* (for cylindrical specimens) are used to investigate the plastic behavior of metals at large plastic strains. In the corresponding stress and strain conditions, necking phenomenon is delayed or absent.

The principle of the hydraulic bulge test is shown in Fig. 5.13 (ISO, 2014). A blank with initial thickness t_0 is clamped to a circular die through a blank holder and it is plastically deformed (bulged) by the hydraulic pressure exerted by a fluid to form a hemispherical cup (Bruschi et al., 2014). The specimen sheet deforms under equi-biaxial conditions; the plastic strains in the sheet are $\varepsilon_1 = \varepsilon_2$ (and: $\varepsilon_3 = -2\varepsilon_1$), and the true stresses are $\sigma_1 = \sigma_2$ (and: $\sigma_3 = 0$) as seen in Table 5.3. In general, interrupted tests are carried out at specific values of the fluid pressure, p , without reaching the conditions of ductile fracture. For each value of p , the corresponding curvature radius at the apex, ρ , and the dome apex thickness, t , are measured. Assuming pure equi-biaxial conditions:

$$\sigma_1 = \sigma_2 = \frac{p \cdot \rho}{t}$$

and under the additional assumption of proportional strain path: $\varepsilon_1 = \varepsilon_2 = -\varepsilon_3/2$, where:

$$\varepsilon_3 = \ln \left(\frac{t}{t_0} \right)$$

Referring to Section 5.3, $\varepsilon_{eff} = 2\varepsilon_1 = -\varepsilon_3$, and $\sigma_{eff} = \sigma_1$. From the experimental points, the $\sigma_{eff} - \varepsilon_{eff}$ curve can be obtained up to large strains. An example is shown in Fig. 5.14 for an AA6061-T6 Al alloy (Mirnia and Vahdani, 2020). In the figure, the results of a simple uniaxial tensile test are included for comparison. It can be seen that the data of the uniaxial tests and those obtained with the hydraulic bulge test can be

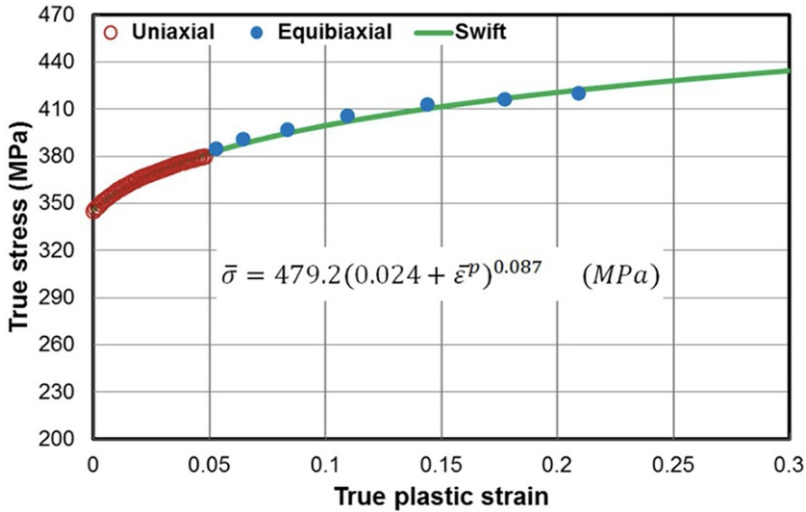


Fig. 5.14 $\sigma_{eff} - \epsilon_{eff}$ curves of a 6061-T6 Al alloy sheet obtained from tensile and hydraulic bulge tests.

From Mirnia, M.J., Vahdani, M., 2020. Calibration of ductile fracture criterion from shear to equibiaxial tension using hydraulic bulge test. *J. Mater. Process. Technol.* 280, 116589, reproduced with permission of the publisher.

fitted with the same interpolating curve (in this case, the Swift equation is used, which has the general form: $\sigma = K \cdot (a + \epsilon)^n$; it is similar to the L-H equation but it fits the experimental data better for small strains). However, it is also seen that the data from tensile testing refer to rather low strains, since only the data up to the diffuse necking are used. It is clear that in such cases, the hydraulic bulge test allows modeling of the stress-strain behavior of metal sheets with better accuracy, particularly at large strains. In addition, this test helps in calibrating the relevant constitutive equations and yielding criteria with higher reliability (Sigvant et al., 2009).

The principle of the torsion test is schematized in Fig. 5.15. A cylindrical specimen with length L and radius a is fixed at one end and subjected to a torque, M_T , at the other end. The specimen deforms under pure shear without necking, and the principal stresses make an angle of 45 degrees with the longitudinal axis of the cylinder, whereas the corresponding maximum shear stress is perpendicular and parallel to this axis. During the test, the applied torque, M_T , and the angle of twist, Θ , are recorded. The effective (von Mises) strain can be obtained from the following relationship (Dieter, 1988):

$$\sigma_{eff} = \frac{\sqrt{3} \cdot M_T}{2 \cdot \pi \cdot a^3} \cdot (3 + m + n)$$

In the case of hot deformation of metals (i.e., for $T > 0.4 - 0.5 T_m$), the strain-hardening coefficient, n , tends to 0 (in the secondary creep stage), whereas the strain rate sensitivity coefficient, m , gains in importance (Chapter 4).

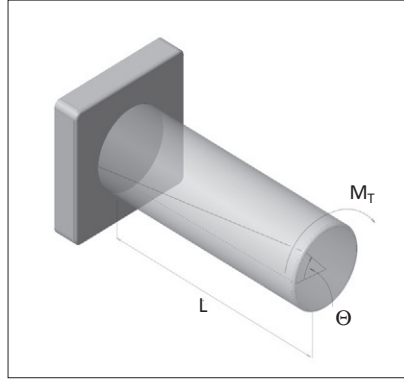


Fig. 5.15 Schematization of the torsion test of a cylindrical specimen.

Since: $\gamma = a \cdot \theta / L$, the effective strain can be calculated with the following relationship:

$$\epsilon_{eff} = \frac{2 \cdot \pi \cdot a}{\sqrt{3} \cdot L} \cdot N$$

where N is the number of turns.

An example is shown in [Fig. 5.16](#) for an AISI M2 tool steels ([Imbert et al., 1984](#)). In the figure, the obtained effective stress-strain curves between 900 and 1100°C and

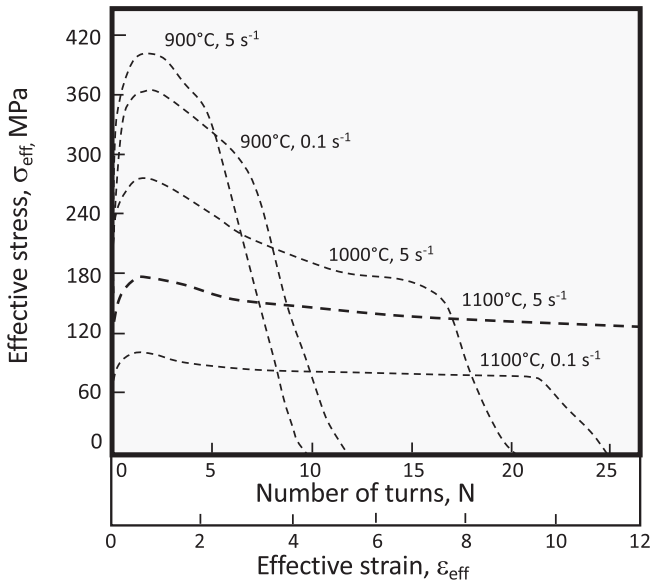


Fig. 5.16 $\sigma_{eff} - \epsilon_{eff}$ curves of an AISI M2 tool steel obtained from the hot torsion test; the number of turns are also indicated.

Redrawn from Imbert, C., Ryan, N.D., McQueen, H.J., 1984. Hot workability of three grades of tool steels. *Metall. Trans. A* 15, 1855–1864.

at 0.1 to 5 s^{-1} are shown. It is seen that very large strains can be obtained, especially at the highest temperatures. It has been observed that the flow curves obtained using the torsion tests are very similar to those obtained by compression.

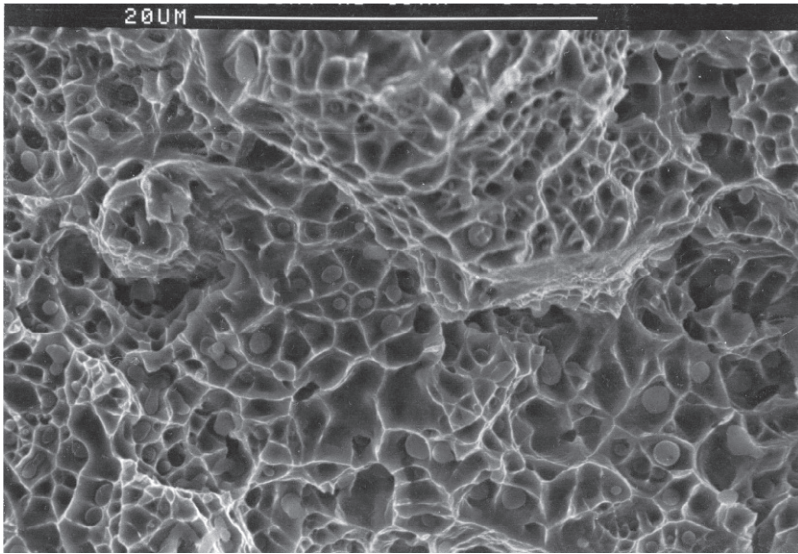
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Mechanisms of plastic damage and ductile fracture

6



Typical dimpled ductile fracture morphology.

Chapter outline

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This and the next two chapters are the core of the book. This chapter is focused on the damage mechanisms responsible for the ductile fracture of metals, which represents their ductility limit. The next chapters present the influence of microstructure, temperature, and strain rate on ductility of metals.

The ductile fracture mechanisms comprise voids nucleation, voids growth, and finally voids coalescence to form a ductile crack. The damage mechanisms depend on the metal microstructure, the applied effective strain, and the local state of stress (in particular the local stress triaxiality). It is advantageous to consider the damaging process as a sequence of distinct stages, even if some nucleation may still continue when some voids have already begun to grow. With this approach, a ductile fracture criterion based on microstructural damage is then proposed and discussed. The criterion uses the Rice and Tracey analysis to model the voids growth.

The knowledge of the damaging mechanisms responsible for the ductile fracture and the use of a proper ductile fracture criterion is paramount to understand and predict the ductility of metals in different conditions. In particular, this is very useful in treating the metal-forming processes by plastic deformation with respect to the possible failure by ductile fracture, considering that for most processes ductile cracking is an important defect and must be carefully avoided, but for some others it is used as an advantage.

6.1 Ductile fracture in metals: An overview

Fig. 1.1C shows the typical stress-strain curve in the case of a cylindrical specimen under uniaxial tension. As already observed, after uniform deformation, necking occurs and the geometry of the specimen changes with the formation of a circumferential notch. Upon increasing tensile deformation, the conditions for fracture are attained in the neck region. In metals, fracture is then called *ductile*, as it typically occurs after a large plastic deformation.

Three main macroscopic ductile fracture processes can be recognized (Noell et al., 2018):

1. *Purely plastic fracture* (necking to a point);
2. *Dimpled ductile fracture*;
3. *Shear ductile fracture*.

Purely plastic fracture is the upper limit of metal ductility, since it occurs when the resistant section vanishes. It typically takes place when no damage mechanisms intervene during plastic deformation, and then in pure metals, or in engineering alloys if the damaging mechanisms are suppressed through a superimposed large hydrostatic pressure and/or DRX phenomena (possibly occurring at high temperature). Two examples are shown in Fig. 6.1.

Dimpled ductile fracture is the most common fracture mechanism in engineering metals at temperatures approximately below $0.7\text{--}0.8T_m$ and is considered in detail in the present chapter. It is characterized by three stages of *plastic damage* before final fracture:

- Voids nucleation;
- Voids growth;
- Coalescence of adjacent voids to form a crack, which will lead to fracture with the typical dimpled morphology.

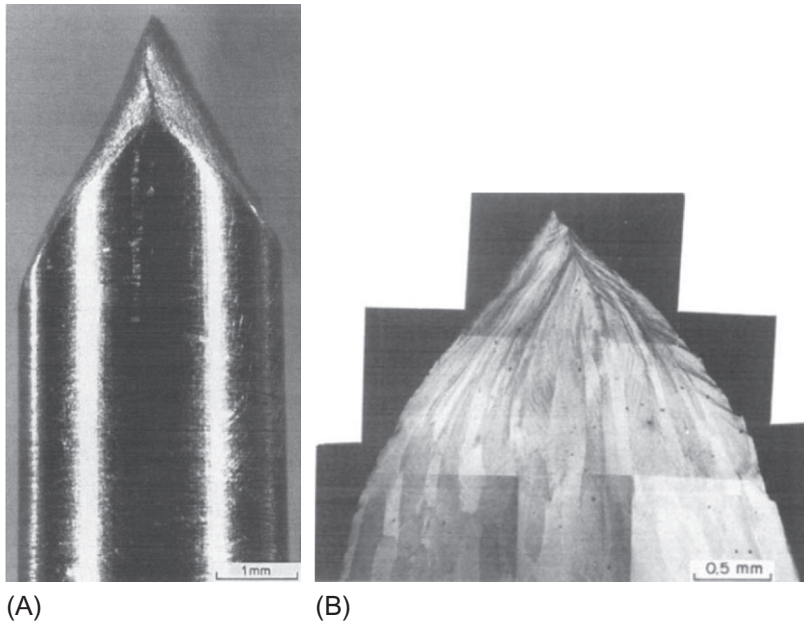


Fig. 6.1 Examples of purely plastic fracture (necking to a point) in two Al alloy specimens broken under tensile testing with a superimposed pressure of (A) 1100 MPa and (B) 690 MPa. Reproduced from Teirlink, D., Zok, F., Embury, J.D., Ashby, M.F., 1988. Fracture mechanism maps in stress space. *Acta Metall.* 36, 1213–1228, with permission of the Publisher.

The main factors responsible for voids nucleation are *particles* in the microstructure. Voids nucleation is not necessary when voids are pre-existing, for example in the form of porosity. Some metals, such as Ti alloys, do not contain nucleating particles, and voids nucleate at phase interfaces as a result of the local accumulation of dislocations (this is seen in the next chapter).

Shear ductile fracture is a special case of dimpled ductile fracture. It occurs when plastic deformation localizes in narrow shear bands. As discussed in [Section 3.6](#), there can be spontaneous or forced shear localization. In every case, the damaging sequence occurs inside the shear bands, where plastic deformation is concentrated and intensified. Of course, shear localization anticipates the attainment of the ductile fracture conditions; therefore, in most cases, the occurrence of shear ductile fracture is rather detrimental.

6.2 Particles in engineering metals

Engineering metals are typically characterized by a heterogeneous microstructure ([Fig. 1.8A](#)) and contain different kinds of particles. These particles are broadly classified into:

- *Inclusions* (such as oxides and sulfides);
- *Intermetallic precipitates* (such as carbides).

The characteristics and number of such particles play an important role in the process of ductile damage and fracture. Important parameters to characterize engineering particles are:

- Volume fraction, f_v (given by the ratio of the total volume of particles and the corresponding volume of material);
- Area fraction, f_a (given by the ratio of the area of particles on a material cross section, and the corresponding area of the cross section);
- Mean particle radius, R ;
- Mean number per unit volume, N_v ;
- Mean number per unit area, N_a ;
- Mean spacing, λ_v .

Generally, it is assumed that $f_v = f_a$. In addition, assuming that the particles have a spherical geometry, the following relationships are obtained (Vander Voort, 1984):

$$f_a = N_a \cdot \pi \cdot R^2 \quad (6.1)$$

$$N_v = \frac{f_v}{\frac{4}{3} \cdot \pi \cdot R^3} \quad (6.2)$$

$$\lambda_v = \left(\frac{1}{N_v} \right)^{1/3} \quad (6.3)$$

Naturally, the particles may have a complex geometry. In addition, their size follows a distribution around a mean value. Using well-established metallographic techniques, the mean values of N_a and R are quite easily determined from microstructural observations (e.g., through optical or scanning electron microscopy). Consequently, the mean values of N_v and λ_v can be estimated with Eqs. (6.1)–(6.3).

The main particles in engineering metals and their characteristics will be described in the next chapter.

6.3 Voids nucleation

Voids nucleation is the first stage in the plastic damage of metal alloys. If the alloy contains porosity, such pores act as pre-existing voids. In this case, this stage is not required. In engineering alloys, voids nucleation generally occurs at particles present in the microstructure. In fact, particles induce a strain and stress concentration in the surrounding matrix that favors voids nucleation, either by *interfacial decohesion* or *particle fracture*. Two examples are shown in Fig. 6.2 for a spheroidized steel containing 0.17%C and 1.29%Mn (Kwon and Asaro, 1990). In the figure, a schematization of the two mechanisms is also shown.

Voids nucleation is attained when the tensile stress at the particle, σ , reaches a critical value, σ_c . In the case of nucleation by interfacial decohesion, σ_c is related to the strength of the particle-matrix interface. In the case of nucleation by particle cracking, σ_c depends on the intrinsic fracture toughness of the particles, their morphology, and size. In fact, particles with an irregular morphology contain defects that favor their

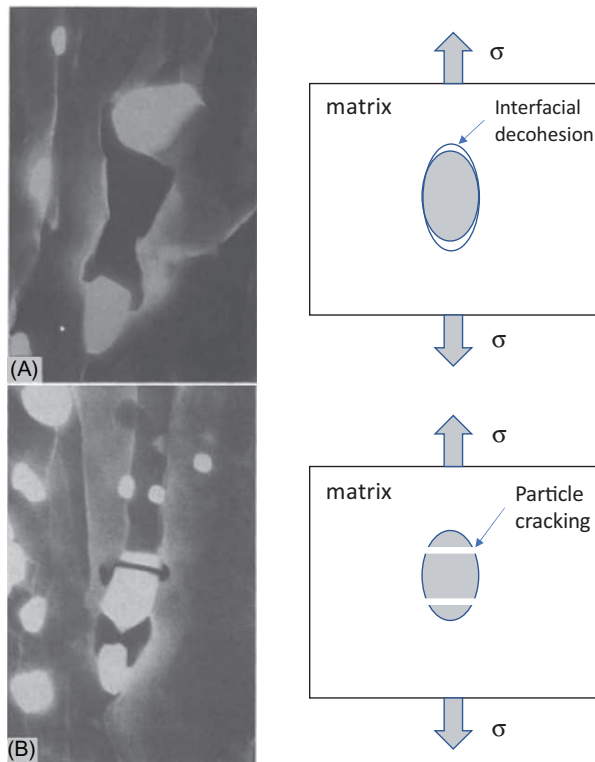


Fig. 6.2 (A) Interfacial decohesion around large inclusions, and (B) fracture of a large elongated particle. The magnification is not known. A schematization of the two mechanisms is shown at the right side of the picture.

Reproduced from Kwon, D.I., Asaro, R.J., 1990. Hydrogen-assisted ductile fracture in spheroidized 1518 steel. *Acta Metall. Mater.* 38, 1595–1606, with permission of the Publisher.

brittle fracture; in addition, in comparatively large particles it is statistically more probable to have critical defects for cracking. The lowest σ_c values will determine the prevailing mechanism of voids nucleation.

Different models have been proposed to evaluate the tensile stress at the particles. A well-known model was proposed by [Argon et al. \(1975\)](#). It follows a continuum approach, and considers that the tensile stress at the particle, σ , is given by the sum of the local average plastic stress, σ_{eff} , and the mean (or hydrostatic) stress, σ_m . Nucleation then occurs when the critical stress, σ_c , is reached:

$$\sigma_{\text{eff}} + \sigma_m = \sigma_c \quad (6.4)$$

Subsequently, Eq. (6.4) was slightly modified to better fit the experimental data ([Needleman, 1987](#)):

$$\sigma_{\text{eff}} + c \cdot \sigma_m = \sigma_c$$

where c is a material constant in the range of 0–1. Setting $\sigma_{\text{eff}} = K \cdot \varepsilon_{\text{eff}}^n$ (Eq. 2.7), the effective nucleation strain, $\varepsilon_{\text{eff},n}$, can be estimated:

$$\varepsilon_{\text{eff},n} = \left(\frac{\sigma_c - c \cdot \sigma_m}{K} \right)^{1/n} \quad (6.5)$$

Eq. (6.5) can be rewritten using the concept of stress triaxiality, $T = \sigma_m/\sigma_{\text{eff}}$:

$$\varepsilon_{\text{eff},n} = \left(\frac{\sigma_c}{K \cdot (1 + c \cdot T)} \right)^{1/n} \quad (6.6)$$

The model is rather simple but it captures some of the most important physical parameters involved in the nucleation of voids. For example, the model predicts a decrease in the nucleation strain as σ_c is decreased. For Fe oxides and MnS inclusions in steels, the σ_c values are comparatively low, in the range of 350–600 MPa for nucleation by decohesion (Testa et al., 2020; Wcislik and Pala, 2021). Correspondingly, the nucleation strain is in the range of 0.01–0.3. In some cases, σ_c may be vanishingly small and then $\varepsilon_{\text{eff},n} \sim 0$. This may happen when the particles are cracked or detached from the matrix during the fabrication process, and before any applied deformation (this could be the case for MnS inclusions). In the case of nucleation by decohesion at cementite particles, the σ_c values are higher, above 1200 MPa, and the nucleation strains are also higher, greater than 0.3. In spheroidized steels containing rounded submicrometric carbides, the nucleation strain is quite high, greater than 0.5 (Le Roy et al., 1981; Straffelini, 2003).

Eq. (6.6) also correctly predicts the role of stress triaxiality, namely a decrease in the nucleation strain with an increase in T . This behavior is confirmed by the experimental data reported for Swedish iron containing Fe oxides (Hancock and Thomson, 1985). Even if the scatter of the experimental data is rather large, it is reported that the nucleation strain decreases from approximately 0.2–0.3 at $T = 1/3$ to ~ 0.05 at $T = 1.25$.

Eq. (6.6) seems to predict an increase in the nucleation strain as the strain-hardening coefficient, n , is decreased. But this is an apparent prediction, since a decrease in n is generally accompanied by an increase in K , as seen in Section 2.3, and therefore by a decrease in $\varepsilon_{\text{eff},n}$. For example, for a mild steel $\sigma_Y = 200$ MPa and $n = 0.22$ (Fig. 2.13). Therefore: $K \sim \sigma_Y/0.002^n = 785$ MPa. Setting $c = 0.2$, $\sigma_c = 400$ MPa, and $T = 1/3$, from Eq. (6.6) it is obtained that: $\varepsilon_{\text{eff},n} \sim 0.035$. In an HSLA steel, coefficient n decreases to 0.12 and σ_Y increases to 500 MPa. In the same conditions, $\varepsilon_{\text{eff},n}$ is predicted to decrease to 0.0002.

In some engineering metals, more than one population of particles may be present in the microstructure. For example, in spheroidal cast iron, large graphite spheroids promote nucleation of primary voids at very low strains (a σ_c value for decohesion of 80 MPa is reported; Dong et al., 1997). In these metals, inclusions and/or oxides may also be present in the microstructure, and they will nucleate voids successively in the regions between the primary voids. A similar phenomenon occurs in steels containing a population of (comparatively large) inclusions and one of (comparatively small and well bonded) precipitates.

It is generally observed that very small particles (in nanometric size) do not act as nuclei for voids nucleation, but they modify the characteristics of geometrically

necessary dislocations (changing λ_g ; Eq. 2.2) and therefore the properties of the metal matrix (such as n) around the nucleating particles. As seen later in this chapter, this markedly influences the mechanism of voids coalescence.

6.4 Voids growth

The second stage in the damaging process leading to ductile fracture is voids growth. In fact, because of the plastic deformation, voids grow in size. When they reach a critical condition, they coalesce to form a crack that leads to final ductile fracture. As already mentioned, the two processes (nucleation and growth) are not completely distinct, but the separation into two stages is useful in modeling the process.

As an example, Fig. 6.3 shows different cross sections taken close to the fracture region of a cylindrical specimen made with a spheroidized steel and tensile tested to fracture (Le Roy et al., 1981). The local plastic strain, ε_1 , is given by the following relationship: $\varepsilon_1 = 2\ln(D/D_0)$, where D_0 is the original diameter of the testing bar and D is the diameter at a given distance from the fracture surface. When moving close to the fracture surface along the loading axis, D decreases and ε_1 increases. Therefore the sequence of damaging events that develop with increasing strain can be recognized. For low values of the strain (i.e., far from the fracture

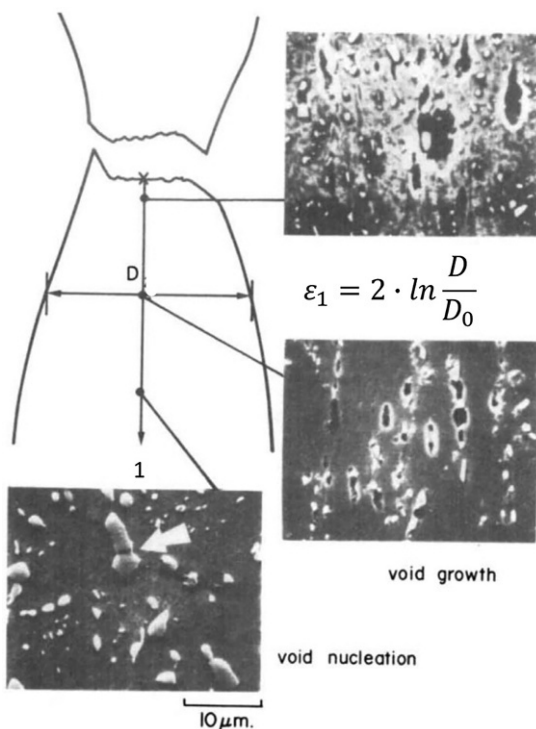


Fig. 6.3 Cross sections showing the voids nucleation and growth stages in a spheroidized steel submitted to tensile testing. Reproduced from Le Roy, G., Embury, J.D., Edward, G., Ashby, M.F., 1981. A model of ductile fracture based on the nucleation and growth of voids. *Acta Metall.* 29, 1509–1522, with permission of the Publisher.

surface), voids nucleation (by particle decohesion and particle cracking) can be appreciated (the arrow in the figure shows an example of particle cracking). As the plastic strain increases (i.e., approaching the fracture surface), voids growth becomes important. Initially, voids mainly grow in the longitudinal direction (in general, voids grow in the direction of maximum strain), but close to the fracture surface, where triaxiality is higher because of necking, voids are also seen to grow in a transverse direction.

Different relationships have been proposed to model the voids growth (Benzerga et al., 2016). A simple and very useful one was developed by Rice and Tracey (Rice and Tracey, 1969). It considers a single spherical void, placed in a metal matrix under plastic deformation, and under a given remote triaxial state of stress. The model gives the growth of the void radius, R , as a function of the effective strain increase, $d\epsilon_{\text{eff}}$, and the triaxiality parameter, T :

$$\frac{dR}{R} = \alpha \cdot d\epsilon_{\text{eff}} \cdot \exp\left(\frac{3}{2} \cdot T\right) \quad (6.7)$$

where α is a constant, which was found to be equal to 0.283. This constant was then re-evaluated by Huang (1991), who proposed higher values than 0.283. In particular, α was found to be 0.427 for $T > 1$ and $0.427T^{0.25}$ for $1/3 < T < 1$. Eq. (6.7) shows that voids growth depends on the increase in the effective strain. It also shows the paramount role of stress triaxiality.

Consider now a voids population with average radius, R . Using Eq. (6.1) or (6.2), we obtain:

$$\frac{df_v}{f_v} = C \cdot \frac{dR}{R} \quad (6.8)$$

From Eq. (6.1), $C = 2$, and from Eq. (6.2), $C = 3$. Considering that voids grow in a condition of diffuse plastic deformation in the surrounding metal matrix, by combining Eqs. (6.7) and (6.8) the following relationship is obtained:

$$\frac{df_v}{f_v} = C \cdot \alpha \cdot \exp\left(\frac{3}{2} \cdot T\right) \cdot d\epsilon_{\text{eff}} \quad (6.9)$$

By integrating Eq. (6.9) from the initial condition at which $\epsilon_{\text{eff}} = 0$ (voids already exist at the beginning of the growth stage), indicating here with f_{v0} the initial volume fraction of voids, and assuming that T does not change during the deformation path:

$$f_v = f_{v0} \cdot \exp(a \cdot \epsilon_{\text{eff}}) \quad (6.10)$$

where $a = C \cdot \alpha \cdot \exp\left(\frac{3}{2} \cdot T\right)$.

Eq. (6.10) can be verified with reference to the porosity evolution during tensile deformation in porous iron, produced by powder metallurgy and containing very spheroidized pores (Spitzig et al., 1988). For these materials $\epsilon_{\text{eff},n} = 0$, i.e., voids nucleation is not required since voids are pre-existing in the microstructure in the form

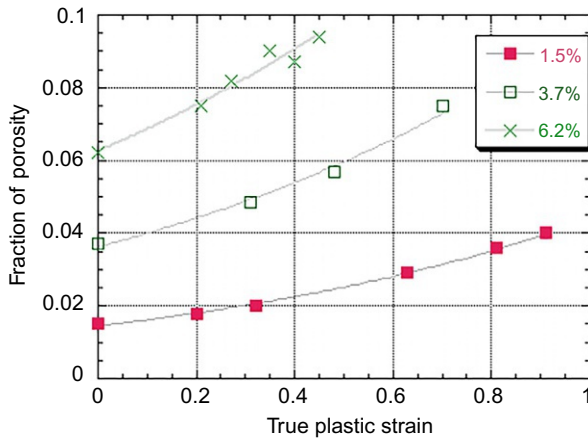


Fig. 6.4 Experimental dependence on the volume fraction of porosity (voids) as a function of the tensile plastic deformation for three materials with different initial porosity—indicated in the box. The experimental points are fitted with Eq. (6.10), setting $\alpha = 1$. Data from Spitzig, W.A., Smelser, R.E., Richmond, O., 1988. The evolution of damage and fracture in iron compacts with various initial porosities. *Acta Metall.* 36, 1201–1211.

of well-rounded pores. In Fig. 6.4, the experimental evolution of the fraction of porosity (i.e., the fraction of voids, f_v) as a function of the tensile plastic deformation and for three different values of initial porosity is shown. It can be noted that the increase in f_v follows an exponential trend, as predicted by Eq. (6.10). By fitting the experimental data, it is seen that $\alpha \sim 1$. Using Eq. (6.10) (setting $T = 1/3$ and $\alpha = 0.32$ from the Huang equation), $\alpha = 1$ for $C = 2$, and 1.57 for $C = 3$, confirming the general validity of the model. Similar results were obtained for a re-sulfurized steel containing a large fraction of Mn sulfides (some of them were pre-cracked, behaving as initial pre-existing voids), through advanced in situ observations using X-ray tomography (Seo et al., 2015).

6.5 Voids coalescence and fracture

Voids grow in size during plastic deformation, up to their coalescence into a ductile crack (or into a large cavity, in the case of a coalescence of few large voids). Fig. 6.5 is a cross section showing the distribution of voids in a ferritic steel subjected to tensile testing (Hancock and Thomson, 1985). A couple of cracks is shown, formed by the coalescence of adjacent voids. Fig. 6.6 represents a top view of the fracture surface of a low carbon steel. The typical *dimpled ductile fracture* is shown, and in several dimples, the original particles that nucleated the voids can be recognized. They are rounded Fe oxides that are rather weakly bonded to the matrix.

Here, a simplified model is first proposed, consisting of spherical voids with initial radius, R_0 . It is supposed that R_0 coincides with the average radius of the particles that are responsible for the nucleation, and that all particles nucleate the voids (Pardo

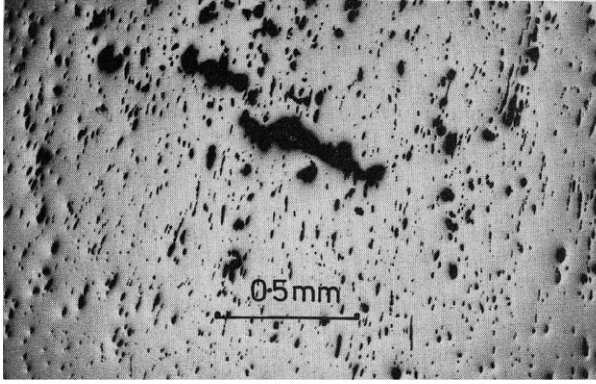


Fig. 6.5 Metallographic cross section showing the coalescence of voids in a ferritic steel (Swedish iron, with $\sim 1\%$ Fe oxide particles) subjected to tensile testing. Reproduced from Hancock, J.W., Thomson, R.D., 1985. Strain and stress concentrations in ductile fracture by void nucleation growth and coalescence. *Mater. Sci. Technol.* 1, 684–690, with permission of the Publisher.

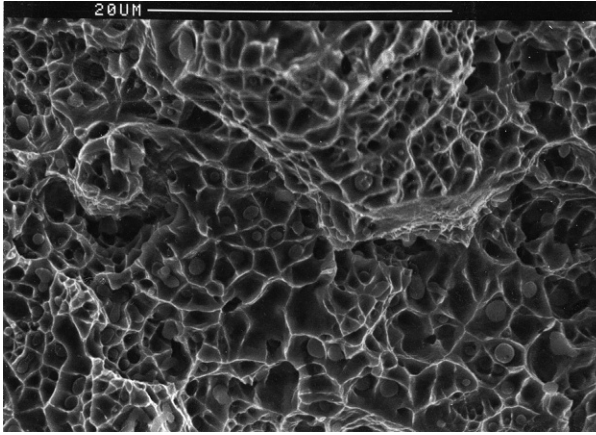


Fig. 6.6 Typical dimpled ductile fracture morphology. In some dimples, the original particles that nucleated the corresponding void can be recognized.

et al., 1998). The model will be a reference framework within which to explain the role of the stress-strain state and microstructure on the ductility of metals.

Under the hypothesis of proportional plastic deformation and constant triaxiality along the deformation path, we may use Eq. (6.7) to estimate the effective plastic strain increment, $\Delta \epsilon_{\text{eff},g-c}$, for growth and coalescence into a ductile crack. This is done under the additional assumption that the voids do not interact, and coalescence takes place when two adjacent voids impinge (Fig. 6.7), i.e., when $R = \lambda_v/2$, where λ_v is the regular voids spacing:

$$\Delta \epsilon_{\text{eff},g-c} = \frac{\int_{R_0}^{\frac{\lambda_v}{2}} \frac{dR}{R}}{\alpha \cdot \exp\left(\frac{3}{2} \cdot T\right)} = \frac{\ln\left(\frac{\lambda_v}{2 \cdot R_0}\right)}{\alpha \cdot \exp\left(\frac{3}{2} \cdot T\right)} \quad (6.11)$$

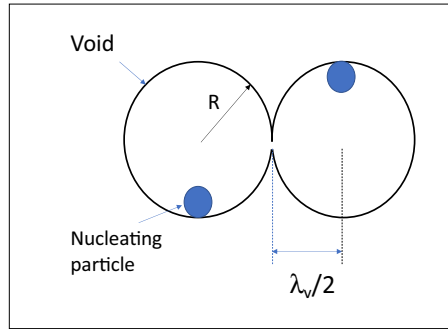


Fig. 6.7 Schematization showing coalescence of two neighboring voids by impingement.

The voids spacing, λ_v , can be evaluated with Eq. (6.3). After the substitution, the following equation is obtained:

$$\Delta \varepsilon_{\text{eff},g-c} = \frac{1}{\alpha} \cdot \ln \left(\frac{0.806}{f_{v0}^{1/3}} \right) \cdot \exp \left(-\frac{3}{2} \cdot T \right) \quad (6.12)$$

which can be generalized by setting:

$$\varepsilon_m = \ln \left(\frac{0.806}{f_{v0}^{1/3}} \right) \quad (6.13)$$

and finally obtaining:

$$\Delta \varepsilon_{\text{eff},g-c} = \frac{1}{\alpha} \cdot \varepsilon_m \cdot \exp \left(-\frac{3}{2} \cdot T \right) \quad (6.14)$$

The parameter ε_m is termed *intrinsic ductility* (Straffelini, 2003). It is graphically shown in Fig. 6.8. The figure shows that ε_m decreases with increasing the initial volume fraction of particles that are responsible for voids nucleation. On the contrary, if f_{v0} tends to vanishingly small values, ductility tends to infinity (and plastic deformation continues up to a point, Fig. 6.1: purely plastic fracture). This condition is approached in special cases only, when voids nucleation is extremely difficult.

However, it is commonly observed that coalescence occurs *before* voids impingement because of the strain localization and intensification in the inter-voids ligaments, i.e., in the matrix between two neighboring voids (Ashby et al., 1979; Thomason, 1990). This process is schematized in Fig. 6.9A. Such localization induces a marked local thinning in the ligaments and the attainment of anticipated coalescence by *localized necking*, as seen in Fig. 6.9B. Voids coalescence is then dependent on the matrix properties. Therefore it is expected that coalescence (and then ductility) depend especially on coefficients n and m .

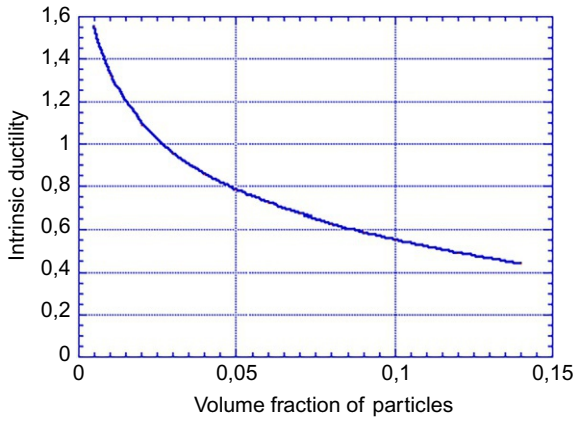


Fig. 6.8 Intrinsic ductility vs volume fraction of particles, i.e., of initial voids (Eq. 6.15).

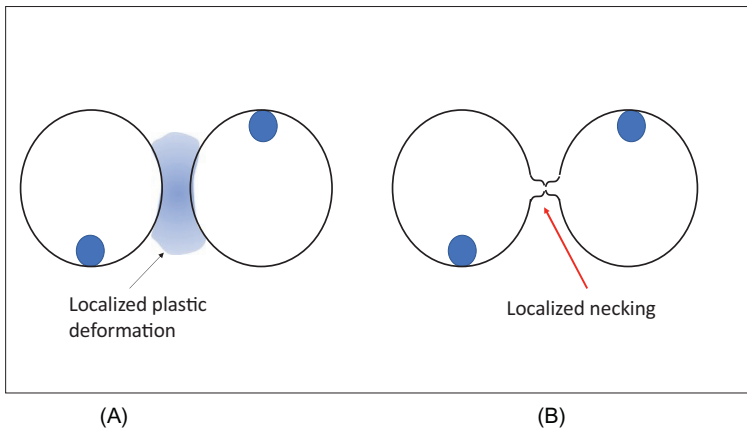


Fig. 6.9 (A) Localization of plastic deformation in the inter-voids ligaments; (B) attainment of anticipated coalescence by localized necking.

As seen in [Chapters 2 and 3](#), from a macroscopic point of view, coefficients n and m have particular and useful physical meaning. Coefficient n represents the strain at necking, ϵ_n , as necking is little influenced by m . Coefficient m represents the attitude of strain localization after necking; if m increases, the strain localization in the neck region becomes more difficult (in this case also n plays a role that is similar to m). In [Section 3.4](#) we used these concepts in a microscopic perspective, to show the tendency toward the formation of shear bands. We can therefore extend this approach to understand how the matrix characteristics influence ductile fracture. To show this, two limiting cases are considered.

1. The matrix is characterized by comparatively high n and m values.

In this case the localization tendency in the inter-voids ligaments is significantly alleviated, since the local flow strength of the material increases (due to the increase in

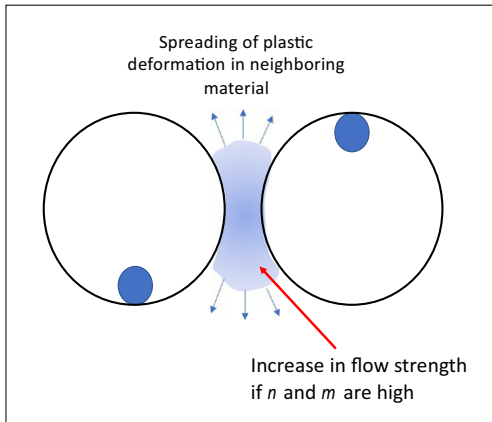


Fig. 6.10 Role of positive n and m values in reducing the intensity of localization.

strain and strain rate), and therefore the plastic deformation spreads in the neighboring regions. Refer to Fig. 6.10. Coalescence becomes more difficult and ductility is expected to increase.

2. The matrix is characterized by comparatively low n and m values (and m may also be negative).

Small or even negative m values favor localization of deformation in imperfections, and hence the formation of shear bands (Section 3.6). As a consequence, the damaging process concentrates in these bands, as schematized in Fig. 6.11A. Local deformation is large and voids elongate in the direction of maximum strain, indicated with direction 1 (Tvergaard, 2009). Voids growth is very limited and voids attain a premature coalescence by *localized shearing*; correspondingly, ductility is expected to decrease. The morphology of the fracture dimples is very typical, assuming a parabolic shape. It is shown, as an example, in Fig. 6.11B.

The matrix may contain two populations of big and small particles. In this case, the secondary small particles may play a double role. First, they contribute to reducing the matrix strain-hardening coefficient, n , as seen, for example, in Section 2.3 for the case of carbonitride particles in HSLA and quenched and tempered steels. Subsequently, they nucleate a secondary population of voids when rather large strains are attained in the inter-voids regions. This process is schematized in Fig. 6.12A. An example is shown in Fig. 6.12B, which shows the fracture morphology of a steel specimen with a tempered martensitic microstructure (Krauss, 2001). The large voids were nucleated at inclusions (sulfides: low in number because of the high cleanliness of the steel), whereas the small voids were nucleated at a secondary population of smaller (and more numerous) carbides. The voids at the inclusions nucleated at small strains (because of the low σ_c values; see Eq. 6.6), and the associated strain increase (up to fracture) is rather large, as demonstrated by the comparatively large size of the dimples. The voids at the carbides nucleated later, at comparatively higher local strains, as said before. They are small since the strain increase up to fracture is also correspondingly small.

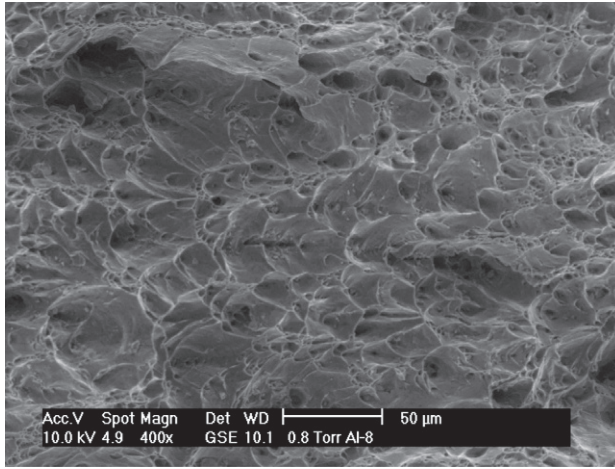
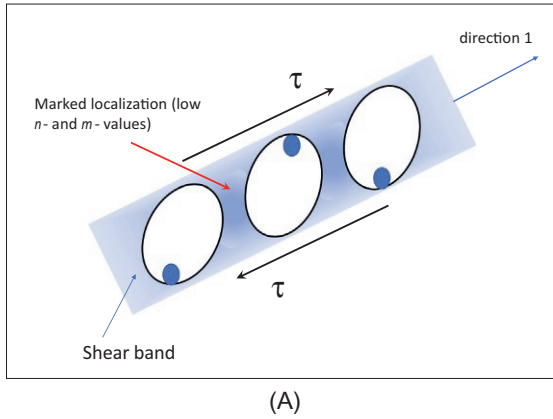


Fig. 6.11 (A) Schematization showing intense localized plastic deformation within a shear band; (B) morphology of the shear ductile fracture with the typical parabolic dimpling. Figure (B) is reproduced from Straffelini, G., Fontanari, V., Zadra, M., 2013. Influence of specimen width on the deformation and fracture behaviour of AA5182 sheets. Eng. Fract. Mech. 109, 262–272, with permission of the Publisher.

6.6 Rice and Tracey ductile fracture criterion

Fig. 6.13 schematizes the damaging phenomena occurring during tensile straining, illustrated in the previous sections. Two contributions were identified: the effective strain for voids nucleation, $\epsilon_{\text{eff},n}$, and the effective strain for voids growth up to coalescence, $\Delta\epsilon_{\text{eff},g-c}$. On this basis, we can say that ductile fracture occurs once the applied effective plastic strain, ϵ_{eff} , reaches (in the most critical sample position) a critical value, $\epsilon_{\text{eff},f}$ (the so-called *effective strain to fracture*) given by:

$$\epsilon_{\text{eff},f} = \epsilon_{\text{eff},n} + \Delta\epsilon_{\text{eff},g-c} \quad (6.15)$$

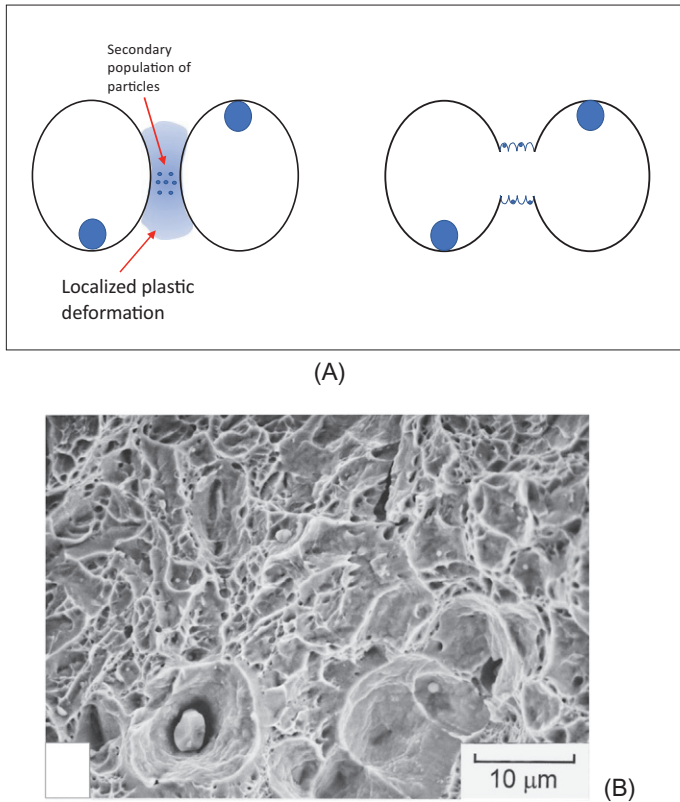


Fig. 6.12 (A) Secondary population of particles in the inter-void ligaments and its role in ductile fracture; (B) fracture morphology of a steel specimen with a tempered martensitic microstructure. The presence of two populations of dimples can be appreciated (large dimples nucleated at sulfides, and small dimples nucleated at carbides).

Figure (B) is reproduced from Krauss, G., 2001. Deformation and fracture in martensitic carbon steels tempered at low temperatures. *Metall. Mater. Trans. A* 32, 861–877, with permission of the Publisher.

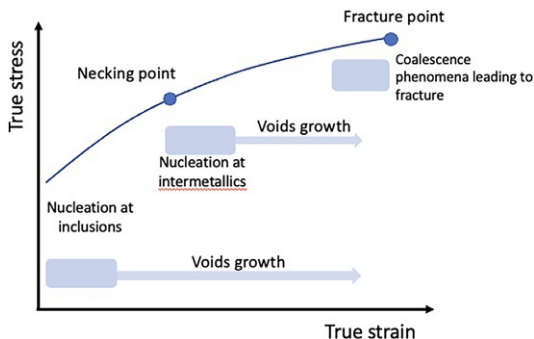


Fig. 6.13 Schematization of true stress-true strain curve (Fig. 1.2), with the ductile damaging processes indicated. (Note that close to the fracture point, the true stress may decrease when strain hardening is not able to counterbalance the internal reduction of the load-bearing section due to voids growth and coalescence.)

A simple evaluation of $\varepsilon_{\text{eff},f}$ can be obtained by assuming that the nucleation strain is negligible ($\varepsilon_{\text{eff},n}=0$), and by evaluating the strain increase for voids growth and coalescence using Eq. (6.14). Hence:

$$\varepsilon_{\text{eff},f} = \frac{1}{\alpha} \cdot \varepsilon_m \cdot \exp\left(-\frac{3}{2} \cdot T\right) \quad (6.16)$$

The intrinsic ductility, ε_m , can be experimentally obtained from a uniaxial tensile test. During tensile testing, T is equal to 1/3 in the uniform deformation but it increases after necking. The value of T at fracture can be estimated following the procedure outlined in Section 5.6. If, for example, $T=0.4$, then: $\alpha=0.34$, and from Eq. (6.16): $\varepsilon_f = 1/0.34 \cdot \varepsilon_m \cdot \exp\left(-\frac{3}{2} \cdot 0.4\right) = 1.61 \cdot \varepsilon_m$, where ε_f is the *tensile true plastic strain at fracture* (refer to Chapter 1; often, it is simply called the *true fracture strain*), which is obtained experimentally (Eq. 1.11). Based on this, Eq. (6.16) can be rewritten in the following way:

$$\varepsilon_{\text{eff},f} = \frac{\alpha'}{\alpha} \cdot \varepsilon_f \cdot \exp\left(-\frac{3}{2} \cdot T\right) \quad (6.17)$$



Metal ductility



Stress triaxiality

where α' is a constant that must be determined considering the triaxiality factor at fracture; if $T=0.4$, α' is equal to $1/1.61=0.62$. Eq. (6.17) is the *Rice and Tracey (R-T) ductile fracture criterion*.

In Table 6.1, the conditions for ductile fracture for some important states of stress are listed.

The R-T criterion for dimpled ductile fracture is of course simplified. In the following, relevant aspects are listed and discussed. Some of them will be reconsidered in the next sections.

1. The assumption $\varepsilon_{\text{eff},n}=0$ may be acceptable when nucleation takes place at inclusions, but in other cases the strain for voids nucleation cannot be neglected (Fig. 6.13). This shortcoming is partially compensated for by the presence of the material ductility, ε_f , in Eq. (6.17). It is obtained experimentally, and hence contains information on the nucleation strain as well;
2. The geometry of the particles is often not spherical (and may also change during plastic deformation). In particular, the presence of elongated inclusions (such as Mn-sulfides) introduces an

Table 6.1 Ductile fracture conditions for some relevant states of stress introduced in Section 5.3.

	T	α	$\varepsilon_{\text{eff},f}/\varepsilon_f$
Uniaxial stress	$1/3=0.33$	0.324	1
Plane strain	$\sqrt{3}/3=0.58$	0.373	0.6
Equi-biaxial plane stress	$2/3=0.67$	0.386	0.51
Pure shear	0	0.28	1.91

- anisotropy in ductility. To account for this effect, the criterion should be used with reference to the specimen directions and by adopting the ϵ_f values determined along these directions;
3. The triaxiality parameter, T , changes during plastic deformation and the deformation process is often not proportional during tensile testing; the criterion should then be used in its integral form (see Section 5.2). In some cases, the initial T value is used in Eq. (6.17), and this might be a noticeable approximation. When using numerical methods, a coupled approach can be adopted, considering both the evolution in the effective strain and the stress triaxiality during plastic deformation;
 4. For T values close to 0, plastic deformation may be forced to develop in imperfections and then to form shear bands, where an anticipated coalescence of the voids by localized shearing can take place (Fig. 6.11A). This phenomenon is particularly relevant in alloys that are sensitive to shear localization, i.e., in alloys containing imperfections in their microstructure, and/or characterized by comparatively low m and n values. This phenomenon is not accounted for by Eq. (6.17);
 5. Ductile fracture is generally not observed in the case of T values equal to or lower than $-1/3$. The condition $T = -1/3$ can be attained during homogeneous compression tests, if no friction between the specimen and die platen is present (this is analyzed in Chapter 11), and T may be lower than $-1/3$ in the case of tensile testing with large superimposed hydrostatic pressure (Bao and Wierzbicki, 2005). In these cases, ductility is expected to increase as fracture tends to necking to a point, but this is not considered in Eq. (6.17).

Numerous experimental investigations have demonstrated that it is almost impossible to describe the entire fracture curve, from low to high triaxialities, with a unique function. In many cases, the R-T criterion must be generalized or even replaced with other equations, which are able to better fit the experimental data in specific T intervals. A number of phenomenological ductile fracture criteria are available in the literature (Wiebicki et al., 2005).

In this text, the R-T criterion is preferred despite its shortcomings, since it is based on the physical damage mechanisms. In addition, with proper corrections, the criterion is sufficiently accurate to depict the dependence of ductility on stress triaxiality in the T intervals that are of interest in metalworking operations.

6.7 Influence of triaxiality on ductility

To better address the role of stress triaxiality on ductility, two representative examples are considered in this section. They are shown in Fig. 6.14, where the effective strain to fracture is plotted as a function of the triaxiality factor for an austenitic high-temperature steel (Trattnig et al., 2008) and a 2024-T351 Al alloy (Bao and Wierzbicki, 2004). The tests were conducted using different specimens having notches or cracks, and at different loading conditions. To obtain high T values, compact tension and circumferentially cracked cylindrical specimens were used. To obtain negative T values, cylindrical upsetting specimens (which are described in Chapter 11) and torsion specimens were used. In all the cases, the authors employed FEM analysis to calculate the T at fracture.

For T in excess of ~ 0.4 , the effective fracture strain is seen to decrease in an almost exponential way in both materials. For the austenitic steel, the exponential trend is also followed for $T < 0.4$. However, the Al alloy shows a drop in ductility for T close to 0.

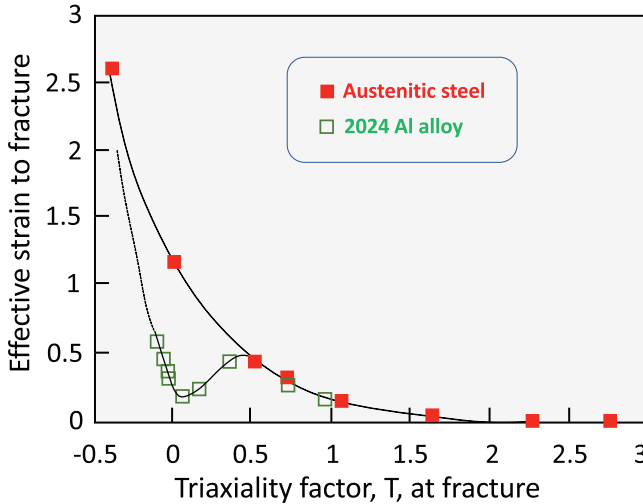


Fig. 6.14 Effective strain at fracture vs the triaxiality factor, T , at fracture for an austenitic steel (data from Trattnig, G., Antretter, T., Pippan, R., 2008. Fracture of austenitic steel subject to a wide range of stress triaxiality ratios and crack deformation modes. *Eng. Fract. Mech.* 75, 223–235) and a 2024-T351 Al alloy (data from Bao, Y., Wierzbicki, T., 2005. On the cut-off value of negative triaxiality for fracture. *Eng. Fract. Mech.* 46, 81–98).

The experimental data for the austenitic steel can be fitted using the R-T criterion (Eq. 6.17). A better fit (able to especially account for the limitations expressed in points 1 and 2 of the list in the previous section) can be obtained by generalizing the R-T criterion as:

$$\varepsilon_{\text{eff},f} = a + b \cdot \exp(-c \cdot T) \quad (6.18)$$

where a , b , and c are experimental constants (Johnson and Cook, 1985). For the data in Fig. 6.14, the authors obtained: $a = 0$, $b = 1.21$, $c = 1.88$. Note that parameter c is a bit greater than $3/2 = 1.5$, which is present in the R-T criterion. Other experimental data are listed in Table 6.2.

For T lower than 0.4, the shortcoming expressed in point 4 of the list in the previous section may gain importance. This is due to the forced transition to shear ductile fracture that may occur for T close to 0 (Table 5.4). This is not the case for austenitic steels. In fact, these metals are characterized by rather high n values and then they do not easily undergo shear localization in quasi-static deformation conditions; this is clearly shown by the trend in $\varepsilon_{\text{eff},f}$, displayed in Fig. 6.14.

However, this may be the case for alloys that are prone to shear localization, i.e., alloys with microstructural imperfections and characterized by comparatively low or negative m values and low n values. One example is provided by some Al alloys, as seen in Section 3.6. This is shown in Fig. 6.14 for a 2024-T351 Al alloy. For T lower than 0.4, the experimental fracture strains are lower than would be predicted by the R-T criterion. As seen, the transition to shear ductile fracture is accompanied by

Table 6.2 Experimental data for the coefficients of Eq. (6.18).

Material	a	b	c	ϵ_f	References
AISI 1045	0.154	0.209	4.862	0.95	Abushawashi et al. (2013)
Ferritic steel (Weldox 460E)	0.075	1.732	0.54		Borvik et al. (2003)
AISI 4340	0.05	3.44	2.12		Guo and Yen, (2004)
API X65 steel (0.08C, 1.45Mn)	0.1	3.29	1.54	1–3	Oh et al. (2007)
Austenitic high-temperature steel	0	1.21	1.88	0.66	Trattnig et al. (2008)
5083 Al alloy	0.0261	0.263	0.349	0.14	Clausen et al. (2004)
Ti-64	−0.078	0.282	0.479	0.14	Kay (2003)

micro-shear strain localizations (Fig. 6.11A) that markedly anticipate the critical conditions for fracture practically avoiding the growth stage (Li et al., 2011). This behavior was observed also in other Al alloys (El-Magd et al., 2001) and also in Ti alloys, as shown in Fig. 6.15A for pure Ti (Zhang et al., 2021) and Ti-64 alloy (Verleysen and Peirs, 2017). In fact, Ti alloys are also rather prone to shear localization, due to their very low n values (see also Fig. 3.31). The general trend for materials that display this behavior is schematized in Fig. 6.15B.

6.8 Ductility and microstructure

Eq. (6.17) clearly shows that the effective strain to fracture depends on **metal ductility** and **stress triaxiality**. The role of microstructure is generally investigated by tensile testing using un-notched specimens, whereas the role of stress triaxiality is often simply determined using circumferentially notched specimens (Section 5.6).

In this final section of the chapter, some aspects of tensile testing for evaluating ductility are outlined and, subsequently, the pivotal microstructural features determining the metal ductility are briefly summarized. These concepts are used extensively in the next two chapters.

Tensile testing and ductility

From an experimental viewpoint, metal ductility is typically evaluated with standard unnotched tensile tests with a cylindrical or rectangular shape (refer to Section 1.1). As already mentioned, the triaxiality factor is 1/3 in the uniform region but greater than 1/3 in the necked region. It increases with plastic straining and is maximum in the center of the tensile specimen. Therefore the ductile crack forms by voids

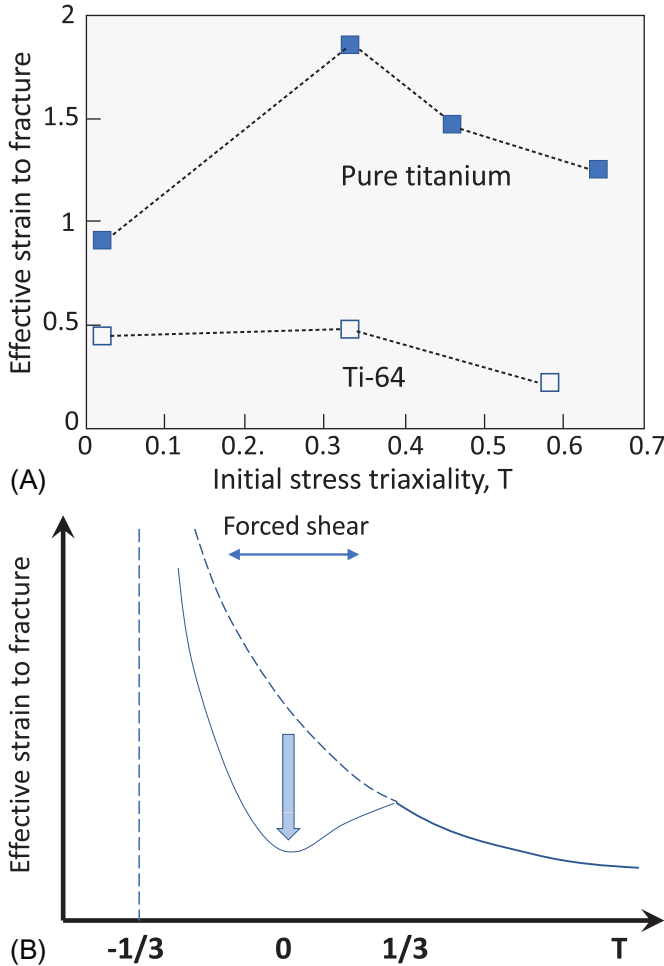


Fig. 6.15 (A) Effective fracture strain to fracture vs initial triaxiality in pure titanium (data from Zhang, H., Li, X., Gao, T., Song, H., Huang, G., 2021. Experimental study on deformation evolution and fracture behaviors of pure titanium at different stress triaxialities. Eng. Fract. Mech. 258, 108127) and a Ti-64 alloy (data from Verleysen, P., Peirs, J., 2017. Quasi-static and high strain rate fracture behaviour of Ti6Al4V. Int. J. Impact Eng. 108, 370–388); (B) schematic showing the dependence of the effective fracture strain on T for metals prone to forced shear localization, for which fracture occurs shortly after voids nucleation.

coalescence in the center of the specimen and then propagates radially. When using axisymmetric specimens or rectangular specimens with low width-to-thickness ratio, a transition in the fracture mode may occur when the crack is sufficiently long. In fact, at the crack tip, two planes forming an angle of 45 degrees and with maximum shear stress are present. Therefore a transition from dimpled ductile fracture (at the center of the specimen) to shear ductile fracture (at the periphery) can occur when the remaining load-bearing section is sufficiently small. As a result, the typical *cup-and-cone* fracture is obtained. The sequence of the events is schematized in Fig. 6.16A and an

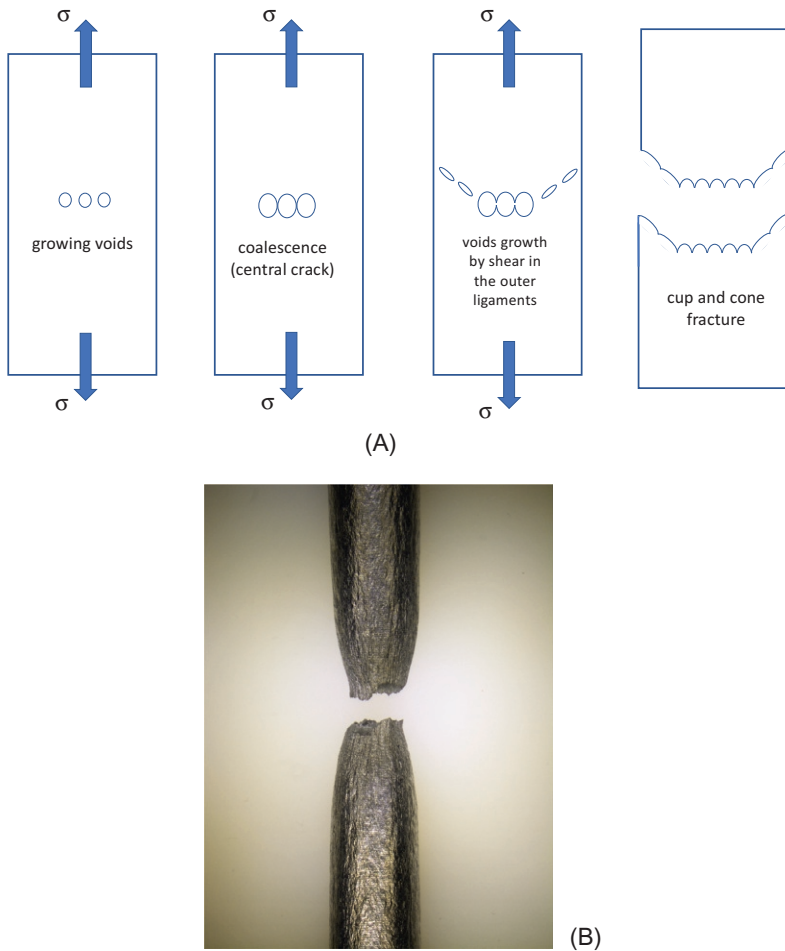


Fig. 6.16 (A) Schematization showing the formation of a center crack by voids coalescence and the subsequent formation and growth of voids by shear in the remaining load-bearing section. As a result, the typical cup and cone fracture is obtained. (B) Example of cup-and-cone fracture in a steel specimen.

example is shown in Fig. 6.16B. The central fracture region generally prevails and the measured tensile true fracture strain, ϵ_f (given by Eq. 1.9), is an adequate representation of ductility (with a proper evaluation of T , as explained in Section 6.6).

Ductility is commonly expressed by three parameters:

1. The true fracture strain, ϵ_f ;
2. The reduction of area at fracture, RA, Eq. (1.11) (often expressed in RA%);
3. The elongation at fracture, EL, Eq. (1.12) (often expressed in EL%).

As seen, RA is strictly related to ϵ_f ($RA = 1 - e^{-\epsilon_f}$), whereas the elongation at fracture, EL, is not formally related to ϵ_f . If fracture occurred in the uniform region, RA and EL coincide, because of the volume constancy during plastic deformation. However,

fracture takes place after necking and the experimental RA values are greater than the EL values, because of the strain localization, even if they are generally proportional. The EL values depend on the original gauge length (l_0 ; see [Section 1.1](#)), and to facilitate the comparison among different data, l_0 is generally taken as four or five times the specimen original diameter.

The concept of the elongation at fracture, EL, is often used when testing sheet specimens, because of its important practical use and because of the experimental difficulty in reliably measuring the extension of the fracture surfaces. The concept of ϵ_f is required when using a ductile criterion, such as the R-T criterion (Eq. 6.17). It is more significant from a scientific point of view, since it is related to the damaging mechanisms, and then it allows comparison of the ductility of metals. In doing this, however, reference to the concept of the intrinsic ductility should be better made (making use of Eq. 6.17). In fact, necking may give rise to different triaxiality parameters in the fracture region, and this also affects the recorded ductility.

In the following chapters, these experimental parameters will be used to represent the ductility of metals, using available literature data.

A final observation regards the influence of specimen geometry, and in particular its size, on the experimental data and their scatter. The damaging mechanisms responsible for ductile fracture are statistical in nature. The characteristics and distribution of engineering particles are not homogeneous and it is common to have microstructural regions where the particle density is much different from the average. Also the dimension, morphology, and bonding strength of the particles may be different from the average in some microstructural regions. These local variations introduce a scatter in the experimental data that increases with increasing specimen volume (roughly given by the product of the cross section and the gauge length). In addition, these microstructural variations introduce a specimen-size effect. In general, the experimental ductility decreases with increasing specimen volume. Both these phenomena (scatter and size effect) can be explained by considering that with increasing the specimen volume, the probability of sampling anomalous microstructural regions increases.

The problem of the scatter of the experimental data is faced using the average of a sufficient number of repeated tests. On the contrary, the size-effect phenomenon is rarely identified in practice. Therefore, in the next chapters it will be not taken into account, even if difficulties in comparing results obtained using specimens with different geometry could be attributed to this phenomenon ([Pineau et al., 2016](#)).

Role of coefficients n and m on ductility

The main microstructural features affecting ductility are coarse engineering particles (such as inclusions and intermetallic particles with size greater than $0.1\text{--}1\text{ }\mu\text{m}$) and the metal matrix surrounding such particles. In particular:

1. *Coarse particles.* Ductility decreases with decreasing the particle bonding strength with the matrix and/or their intrinsic brittleness (which facilitates voids nucleation) and with increasing their volume fraction, f_v (which decreases the intrinsic ductility).
2. *Metal matrix (surrounding the particles).* Ductility increases as both the strain-hardening coefficient, n , and the strain rate sensitivity, m , increase (an increase in n is generally accompanied by a decrease in the strength coefficient, K , and this makes nucleation more difficult;

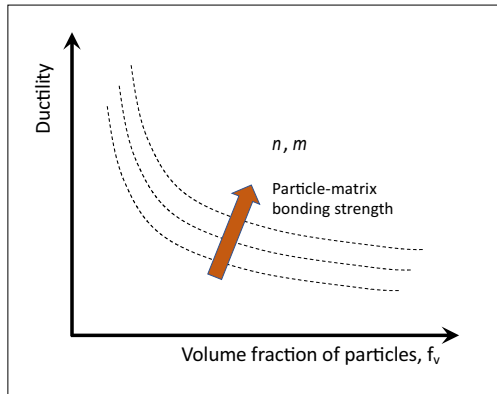


Fig. 6.17 Influence of f_v , particle-matrix bonding strength, and n and m on ductility.

in addition, high n and m values facilitate spreading of plastic deformation far from the intervoids ligaments, decreasing the intensity of the damaging phenomena involved in voids coalescence).

Fig. 6.17 schematically shows the influence of f_v , particle bonding, and n and m on ductility, as obtained from the analysis of the micro-mechanisms of ductile damage and fracture.

As seen, coefficients n and m also have an important engineering meaning with regard to the *macroscopic* necking and post-necking behavior of metals under tensile deformation. Therefore, in the analysis of ductility of metals, the *macroscopic* and *microscopic* levels (ultimately interrelated, being dependent on the same deformation mechanisms) must fruitfully be considered together, with the awareness that sometimes the experimental data are more easily explained making reference to one level or the other. We anticipate here some significant cases to be extensively considered in the next two chapters.

1. In bcc metals and, in particular, in ferritic steels, both n and m are positive and comparatively high; in addition, the m values are broadly proportional to the corresponding n values (see Fig. 3.15B). Therefore, in metals containing approximately the same fraction of particles, ductility variations are expected to be proportional to variations in the coefficient n only. In these metals, the strain for necking is proportional to n (Considère criterion), and the strain for ductile fracture is also expected to be proportional to n .
2. In metals with comparatively high n values and low (or negative) m values, such as the austenitic stainless steels, uniform plastic deformation is rather diffuse and necking takes place at large strains. However, with the onset of necking the local strain rate (in the neck region) considerably increases, and the incapability of the metal to locally increase the flow strength (because of the low or negative m value, see Fig. 3.17) induces a strong strain localization that will quickly conduct to fracture.
3. If m is negative and also n is low, the formation of shear bands is rather easy. Necking takes place at small strains, and then plastic deformation may easily localize in microstructural or geometrical imperfections. This induces a significant strain intensification that in turn causes an early coalescence and fracture within the bands.
4. When temperature is in excess of $0.4\text{--}0.5T_m$, macroscopic necking takes place rather early, since thermal softening induces a rapid decrease in n (that tends to 0 approaching the

secondary creep region; [Section 4.5](#)). Since also the strength coefficient, K , markedly decreases with temperature, because of the decrease in the flow strength, voids nucleation is made difficult with beneficial effects on ductility. In addition, with the absence of strain hardening, the possibility of spreading plastic deformation away from the inter-voids regions when approaching the coalescence stage relies on the flow stress increase due to the local rise in the strain rate. This means that ductility is expected to markedly depend on m .

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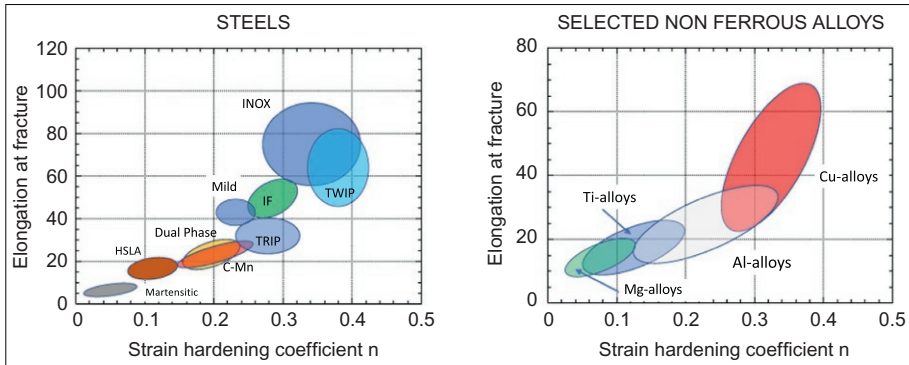
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Influence of microstructure on ductility

7



The ductility of metals mainly depends on the average strain hardening coefficient, n .

Chapter outline

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In this chapter, the concepts introduced in the previous chapter are used to present the ductility of metals as influenced by their microstructural characteristics.

Microstructural particles play a pivotal role in the plastic damage of metals since they are responsible for the voids' nucleation. Therefore the main characteristics of particles in steels and Al alloys, along with the typical microstructures of these alloys, are first presented. Then, the ductility of these alloys is analyzed in reference conditions (RT—or close to RT—and in quasi-static loading), considering the combined role of particles and matrix properties, expressed by the corresponding average n (strain-hardening coefficient) and m (strain rate sensitivity coefficient) values. The main ductility characteristics of other non-ferrous alloys are also briefly outlined. In any case, selected examples with particular engineering and metallurgical

relevance are reviewed. A final comparative analysis is also reported to understand the ductility-to-strength relationships in metallic alloys. This is a useful guideline for a ductility-oriented microstructural design.

The information reported in this chapter is pivotal for understanding and predicting the attitude of ferrous and non-ferrous alloys toward plastic deformation at RT, i.e., to understand the conditions for producing sound parts, free from ductile cracks, in cold working conditions. The ductility characteristics at high temperatures (typical of warm and hot working) and in special conditions of high strain rates, are discussed in the next chapter.

7.1 Particles in steels and Al alloys

Particles in engineering metals can be subdivided into *inclusions* and *intermetallic compounds*. In **steels**, the most common inclusions are *oxides* and *sulfides*. Oxides form in the melt as a result of the reaction of oxygen with elements such as Si, Mn, and Al (mostly employed for deoxidation). Therefore typical oxides in steels are: SiO_2 , MnOSiO_2 , MnAl_2O_3 , and Al_2O_3 and most of the oxygen in steel is present as oxides. Most oxides float from the liquid steel before solidification, but some of them remain in the microstructure. These particles are small in size, with a diameter typically lower than $10\text{ }\mu\text{m}$ (most have a diameter of few micrometers), with a rough equiaxial morphology. Typically, the volume fraction of oxides, f_v , is in the range of 5×10^{-3} to 2×10^{-2} and the mean number of oxides per unit volume, N_v , is $\sim 10^5\text{ mm}^{-3}$. The mean spacing is then typically around $15\text{--}20\text{ }\mu\text{m}$ (Eq. 6.3). Oxides have a thermal expansion coefficient that is smaller than that of steel. Therefore they are firmly attached to the steel matrix after solidification.

Fe and *Mn sulfides* form in the interdendritic regions during the solidification of the melt because of the very low solubility of S in the steel matrix. In general, the Mn/S ratio is calibrated to obtain only MnFe sulfides poor in Fe, since MnFe sulfides rich in Fe have adverse effects on the high-temperature ductility of steels (Section 8.4). The solubility of S in solid steel is $\sim 0.003\text{ wt\%}$. Therefore f_v can be estimated with the following relationship (Grong, 1997):

$$f_v = 5.4 \cdot 10^{-2} \cdot (\text{wt\%S} - 0.003) \quad (7.1)$$

If the deoxidation degree is high, sulfides solidify into an equiaxial shape. Otherwise, they assume an irregular morphology. They have an average size in the range of $1\text{--}10\text{ }\mu\text{m}$, and a rather large spacing, in the range of $80\text{--}100\text{ }\mu\text{m}$. At high temperatures, sulfides are soft and they undergo easy plastic deformation during hot working. Because of this, they assume a typical elongated geometry in the direction of plastic flow (typical length: $30\text{--}100\text{ }\mu\text{m}$). Contrary to the oxides, the sulfides exhibit a thermal expansion coefficient that is larger than that of steel. Considering the weak bonding between the sulfide particles and the steel matrix, during cooling, they detach easily from the steel matrix. However, after plastic deformation (rolling or forging), their adhesion to the steel matrix increases.

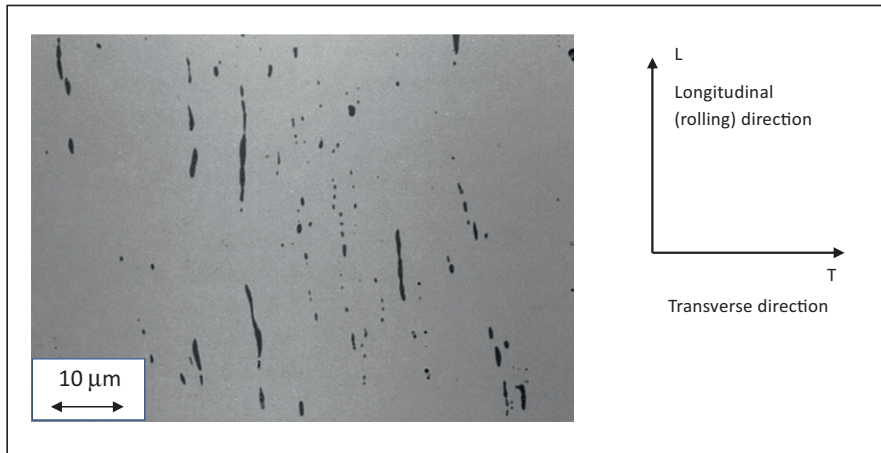


Fig. 7.1 Elongated Mn sulfides along the rolling direction in a resulfurized steel.

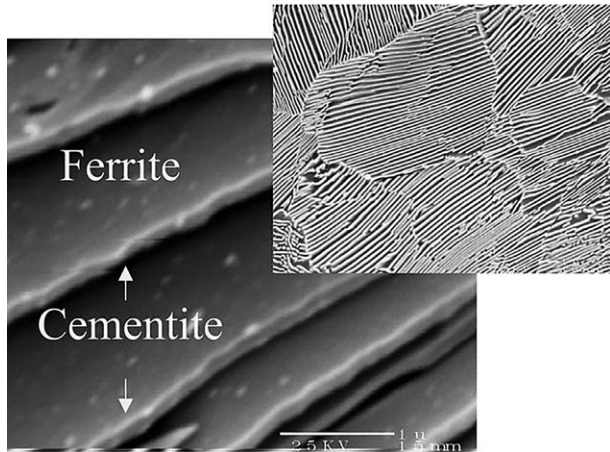
Most C-Mn steels contain approximately 0.04% S and therefore up to 0.2% of sulfides. Steels with better quality contain a typical value of 0.02% S, and f_v is then less than 0.1%. With special desulfurizing treatments, the S content can be decreased well below 0.01%, and particularly low values are displayed by *clean* steels that require high properties (sulfides impair ductility and also fatigue strength). Most steels are also submitted to a *calcium treatment* before casting, to improve their castability under continuous casting. In this case, a different kind of inclusion is formed in place of oxides and sulfides. In fact, a minute volume fraction of rounded inclusions is formed during solidification, with a diameter in the range of 1–3 μm .

A special role is played by *resulfurized steels*, i.e., steels with high content of S (approximately 0.2–0.3 wt% and $f_v \sim 1.5\%$). These steels are used when high machinability (per chip removal) is required (Section 16.4), and the reduction of mechanical properties induced by S is of less importance. As an example, Fig. 7.1 shows elongated Mn sulfides in a resulfurized steel.

The *intermetallic precipitates* are typically smaller than the inclusions, higher in number, and better anchored to the steel matrix. In steels, precipitates are typically Fe carbides (such as FeC: cementite) and nitrides. They form because C and N have a low solubility in the matrix. In ferritic steels (with less than 0.1% of C), small carbides precipitate at the grain boundaries, possibly together with small pearlite nodules. They have a submicrometric dimension that is proportional to the grain size, since both cementite and grain growth depend on the same diffusional phenomena (Curry and Knott, 1979).

In steels with medium-high C content and in normalized conditions, no discrete carbide particles are present in the microstructure. On the contrary, pearlite colonies are present. They are made of alternate layers of cementite and ferrite. The interlamellar spacing is typically below 0.25 μm . In eutectoid steels (C content in the range of 0.7%–0.8%), the microstructure is fully pearlitic, and pearlite forms colonies within the prior austenitic grains. Fig. 7.2 shows the pearlite morphology in a fully pearlitic steel (Zelin, 2002).

Fig. 7.2 Pearlite morphology in a fully pearlitic steel. Reproduced from Zelin, M., 2002. Microstructure evolution in pearlitic steels during wire drawing. *Acta Mater.* 50, 4431–4447, with permission of the Publisher.



Heat-treated steels contain a distribution of very fine carbides after tempering. At tempering temperatures below 200°C, tiny (2–4 nm in size) ϵ - and η -transition carbides precipitate, whereas at higher tempering temperatures, larger cementite particles precipitate (Krauss, 2001). These particles have a typical size around 0.1 μm (Bowen et al., 1986), and confer high strength to the steel. As an example, for the very fine carbides precipitated in a Cr-Mo steel (chemical composition: 0.09%C, 0.51%Mn, 9.36%Cr, 0.99%Mo) the following parameters were obtained: equivalent diameter = $\sim 0.07 \mu\text{m}$, $f_v = 0.015$, $N_v = 9.08 \cdot 10^{10} \text{ mm}^{-3}$, $\lambda_v = 0.302 \mu\text{m}$ (Noble et al., 1990).

In HSLA steels, elements such as Nb, V, Ti, Al, and N are added to favor the solid-state precipitation of nitrides and carbonitrides. Such particles are very hard, and their characteristics are like those of the carbides precipitated in steels after tempering. Only in Ti-based HSLA steels can larger particles form during the solidification process (Gladman, 1997).

After high-temperature annealing, carbides grow in size and assume a spheroidal morphology. This process, called *spheroidization annealing*, is carried out to increase the ductility and plastic formability of the steels, as will be shown later. In Table 7.1, the typical characteristics of spheroidal carbides in different steels, with C-content from 0.12% to 0.92%, are listed (Le Roy et al., 1981). In the case of high C content and annealing at high temperature and over long times, well-rounded and micrometric carbides can be obtained.

Table 7.1 Main characteristics of spheroidal carbides in different steels.

Steel	%C	f_v	R (μm)	λ_v (μm)	N_a (mm^{-2})	N_v (mm^{-3})
1015	0.12	0.017	0.42	7.02	30,150	$5.4 \cdot 10^7$
1035	0.34	0.045	0.45	4.2	74,000	$1.2 \cdot 10^8$
1045	0.46	0.0768	0.35	2.6	200,000	$4.2 \cdot 10^8$
1090	0.92	0.136	0.36	1.98	334,000	$6.9 \cdot 10^8$

Data from Le Roy, G., Embury, J.D., Edward, G., Ashby, M.F., 1981. A model of ductile fracture based on the nucleation and growth of voids. *Acta Metall.* 29, 1509–1522.

In tool steels, large and sometimes irregular carbides are present in the as-cast microstructure. Generally, with the application of large plastic deformation (e.g., with a proper rough forging treatment; see [Chapter 11](#)), the eutectoid carbides are fragmented and homogeneously distributed in the microstructure. Following the final heat treatment, small secondary carbides also precipitate in the matrix bulk in the case of high-speed tool steels.

In some steels, hardening can be obtained by the precipitation of particles that are not carbides or nitrides. An example is an addition of 1%–3% of Cu in ferritic steels, such as IF and HSLA steels, and the production of a fine precipitation of Cu nanometric particles in the matrix. The mean size is rather small, like that of transition carbides; the volume fraction is typically in the range of $\sim 10^{-3}$ – 10^{-2} and the mean spacing is approximately 0.04 μm ([Aroztegui et al., 1989](#)).

In general, alloy steels, such as tool steels, heat-treated steels, stainless steels, and HSLA steels, are high-quality steels. This means that they contain low amounts of O and S, leading to a low fraction of oxides and sulfides. The main particles in these steels are the intermetallics. On the contrary, in low alloy steels, such as C-Mn steels for structural applications, the content of inclusions is comparatively higher.

In **Al alloys**, oxide particles may be present in the microstructure. They may be Al_2O_3 , MgO , or spinels such as MgAl_2O_4 . In general, they form on the melt surface due to the exposure of the liquid to the surrounding atmosphere. They may be entrapped in the microstructure, but most of them float or are eliminated using special filtering systems before solidification.

Metallic and intermetallic compounds are the most important particles in Al alloys. There may be a presence of four kinds of these particles ([Totten and Scott, 2003](#)):

1. Primary particles;
2. Intermetallic particles (also called intermetallic constituents);
3. Dispersoids;
4. Precipitates.

Primary particles form during solidification and they are mainly found in castings. An example is the Si particles that form in hypereutectoid Al-Si alloys. In general, they are large and have a blocky geometry, which significantly reduces ductility. Due to this, such alloys are treated with special inoculants before solidification, in order to reduce the size of the particles and make them more rounded.

The comparatively coarse *intermetallic constituents* form during solidification and cooling of the liquid metal to RT, mainly at the grain boundaries. An important class of constituents is the Fe-Al compounds, such as FeAl_6 or FeAl_3 . These compounds form because of the presence of Fe in the starting composition. Fe is introduced in some alloys for die casting and then is present in the scrap that is possibly used to produce specific alloys. However, in general, it is an impurity. As an example, [Fig. 7.3A](#) shows the microstructure of a 5182 Al alloy containing 1.1% of $(\text{Fe}, \text{Mn})\text{Al}_m$ precipitates ([Zadra et al., 2008](#)). In Al-Mg alloys (5000 class), $\text{Al}_m(\text{Fe}, \text{Mn})$ particles (with $m = 3$ or 4), and $\text{Al}_{12}(\text{Fe}, \text{Mn})_3\text{Si}$ and Mg_2Si particles are present, depending on the chemical composition and thermal treatment. [Table 7.2](#) shows the main characteristics of FeAl_6 constituents in a 5754 Al alloy ([Spencer et al., 2002](#)). In Al-Cu alloys (2000 class), typical intermetallic constituents are Al_7CuFe . In alloys of the 6000 class,

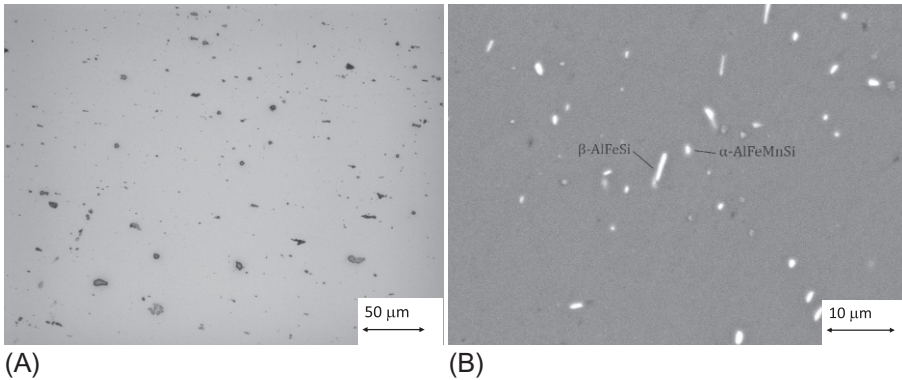


Fig. 7.3 (A) Microstructure of a 5182 Al alloy containing 1.1% of Al_m (Fe, Mn) precipitates. (B) Microstructure of a 6060 Al alloy containing elongated plate-like β -AlFeSi particles and rounded α -AlFeMnSi particles.

Reproduced from (A) Zadra, M., Straffelini, G., Molinari, A., 2008. Tensile ductility of AA5182: effect of impact loading. *Mater. Sci. Technol.* 24, 1259–1264, with permission of the Publisher, (B) Menapace, C., Bernard, F., Lusa, M., Ischia, G., Straffelini, G., 2021. Effect of dual-rate ageing treatment on the tensile properties of a commercial 6060 Al alloy. *Mat. Sci. Eng. A*, 819, 141468, with permission of the Publisher.

Table 7.2 Main characteristics of $FeAl_6$ precipitates in a 5754 Al alloy (data from [Spencer et al., 2002](#)).

Fe content, wt. %	f_a	λ_v , μm
0.08	0.0014	19.5
0.16	0.0049	13.3
0.3	0.0088	9.8

Data from Spencer K., Corbin S.F., Lloyd D.J., 2002. The influence of iron content on the plane strain fracture behaviour of AA 5754 Al-Mg sheet alloys. *Mater. Sci. Eng. A* 325, 394–404

elongated plate-like β -AlFeSi particles are present in the microstructure, together with rounded α -AlFeMnSi particles (Fig. 7.3B). In general, intermetallic constituents are quite large and have an ellipsoid shape, with a dimension in the range of 1–30 μm and volume fractions lower than 0.01. They do not induce an appreciable hardening in the alloys, but they play an important role in the ductile fracture by reducing ductility.

After solidification, Al alloys are often submitted to a *homogenization* treatment. This treatment is aimed at modifying the morphology of intermetallic precipitates and increasing ductility. For example, long treatments (some hours) at 585°C are carried out for the 6000 alloys to induce a transformation of elongated plate-like β -AlFeSi particles into rounded α -AlFeMnSi particles. Other particles may also dissolve into the matrix during the homogenization treatment. After the subsequent plastic deformation processes, such as extrusion or forging, these particles can be fragmented and oriented in the direction of plastic flow. In some Al alloys that contain transition

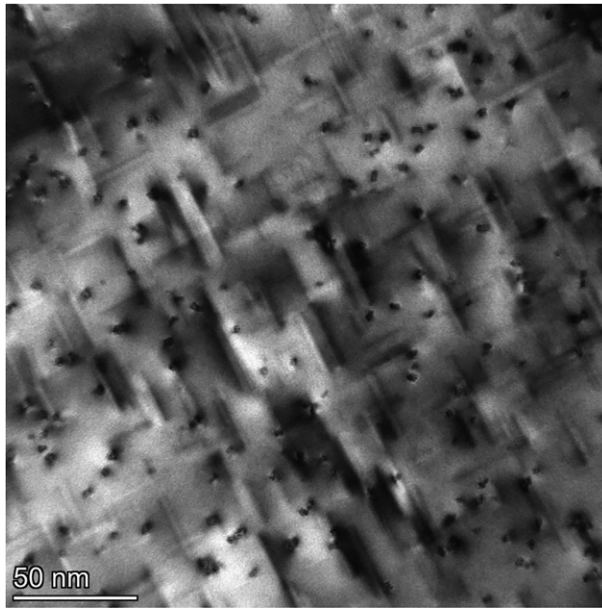


Fig. 7.4 Needle-shaped β'' fully coherent precipitates in a 6060 Al alloy.

Reproduced from Menapace, C., Bernard, F., Lusa, M., Ischia, G., Straffelini, G., 2021. Effect of a dual-rate ageing treatment on the tensile properties of a commercial 6060 Al alloy. *Mater. Sci. Eng. A* 819, 141468, with permission of the Publisher.

elements like Zr, Mn, and Cr, particular precipitates (such as Al_3Zr , $\text{Al}_{20}\text{CuMn}_3$, or $\text{Al}_{12}\text{Mg}_2\text{Cr}$) are formed during the homogenization treatment. They are called *dispersoids* and are small, in the range of 0.01–0.1 μm . These particles inhibit recrystallization and grain growth during the subsequent thermal treatments.

Precipitates form during the aging stage of the solution treatment, realized in many Al alloys (generally after forming) to increase their mechanical strength. Such precipitates are nanometric in size and they hinder the dislocation motion. Alloys in the under-aged state contain very small coherent particles, whereas alloys in the over-aged state contain small incoherent particles. As an example, Fig. 7.4 shows the microstructure (obtained by TEM) of a heat-treated and aged 6060 Al alloy (Menapace et al., 2021). The fine distribution of closely spaced nanometric fully coherent β'' (Mg_5Si_6) precipitates can be appreciated. The mean size of the precipitates is $4\text{ nm} \times 4\text{ nm} \times 50\text{ nm}$, and the mean number per unit area is $3600\text{ }\mu\text{m}^2$. The calculated mean spacing, λ_v , is rather small, on the order of 0.01 μm .

A particular microstructural feature of Al alloys is constituted by *porosity*. Al castings observe the presence of both solidification and gas porosity. In wrought alloys, only gas porosity is present. Gas porosity is created by the formation of molecular hydrogen during solidification, since the solubility of hydrogen in solid aluminum is very low. Such excess hydrogen gives rise to the formation of micropores in the microstructure. The typical size of the micropores is in the range of 1–5 μm , and their number, N_v , is in the range of 8000–70,000 mm^{-3} . The volume fraction of such pores

is rather low, in the range of 0.001 and 0.003, and the calculated mean spacing is in the range 24–50 μm . They typically form adjacent to the precipitates, and together with intermetallic particles, micropores play an important role in the ductile fracture of Al alloys (Toda et al., 2014).

To summarize:

- Particles in steels are inclusions or precipitates;
- Inclusions are mainly oxides and sulfides. They are comparatively large (1–10 μm) and their mean spacing is also large (15–100 μm);
- Precipitates are carbonitrides (with a size less than 0.1 μm and a mean spacing less than 0.3 μm), and fine carbides precipitated during tempering and after quenching (size less than 0.01 μm and small spacing, approximately 0.05 μm);
- The main particles in Al alloys are intermetallics, dispersoids, and precipitates;
- Intermetallic particles have a size of less than $\sim 5 \mu\text{m}$, and a mean spacing of 10–20 μm ; dispersoids are like intermetallic particles;
- Precipitates formed during aging are nanometric in size and their mean spacing is rather small (on the order of 0.01 μm);
- In Al alloys, gas micropores are also present, with a size of 1–5 μm and a mean spacing of 20–50 μm ;
- In Table 7.3, the main characteristics of the particles present in steels and Al alloys analyzed in the present section are summarized;
- In Fig. 7.5, the main role of the particles in the ductile damage process is anticipated; it is discussed in detail in the next sections.

Table 7.3 Summary of main characteristics of the particles in steels and Al alloys outlined in this section.

Type of particles	Size	Volume fraction, f_v	Mean spacing, λ_v (μm)
Steels			
Oxides	<10 μm	10^{-3} – 10^{-2}	15–20
Sulfides	<10 μm (Elongated inclusions: width: $\sim 5 \mu\text{m}$ length: 20–30 μm)	< 10^{-3} (Higher content in resulfurized steels)	80–100
Fine carbides and carbonitrides (high temperature tempering, HSLA steels)	<0.1 μm	$\sim 10^{-3}$ – 10^{-2}	<0.3
Fine transition carbides	<0.01 μm	$\sim 10^{-3}$ – 10^{-2}	<0.05
Spheroidal carbides	$\sim 1 \mu\text{m}$ (or less)	$\sim 10^{-2}$	2–10
Al alloys			
Intermetallic constituents	1–5 μm	$\sim 10^{-2}$	10–20
Porosity	1–5 μm	$\sim 10^{-3}$	20–50
Dispersoids	<0.1 μm	$\sim 10^{-3}$	<1
Precipitates (after aging)	<0.02 μm	10^{-2} – 10^{-1}	~ 0.01

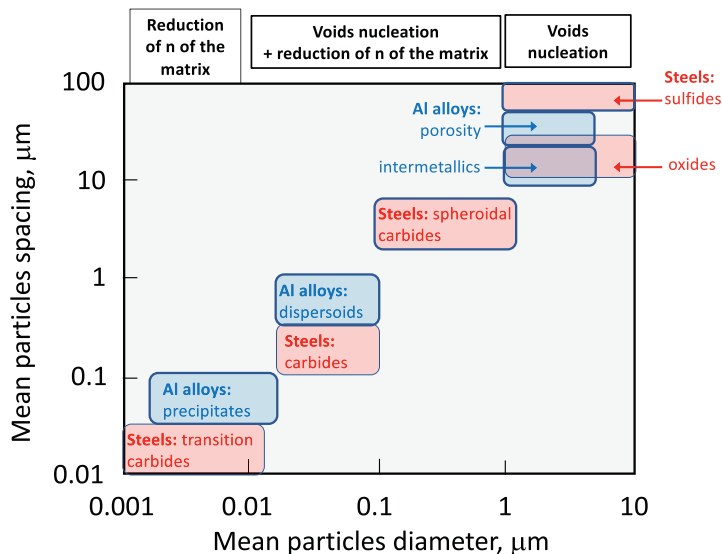


Fig. 7.5 Schematization showing the role of engineering particles on the ductile damage process (only broad intervals are indicated).

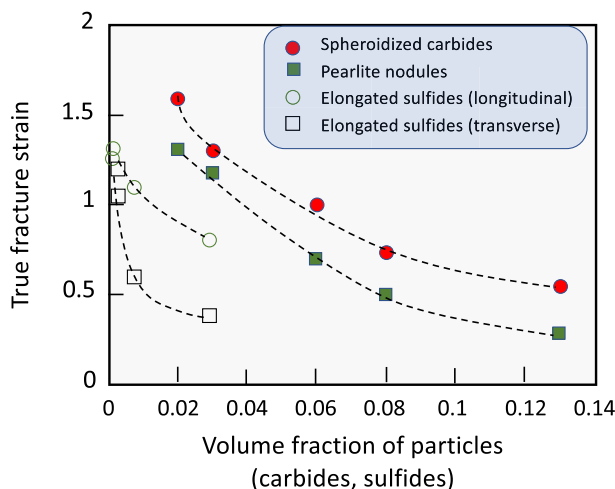


Fig. 7.6 Experimental relationship between the true fracture strain and the volume content of particles (either sulfides or carbides) in some steels.

Redrawn from Pickering, F.B., 1978. *Physical Metallurgy and the Design of Steels*, Applied Science Publishers.

7.2 Ductility of steels

We first consider the role of coarse particles, with a size in the range of 1 μm , on ductility. Fig. 7.6 depicts the experimental tensile true fracture strain, ϵ_f , for some **low-C steel**, **pearlitic** and **spheroidized** steels as a function of the fraction of particles in the

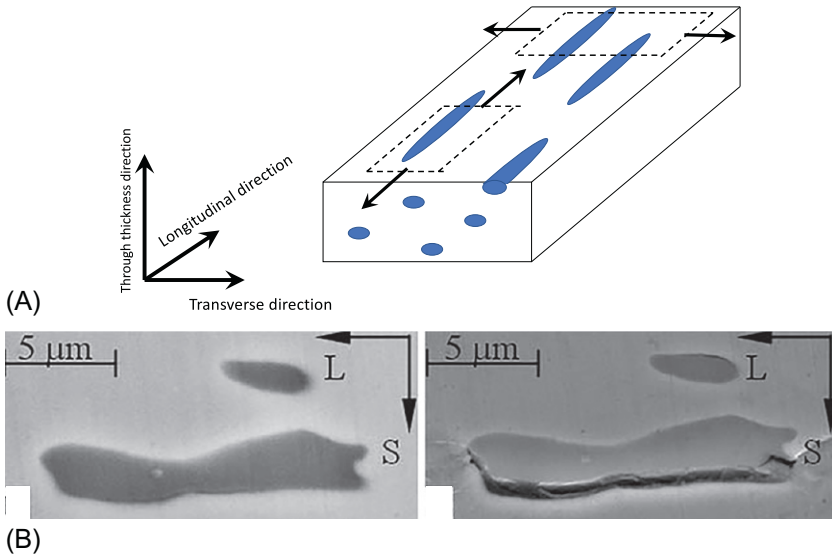


Fig. 7.7 (A) Schematization showing the orientation of elongated inclusions and obtaining of tensile specimens in the longitudinal (rolling) and transverse direction (for the sake of simplicity, the size of the inclusions is exaggerated). (B) SEM images showing the detachment of MnS elongated from the matrix, tested perpendicularly to the tensile axis.

(B) Reproduced from Hosseini, S.B., Temmel, C., Karlsson, B., Ingsten, N.-G., 2007. An in-situ scanning electron microscopy study of the bonding between MnS inclusions and the matrix during tensile deformation of hot-rolled steels. *Metall. Mater. Trans. A* 38, 982–989, with permission of the Publisher.

microstructure. Four types of steels are considered. Two steels are low-C steels containing approximately 0.04% S and therefore up to 2% of sulfides. For these steels voids nucleation is due to the sulfides (and possibly to other inclusions such as oxides). The other two steels are spheroidized and pearlitic steels, with different C content, i.e., different fractions of rounded or pearlitic carbides. These steels contain a low fraction of inclusions and high fraction of carbides (or pearlitic carbides) that are mostly responsible for the nucleation of damaging voids. As expected, for all the steels ductility is seen to decrease with increasing the initial fraction of particles.

Fig. 7.6 shows that the ductility of low-C steels is lower for the specimens taken in the transverse direction. As a matter of fact, the Mn sulfides are elongated in the longitudinal (rolling) direction, as shown in Fig. 7.1, and when the tensile specimens are taken in the transverse direction these sulfides are perpendicular to the loading direction (Fig. 7.7A), and then induce the formation of very large voids on the resisting section (Fig. 7.7B), whose growth quickly leads to coalescence and fracture. Another example is presented in Fig. 7.8. It shows the engineering stress-strain curves of a medium C steel (0.36%–0.38% C) for hot forging, with a comparatively high S content (0.03%). The tensile specimens had different orientations with respect to the rolling direction (in this case, 90 degrees means that the specimen is parallel to

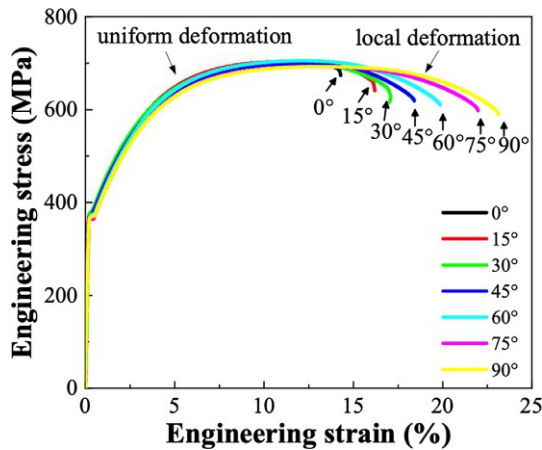


Fig. 7.8 Engineering stress-strain curve for a medium carbon steel for hot forging. The tensile specimens had different orientations with respect to the rolling direction (90 degrees means that the specimen is parallel to the rolling direction, and 0 degree that it is perpendicular). Reproduced from Wu, M., Fang, W., Chen, R.M., Jiang, B., Wang, H.B., Liu, Y.Z., 2019. Mechanical anisotropy and local ductility in transverse tensile deformation in hot rolled steels: the role of MnS inclusions. *Mater. Sci. Eng. A* 744, 324–334, with permission of the Publisher.

the rolling direction, and 0 degree that it is perpendicular). It is observed that ductility markedly decreases in passing from 90-degree orientation (EL: 40%; RA: 22%) to 0 degree (EL: 20%; RA: 13%). In passing, it is also noted that the flow curves in the uniform region were not influenced by the characteristics of the sulfides; as already observed in [Section 2.3](#), uniform deformation (and the strain-hardening coefficient) is little affected by large and widely dispersed particles.

The ductility of these steels can be noticeably improved by reducing the S content and in particular by adopting the calcium treatment before casting. It induces the formation of spherical particles, most of which float out before solidification. As a result, a low fraction of rounded inclusions remains in the steel microstructure, with a limited effect on ductility.

The spheroidized and pearlitic steels in [Fig. 7.6](#) have different C content, i.e., different fractions of rounded or lamellar carbides (it is recalled that the fraction of carbides is proportional to the C content since the solubility of C in the ferrite is negligible). Also, for these steels, ductility decreases with increasing the fraction of particles. However, when comparing with the other steels, it is clear that ductility is much higher for the same fraction of particles. This behavior can be explained by considering that plastic strain for voids nucleation is not negligible in these steels (whereas it is very small in steels containing the sulfides, as already explained), and it is comparable with the strain for voids growth and coalescence. Therefore it contributes significantly to ductility.

It can also be observed that for a given content of carbides, the pearlitic steels display a lower ductility than the spheroidized ones. This is because pearlite is made by ferrite and cementite lamellae. Plastic deformation concentrates in the ferrite lamellae,

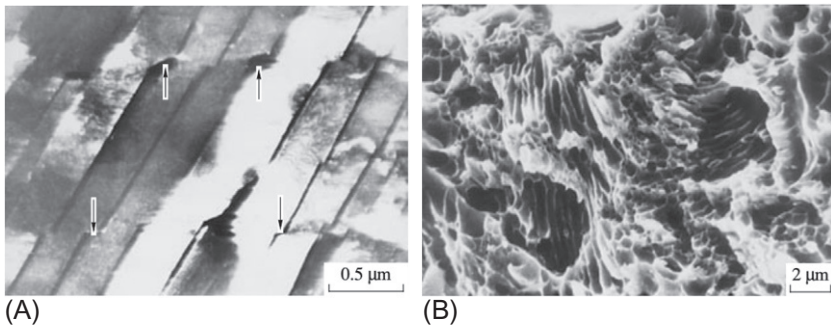


Fig. 7.9 (A) Micro-shear bands (indicated by the *arrows*) in the ferrite lamellae of a coarse pearlite colony; (B) fracture morphology showing the typical oblong dimples.

Reproduced from Izotov, V.I., Pozdnyakov, V.A., Luk'yanenko, E.V., Usanova, O.Y., Filippov, G.A., 2007. Influence of pearlite fineness on the mechanical properties, deformation behavior, and fracture characteristics of carbon steel. *Phys. Metals Metallogr.* 103, 519–529, with permission of the Publisher.

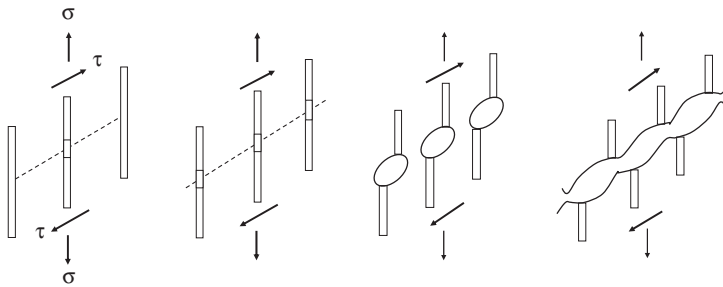


Fig. 7.10 Miller-Smith mechanism for ductile fracture of pearlite. The cementite lamellae parallel to the loading direction are subjected to the maximum shear that induces local micro-shearing in the ferrite lamellae and then fractures of cementite with the subsequent formation of voids.

Redrawn from Miller, L.E., Smith, G.C., 1970. Tensile fractures in carbon steels. *J. Iron Steel Inst.* 208, 998–1005.

where dislocations form cell substructures. Subsequently micro-shear bands are formed; during plastic deformation they rotate and cut the cementite lamellae, as shown in Fig. 7.9A (Izotov et al., 2007). Cementite lamellae can then prematurely break, inducing the formation of voids, following the mechanism schematized in Fig. 7.10, originally suggested by Miller and Smith. Such voids then have a size like the corresponding pearlitic colonies, whose volume fraction is greater than the volume fraction of the sole carbides. The resulting fracture morphology exhibits the typical fibrous morphology characterized by the presence of oblong dimples (Fig. 7.9B).

On this basis, the importance of spheroidization annealing in alloy steels can be highlighted. This treatment has two notable effects: it increases ductility and decreases the flow strength. For this reason, it is commonly adopted for medium-high C steels that must be formed by cold plastic deformation.

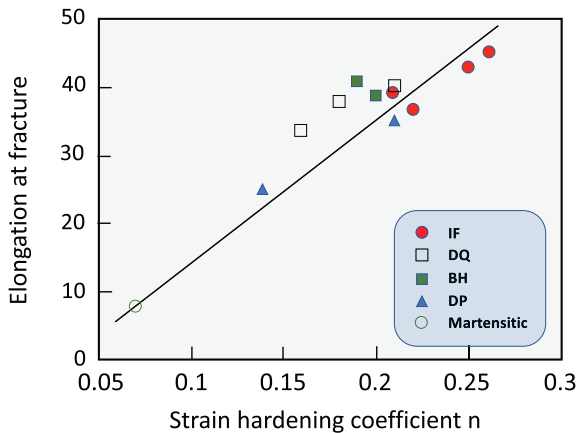


Fig. 7.11 Experimental relationship between the elongation at fracture and the strain-hardening coefficient for some ferritic steels for sheet metal working operations (data collected from: [Rana et al., 2009](#); [Talyan et al., 1998](#); [Omar, 2011](#)).

Consider now the ductility of advanced **ferritic steels** for sheet metal forming applications. These steels are rather *clean* since they are submitted to the calcium treatment. The corresponding S content is low (below 0.02%) to have a very low fraction of inclusions (on the order of 0.1%), to maximize ductility, and avoid penalizing the surface appearance of the products (that would be damaged by the surface inclusions). Dimpled ductile fracture is the result of nucleation of voids at the residual inclusions. The fraction of inclusions is low, and it is almost the same in these steels. Consequently, the ductility is expected to be high and dependent on the matrix properties, in particular on its ability to resist strain localization in the coalescence stage. [Fig. 7.11](#) shows the experimental relationship between the tensile elongation at fracture and the strain-hardening coefficient, n , for some IF, DQ (draw quality), BH, and dual phase steels. It can be clearly seen that the elongation at fracture increases with the strain-hardening coefficient, n , since ductility is mainly given by the attitude of the metal matrix toward micro-necking.

For microalloyed steels (**HSLA steels**), the picture is similar. These steels are also subjected to the calcium treatment and then contain a small fraction of equiaxed inclusions. However, they also contain precipitated carbonitrides. [Fig. 7.12](#) shows the fracture morphology, obtained by tensile testing, of a Ca-treated HSLA-100 steel (with only 0.003% S). The fracture morphology is very similar to that shown in [Fig. 6.12B](#), and then characterized by the presence of two populations of dimples. Together with large dimples, formed early at the small spherical inclusions, small dimples nucleated at the carbonitrides can be seen.

The HSLA steels considered in [Fig. 2.9](#) have n values in the range of 0.09–0.14. The corresponding elongations at fracture are in the range of 15%–20% ([Silva et al., 2018](#)). These experimental data fall in the interpolating line given by the data in [Fig. 7.11](#). Also, for these steels, the ductile fracture is typically initiated with the nucleation at the comparatively coarse sulfides. However, the final ductility is proportional to the average strain-hardening coefficient, n , which is related to the presence of the incoherent precipitates. Such particles then play a double role. First, they decrease n (and then they accelerate the growth of (primary) voids); second, they favor final

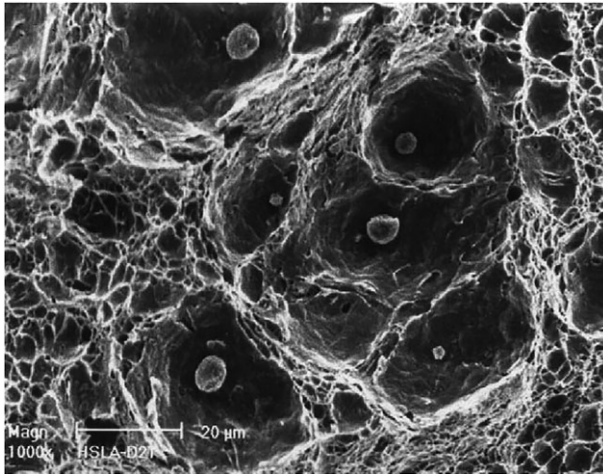


Fig. 7.12 Fractograph of a (clean) HSLA steel subjected to tensile testing. The presence of large dimples (nucleated at coarse inclusions) and very small dimples (nucleated at the precipitated carbonitrides) can be appreciated.

Reproduced from Chae, D., Koss, D.A., 2004. Damage accumulation and failure of HSLA-100 steel. *Mater. Sci. Eng. A* 366, 299–309, with permission of the Publisher.

coalescence by the nucleation of a population of secondary voids. In any case, they are responsible for the low n values and the corresponding low elongations at fracture.

Martensitic steels are not subjected to working processes by plastic deformation. However, they also require a minimum level of ductility to guarantee a sufficient reliability in service and fracture toughness. In general, these steels are rather clean (to maximize fatigue resistance). The voids nucleate at inclusions but the ductility is mainly determined by the matrix properties, such as the presence of fine carbides (that are firmly attached to the matrix), high dislocations density, and the boundaries of martensite substructures (lath martensite at low C content, plate martensite otherwise), that modify the strain-hardening coefficient, n . It is very important to distinguish between steels submitted to low-temperature tempering and steels submitted to high-temperature tempering. In the first case, the fine precipitates act together with the high density of dislocations in reducing the strain-hardening coefficient and favoring micro-necking in the ligaments among growing voids. However, the resulting ductility is generally not too low, ϵ_f being in the range of 0.22–0.9, depending on the C content and the austenitization temperature (Krauss, 2001). This is a result of the typical low inclusion content of these steels and the presence of residual austenite that induce the first yielding phenomenon (see Section 1.3). This induces quite a high initial strain-hardening rate, which delays necking and micro-necking to comparatively large strains.

In the case of high-temperature tempering, coarse cementite particles form (Table 7.3) and they may serve as sites for secondary voids nucleation, which reduces ductility. However, the density of dislocation is decreased by tempering. This, together with the disappearance of the transition carbides, increases the strain-

hardening coefficient, n , contributing to an increase in ductility. In the end, the yield strength decreases with the intensity of tempering, but ductility increases. A typical value for n in tempered martensitic steels is 0.07 and the corresponding elongation at fracture is 8%. This value is also included in Fig. 7.11.

A comparatively high tempering temperature may also cause embrittling phenomena, such as *tempered martensite embrittlement* (TEM) and *temper embrittlement* (TE). Such phenomena may induce a transition from ductile to brittle behavior, with a marked decrease in ductility, especially at high strain rates and in the presence of a high stress triaxiality (these aspects are further explored in Chapter 9).

A particular behavior is displayed by the **austenitic stainless steels**. In these steels (at least in the best grades), the inclusions content is limited, and ductility variations are mainly controlled by the matrix properties. In metastable grades, plastic deformation is characterized by the TRIP effect. The local formation of martensite during plastic deformation induces a local strengthening that favors the spreading of plastic strain. In Section 3.2 it is shown that the uniform strain (representative to the strain-hardening coefficient at necking) displays a maximum in the temperature range -10°C to 0°C for two AISI 304 steels. See also the behavior (C) in the schematization of Fig. 3.14. This behavior is reflected in a similar trend for the ductility, as shown in Fig. 7.13 for types 304 and 316 stainless steels (note that the maximum for the 316 steel is below RT since this steel has a higher SFE than 304). Similar trends are shown by different kinds of austenitic steels (Llewellyn and Hudd, 2000).

In this respect, it must be noted that in metastable austenitic stainless steels, the strain rate sensitivity coefficient, m , assumes negative values at comparatively large strains, in particular for temperatures below RT. This is shown, for example, in Fig. 3.17. As explained in Section 3.3, this behavior is rather typical in alloys that display phase transformations during plastic deformation such as the martensitic transformation. During plastic deformation, the strain-hardening rate and the instantaneous n value are rather high, and this means that strain localization is very limited. But when local plastic deformation is so intense as to involve the martensitic regions,

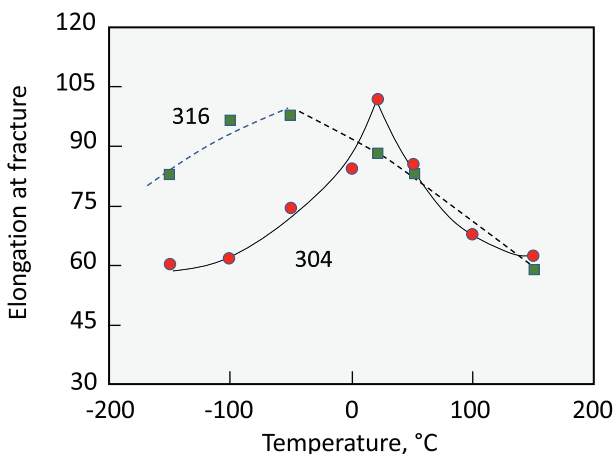


Fig. 7.13 Elongation at fracture of types 304 and 316 austenitic stainless steels in the temperature range of -150°C to 150°C . Data obtained from Byun, T.S., Hashimoto, N., Farrell, K., 2004. Temperature dependence of strain hardening and plastic instability behaviors in austenitic stainless steels. *Acta Mater.* 52, 3889–3899.

the strain-hardening rate, and n , decrease. This induces the attainment of macroscopic necking (when the macroscopic strain is equal to the instantaneous n value). At this point, the negative m value favors further localization of deformation, and thus the quick occurrence of ductile fracture.

The TRIP effect is also used to enhance ductility in high-strength steels for sheet metal forming operations (**TRIP steels**). In general TRIP steels display average n values in the range of 0.2–0.35, resulting from their strain-hardening behavior (Fig. 2.11A). The corresponding ductility is comparatively high, in the range of 20%–35%, in substantial agreement with the data of Fig. 7.11.

Rather high ductility values are displayed by **TWIP steels**. As seen in Section 2.4, the strain-hardening rate of these steels and the corresponding n values are very high, because of the contribution of twinning to plastic deformation. As for the TRIP effect, the formation of twins in the highly strained regions (note that also for twinning a critical strain must be reached) increases the metal strength locally, favoring the spreading of plastic deformation. As a result, the tensile plastic strain at fracture is very large. Fig. 2.16 shows the stress-strain behavior of three TWIP steels with different SFEs. The elongation at fracture is lower for the steel with 6%Al, which displays an insignificant contribution of twinning to plastic deformation. It is maximum for the steel with 3%Al (greater than 80%), characterized by an abundant twin formation, in particular at high strains. A striking example is shown in Fig. 7.14 that compares the stress-strain behavior of a typical TRIP steel, an FBDP steel (Ferrite-Bainite Dual Phase steel), and a TWIP steel (Cai et al., 2011). Tensile elongation at fracture of more than 100% was measured for the TWIP steel.

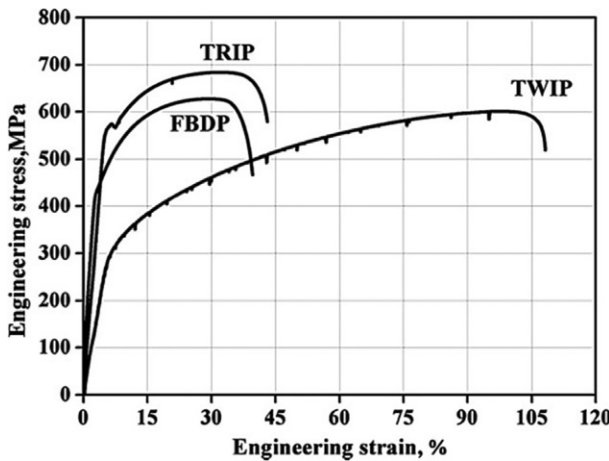


Fig. 7.14 Engineering stress-strain curves for three FBDP, TRIP, and TWIP steels. Reproduced from Cai, M.-H., Ding, H., Tang, Z.-Y., Lee, H.-Y., Lee, Y.-K., 2011. Strain hardening behavior of high performance FBDP, TRIP and TWIP steels. *Steel Res. Int.* 82, 242–248, with permission of the Publisher.

However, TWIP steels also display negative m values. As seen in Section 3.4, this is mainly due to the DSA phenomenon (serrations are also shown during tensile plastic deformation of the TWIP steel in Fig. 7.14), even if a contribution may also arise from a decrease in intensity of the TWIP effect with increasing strain rate (as observed for the TRIP effect). This means that fracture may occur soon after necking, as also encountered in the austenitic stainless steels. In addition, TWIP steels are very sensitive to geometrical imperfections, such as surface defects originating from the manufacturing process (Chung et al., 2011). This means that these steels are rather prone to shear localization during sheet tensile deformation, as already seen for some Al and Ti alloys. The formation of shear bands leads to shear fracture rather early during plastic straining. As is seen in Chapter 15, devoted to sheet metal forming, this phenomenon does not impair the optimal formability properties of these alloys (the strains to the appearance of shearing are comparatively high, as confirmed by the very high elongations at fracture), but nevertheless it must be properly taken into account.

To summarize:

- In C-Mn (ferritic-pearlitic) steels, coarse inclusions are responsible for the nucleation of voids that lead to ductile fracture.
- In high-quality steels (containing pearlitic or spheroidized carbides), ductility is determined by the volume fraction of the carbides.
- In sufficiently clean steels (S less than, say, 0.02%, generally because of the Ca treatment), f_v of the inclusions is rather small, close to 0.1%, and very similar among the steels; voids are nucleated at the residual inclusions but ductility is determined by the matrix properties, and in particular by the strain-hardening coefficient, n , as schematically shown in Fig. 7.15.
- Steels that display the TRIP and TWIP effects are characterized by high n values but negative m values. Ductility is rather high, but fracture occurs soon after macroscopic necking.

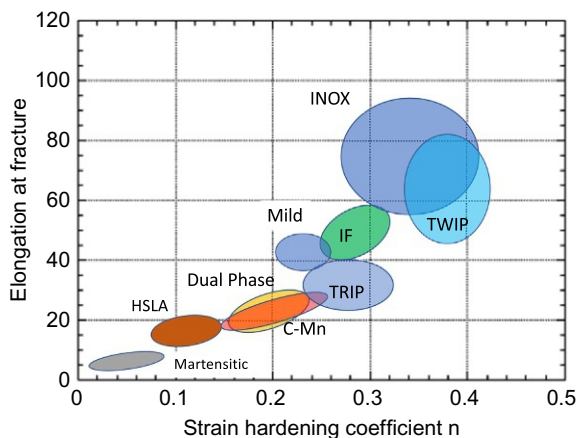


Fig. 7.15 Experimental dependence of elongation at fracture of selected steels with the average strain-hardening coefficient n .

7.3 Ductility of Al alloys

The alloys for plastic forming processes, i.e., the so-called *wrought alloys*, are considered. As mentioned in Section 7.1, intermetallic particles are mainly responsible for voids nucleation and play a primary role in reducing the ductility of Al alloys. In general, voids nucleation occurs by particle cracking, particle decohesion being mainly favored at high temperatures. Fig. 7.16 shows an example of voids nucleation by particle cracking. Fig. 7.17 shows a typical dimpled fracture surface. The presence of particles at the bottom of the dimples can be appreciated; they were identified as Fe- and Mn-rich particles (Smerd et al., 2005). Fe-containing particles are particularly deleterious in reducing ductility. In Al alloys gas porosity also plays an important role in the ductile fracture process. As highlighted in Section 7.1 wrought alloys contain some dissolved hydrogen that forms micropores at the intermetallic precipitates. Such micropores act as pre-existing voids that grow during plastic deformation. Therefore the actual fraction of voids that are involved in the plastic damage may be higher than the fraction of the particles. Fig. 7.18 shows the hydrogen desorption during tensile testing of a 2024 Al alloy (Toda et al., 2014). The tests were carried out in an ultra-high vacuum atmosphere and hydrogen desorption was measured as ion current values. A peak in the hydrogen desorption can be clearly seen at the final fracture, showing that a large fraction of dimples were made by gas porosity. Iron and hydrogen are then impurities that should be reduced to a minimum to avoid penalizing ductility.

Fig. 7.19 shows the reduction of ductility with increase in the intermetallic constituents in pure Al, Al-4%Mg alloys (class 5000) and different Al-Cu alloys (class 2000). For pure Al, containing FeAl and FeNi particles, the fracture strain is comparatively

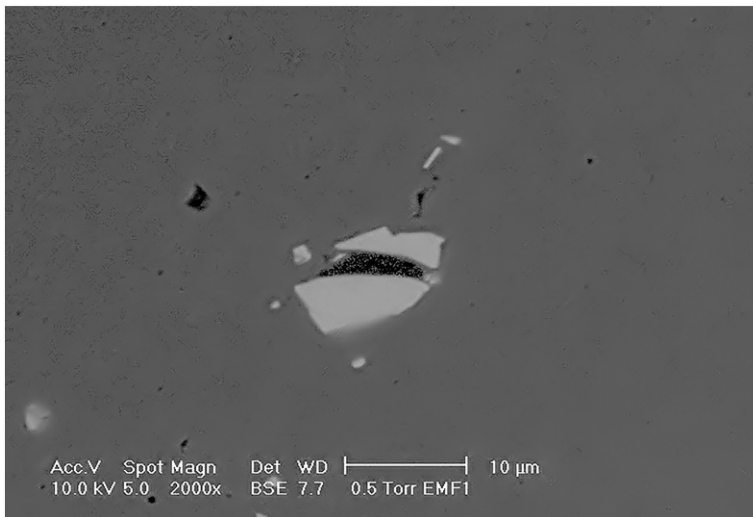


Fig. 7.16 Micro-void nucleation by particle cracking in a 5182 Al alloy.

Reproduced from Zadra, M., Straffelini, G., Molinari, A., 2008. Tensile ductility of AA5182: effect of impact loading. Mater. Sci. Technol. 24, 1259–1264, with permission of the Publisher.

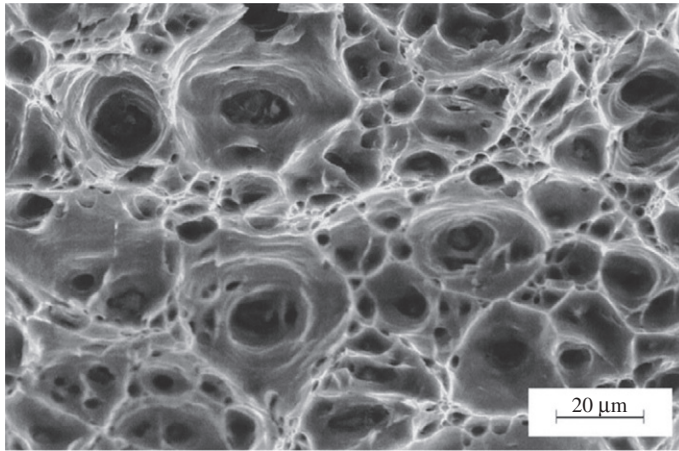


Fig. 7.17 Fracture morphology of a 5754 Al alloy. Some of the largest dimples contain Fe- and Mn-rich constituents.

Reproduced from Smerd, R., Winkler, S., Salisbury, C., Worswick, M., Lloyd, D., Finn, M., 2005. High strain rate tensile testing of automotive aluminum alloy sheet. *Int. J. Impact Eng.* 32, 541–560, with permission of the Publisher.

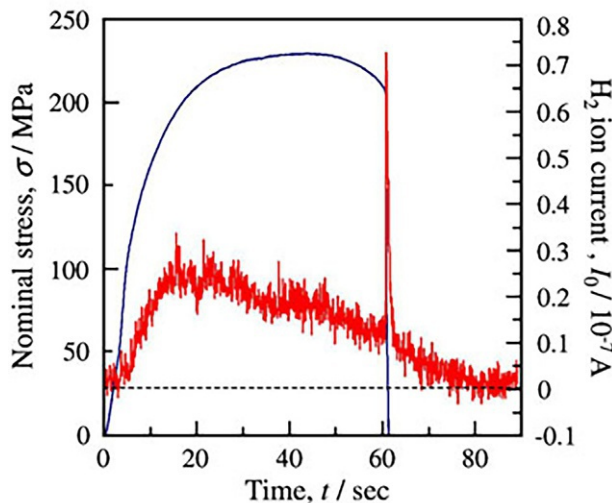


Fig. 7.18 Hydrogen desorption during tensile testing of a 2024 Al alloy. Reproduced from Toda et al. (2014), with permission of the Publisher.

high. In the commercial alloys, FeAl_m particles are responsible for voids nucleation and the strain for nucleation is negligible, since the cracking of intermetallic constituents occurs at the beginning of plastic deformation (Lassance et al., 2007). In addition, the possible presence of dispersoids and gas porosity in the commercial alloys may have contributed to decreasing ductility.

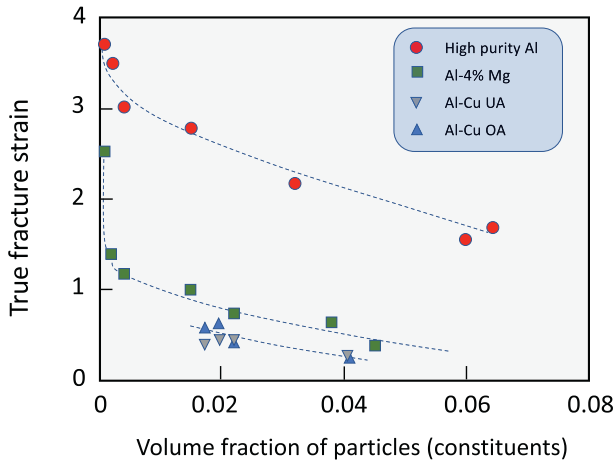


Fig. 7.19 Influence of volume fraction of intermetallic particles on the ductility of pure Al and an Al-4%Mg alloy.

Data from ASM International, 1991. ASM Handbook, Vol. 2 – Properties and Selection: Nonferrous Alloys and Special-Purpose Materials. ASM International; Walsh, J.A., Jata, K.V., Starke, Jr E.A., 1989. The influence of Mn dispersoid content and stress state on ductile fracture of 2134 type Al alloys. *Acta Metall.* 37, 2861–2871.

In heat-treatable Al alloys, ductility depends greatly on the aging treatment. In general, voids nucleation occurs at constituent particles and gas porosity, and the particles precipitated after the heat treatment influence ductility by modifying the strain-hardening coefficient, n . We have seen in [Section 2.5](#) that in the under-aged condition (UA), the strain-hardening coefficient decreases with aging time up to the peak stress, and in the over-aged condition (OA) the strain-hardening coefficient increases with aging time. Sometimes the UA condition is preferred in practice, since shorter treatment times are involved, and this is advantageous from an industrial point of view. [Fig. 7.20A](#) shows the evolution of the strain-hardening coefficient, n , for a 6060 Al alloy subjected to different aging treatments in the UA condition. As expected, n is seen to decrease with increasing the treatment time. [Fig. 7.20B](#) shows the corresponding experimental dependence of the elongation at fracture on the coefficient n . It is confirmed that ductility is proportional to n , irrespective of the selected treatment cycle. Consider, in this respect, that the aging cycle does not modify the constituent particles (including their fraction). Of course, UA alloys are not typically subjected to subsequent working processes by plastic deformation. However, as for heat-treated steels, they also require a minimum level of ductility to guarantee a sufficient reliability in service and an adequate fatigue behavior.

A complete picture of the influence of aging duration on the ductility of an Al-Cu-Mg alloy (class 2000) and an Al-Mg-Si alloy (class 5000), the first aged at 240°C and the second at 190°C for different times, is shown in [Fig. 7.21](#) ([Liu et al., 2004](#)). Here, the ductility (expressed in terms of the percent reduction of area at fracture) is plotted versus the aging time. As expected, for both alloys ductility

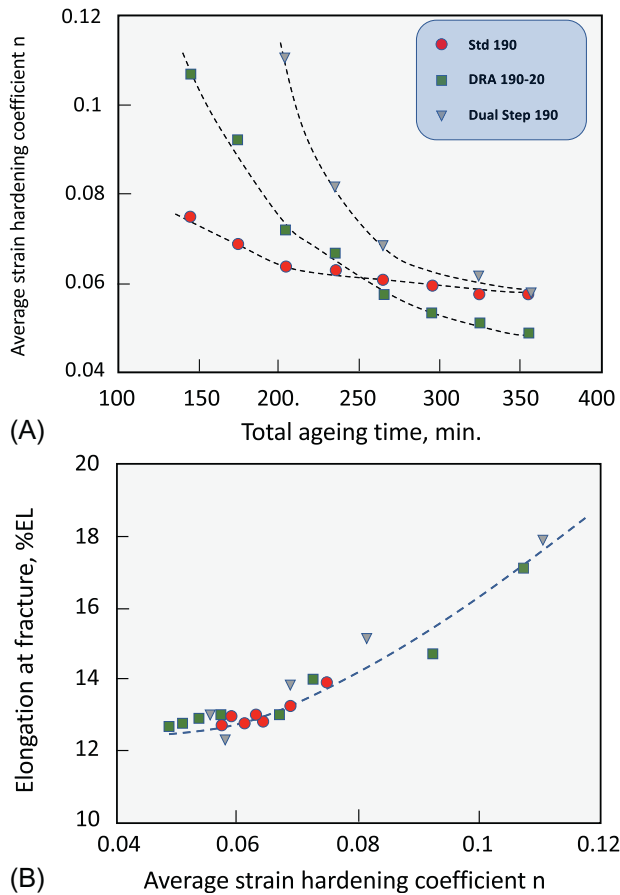


Fig. 7.20 (A) Experimental evolution of the average strain-hardening coefficient, n , in 2021 Al alloy subjected to different aging treatments (Std 190 is a standard solution hardening with aging at 190°C; the other two treatments are described in the cited reference); (B) corresponding experimental dependence of the tensile elongation at fracture with n .

Data from Menapace, C., Bernard, F., Lusa, M., Ischia, G., Straffelini, G., 2021. Effect of a dual-rate ageing treatment on the tensile properties of a commercial 6060 Al alloy. Mater. Sci. Eng. A 819, 141468.

decreases with aging time up to the peak aging condition (indicated by two arrows), and then starts increasing again in the overaged conditions, following the trend of the average strain-hardening coefficient.

Al alloys of the class 5000 are typically non-heat-treatable alloys. They contain Mg in solid solution and the presence of this element induces the appearance of the DSA phenomenon during plastic deformation. As seen in Section 3.4, in the temperature interval where DSA is present, the m values are negative. We have seen that in the case of steels, negative m values penalize ductility, since fracture is quickly attained after necking. The same occurs for Al alloys. We have already seen that negative m

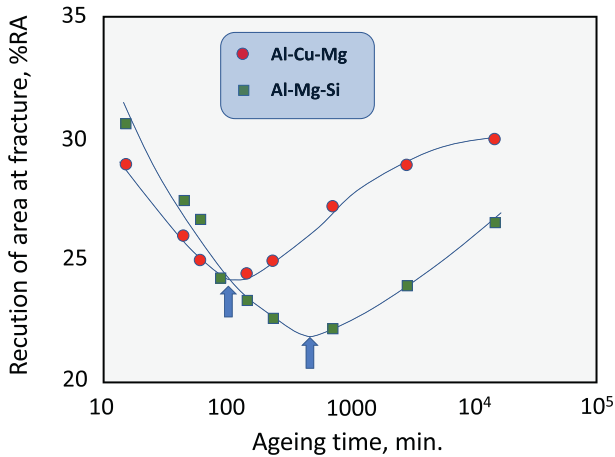


Fig. 7.21 Influence of aging duration on the ductility of an Al-Cu-Mg alloy (class 2000) and an Al-Mg-Si alloy (class 5000), the first aged at 240°C and the second at 190°C for different times. Data obtained from Liu, G., Zhang, G.J., Ding, X.D., Sun, J., Chen, K.H., 2004. The influences of multiscale-sized second-phase particles on ductility of aged aluminum alloys. *Metall. Mater. Trans. A* 35, 1725–1734.

values favor the formation of shear bands that anticipate localized necking in the tensile deformation of sheet specimens (Section 3.6). In the shear band, plastic deformation increases significantly and the negative m values accelerate the strain intensification up to the quick attainment of the fracture condition ($\epsilon_{\text{eff}} = \epsilon_{\text{eff},f}$). As an example, Fig. 7.22 shows the experimental dependence of ductility of a 5182-O Al alloy on temperature (close to RT) (Picu et al., 2005). The evolution of the strain rate sensitivity coefficient, m , with temperature is shown in Fig. 3.25B. Ductility approximately follows the same trend of m , and therefore reaches a minimum in the temperature range of 0–50°C. Similar to TWIP steels, ductility remains rather high, despite the negative role of m . As a result, these materials are still rather attractive to plastic-forming operations. However, because of DSA, the surface appearance is penalized, due to the formation of band markings (see Fig. 3.19B).

To summarize:

- In Al alloys, ductile damage is initiated at rather coarse intermetallic constituents (mainly FeAl_m intermetallics) and at gas porosity; ductility decreases with increasing the fraction of intermetallic constituents and hydrogen content. Iron and hydrogen are then impurities that should be reduced to a minimum;
- In heat-treatable alloys, ductility is proportional to the strain-hardening coefficient (which generally decreases with time in the underaged condition, and increases with time in the overaged condition);
- In non-heat-treatable alloys that display the DSA phenomenon (such as the alloys of the 5000 class), ductility is influenced by the negative m values, and fracture quickly occurs after localized necking (by shear localization).

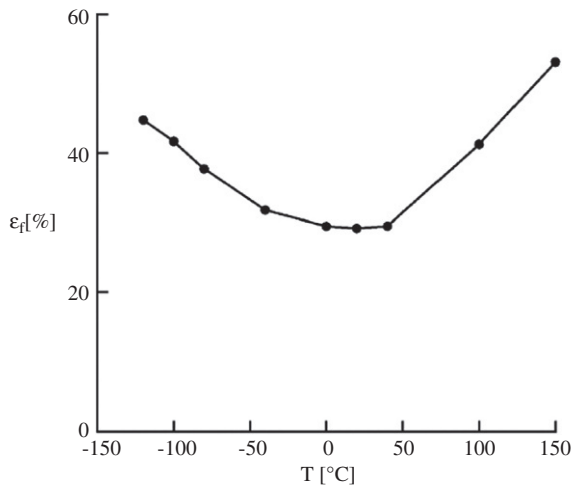


Fig. 7.22 Experimental dependence of the percentage elongation at fracture of a 5182-O Al alloy on temperature.

Reproduced from Picu, R.C., Vincze, G., Ozturk, F., Gracio, J.J., Barlat, F., Maniatty, A.M., 2005. Strain rate sensitivity of the commercial aluminum alloy AA5182-O. *Mater. Sci. Eng. A* 390, 334–343, with permission of the Publisher.

7.4 Ductility of other selected non-ferrous alloys

In this section, some peculiar aspects of the ductile fracture behavior of Cu, Mg, and Ti alloys in reference conditions are briefly presented and discussed.

Cu and Cu alloys

In Cu and Cu alloys, the main particles responsible for voids nucleation are inclusions and precipitates. The residual oxygen forms globular Cu oxides (such as CuO₂), whose typical volume fraction is in the range of 0.002–0.005, and the interparticle spacing is in the range of 10–15 μm (Melander, 1980; Pardoen, 2006). In brasses (Cu–Zn alloys), ZnS may also be present (Weinrich and French, 1976). In addition, intermetallic precipitates may also be present in the microstructure, depending on the chemical composition of the alloy. Voids nucleate at these particles rather soon after the onset of plastic deformation.

In sufficiently clean Cu, the inclusions content is rather low and the ductility (in particular its variation among materials of the same type) is mainly controlled by the matrix properties, as observed in clean steels. As an example, Fig. 7.23 shows the true stress–true strain curves of pure Cu (99.97%) wires, in the cold-worked (as-received) condition, and in the annealed condition. The average strain-hardening coefficient (obtained using the procedure of Fig. 2.7) for the annealed condition is 0.46. The corresponding fracture strain is $\epsilon_f = 1.27$ (RA = 72%). In general, in

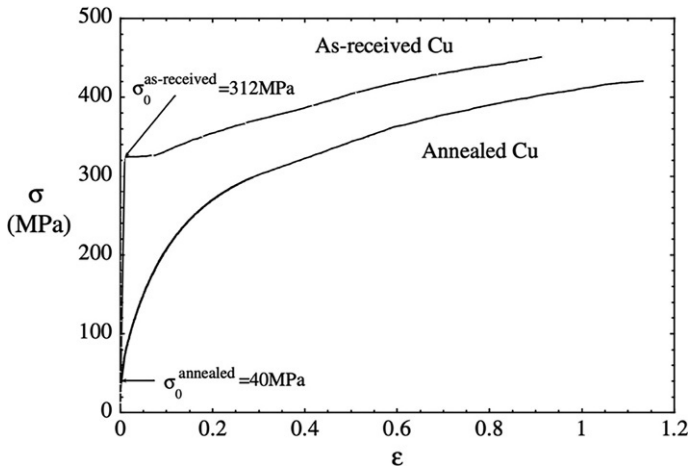


Fig. 7.23 True stress-true strain curves for pure Cu in the as-received (as received/cold drawn) and annealed state.

Reproduced from Pardoen, T., 2006. Numerical simulation of low stress triaxiality ductile fracture. *Comput. Struct.* 84, 1641–1650, with permission of the Publisher.

annealed OFHC (oxygen-free high-conductivity) Cu, the elongation at fracture is around 50%. The very high ductility of annealed Cu is clearly due to the high strain-hardening coefficient. For the cold-worked Cu in Fig. 7.23, the average strain-hardening coefficient is 0.12 and fracture strain decreases to 0.95 (RA = 61%).

In Cu-Zn and Cu-Al alloys, the SFE decreases with increasing alloying content (see Table 2.1). Correspondingly, the ductility increases because of the additional contribution to strain hardening exerted by twinning. An example is shown in Fig. 7.24, which compares the true stress-true strain curves, and the corresponding strain-hardening curves, for two Cu-Al alloys (Tian et al., 2016). The first one is Cu-4% Al with a high SFE, and the second one is Cu-11%Al with a lower SFE. For the Cu-11Al alloy, twinning dominates plastic deformation and this results in a higher

Fig. 7.24 True stress-true strain curves and the corresponding strain rate evolutions of two Cu-Al alloys, containing 4% and 11% of Al (Tian et al., 2016). The tests were carried out at RT with an initial strain rate of $8.3 \cdot 10^{-4} \text{ s}^{-1}$.

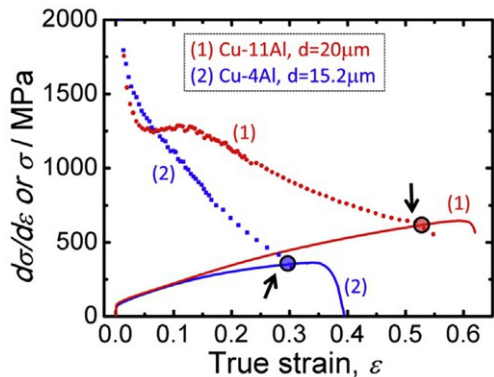


Table 7.4 Yield strength and percent elongation at fracture for selected commercial Cu alloys.

Common name	Composition (wt%)	σ_Y (MPa)	EL (%)
Electrolytic tough-pitch	99.9min Cu	83	48
Gilding, 95%	Cu-5Zn	97	45
Red brass, 85%	Cu-15Zn	110	45
Cartridge brass	Cu-30Zn	179	48
Phosphor bronze, 1.25% E	Cu-1.4Sn	124	47
Phosphor bronze, 5%	Cu-5Sn	165	50
	Cu-8Al	207	65
Copper nickel, 10%	Cu-10Ni	124	38
Copper nickel, 30	Cu-30Ni	172	32

Data from ASM International, 1991. ASM Handbook, Vol. 2 – Properties and Selection: Nonferrous Alloys and Special-Purpose Materials. ASM International.

flow strength and, especially, a higher strain-hardening rate, with a corresponding increased ductility (a similar behavior is displayed by the TWIP steels in Fig. 2.16). Selected experimental data are also summarized in Table 7.4. In Cu-Zn alloys, ductility decreases when the Zn content exceeds 37% (α - β brasses). In fact, the β -phase has a bcc structure and is characterized by a lower strain-hardening ability that penalizes ductility.

The low ductility of α - β brasses is useful with regard to the machinability of the alloys. In fact, a low ductility favors the breaking of the chip (see Chapter 16). Machinable brasses contain lead that is dispersed in the microstructure, preferentially at the grain boundaries, in the form of particles with an average size close to 1 μm . Such particles further reduce ductility by greatly favoring voids nucleation. In addition, they also serve as a lubricant at the tool-workpiece interface. The commonly used Cu-37Zn-3Pb alloys (37%Zn and 3%Pb) have an elongation at fracture of 24%, and a true fracture strain of 0.43 (Pantazopoulos and Toulfatzis, 2012).

Mg alloys

In Mg alloys, the main particles responsible for voids nucleation are intermetallics and oxides. They are characterized by a rather irregular morphology that greatly helps nucleation. Fig. 7.25A shows some Mn-Al and Mg-Al particles in a AZ91 alloy (with $f_v = 0.004$). It was observed that voids nucleation by particle cracking started from the beginning of plastic deformation ($\epsilon_n \approx 0$). Fig. 7.25B shows the dimpled morphology of an AZ31B alloy after fracture. In this case, the particles at the bottom of the dimples (responsible for voids nucleation) were found to be Mg-oxides. It has also been observed that voids can nucleate due to twinning, in particular as a result of twin/grain boundary interactions (Kondori and Benzerga, 2014). As an example, Fig. 7.26 shows twinning-induced voids in a AZ31B alloy (Rodriguez et al., 2016).

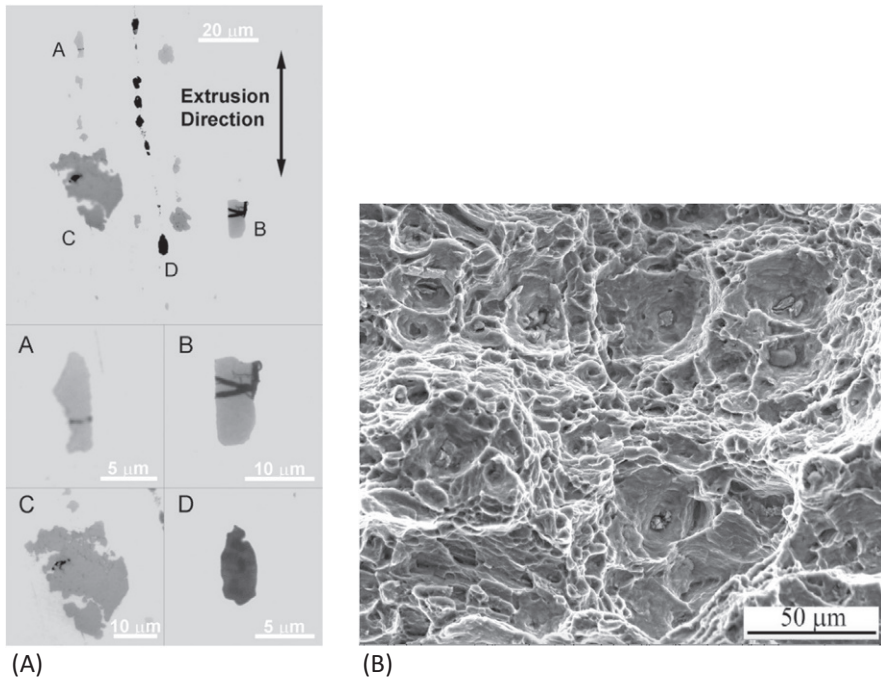


Fig. 7.25 (A) Cracked particles in an AZ61 Mg alloy; the cracks are perpendicular to the (tensile) loading direction that corresponds to the extrusion direction (reproduced from Lugo, M., Tschopp, M.A., Jordon, J.B., Horstemeyer, M.F., 2011. Microstructure and damage evolution during tensile loading in a wrought magnesium alloy. *Scr. Mater.* 64, 912–915, with permission of the Publisher); (B) dimpled fracture morphology in a AZ31B Mg alloy (reproduced from Kondori, B., Benzerga, A., 2014. Effect of stress triaxiality on the flow and fracture of Mg alloy AZ31. *Metall. Mater. Trans. A* 45, 3292–3307, with permission of the Publisher).

However, the matrix properties also play an important role in determining the ductility of Mg alloys. As already discussed, Mg alloys are characterized by a very low strain-hardening rate, and then a rather low average strain-hardening coefficient, n . Therefore, during plastic deformation coalescence of voids is rather easy, and it has been observed that localized shearing (promoted by the low n values) plays a very important role. As a result, especially in plastic deformation of sheet samples (high width-to-thickness ratio), shear localization and formation of shear bands is rather easy. Correspondingly, the resulting ductility is very low. For example, the elongation at fracture for an AZ61 alloy is 16% (ASM International, 1991) and the true fracture strain is generally lower than 0.5.

Different methods were proposed (and are under continuous development), for enhancing the ductility of Mg alloys at RT. As explained in Section 2.5, a reduction in the grain size is a viable method, even if the associated process costs are increased.

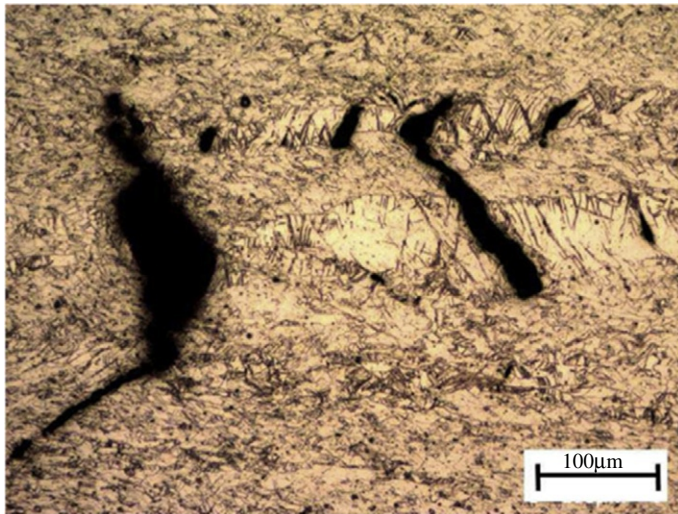


Fig. 7.26 Formation of twinning-induced voids in a AZ31B alloy.

Reproduced from Rodriguez, A.K., Ayoub, G.A., Mansoor, B., Benzerga, A.A., 2016. Effect of strain rate and temperature on fracture of magnesium alloy AZ31B. *Acta Mater.* 112, 194–208, with permission of the Publisher.

Another possibility is micro-alloying with specific rare-earth (RE) elements that randomize the grains' orientations and modify the activities of the different slip planes (Hirsch and Al-Samman, 2013). However, the improvements are still limited, and most of the plastic deformation processes of Mg alloys are carried out at temperatures beyond 300°C, i.e., in the high-temperature creep regime (see Section 11.5).

Ti alloys

A very peculiar behavior is displayed by Ti alloys. Because of the rather high solubility of the interstitial elements like oxygen, these alloys do not contain oxide particles. They also do not typically contain many precipitates or secondary phases that could nucleate the voids. Specific melt-related defects can be present, such as TiN compounds, Al rich regions, or W rich inclusions, but their content is generally kept to a minimum, through optimized melting and solidification practice (Lutjering and Williams, 2003). Nevertheless, ductile fracture is characterized by the typical dimpled morphology. In α - β alloys (like Ti-64), voids were found to mainly nucleate at the α - β interfaces, as shown in Figs. 7.27 and 7.28. In other alloys, voids were also found to nucleate inside the α - or β -grains. In general, voids nucleation is favored by local strain incompatibilities, provided by grain boundaries, twins or other microstructural imperfections. The nucleation strain is not negligible in these cases, and it contributes to determining the metal ductility.

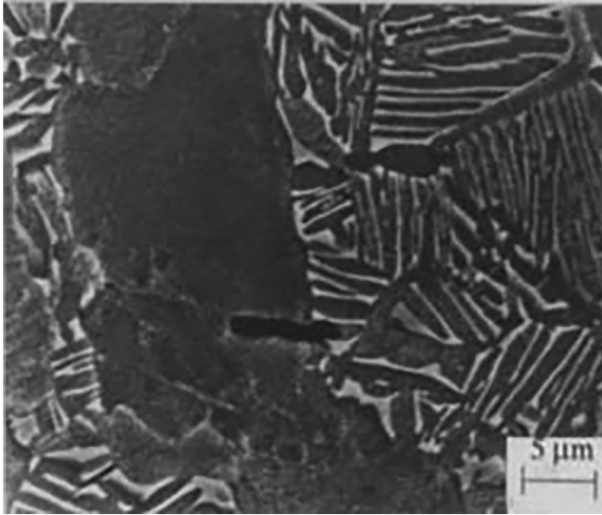


Fig. 7.27 Void nucleated at the α - β interface in a α - β Ti alloy. Reproduced from Helbert, A.L., Feaugas, X., Clavel, M., 1998. Effects of microstructural parameters and back stress on damage mechanisms in a/b titanium alloys. *Acta Mater.* 46, 939–951, with permission of the Publisher.

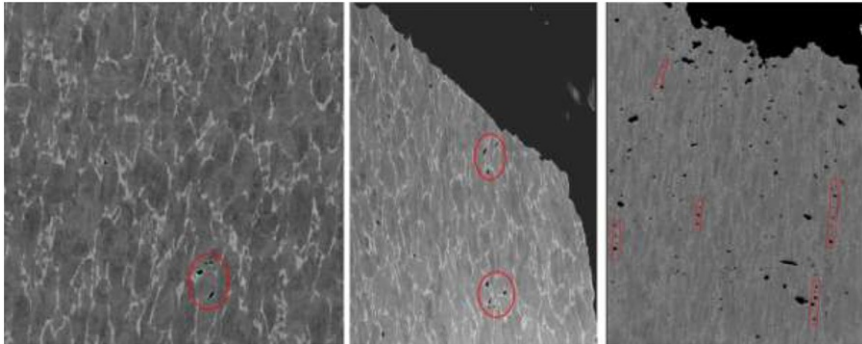


Fig. 7.28 Cross sections of a Ti-6Al4V alloy deformed to fracture, showing the voids at the α - β interfaces. Voids display a heterogeneous distribution (see the red circles) that favors coalescence.

Reproduced from Verleysen, P., Peirs, J., 2018. Quasi-static and high strain rate fracture behaviour of Ti6Al4V. *Int. J. Impact Eng.* 108, 370–388, with permission of the Publisher.

As seen in [Section 2.5](#) (see also [Table 2.3](#)), Ti alloys are characterized by rather low n values that reflect a pronounced tendency towards voids coalescence by localized shearing ([Fig. 6.11](#)), hence reducing ductility. In the case of plastic deformation of sheet samples (high width-to-thickness ratio), shear localization and formation of shear bands are rather easy, as observed in Al and Mg alloys.

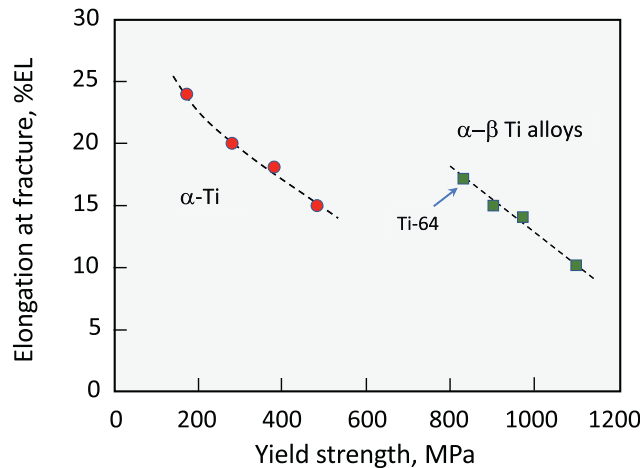


Fig. 7.29 Experimental dependence of the elongation at fracture on yield strength for some Ti alloys.

Data from ASM International, 1991. ASM Handbook, Vol. 2 – Properties and Selection: Nonferrous Alloys and Special-Purpose Materials. ASM International.

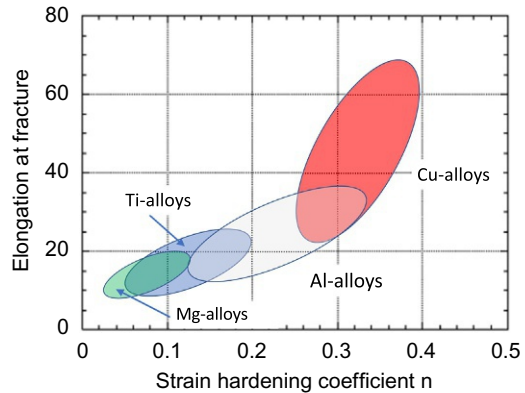
Fig. 7.29 show typical ductility values (expressed in terms of the elongation at fracture) for two Ti alloys, plotted as a function of the yield strength.

Metastable β alloys generally display a behavior similar to that of α - β alloys. However, if the stability of the β phase is properly lowered, these alloys may display twinning or martensitic transformation during plastic deformation. An exemplary case is that provided by the novel Ti β alloy, whose strain hardening behavior is shown in Fig. 2.24 (Ren et al., 2020). The yield strength of the alloy is 870 MPa and its elongation at fracture is 41%, i.e., much greater than the typical values of traditional Ti alloys. The tensile test shown in Fig. 2.24 was carried out at a strain rate of $5 \times 10^{-4} \text{ s}^{-1}$. The role of strain rate should be considered, since Ti alloys have a low thermal conductivity. If the strain rate increases, Ti alloys undergo significant temperature rise during plastic deformation. This may induce a suppression of twinning and a reduction of ductility (Ji et al., 2017).

To summarize:

1. In Cu and Mg alloys, voids nucleate at inclusions and intermetallic compounds, whereas in Ti alloys, voids nucleate at microstructural heterogeneities, due to local strain incompatibilities;
2. Even in these alloys, ductility is proportional to the average strain-hardening coefficient, n , as schematically shown in Fig. 7.30; note that with cold working, n decreases, and the elongation at fracture accordingly decreases;
3. In Mg and Ti alloys, ductility is rather low, reflecting a marked attitude of these alloys towards early coalescence by localized shearing.

Fig. 7.30 Experimental dependence of elongation at fracture of selected non-ferrous alloys with the average strain-hardening coefficient n . Note that with cold working, n decreases and the elongation at fracture accordingly decreases.



7.5 Final remarks on ductility of metals

It is generally recognized that ductility decreases with increasing strength, following a typical *banana-shaped* curve. This is shown in Fig. 7.31 for selected steels and non-ferrous alloys. The motivation of this trend relies on the microstructural characteristics of the alloys and their influence on the deformation mechanisms. In general, the increase in the yield strength of engineering metals is due to the increase in the density of geometrically necessary dislocations, ρ_G . An increase in ρ_G also determines a decrease in n , as explained in Chapter 2, which in turn is reflected by a decrease in ductility.

Fig. 7.31A shows the experimental dependency of the elongation at fracture of different steels as a function of yield strength (all in the annealed condition, with the exception of martensitic steels). Steels show the typical banana-shaped trend, with two exceptions. The first is that of TRIP steels: their ductility is like that of mild/IF steels, but they display a much greater strength. The second regards fcc austenitic

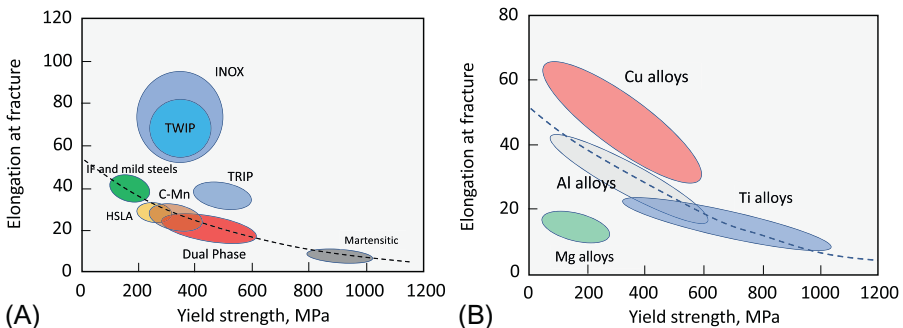


Fig. 7.31 Experimental dependence of elongation at fracture of steels (A) and selected non-ferrous alloys (B) with the yield strength. Note that with cold working, n decreases and the elongation at fracture accordingly decreases.

stainless steels, in particular the metastable grades and high-Mn TWIP steels. They are characterized by TRIP/TWIP effects during plastic deformation and display a rather high ductility.

Fig. 7.31B shows the experimental dependency of the elongation at fracture on the yield strength for selected non-ferrous alloys. The dotted line is the banana-curve of steels, taken from Fig. 7.31A. The figure shows that Al and Ti alloys almost follow the trend of the steels, whereas Cu alloys display a better strength-to-ductility ratio (similarly to inox and TWIP/TRIP steels), and Mg alloys display a worse strength-to-ductility ratio. Cu alloys take advantage of their fcc structure and intermediate SFE (and then of possible TWIP effect during plastic deformation); Mg alloys are significantly affected during plastic deformation by the limitations provided by their hcp lattice.

To summarize, the *dotted banana curve* represents a reference behavior that can be related to the fundamental characteristics of dislocations (in particular, geometrically necessary dislocations), and their role in plastic deformation. The materials that lie outside of this trend are characterized by particular deformation mechanisms. In particular, the occurrence of TRIP/TWIP effects is beneficial for increasing ductility without sacrificing strength. An exemplary case is that provided by the novel Ti β alloy, whose strain-hardening behavior is shown in Fig. 2.24 (Ren et al., 2020). The yield strength of the alloy is 870 MPa and its elongation at fracture is 41%, i.e., much greater than the value predicted by the reference curve.

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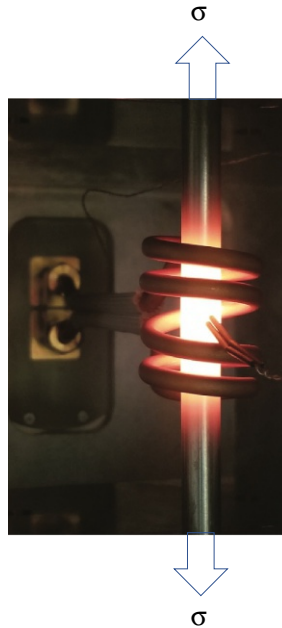
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Ductility of metals at high temperature

8



High-temperature tensile testing, with induction heating of the sample.
Courtesy of Dynamic Systems Inc.

Chapter outline

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 - 8.2 Ductile fracture for $T < 0.4-0.5T_m$ 221
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In this chapter, the ductility of metals for temperatures up to the melting point is considered, using many of the concepts introduced in the last two chapters.

Initially, the ductility of metals up to $0.4\text{--}0.5T_m$ is considered. A particular emphasis on the ductility loss occurring in the temperature range $0.15\text{--}0.4T_m$ is given (blue brittleness, connected with the DSA phenomenon). In addition, the role of strain rate is considered, introducing the well-known Johnson-Cook equation. Subsequently, the role of temperature on ductility in the high-temperature regime (above $0.4\text{--}0.5T_m$) is presented, focusing first on the high-stress (dislocation) creep regime. This is the temperature region of both hot and warm working processes. Also in this case, special emphasis is placed on particular embrittling phenomena taking place at $\approx 0.6T_m$ (the so-called: ductility trough). A unifying view is presented, considering how these phenomena are important in different alloys. To complete the picture, the concept of peak ductility temperature and the related hot shortness phenomenon are discussed.

At the end of the chapter, two other phenomena are presented. First, the deformation and fracture in the low-stress creep regime is treated, focusing on the conditions able to promote a sufficiently high strain rate and a corresponding high ductility, i.e., the conditions of superplasticity. This is a special material behavior that is suitable for sheet forming operations. Then, the damaging and fracture behavior of metals when forming adiabatic shear bands (ASBs) is considered. This is an important phenomenon for different metals in specific working operations.

8.1 A general overview

The fracture behavior of metals reflects their deformation behavior, which is typically represented by the deformation mechanism map schematized in Fig. 4.17. In particular:

- *Dimpled ductile fracture* (seen in detail in the previous chapters) typically occurs at temperatures below $0.7\text{--}0.8T_m$, and at comparatively high strain rates, typically above $5 \times 10^{-4} \text{ s}^{-1}$. For temperatures below $0.4\text{--}0.5T_m$, it occurs at high σ/E -values and is assisted by the dislocation slip (refer to Chapter 3); for temperatures above $0.4\text{--}0.5T_m$, it occurs at lower σ/E values and is mainly assisted by dislocation climb and DRV phenomena (refer to Chapter 4).
- *Ductile fracture by necking to a point* (see Fig. 6.1) occurs at very high temperatures, approximately above $0.8T_m$, and it is largely assisted by DRX phenomena.
- *Intergranular dimpled ductile fracture* occurs at low strain rates (typically below 10^{-6} s^{-1}) and low stresses, i.e., in correspondence to the region of grain boundary sliding (GBS) and diffusion creep. It is also known as *cavitation* and is very similar to dimpled ductile fracture (further details are provided in Section 8.6).

Intergranular dimpled fracture may also occur in the region of dislocation creep if the metal is embrittled, i.e., if it displays a ductility trough following particular damaging phenomena.

8.2 Ductile fracture for $T < 0.4\text{--}0.5T_m$

As seen in Chapter 3, for temperatures below $0.4\text{--}0.5T_m$, the plastic deformation of metals can be rationalized in the frame of the thermally activated slip of dislocations, and the influence of temperature and strain rate on plastic behavior is rather marked at comparatively low temperatures (lower than, say, $\sim 0.2T_m$). Since the strain rate sensitivity index, m , is very small and slightly affected by temperature, it is expected that ductility of metals in this temperature interval is mainly dependent on the evolution of coefficient n with temperature. Three important cases are found in metals, and they are depicted in Fig. 3.14.

1. In bcc metals, such as Fe and ferritic steels, the yield and flow strength significantly increase with decreasing temperature for temperatures lower than RT ($\sim 0.2T_m$) and, correspondingly, n decreases as temperature (and strain rate) are decreased; this is behavior A in Fig. 3.14.
2. In fcc metals, such as Cu and Al alloys, yield strength increases slightly with decreasing temperature and, differently from bcc metals, coefficient n increases with decreasing temperature; this is behavior B in Fig. 3.14.
3. In metastable fcc metals, such as austenitic stainless steels (characterized by the TRIP/TWIP phenomenon during plastic deformation), a peak value of n at temperatures close to $0.1\text{--}0.2T_m$ (depending on the chemical composition and hence on the SFE) is observed; this is behavior C in Fig. 3.14.

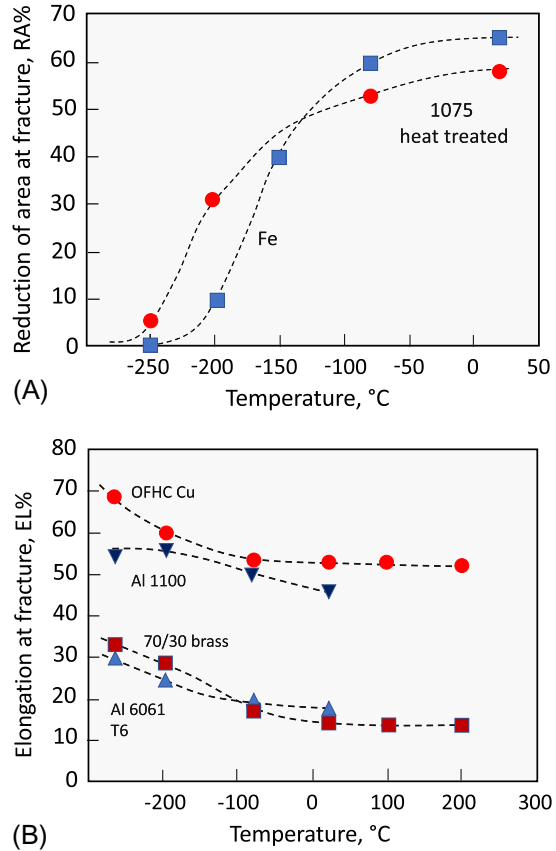
The experimental evolution of ductility of metals with temperature reflects the described evolution of the strain-hardening coefficient n . As an example, in Fig. 8.1 the ductility at low temperatures of Fe and a high-strength steel (behavior A), and of selected Cu and Al alloys (behavior B) is shown (see also Figs. 3.7 and 3.8). Ti alloys with a biphasic $\alpha + \gamma$ microstructure, such as Ti-64, display a behavior that is approximately intermediate between that of Cu and Fe.

As far as the evolution of ductility in metastable austenitic stainless steels is concerned (behavior C), reference can be made to the data shown in Fig. 7.13. Also, for these steels, the variation of ductility with the temperature follows the same trend shown by the strain-hardening coefficient, n .

The influence of temperature and strain rate on ductility decreases in intensity with increasing temperature above $\sim 0.2T_m$, and approaching $0.4\text{--}0.5T_m$, as predicted by the trends in n depicted in Fig. 3.14. However, a particular behavior is observed when the DSA phenomenon occurs during plastic deformation. As seen in Section 3.4, this phenomenon occurs in specific temperature and strain rate intervals (generally in the temperature interval: $0.15\text{--}0.4T_m$). We have already seen the case of Al—Mg alloys (class: 5000) that display DSA for temperatures around RT. Fig. 7.22 shows a drop in ductility in the temperature region characterized by DSA, due to the negative m values. In Section 3.4, we have also seen that, for a given temperature, DSA disappears when the strain rate is sufficiently high (i.e., above a certain value). For Al—Mg alloys (class 5000), an increase in the strain rate above approximately 10 s^{-1} is sufficient to exit the DSA region at RT (Picu et al., 2005). This is schematized in Fig. 8.2. For a 5083 Al alloy, it was obtained that m increases from approximately -0.01 in the DSA region to approximately 0.02

Fig. 8.1 (A) Reduction of area at fracture as a function of temperature for Fe and a heat-treated steel with $\sim 0.75\%C$; (B) elongation at fracture as a function of temperature for two Al and Cu alloys.

A: Data from Dieter, G., 1988. *Mechanical Metallurgy*, SI Metric Edition, McGraw Hill; Warren, K.A., Reed, R.P., 1963. *Tensile and Impact Properties of Selected Materials From 20 to 300K*, US National Bureau of Standards, Monograph 63; B: Data from Warren, K.A., Reed, R.P., 1963. *Tensile and Impact Properties of Selected Materials From 20 to 300K*, US National Bureau of Standards, Monograph 63.



when strain rate is in excess of $10\text{--}100\text{ s}^{-1}$ (Clausen et al., 2004). Correspondingly, the fracture mode changes from shear fracture to cup-and-cone fracture and the fracture strain also increases. For a 5182 Al alloy, it is reported that the fracture strain increases from 0.8 at $1 \times 10^{-3} \text{ s}^{-1}$ to 0.9 at 170 s^{-1} (Zadra et al., 2008).

Fig. 8.2 also shows that DSA disappears, at a given strain rate, with properly increasing temperature above RT, or decreasing below RT. Fig. 8.3 shows the fracture morphology of a number of tensile specimens made of Al 5083 alloys, tested at a strain rate of $1.7 \times 10^{-3} \text{ s}^{-1}$ and at increasing temperature (Clausen et al., 2004). At comparatively low temperature, deformation localizes in spontaneous shear bands and a shear ductile fracture then results. With increasing temperature, the DSA region is exited and m becomes positive. Consequently, the fracture mode changes, and a typical ductile cup-and-cone fracture occurs.

Also, a number of steels display the DSA phenomenon, but of course at a higher temperature than Al alloys. Some examples are reported in Section 3.4. DSA is typically present in the temperature interval $230\text{--}600^\circ\text{C}$ (i.e., close to $\approx 0.3T_m$). In this interval, negative m values are recorded. The corresponding drop in ductility is shown

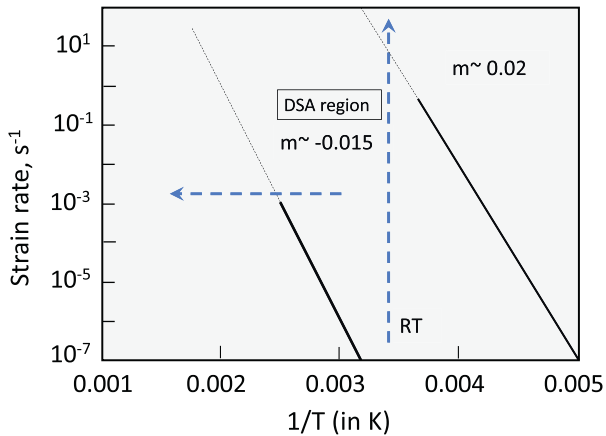


Fig. 8.2 Schematicization showing the DSA region for an Al—Mg alloy (class 5000) (data collected from different sources, indicated in the text). In the DSA region, shear ductile fracture occurs, and ductility is low.

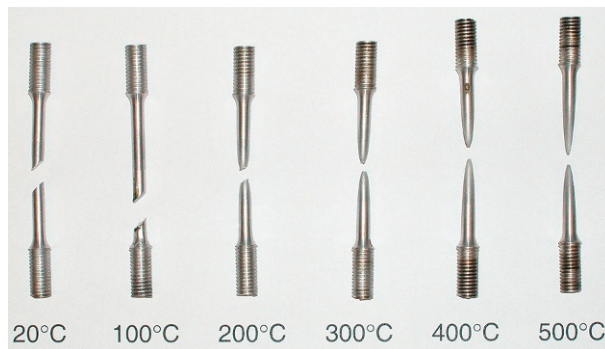


Fig. 8.3 Macroscopic fracture morphology of smooth tensile samples exhibits fracture at different temperatures.

From Clausen, A.H., Borvik, T., Hopperstad, O., Benallal, A., 2004. Flow and fracture characteristics of aluminium alloy AA5083-H116 as function of strain rate, temperature and triaxiality. *Mater. Sci. Eng. A* 364, 260–272, reproduced with permission of the publisher.

in Fig. 8.4 in the case of a dual-phase steel. This phenomenon is also called *blue brittleness*, since at these temperatures a bluish color is produced on the steel surfaces. This phenomenon can be alleviated or eliminated with the addition of elements such as Ti and Al, which are able to tie up N atoms and avoid DSA.

In the case of very high strain rates, in excess of 10^2 – 10^3 s⁻¹, the picture changes and strain rate plays an important role in ductility. In particular, fracture strains increase by about 20%–50% with respect to quasi-static conditions. The DSA phenomenon disappears and, as seen in Sections 3.2 and 3.3, both coefficients n and m considerably increase at high strain rates. At the same time, stresses and strains are not uniform along the sample length, as the deformation process has a short duration. Therefore inertial effects (due to the acceleration of any volume element of the sample during deformation) play an important role. Together with the high n and m values,

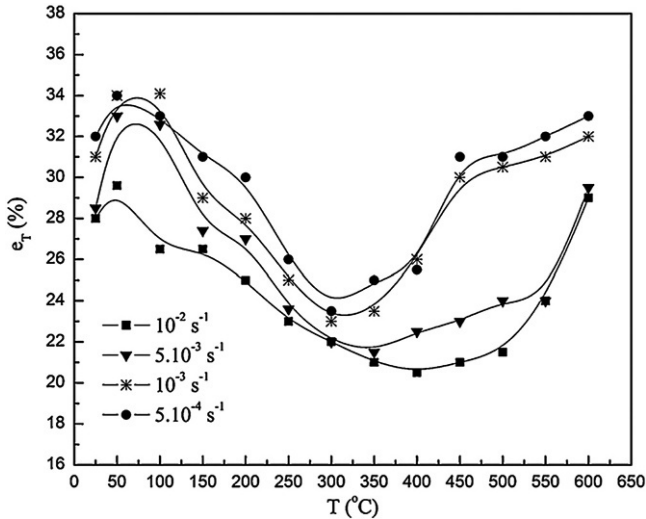
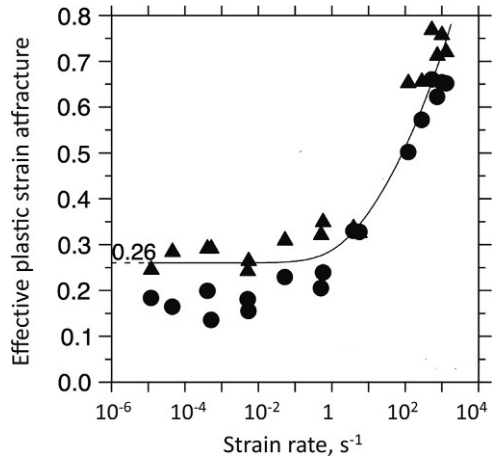


Fig. 8.4 Blue brittleness phenomenon in a dual phase steel (e_T indicates the percent elongation at fracture).

From Queiroz, R.R.U., Cunha, F.G.G., Gonzales, B.M., 2012. Study of dynamic strain aging in dual phase steel, Mater. Sci. Eng. A 543, 84–87, reproduced with permission of the publisher.

Fig. 8.5 Experimental fracture strains for specimens (parallel to the rolling direction) tested at RT and for different strain rates. The experimental data points indicated with *triangles* were obtained using a micrometer screw to measure the area of the fracture surface, whereas those indicated with *circles* were obtained using microscope measurements.

From Clausen, A.H., Borvik, T., Hopperstad, O., Benallal, A., 2004. Flow and fracture characteristics of aluminium alloy AA5083-H116 as function of strain rate, temperature and triaxiality. Mater. Sci. Eng. A 364, 260–272, reproduced with permission of the publisher.



the inertial effects contribute to spreading plastic deformation over the sample gauge and to lowering the stress triaxiality in the (developing) neck region, thus avoiding any intense localization of plasticity and decreasing the plastic damage evolution (Hu and Dahen, 1996). As an example, Fig. 8.5 shows the dependency of ductility with strain rate for a 5083 Al alloy (Clausen et al., 2004). As expected, ductility is almost unaffected by strain rate for comparatively low strain rates, up to approximately 10 s^{-1}

(in agreement with the small variation of n and m with strain rate, and the absence of inertial effects). Then it starts increasing, and at 10^3 s^{-1} it is approximately three times that at low strain rates.

Only when the deformation velocity is too high (in excess, e.g., of $\sim 80 \text{ m/s}$ in ferritic steels, corresponding to a strain rate of $\sim 10^4 \text{ s}^{-1}$ in the case of a gauge length of 10 mm), a drastic drop in ductility is observed. This is the result of wave effects. In fact, when deformation velocity is too high, plastic waves propagate at a velocity that is lower than the displacement velocity of the specimen. This results in the specimen fracturing, due to the development of local huge strains (Hu and Dahen, 1996). In most cases the fracture process is also accompanied by intense local heating and the formation of adiabatic shear bands (Section 8.7).

For specific calculations, it is useful to have a relationship capable of predicting the ductility of a metal as a function of temperature, T , and strain rate, $\dot{\epsilon}$. A rather common approach is based on the *J-C model* (Johnson and Cook, 1985):

$$\epsilon_{eff, f} = \epsilon_{eff, f, RT} \cdot \left(1 + D_1 \cdot \ln \frac{\dot{\epsilon}_{eff}}{\dot{\epsilon}_0} \right) \cdot \left(1 + D_2 \cdot \frac{T - T_0}{T_m - T_0} \right) \quad (8.1)$$

where

- $\epsilon_{eff, f, RT}$, is the effective fracture strain at RT and in quasi-static conditions (given, e.g., by the R - T criterion, Eq. 6.17 or Eq. 6.18);
- $\dot{\epsilon}_{eff}$ is the effective plastic strain rate, and $\dot{\epsilon}_0$ is usually set to 1 s^{-1} ;
- T_0 is the absolute RT , and T_m is the absolute melting point of the alloy;
- D_1 and D_2 are material constants.

Coefficients D_1 and D_2 must be obtained by specific experimental tests. Some examples are listed in Table 8.1.

8.3 Ductile fracture for $T > 0.4\text{--}0.5T_m$

When the temperature is greater than $0.4\text{--}0.5T_m$ and the strain rate is comparatively large (in the range of $5 \times 10^{-4}\text{--}10^2 \text{ s}^{-1}$), plastic deformation takes place by high-stress creep (dislocation creep). Fig. 4.16 schematizes the typical hot true stress-true strain curve occurring with and without DRX. In the figure, the contribution to softening exerted by the nucleation and growth of voids during tensile plastic deformation is also highlighted.

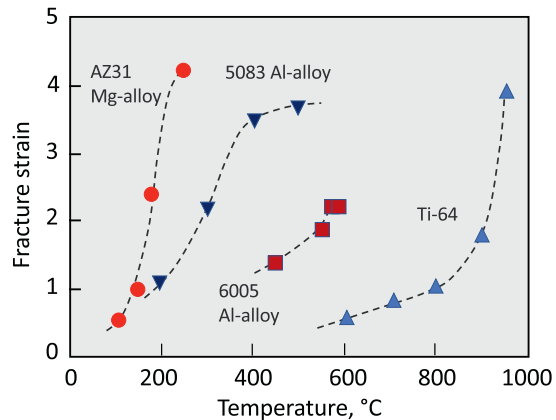
Table 8.1 Experimental materials constants to be used in Eq. (8.1).

Alloy	D_1	D_2	Reference
Weldox 460E steel	−0.015	0	Borvik et al. (2003)
AISI 4340	0.002	0.61	Guo and Yen (2004)
2024-T3	0.011	0	Lesuer (2000)
Ti-64	0.014	3.87	Lesuer (2000)

As seen in Section 4.5, in this temperature interval necking occurs early during plastic deformation, since thermal softening induces a reduction of the strain-hardening coefficient, n , that tends to 0 at the end of the primary creep region. Subsequently, very diffuse necking takes place because of the comparatively high m values. As usual, plastic damage is characterized by voids nucleation, growth, and coalescence. Voids are nucleated at microstructural particles (often located at the grain boundaries, especially when grain boundary sliding plays a role). At high temperature, nucleation is much more difficult than at RT, because DRV/DRX phenomena reduce the stress intensification at the particles that is required for debonding or cracking (Section 6.3). As predicted by Eq. (6.6), $\epsilon_{eff,n}$ is expected to markedly increase as both n and K decrease, as it occurs in this temperature interval. Therefore voids nucleate at comparatively large strains. To attain ductile fracture, voids must then grow and coalesce, but coalescence is made difficult by the high m values, which favor the spreading of plastic deformation away from the inter-voids regions. As seen in Fig. 4.11, coefficient m markedly increases with temperature and this is reflected in the corresponding increase in ductility displayed by metals in this temperature interval (if, of course, other phenomena do not intervene). Therefore above $0.4\text{--}0.5T_m$ a steep increase in ductility is generally observed due to the increased difficulty in voids nucleation and the subsequent difficulty in voids coalescence.

Fig. 8.6 collects some significant examples. For Mg alloys, ductility is very low at RT, as seen in the previous chapter. The true fracture strain is below 0.5 and fracture is typically within shear bands, because of the very low strain-hardening coefficient. However, as temperature is increased the situation changes. As anticipated in Section 2.5, plastic deformation involves more slip systems, and this facilitates plastic deformation. In addition, the strain rate sensitivity coefficient, m , is seen to increase from 0.004 at RT, to 0.04 at 100°C, 0.09 at 200°C, and 0.14 at 300°C (Rodríguez et al., 2016) with a corresponding marked increase in ductility. The increase in ductility is also accompanied by a change in the fracture mode. Shear fracture is observed at low temperatures, but at 300°C the typical cup-and-cone fracture morphology is observed. The two reported Al alloys also observe an increase in ductility with increasing

Fig. 8.6 Experimental variation of ductility with temperature (for temperatures above $0.45\text{--}0.5T_m$) for different non-ferrous alloys (data obtained from Barnett (2007) for Mg alloy, Clausen et al. (2004) for the 5083 Al alloy, Lassance et al. (2007) for the 6005 Al alloy, Semiatin et al. (1998) for Ti-64; the tests were carried out at strain rates in the range of $\sim 10^{-2}$ to 1 s^{-1}).



temperature. The 6005 Al alloy was submitted to a homogenization treatment for 2 h 30 min prior to testing (Lassance et al., 2007). With a longer homogenization treatment of 4 h 30 min, ductility increases two to three times. In the case of the Ti-64 alloy (with biphasic $\alpha + \gamma$ microstructure), the increase in ductility is moderate up to 800°C, as long as the fraction of β phase is lower than 0.2, but it is rather marked above 800°C with a rapid increase in the fraction of β , which is characterized by a greater deformability and higher m value (Valoppi et al., 2017). Of course, steels also observe an increase in ductility with temperature, but they often show a special behavior in specific temperature intervals (in particular when they assume a biphasic, $\alpha + \gamma$, microstructure), which is discussed in the next section.

As seen in Chapter 4, at comparatively high temperatures and low strain rates, DRX phenomena play an important role during plastic deformation. They also contribute to increasing ductility. In fact, the continuous formation of new dislocation-free grains removes any strain (and stress) concentration at the microstructural heterogeneities (comprising the particles), and strongly reduce the possibility of nucleation of voids. Therefore only a few voids nucleate and grow, and the fracture surfaces are hence characterized by the presence of few and big equiaxed dimples. An example is shown in Fig. 8.7, which shows the fracture surfaces of a low-alloyed ultrahigh strength steel tensile tested at 1000°C ($\sim 0.72T_m$), and at 0.1 and 1 s⁻¹. The initial diameter of the tensile specimen was 8 mm, and the final diameters were ~ 0.58 and ~ 0.69 mm, respectively. The corresponding fracture strains were then rather high, amounting to 5.23 and 4.88 at the two strain rates, showing that at high temperature and in the presence of DRX, ductility becomes extremely high. At the limit, DRX may induce necking to a point (no dimples at all), with a reduction of area at fracture (RA%) close to 100%.

The dependency of ductility with strain rate is rather important from a practical viewpoint, affecting the plastic formability of metals in hot conditions. The strain rates

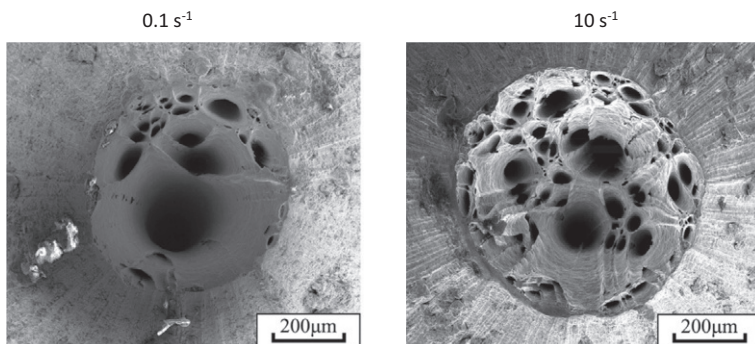


Fig. 8.7 Fracture surfaces of a low-alloyed ultra-high-strength steel tensile tested at 1000°C, and at two strain rates.

From Wen, D.X., Gao, C.X., Zheng, Z.Z., Wang, K., Xiong, Y.B., Wang, J.K., Li, J.J., 2021. Hot tensile behavior of a low-alloyed ultrahigh strength steel: fracture mechanism and physically-based constitutive model. *J. Mater. Res. Technol.* 13, 1684–1697, reproduced with permission of the publisher.

are typically in the range 5×10^{-4} to 10^2 s^{-1} (Fig. 4.17). However, it is very difficult to recognize general trends, because of the large number of parameters that are involved, including: DRV and DRX phenomena; possible phase transformations; additional heating due to the increased adiabaticity of the deformation process. Here just two cases are considered, as significant examples, in Fig. 8.8A. The first regards the steel considered in Fig. 8.7. In this case DRX is present during plastic deformation, and ductility is seen to decrease with strain rate. This can be attributed to the decreasing role of DRX as strain rate is increased. The second example regards the 6005 Al

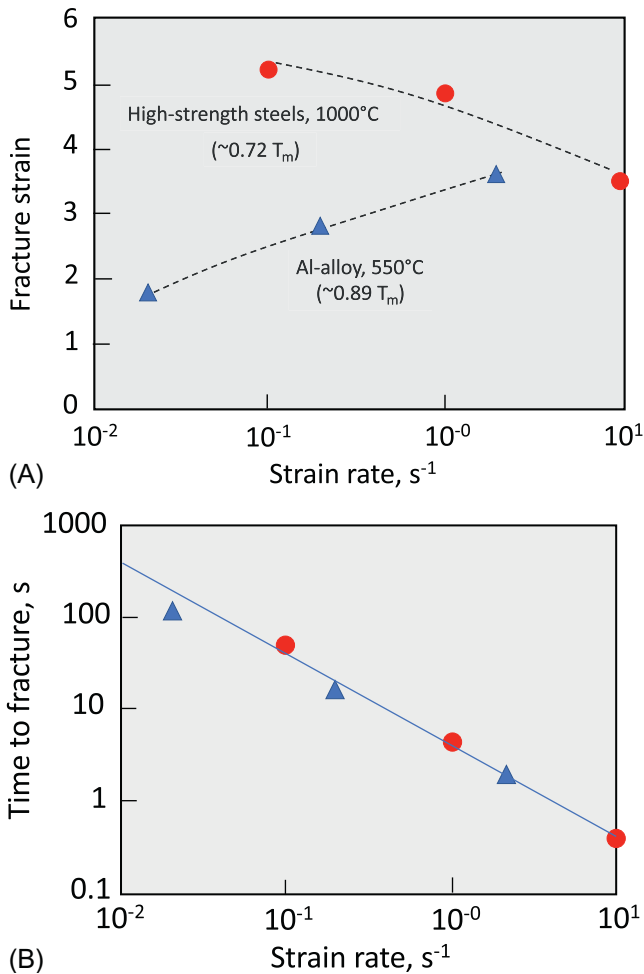


Fig. 8.8 (A) Evolution of fracture strain with strain rate for different alloys (High-strength steel: data from Wen et al. (2021); Al alloy: data from Lassance et al. (2007); (B) corresponding evolution of the time-to-fracture values.

alloy considered in Fig. 8.6. Despite the comparatively high test temperature (at which $m \approx 0.2$), no DRX was observed during plastic deformation, as rather typical in Al alloys characterized by a high SFE (Table 4.2). In this case, the positive effect of the high strain rate sensitivity coefficient, m , on ductility is further amplified with increasing strain rate, and the spreading of plastic deformation (with the corresponding reduction of strain localizations) increases in importance.

In Fig. 8.8B, the calculated *time-to-fracture* values, t_f , for the data of Fig. 8.8A are shown. This parameter is rather important for hot working operations, since the deformation process should be shorter than t_f to avoid any occurrence of ductile fracture in the workpiece. Considering the reported data, when strain rate is approximately 10^{-1} s^{-1} , t_f is in the range of 20–80 s; with a strain rate of 4 s^{-1} , t_f drops to $\sim 1 \text{ s}$.

The ductility data reported in this section refer to wrought metals, i.e., metals that were already submitted to a hot roughing operation, able to break down the as-cast microstructure. The as-cast metals may contain defects (such as voids, segregations, large precipitates, large columnar grains) that appreciably reduce ductility. In general, the ductility of metals with an as-cast structure is approximately half that of metals in the wrought condition. Nevertheless, as-cast metals should have a sufficient ductility to properly allow the continuous casting operations and the roughing operations (see next section and Chapter 10). As an example, Fig. 8.9 shows the hot ductility of an austenitic steel slab with a solidification microstructure (Hou and Zhu, 2013). The microstructure is made by columnar grains, with the presence of δ ferrite in the shell.

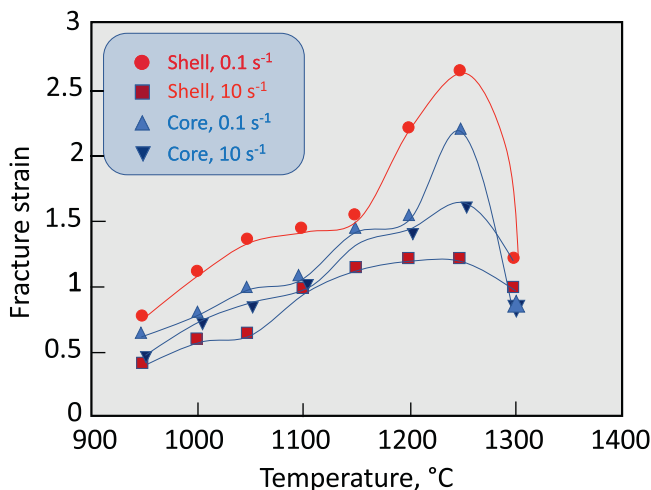


Fig. 8.9 Hot ductility of an austenitic steel slab (Cr15Mn9Cu2NiN); the tensile specimens were obtained parallel to the casting direction.

Redrawn from Hou, G.Q., Zhu, L., 2013. Influence of strain rate on hot ductility of austenitic stainless steel slab. *Mater. Sci. Technol.* 29, 568–572.

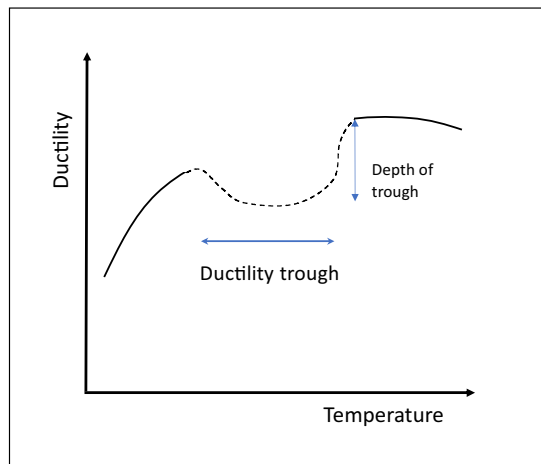
In addition, the grain size of austenite is larger in the core. It is seen that ductility increases with temperature up to a peak value at 1250°C. In addition, ductility is seen to decrease as the strain rate is increased, in agreement with the data in Fig. 8.8A. Such decrease is by ~20% in the shell and only ~5% in the core. In general, the ductility should be at least 0.5 to avoid formation of ductile cracks during continuous casting and roughing operations; the process parameters should be accordingly selected (see Chapter 10).

8.4 High-temperature ductility drop (or ductility trough)

For temperatures greater than $0.4\text{--}0.5T_m$ ductility increases up to a peak temperature value and then decreases. However, in this temperature interval special embrittling phenomena may induce a drop in ductility, as schematized in Fig. 8.10. The phenomenon is also known as *ductility trough*.

In **steels** this phenomenon is particularly important and is also called *red brittleness*, since it takes place in the temperature interval 600–900°C ($0.55\text{--}0.7T_m$) in which a reddish color is produced on the steel surfaces. This phenomenon plays a paramount role in different situations. One regards the straightening stage during continuous casting. In fact, the strains induced by the unbending of the strand (during the transition from vertical to horizontal motion) may induce formation of dangerous *transversal cracks* on the solidified product. This process occurs at strain rates around $10^{-3}\text{--}10^{-2}\text{ s}^{-1}$. Another one refers to the hot forging or rolling of ingots or billets/slabs. In this case, surface cracks may form under comparatively high strain rates, in the range $1\text{--}10^2\text{ s}^{-1}$. In the first case, the metals are deformed during solidification, and therefore with an *as-cast microstructure*. In the second case, they are deformed after cooling and re-heating. In general, the so-called *on-cooling* tests are carried out

Fig. 8.10 Schematization of the high-temperature embrittlement (ductility trough) phenomenon.



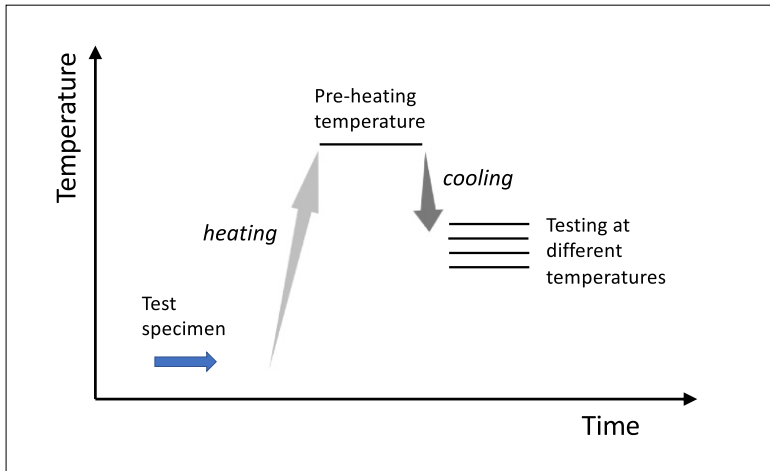


Fig. 8.11 Schematization of the thermal cycle typical of the on-cooling tests.

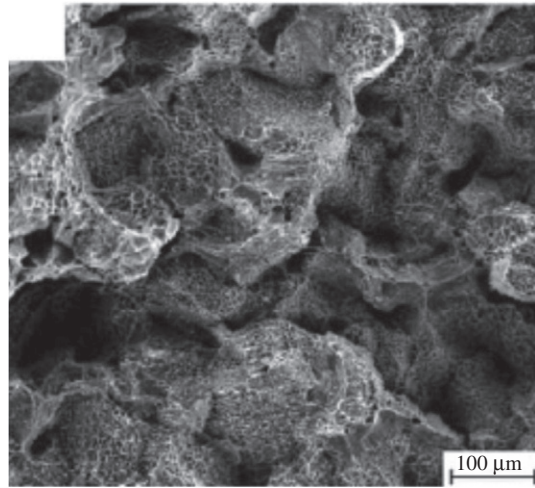
to investigate the ductility trough phenomenon (Fig. 8.11). The solidified ingots or billets/slabs are the starting products, from which the test specimens are machined. The specimens are then heated to high temperature, for example in the range 1050–1300°C in the case of steels. At this temperature, the microstructure of steels is fully austenitic. Subsequently, the specimens are cooled (with a cooling rate similar to that encountered in real conditions) to the testing temperature, and the tensile test is carried out with the selected strain rate.

At high temperature and depending on the chemical composition, steel microstructure is fully austenitic and DRX occurs easily during plastic deformation. As seen, ductility is rather high, also in metals with an as-cast microstructure (Fig. 8.9). As temperature is decreased, a decrease in ductility is observed, because DRX phenomena decrease in intensity, and also coefficient m decreases. A ductility trough may then be observed. In this case, the metal undergoes *intergranular dimpled fracture*. An example is shown in Fig. 8.12. It refers to a HSLA steel fractured at 710°C and $3 \times 10^{-3} \text{ s}^{-1}$ under tensile testing. The fracture morphology is characterized by very small dimples covering the austenite facets (that were present at the testing temperature). Fracture is ductile but ductility is low since plastic deformation is highly localized at the prior austenite grain boundaries where intense voids nucleation occurs and a low additional strain is required for voids growth and coalescence.

To understand the mechanisms of the ductility trough and under which conditions they occur, we have to first consider that in most steels (depending on chemical composition and cooling rate), part of the austenite transforms into ferrite on cooling. The transformation temperature is reduced with increasing the cooling rate but it is raised when cooling takes place together with plastic deformation. In this case, ferrite is called *deformation induced ferrite* (DIF) and it forms as a film at the grain boundaries.

Fig. 8.12 Fracture surface of an HSLA steel tensile tested at 710°C and $3 \times 10^{-3} \text{ s}^{-1}$.

From Großeiber, S., Ilie, S., Poletti, C., Harrer, B., Degischer, H.P., 2012. Influence of strain rate on hot ductility of a V-microalloyed steel slab. *Steel Res. Int.* 83, 445–455, reproduced with permission of the publisher.



Different metallurgical phenomena play an important role in favoring the ductility trough:

1. The previously mentioned formation of a ferrite film at the grain boundaries. Ferrite is weaker than austenite at high temperature, and plastic deformation tends to localize in the ferritic film (plastic strain localization favors voids nucleation, growth, and coalescence);
2. The presence of grain boundary sulfides. Some sulfides are present from the solidification stage. However, a certain amount of S dissolves into the austenitic matrix at high temperature and subsequently globular sulfides precipitate at the grain boundaries during the plastic deformation process (sulfides are responsible for the voids nucleation);
3. A decrease in the contribution of DRX and a decrease in m as temperature is decreased (this favors voids nucleation and coalescence);
4. The precipitation of intermetallic particles (or other second phase particles), whose kinetics are maximum in specific temperature intervals. For example, Nb(CN) particles precipitate fast (in Nb-based HSLA steels) when temperature is in the interval 700–950°C (sub-micrometric particles anticipate the drop at the right side of the trough since they retard DRX; in addition, following grain boundary precipitation, a precipitate-free region is formed in the regions close to the boundaries and in these regions plastic deformation can intensively localize).

If temperature is further decreased, the fraction of ferrite increases. When the ferrite fraction is greater than approximately 40%–50%, plastic deformation becomes homogeneous in the material, involving both ferritic and austenitic areas (Mintz, 2008). As a consequence, the left side of the critical temperature interval is reached, and ductility increases again (Fig. 8.10).

The metallurgical phenomena responsible for the ductility trough are often inter-related and several parameters play an important role. First, we consider the chemical composition of the sulfides. If the ratio of the Mn to S content is comparatively low (this ratio is also indicated with: *Mn/S-ratio*), low-melting point *Fe-rich Mn sulfides* form upon solidification. This would strongly penalize hot ductility. In fact, Fe-rich sulfides melt (or greatly soften) in the critical temperature interval and behave as

pre-existing voids. They may also induce a macroscopic brittle behavior, with RA values close to 0. Therefore steels are typically produced with a sufficiently high Mn/S ratio to allow formation of *Fe-poor Mn sulfides* (with higher melting point) and avoid this phenomenon. The critical ratio depends on the S content and it is about 20 when S is lower than 0.02% and 3.5 in re-sulfurized steels, when S is close to 0.3% (Turkdogan, 1996).

In the proper Mn/S conditions, the temperature extension and the depth of the ductility through depend markedly on the volume fraction of Mn sulfides at the grain boundaries. This in turn depends on both the S and Mn contents, and makes the picture rather complicated. During the permanence at a high pre-heating temperature before cooling to the plastic deformation temperature, a low fraction of S dissolves in the austenitic matrix if Mn is high, and vice-versa. For example, in steels with 1.4% Mn, the solubility of S in austenite is in the range 0.001%–0.002%, whereas in steels with 0.5% Mn, the solubility is higher, in the range 0.006%–0.007% (note, in this respect, that in Ca-treated steels, Ca further reduces the solubility of S in austenite at high temperature). After cooling to the testing (or process) temperature, S forms globular Mn sulfides at the grain boundaries, and the fraction of such sulfides depends on the amount of S solubilized in austenite. Therefore, it increases, for a given total S content, with a decrease in Mn. Ductility then decreases with increasing the S content and decreasing the Mn content, i.e., with decreasing the Mn/S ratio.

Figs. 8.13 and 8.14 show some representative examples. Fig. 8.13 refers to clean steels with low S content. Fig. 8.13A shows that the intensity of the trough depends clearly on the S content. Fig. 8.13B shows the influence of the austenite grain size (at the testing temperature) for a 0.19%C steel (Mintz, 2008). With increasing the grain size, the intensity of the ductility through increases. In fact, for a given volume fraction of sulfides (related to the available S), the higher the grain size, the lower the surface area available for the sulfides to precipitate, producing a higher density of sulfides at the grain boundaries, and hence a lower particle spacing; this in turn produces an increased density of voids and an increased facility for their coalescence.

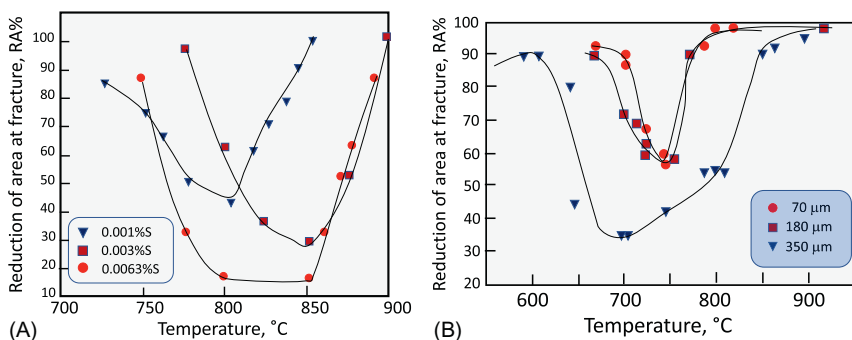


Fig. 8.13 (A) Influence of S content on the intensity of the ductility trough in some clean steels, with an austenitic grain size of $\sim 400 \mu\text{m}$ and tested at 10^{-3} s^{-1} ; (B) influence of grain size on the red ductility phenomenon for a 0.19%C steel.

Data from Mintz, B., 2008. Understanding the low temperature end of the hot ductility trough in steels. Mater. Sci. Technol. 24, 112–120, reproduced with permission of the publisher.

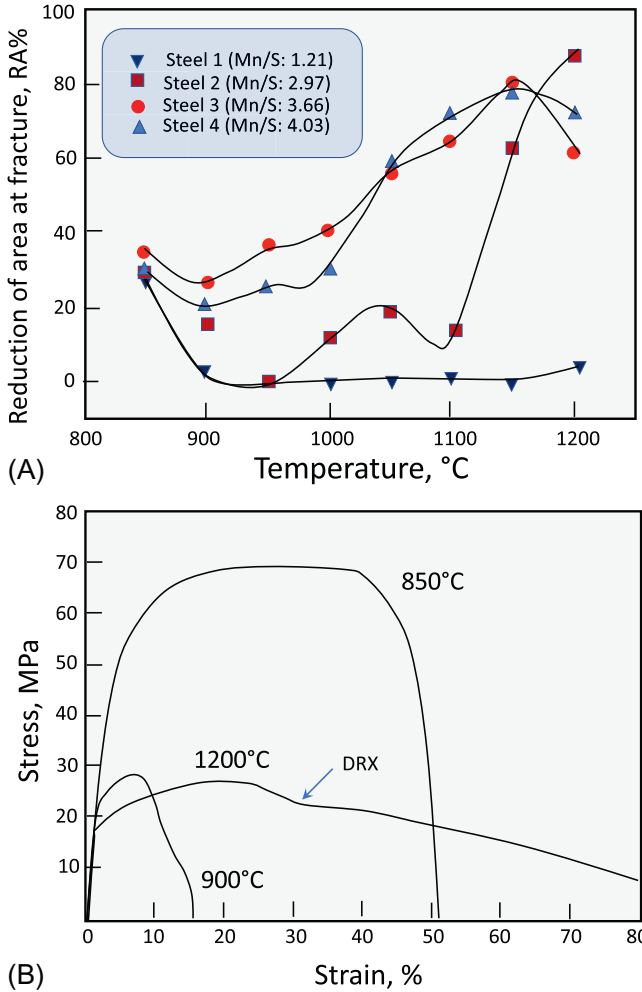


Fig. 8.14 (A) Ductility of resulfurized steels containing 0.31–0.35%S (and some Bi) and with different Mn/S ratios. (B) Hot tensile stress-strain curves of steel indicated with 2 in (A). Redrawn from Liu, H., Yu, Y., Chen, W., Wang, Q., Wang, G., 2013. Effect of Mn/S ratio on the hot ductility of eco-friendly bi-S based free cutting steel. *High Temp. Mater. Processes* 33, 553–561.

Fig. 8.14A refers to resulfurized steels, containing 0.31–0.35%S. These steels (indicated with: 1, 2, 3, and 4 in the figure) also contain some bismuth (0.1%–0.17%) that forms particles that further reduce ductility (and favor machinability; Bi is proposed in place of Pb, which poses health issues) and a low C content. For the two steels indicated with 1 and 2, the Mn/S ratio is below the critical value (3.5) and the corresponding ductility drop is very marked. For steel 1 ductility approaches 0, and fracture is macroscopically brittle. For steels 3 and 4, Mn/S is above the critical value, and the ductility drop is

limited (Steel 4 shows a higher drop because of a higher Bi content). Fig. 8.14B shows the stress-strain curves of steel 2 obtained at different temperatures. The curves at temperatures in the range 950–1100°C show an abrupt drop after the occurrence of necking (with a corresponding ductility drop) because of the intense nucleation and growth of voids, whereas the curves at the highest temperatures (1100–1200°C) display a high ductility because of the occurrence of DRX (indicated by the arrow).

An increase in the strain rate generally reduces the intensity of the ductility drop, anticipating the recovery of ductility at the left side of the trough (i.e., at the low-temperature side). In fact, at these temperatures, no DRX takes place, and an increase in the strain rate results in a better distribution of plastic deformation, decreasing the intensity of its localization in ferrite.

The ductility trough may play an important role also during hot rough forging or rolling of ingots or billets, even if these operations occur at rather high strain rates, on the order of 10 s^{-1} (Chapter 10). As an example, Fig. 8.15 shows the effect of S on the ductility trough of different mild steels containing 0.14–0.16%Mn, tensile tested at 22 s^{-1} (Kizu and Urabe, 2009). The steels were reheated to 1300°C and cooled to the testing temperature with a velocity of 30°C/s. Given the comparatively low Mn content (Mn was kept low to decrease the hot strength of the steels), the ductility loss is rather significant even when the S content is rather low, since the Mn/S ratio decreases from 17.5 to 10 and it is then always below 20.

The ductility trough phenomenon is also influenced by the precipitation of grain boundaries nitrides or carbonitrides, such as AlN, NbCN, and TiCN. Therefore it is quite typical in micro-alloyed HSLA steels, and in steels deoxidized with Al and containing N. The precipitation kinetic of these intermetallic phases is maximum in the interval 700–950°C, and plastic deformation can accelerate the precipitation. The phenomenon can be partly reduced in intensity by controlling the chemical composition of the steel, reducing the Al and N contents (however, this task is difficult to attain,

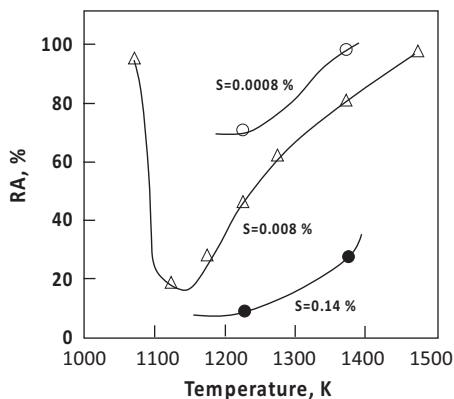


Fig. 8.15 Effect of S on the ductility trough of different low-Mn (0.14%–0.16%) mild steels, tensile tested at 22 s^{-1} .

Redrawn from Kizu, T., Urabe, T., 2009. Hot ductility of sulfur-containing low manganese mild steels at high strain rate. *ISIJ Int.* 49, 1424–1431.

since N is often added to increase the steel strength and some residual Al is useful for a proper desulfurization stage). A strategy adopted to avoid formation of transversal cracks in continuous casting is to assure the strand unbending takes place in a temperature interval that is above or below the ductility trough.

The phenomenon of the ductility trough is paramount in steels but it is not restricted to steels. It is observed in other alloys as well, including Al, Cu, and Ti alloys.

In case of **Al alloys**, two important situations can be recognized. The first concerns the Al—Mg alloys (class: 5000), and the second the Al—Mg—Si alloys (class: 6000). [Fig. 8.16A](#) shows the ductility trough in an Al—Mg alloy at a temperature interval around $0.7T_m$, due to the presence of hydrogen (and the consequent pores formation at grain boundaries during high-temperature exposure, as seen in [Section 7.3](#)), and residual Na. In fact, Na segregates at the grain boundaries (with a kinetic that is maximum at around 300°C) and favors the local nucleation and growth of voids and pores, and then the ductile intergranular fracture ([Okada and Kanno, 1997](#)).

In Al—Mg—Si alloys, ductility displays a drop at a higher temperature for a different reason. An example is shown in [Fig. 8.16B](#), which shows the hot ductility of a 6060 Al alloy ([Lassance et al., 2007](#)). A marked drop in ductility at around 570–580°C is observed. In these alloys, Mg forms different intermetallics, such as Mg_2Si , that are present in the microstructure together with $\beta-Al_5FeSi$ and other intermetallics. If the homogenization treatment is insufficient, a eutectic reaction takes place at high temperature, involving $\beta-Al_5FeSi$, Mg_2Si precipitates, and Al, forming a liquid film at the grain boundaries. The presence of such liquid regions is similar to the presence of large voids from the beginning of the deformation, and markedly reduces ductility. This phenomenon plays an important role in hot working operations, such as in hot rolling and extrusion.

Also, Cu and **Cu alloys** display a ductility trough in the temperature interval 300–600°C (0.4 – $0.7T_m$). The behavior is actually typical for these alloys. For lower temperatures, below 0.4 – $0.5T_m$, ductility is rather high, as seen in [Fig. 8.1B](#). But above this temperature interval, ductility does not increase further, as typically observed in all other metals, but displays a severe drop. An example is shown in [Fig. 8.17](#) for an industrial CuSn6 alloy with different grain size. Up to $\sim 100^\circ\text{C}$, ductility remains high (compare with [Fig. 8.1B](#), remembering that the RA% is always higher than the corresponding EL%). The ductility drop is then rather sudden and the intensity of the trough is seen to increase with increasing grain size ([Ozgowicz, 2005](#)). In addition, the intensity of the trough is reported to increase with decreasing the strain rate ([Laporte and Mortensen, 2013](#)). The comparison of these results with those obtained for steels indicates that the trough is most probably caused by an embrittlement of the grain boundaries. S has a very low solubility in Cu and several investigations have shown that its grain boundary segregation is responsible for the trough. It has also been observed that in the critical temperature interval the specimens fail by mixed intergranular dimpled fracture and intergranular brittle fracture. In the last case, segregations reduce the grain boundary cohesion and favor brittle fracture under the action of grain boundary sliding (GBS). As seen in [Section 8.5](#), GBS is typically observed in the low stress creep regime, but in Cu alloys it is also observed at the temperatures and strain rates that generally pertain to the dislocation creep regime (Laporte and

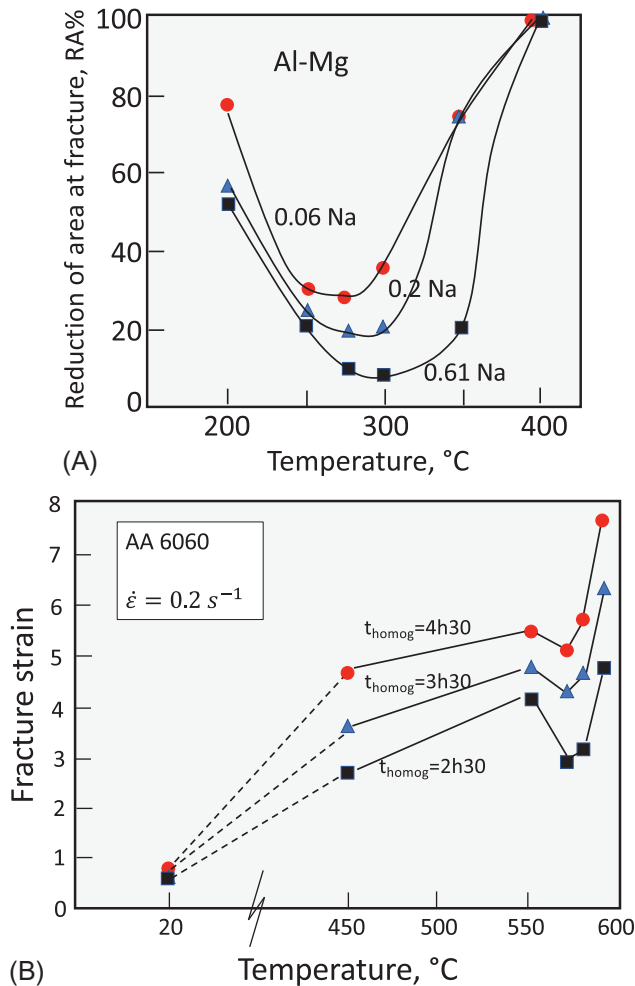


Fig. 8.16 (A) Effect of sodium on the ductility of an Al—Mg alloy (also containing residual hydrogen); (B) true fracture strain vs. testing temperature for a 6060 Al alloy homogenized at 585°C and different times.

(A) Redrawn from Okada, H., Kanno, M., 1997. Hot ductility of Al-Mg and Al-Mg-Y alloys impaired by trace sodium. *Scr. Mater.* 37, 781–786, reproduced with permission of the publisher; (B) Redrawn from Lassance, D., Fabregue, D., Delannay, F., Pardoën, T., 2007. Micromechanics of room and high temperature fracture in 6xxx Al alloys. *Prog. Mater. Sci.* 52, 62–129, reproduced with permission of the publisher.

Mortensen, 2009). This is demonstrated, for example, in Fig. 8.18, which shows an intergranular wedge crack formed by GBS in a CuSnP alloy deformed at 400°C and at $3.1 \times 10^{-5} \text{ s}^{-1}$ (Ozgowicz, 2005). In other words, when entering the creep region, i.e., for temperatures in excess of $0.4\text{--}0.5T_m$, GBS influences the deformation behavior of the Cu alloys and becomes so intense as to dramatically anticipate the

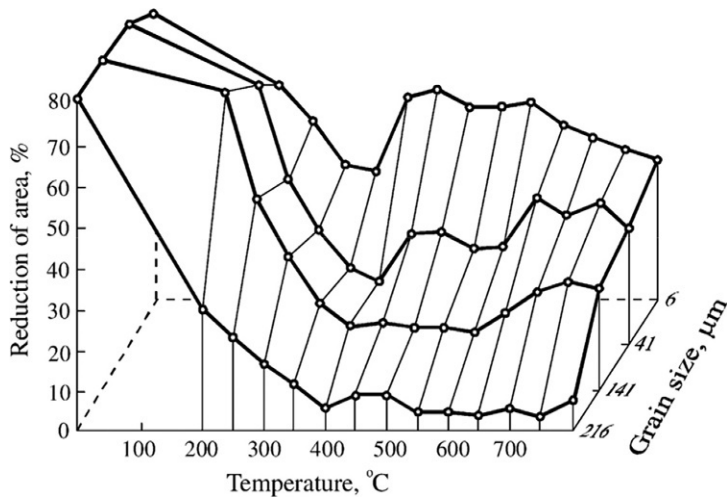


Fig. 8.17 Ductility variation of an industrial CuSn6 alloy with different grain size. From Ozgowicz, W., 2005. The relationship between hot ductility and intergranular fracture in a CuSn6P alloy at elevated temperatures. *J. Mater. Process. Technol.* 162–163, 392–401, reproduced with permission of the publisher.

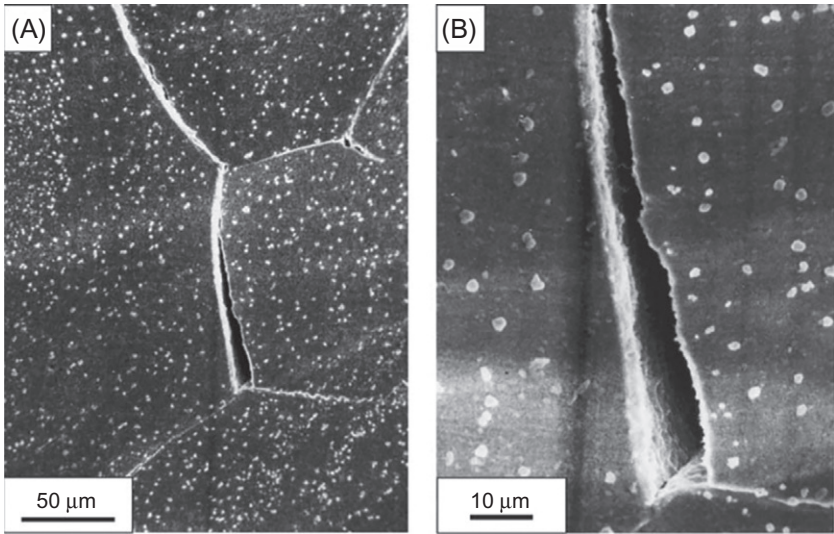


Fig. 8.18 Intergranular wedge crack formed by GBS in a CuSnP alloy deformed at 400°C and at $3.1 \times 10^{-5} \text{ s}^{-1}$. From Ozgowicz, W., 2005. The relationship between hot ductility and intergranular fracture in a CuSn6P alloy at elevated temperatures. *J. Mater. Process. Technol.* 162–163, 392–401, reproduced with permission of the publisher.

fracture conditions by intergranular fracture. As shown by Fig. 8.17, only for temperatures greater than $\sim 700^\circ\text{C}$ and at comparatively low grain size and high strain rates, the ductility of Cu alloys returns to be adequately high.

At the end of this section, we consider the case of **Ti alloys**, focusing on the ($\alpha + \beta$) Ti-64 alloy. As shown in Fig. 8.6, ductility increases very steeply in passing from 900°C to $950/1000^\circ\text{C}$. In fact, ductility is high when the alloy is in the β -region, i.e., above β -transus (T_β), and displays a drop for temperatures lower than T_β , i.e., with increasing the fraction of α phase in the microstructure (this drop in ductility is actually observed in other $\alpha + \beta$ alloys, and also in α alloys). In some cases, the ductility drop is particularly intense and is then followed by a recovery in ductility for temperatures in the interval of $600\text{--}700^\circ\text{C}$. This means that, in specific cases, a ductility trough can be identified in the temperature range $700\text{--}980^\circ\text{C}$, which corresponds to an interval from T_β and a temperature $\sim 250^\circ\text{C}$ lower than T_β (Suzuki and Eylon, 1998). It has been shown that the grain boundary segregation of elements such as S could be responsible for the very intense ductility drop below T_β . A role is also played by the cooling rate from the β region, which determines the microstructural characteristics of the alloy.

8.5 Peak ductility temperature

As already seen in Fig. 8.9, at very high temperatures (close to $\sim 0.85T_m$), ductility reaches a maximum and subsequently declines. The temperature at which ductility is maximum is called *peak ductility temperature* (PDT). The subsequent ductility drop is called *hot shortness*. Additional examples are shown in Fig. 8.19.

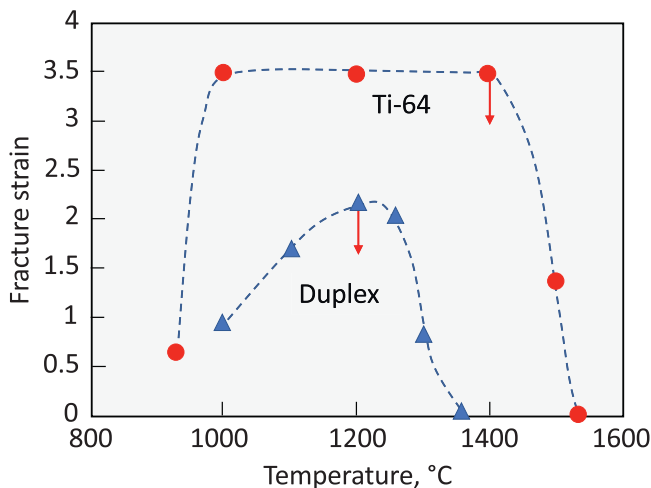


Fig. 8.19 Hot shortness phenomenon (and corresponding PDT, indicated with the arrows) for Ti-64 and a duplex stainless steel (data from Suzuki and Eylon for Ti-64; Fan et al. (2009) for the duplex steel).

Hot shortness (also called *liquid metal embrittlement*) is related to the *liquation* phenomenon, which involves the formation of thin liquid films at the grain boundaries. Such films act as large voids for ductile fracture. The formation of this liquid is due to the grain boundary segregation of impurities (such as: S, P, Sb, Cu, Pb, Sn, in steels) and alloying elements. This segregation decreases the melting temperature of the alloy close to the grain boundaries, and this favors the early local melting. The phenomenon is important in the case of as-cast and wrought microstructures. It is clearly more severe in the case of as-cast microstructures, involving the micro-segregations phenomena in the intercolumnar grains.

A special kind of hot shortness involves the surface of metals, called *surface hot shortness*. It originates from surface oxidation phenomena. In steels, a special role is played by residual Cu. Cu enters the steel charge of the electric arc furnace with the scrap. It is not eliminated from the melt during the subsequent steelmaking process, and hence it remains in the solidification products. During the stay at high temperature in an oxidizing atmosphere (e.g., before hot forging or hot rolling), the surface of the steel billets or ingots is oxidized, and Fe-oxides are formed. Oxides also form in the subsurface regions because of the oxygen diffusion along the grain boundaries. Cu does not oxidize and Cu atoms migrate towards the metal/Fe-oxide interface, since they have a very limited solubility in Fe-oxides. This causes the formation of a thin layer of Cu at the oxide-matrix interface, that can easily soften and melt, since the melting point of Cu is 1085°C, and it decreases to values as low as 900°C if elements like Sn are present (Garza and Van Tyne, 2006). Therefore Cu (in a content as low as 0.15%–0.2%) induces a decrease in the ductility close to the surface of the billet, i.e., right in the region where the largest tensile strains are typically present during processing operations.

In Chapter 11 it is also shown that the hot upsetting (compression) test is particularly suited to reveal the attitude of steels towards surface hot shortness.

To summarize:

- For temperatures lower than $0.4\text{--}0.5T_m$, ductility is mainly dependent on the evolution of the strain-hardening coefficient, n , with temperature (Fig. 3.14).
- In particular, for fcc alloys (such as Al, Cu), ductility increases as temperature is decreased below RT (this opens to the possibility of plastic forming at cryogenic temperatures).
- In this temperature range, a ductility drop is observed in alloys that display the DSA phenomenon.
- For temperatures higher than $0.4\text{--}0.5T_m$, ductility is mainly dependent on the strain rate coefficient, m , and then increases with temperature.
- In this temperature range, a ductility drop (also called ductility trough) is observed in some alloys (including steels, and Al, Cu, and Ti alloys).
- At very high temperatures, above $\sim 0.85T_m$, another ductility drop is observed, called hot shortness.
- In Fig. 8.20, the main phenomena are summarized as a function of the homologous temperature.

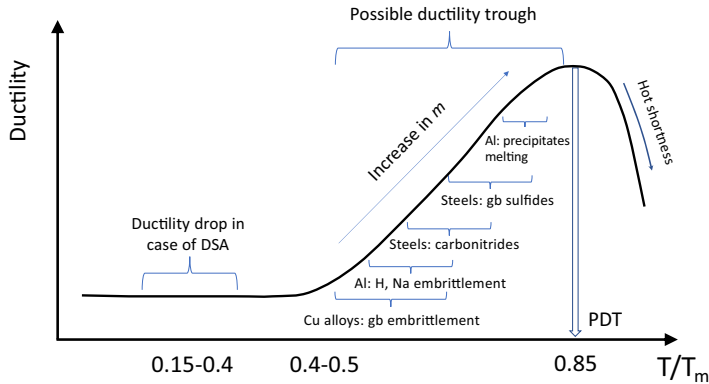


Fig. 8.20 Ductile evolution of metals with homologous temperature.

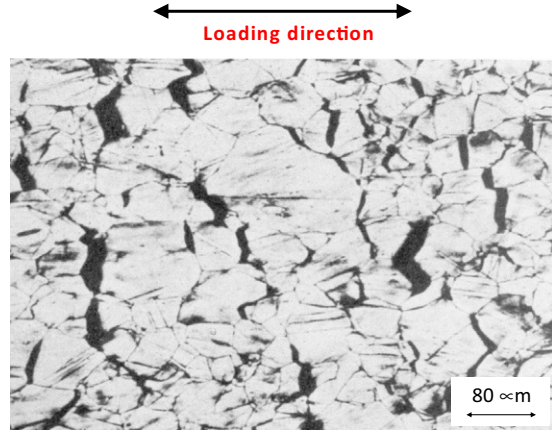
8.6 Superplasticity

When temperature is greater than $0.4\text{--}0.5T_m$ but the applied stress is low (typically: $\sigma/E < 10^{-3} - 10^{-4}$), plastic deformation take place by GBS and diffusion creep, as explained in Section 4.6 (and Fig. 4.17). Correspondingly, the strain rate is very low, in the range $10^{-10}\text{--}10^{-6} \text{ s}^{-1}$, and the time to fracture is extremely long, even if the involved plastic strains are small. For example, the design life of structural parts of a power station, operating in these creep conditions, is approximately 250,000h.

The final fracture is ductile, but typically with an intergranular path. A peculiarity of *intergranular dimpled ductile fracture* is related to voids nucleation, which involves particles (such as inclusions or precipitates) located at the grain boundaries. The presence of GBS may activate other voids nucleation mechanisms (Jiang et al., 1994). In particular, GBS may induce stress concentrations at grain boundary irregularities (such as the particles, or grain boundary ledges), which in turn force vacancies to cluster and form voids (Raj and Ashby, 1975). Dislocation pile-ups at the grain boundaries can help in this process. Another peculiarity is the possible formation of *wedge-cracks* at grain boundaries, in particular at triple junctions, due to GBS. An example is shown in Fig. 8.18 for a CuSnP alloy; another example is displayed in Fig. 8.21, which represents typical *wedge-cracks* on the damaged surface of an austenitic stainless steel (Lai and Wickens, 1979).

The subsequent voids growth may be due to different mechanisms. One is related to the plastic deformation by dislocation motion (by slip and, most importantly, by climb and cross-slip). Another one is related to the diffusion of vacancies into the nucleated voids. As a result, the duration of the fracture process is rather long, and the ductile fracture path is intergranular and typically normal to the loading direction. Specific embrittling phenomena (occurring, for instance, during service), such as the grain boundary precipitation of embrittling phases or the segregation of impurity atoms (such as S), may induce anticipated intergranular fracture and markedly reduce the time to fracture.

Fig. 8.21 Damaged surface of a sample made of austenitic stainless steel subjected to intergranular ductile fracture. From Lai, J.K.L., Wickens, A., 1979. Microstructural changes and variations in creep ductility of 3 casts of type 316 stainless steel. *Acta Metall.* 27, 217–230, reproduced with permission of the publisher.



In the case of low stress creep, the m values are rather high, theoretically ranging from 0.5 (deformation by GBS) and 1 (diffusion creep). As seen, ductility markedly increases with m , since high m values reflect a high diffuse necking capability and, in the case of tensile deformation of sheet specimens, the suppression (or delay) of localized necking. High m values also delay the conditions for voids coalescence (Langdon, 1982). A large number of experimental studies involving different metals have reported values of the elongation at fracture in excess of 400% when the strain rate sensitivity coefficient, m , is greater than 0.3 and the deformation mechanism is dominated by GBS. In these cases, the metal is said to have a superplastic behavior, and the corresponding phenomenon is called *superplasticity*.

As seen, the involved strain rates in this creep regime are typically very low, and therefore the time required to reach a given strain can be extremely long. The question is: may superplasticity be a useful phenomenon in plastic forming processes?

The answer is positive, but only if the materials microstructure is properly modified to raise the strain rates to sufficiently high values. Eq. (4.15) indicates the way to increase $\dot{\epsilon}$, while remaining in the diffusion creep/GBS deformation mechanism region, which is required to have high m values and then a superplastic behavior: a decrease in the initial grain size, d . Hence, to have a reasonable strain rate (in the range of 10^{-3} – 10^{-2} s^{-1} with corresponding forming times on the order of 20–30 min) an average grain size of less than approximately $10 \text{ }\mu\text{m}$ is required. To obtain alloys with a very small grain size, different methods are available. Generally, a thermo-mechanical treatment is adopted. First, the metal is subjected to severe plastic deformation (such as heavy cold rolling) to obtain products, such as sheets, with a high density of dislocations, also forming sub-grain boundaries. Then, the metal is properly annealed to have static recrystallization without grain growth, and produce very small and equiaxed grains. Other methods comprise powder metallurgy and severe plastic deformation by equal-channel angular pressing, ECAP. They are proposed and optimized with the aim of further decreasing the grain size and then possibly increasing the superplastic strain rate in order to decrease the forming time (Charit and Mishra, 2007).

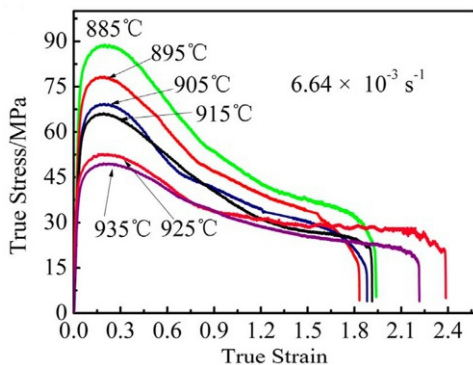


Fig. 8.22 True stress-strain curves of sheet specimens made with a near- α Ti-alloy (Ti-55) with initial grain size of $\sim 2.15 \mu\text{m}$ (Liu et al., 2017).

Different alloys with properly small grain sizes display a superplastic behavior. Up to now, most of the research has been focused on specific Al and Ti alloys, and some of these alloys are currently used in the superplastic sheet forming of different complex components, mainly in aerospace and building design. The Al alloys typically contain elements such as Sc and Zr that form nano-sized dispersoids. They are beneficial, as they retard the grain growth during annealing and also during dynamic recrystallization, which may occur during plastic deformation. As far as Ti alloys are concerned, the widely used Ti-64 alloy is biphasic ($\alpha + \beta$) and this is also beneficial in avoiding grain growth. In any case, the microstructure must be optimized in order to inhibit, or reduce, grain growth during the high-temperature deformation.

As an example, Fig. 8.22 shows the true stress-strain curves of sheet specimens made with a near- α Ti alloy (Ti-55) with initial grain size of $\sim 2.15 \mu\text{m}$ (Liu et al., 2017). The tests were conducted at a strain rate of $6.64 \times 10^{-3} \text{ s}^{-1}$ and at different temperatures, from 885 to 935°C. By comparing with Fig. 4.16, it is observed that thermal softening by DRX occurred after reaching the peak stress. However, the observed increase in the grain size with increasing strain was rather limited (from ~ 2.15 to $\sim 3 \mu\text{m}$) and the grain size remained below $10 \mu\text{m}$. It is also seen that at 925°C, the true fracture strain is close to 2.4, with a corresponding elongation at fracture greater than 900%. It is reported that the strain rate sensitivity index is equal to 0.36. During plastic deformation, and after the initial diffuse necking, the material was able to delay the onset of localized necking that is typical in the case of tensile deformation of sheet specimens (Section 2.8); this allowed for obtaining the observed superplastic behavior. To summarize, in the case of negative m values, shear localization (with the formation of shear bands) is anticipated (Section 3.6) and ductility is low; in the case of positive and high m values, localized necking is delayed, and ductility is high.

The experimental data show that superplastic behavior occurs in specific, and inter-related, temperature and strain rate intervals. This is schematically shown in Fig. 8.23A (temperature interval, fixed strain rate) and Fig. 8.23B (strain rate interval, fixed temperature).

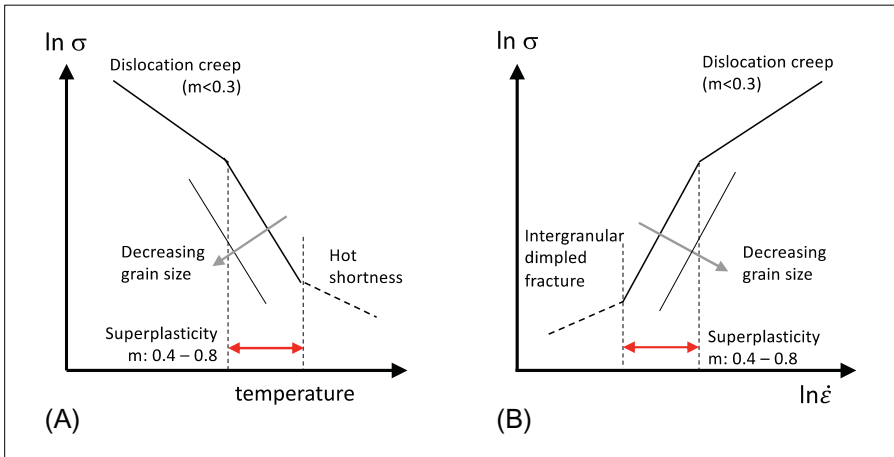


Fig. 8.23 Representation of stress versus temperature (A) and strain rate (B), showing superplasticity at both the intermediate regions.

Consider first the *temperature interval*. As seen, temperature must be in excess of $0.4-0.5T_m$. If temperature is too low, the tensile plastic deformation behavior is that typical of dislocation (high-stress) creep. Hence $m < 0.3$ and no superplastic behavior is attained. Increasing temperature, steady-state (secondary) creep is reached after primary creep, and plastic deformation is dominated by GBS (accompanied by diffusion and dislocation-based local strain accommodation phenomena and, possibly, DRX phenomena) and m is high ($m \approx 0.5$). This is the superplastic region, during which plastic deformation ends when ductile damaging phenomena (comprising grain boundary voids nucleation, voids growth, and intergranular dimpled fracture) take place. But if the temperature is too high, intense DRX and grain growth may occur (depending on the microstructure stability, i.e., on its ability to inhibit or slow down DRX and grain growth) and this causes the disappearance of superplasticity. In addition, hot shortness may occur, with the melting at the grain boundaries due to the presence of segregations, and the easy and premature formation of voids. In any case, fracture is facilitated and the elongation at fracture decreases. Fig. 8.24 shows the elongation at fracture of different Al alloys with a small grain size as a function of temperature (Wang et al., 2018). In all cases, ductility increases with temperature and a superplastic behavior, with elongations in excess of 400%, is displayed when temperature is sufficiently high. However, a drop in ductility is observed when temperature becomes too high.

Consider now the *strain rate interval* (Fig. 8.23B). If the metal is deformed with a too-high strain rate, the tensile plastic deformation behavior is that typical of dislocation (high-stress) creep. As a consequence, $m < 0.3$ and no superplastic behavior is observed. The situation is very similar to that encountered when deformation temperature is too low. By decreasing strain rate, steady-state creep is attained with a plastic deformation dominated by GBS; a superplastic behavior is then obtained ($m \approx 0.5$).

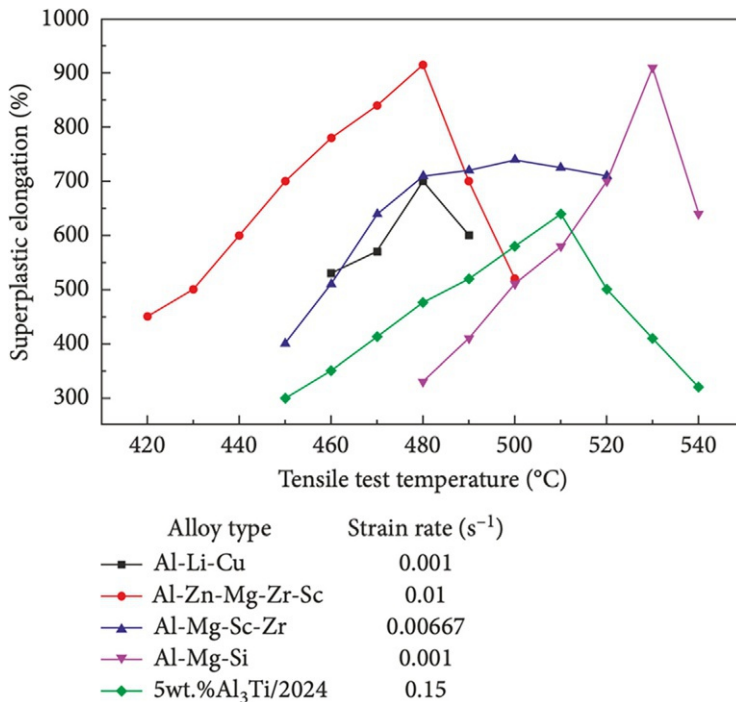


Fig. 8.24 Tensile elongations at fracture of different Al alloys, at different temperatures and for given strain rates (Wang et al., 2018).

But if the strain rate is too low, there is enough time for the occurrence of DRX and grain growth and the prevailing of grain boundary diffusion. Together with the possible presence of impurities at the grain boundaries, grain boundary diffusion favors voids nucleation and inhibits GBS, leading the premature intergranular dimpled ductile fracture (Kawasaki and Langdon, 2016). As an example, Fig. 8.25 shows samples of an Al-3Mg-0.2Sc alloy after tensile test, up to fracture, at 400°C and at different strain rates (Horita et al., 2000). The alloy was produced with an ECA process and had a very small grain size, $\sim 0.2 \mu m$. At a strain rate slightly above $10^{-2} s^{-1}$, m is ~ 0.5 , and correspondingly the observed tensile elongation at fracture is maximum, and in particular in excess of 2200%.

In the schematizations of Fig. 8.23, the role of the grain size is also specified. A decreasing grain size is beneficial for superplasticity, since it facilitates deformation by GBS, and then a lower stress is required for plastic deformation at a given temperature and/or at a given strain rate.

In Table 8.2, the temperature and strain rate intervals for superplastic behavior of selected Al and Ti alloys are listed (data collected from Wang et al., 2018, and Leyens and Peters, 2003).

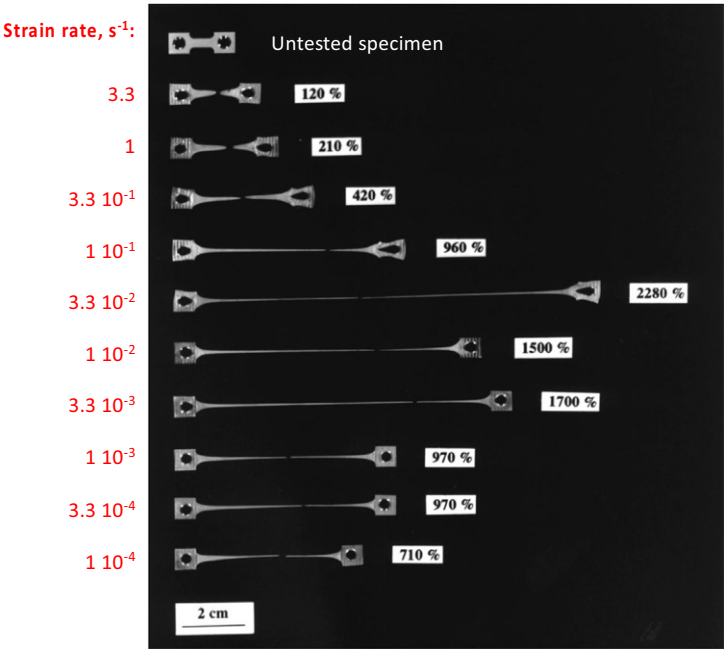


Fig. 8.25 Samples of an Al-3Mg-0.2Sc alloy after tensile test, up to fracture, at 400°C and at different strain rates.

From Horita, Z., Furikawa, M., Nemoto, M., Barnes, A.J., Langdon, T.G., 2000. Superplastic forming at high strain rates after severe plastic deformation. *Acta Mater.* 48, 3633–3640, reproduced with permission of the publisher.

Table 8.2 Temperature and strain intervals for superplastic behavior of selected Al and Ti alloys.

Alloy	Temperature interval (°C)	Strain rate interval (s^{-1})	m (average)
Al-Mg-Sc-Zr	450–525	$1.67 \times 10^{-3} - 10^{-1}$	0.39
Al-Zn-Mg-Sc	450–500	$10^{-3} - 10^{-1}$	0.37
Al-Zn-Mg-Sc	450–500	$10^{-3} - 10^{-2}$	0.5
Al-Mg-Li	430–570	$5 \times 10^{-4} - 10^{-3}$	0.61
Ti-6Al-4V	840–900	$1.3 \times 10^{-4} - 10^{-3}$	0.75
Ti-13V-11Cr-3Mo or B120VCA	800	2×10^{-4}	0.5

Data compiled from Wang, X.-G., Li, Q.-S., Wu, R.-R., Zhang, X.Y., Ma, L., 2018. A review on superplastic formation behavior of Al alloys. *Adv. Mater. Sci. Eng.* Article ID 7606140; Leyens, C., Peters, M., 2003. *Titanium and Titanium alloys*, Wiley-VCH.

8.7 Damage and fracture in ASBs

As mentioned in Section 3.6, adiabatic shear bands (ASBs) may form in metals as a consequence of forced shear deformation in a narrow region and at high strain rates, i.e., in adiabatic conditions. In the ASB, the temperature rise is very significant (in the range of $0.5\text{--}1T_m$), and softening phenomena, such as DRX, can occur. ASB formation is favored in metals with a low thermal conductivity and a high strength. Ti alloys and high-strength steels (including high alloy stainless steels) are rather prone to the formation of ASBs. In general, ASBs (like fracture in shear banding phenomena) are generally detrimental in forming operations and in service performance of a structural part, since they clearly anticipate the attainment of the fracture conditions.

As an example, Fig. 8.26 shows the stress-strain curve of a Ti-64 alloy subjected to dynamic torsional loading at average strain rates of about $3.1 \times 10^3 \text{ s}^{-1}$ (Xu et al., 2008). Four interrupted tests were carried out, and they are labeled with: 1, 2, 3, and 4. Metallographic observations of the four specimens showed that plastic deformation was homogeneous up to a nominal strain of 0.16 (specimen 1), and shear

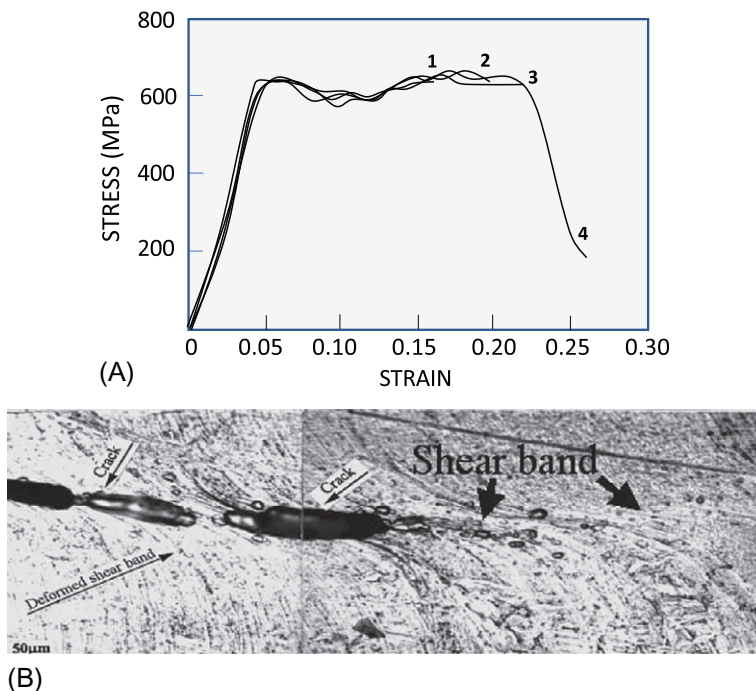


Fig. 8.26 (A) Stress-strain curve of a Ti-64 alloy subjected to dynamic torsional loading at average strain rates of about $3.1 \times 10^3 \text{ s}^{-1}$; (B) formation of voids that coalesced into a crack after point 4 in the stress-strain curve.

(A) Redrawn from and (B) From Xu, Y., Zhang, J., Bai, Y., Meyers, M.A., 2008. Shear localization in dynamic deformation: microstructural evolution. *Metall. Mater. Trans. A* 39, 811–843, reproduced with permission of the publisher.

localization was developed at a strain of 0.196 (specimen 2). At a strain of 0.219 (specimen 3), severe shear banding appeared, with the formation of voids inside the bands that coalesced into a crack. The load drop after point 3 was then due to the ductile damage (geometrical softening) associated with the nucleation, growth, and coalescence of voids within the ASB up to the final fracture.

These observations are general and show that ductile fracture takes place within the ASB when a critical strain is locally achieved, as usual in the ductile fracture process. Therefore the formation of an ASB is a very crucial phenomenon that is characterized by a large (shear) strain intensification; when such strain becomes critical, ductile fracture is triggered inside the ASB. The sensitivity of a metal to anticipated ductile fracture in ASBs, depends on its attitude to the formation of ASBs (discussed in [Section 3.6](#)) and on the facility of voids nucleation and growth (mainly related to the microstructural features responsible for voids nucleation).

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Impact testing and ductile-to-brittle fracture transition

9



Charpy V-notch specimens broken after impact testing.

Chapter outline

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The simple and common Charpy V-notch (CVN) impact test is based on the impact deformation and fracture of a notched specimen with a standard geometry. It is able to provide valuable information on the deformation and fracture behavior of metals. The basic information is the so-called fracture energy that is a quick evaluation of the ductility and fracture toughness of a metal at a given temperature (usually RT) and at a comparatively high strain rate ($\sim 10^3 \text{ s}^{-1}$). In addition, from the recorded impact load-deflection curve it is possible to obtain complete information on the plastic behavior of the material under study.

With a proper procedure, tests at different temperatures (below and above RT) can also be carried out, thus obtaining information on the plastic deformability of metals in conditions that can be close to those encountered in different plastic forming operations. Impact testing is also paramount for performance control of tool steels, to be used in the manufacturing of the forming tools.

Metals typically fracture in a ductile manner, but in special conditions they may fracture in a brittle manner. Four factors may favor a ductile-to-brittle fracture transition:

1. A decrease in testing temperature;
2. An increase in the strain rate;
3. An increase in the local stress triaxiality;
4. A specific microstructural embrittlement.

With impact testing it is possible to properly investigate the ductile-to-brittle transition of metals, revealing the role of microstructure on the fracture mechanisms.

The evaluation of the conditions leading to brittle fracture is very useful for different reasons. The most important one, for the objectives of this text, is to detect possible embrittling or microstructural degradation phenomena induced by the production and manufacturing processes. Such phenomena could impair the ductility of metals in view of subsequent manufacturing operations, or in view of their use in specific service conditions.

9.1 Charpy impact testing

In an impact test, a weighted pendulum strikes against a specimen with a given impact velocity and a given available energy, as schematized in Fig. 9.1A. The initial impact velocity, v , can be varied by changing the initial height of the mass (h_0) with respect to the specimen level. It is typically close to 5 m/s. The specimen contains a notch to favor the deformation and fracture processes. During the test, the pendulum breaks the specimen, and the amount of energy required to break it is recorded. It is called the fracture or impact energy and is indicated with E_f in the figure.

Different specimens can be used, characterized by different dimensions and notch geometry. In a *Charpy V-notch* (CVN) impact test, the specimen contains a notch with a depth of 2 mm and a root radius, ρ , of 0.25 mm, shown in Fig. 9.1B. The corresponding fracture energy is usually indicated with E_{CVN} . The velocity of the mass after the impact is lower than the initial velocity, depending on the absorbed energy. The average strain rate at the notch root is given by: $\sim \bar{v}/2\rho$, where $\bar{v}$ is the average impact velocity. With $v = 5$ m/s, the average strain rate at the notch region is on the order of 10^3 s^{-1} .

Sub-size specimens are sometimes used (for example, with a thickness that is half the usual one, i.e., 5 mm instead of 10 mm), when the fracture energy is too high with respect to the available energy (E_0), or because of limitations given by the geometry of the material from which the specimen is machined.

In the case of ductile fracture, the energy required to break the specimen is high, whereas in the case of brittle fracture it is comparatively small, as the deformation to trigger the fracture process is small. At the notch root of the impact specimen, the state of stress is initially close to plane strain. In fact, the notch exerts a constraint, impeding plastic deformation along the notch front (refer to Section 5.5). In the case of brittle fracture, this state of stress does not change during the fracture event, as the fracture

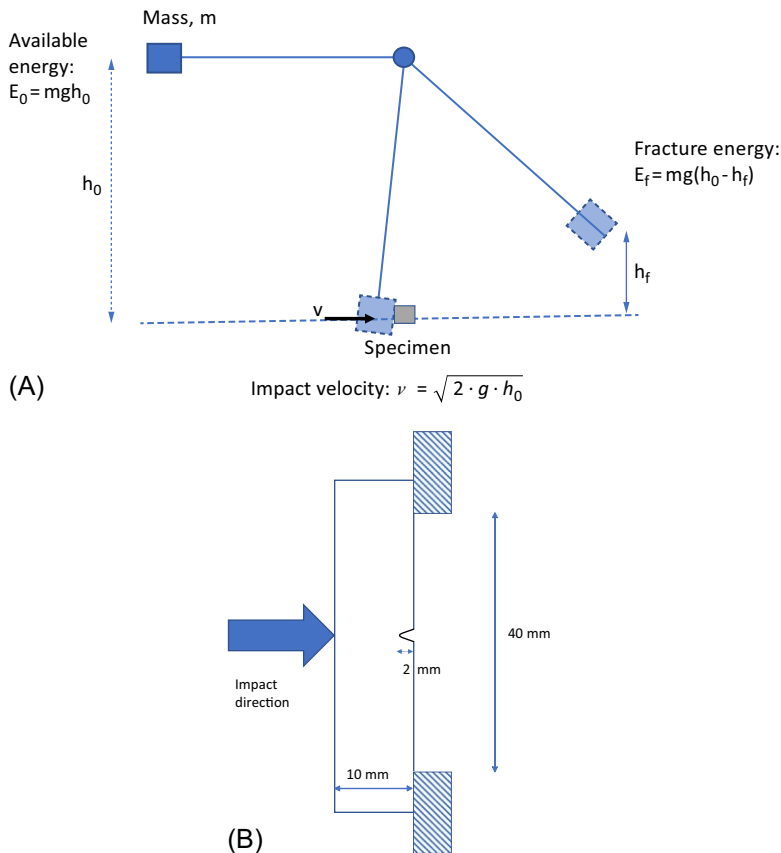


Fig. 9.1 (A) Schematic of the CVN impact test; (B) geometry of a Charpy V-notch (CVN) specimen. The length of the specimen is 55 mm and the thickness is 10 mm.

occurs after a small plastic deformation at the notch. On the contrary, in the case of ductile fracture, the specimen deflection at fracture is large, and so are the deformations in the notch region. Since the notch is not able to impede the material contraction, fracture develops in conditions moving towards uniaxial deformation.

The tests are most often carried out at RT. However, it is also common to carry out tests at low temperatures (below RT), to detect the transition temperature towards brittle behavior. Special procedures are used to bring the specimens to a given temperature, using different temperature conditioning media (generally a cryogenic liquid), and quick transfer and placement in the testing equipment (within 5 s). Tests can also be carried out at high temperature (above RT). In this case, the specimen is heated to the test temperature. It can also be heated on the specimen support, just before the impact.

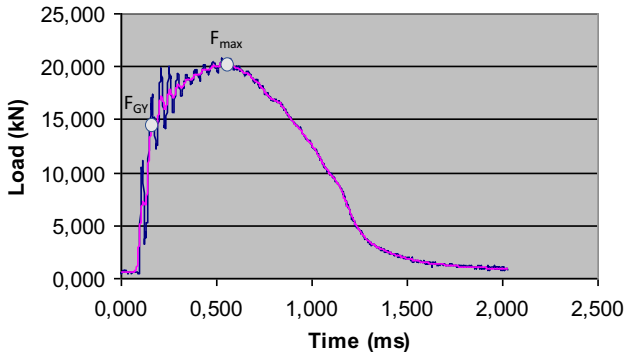


Fig. 9.2 Typical CVN impact load-deflection curve for a ductile metal.

With an *instrumented* impact testing rig, it is possible to record the load applied to the specimen during the impact and obtain the *impact load-deflection curve*, in which the load is plotted as a function of time or specimen deflection. Since the load is applied in a very short time, inertial effects play an important role. Correspondingly, the load signal obtained from the instrumented tup oscillates around the actual value. Particular procedures to smooth the inertia oscillations are then required. Fig. 9.2 shows an impact load-deflection curve for a ductile fracture. The first part of the curve is similar to that shown in Fig. 5.8A. The load is seen to increase up to yielding at the notch root and subsequently general yielding on the load-bearing section is attained. F_{GY} indicates the general yield load. Using a relation proposed by Server (1978) for CVN specimens, it is possible to estimate the dynamic yield strength, σ_{Yd} (i.e., the yield strength in the impact conditions, namely at approximately 10^3 s^{-1}):

$$\sigma_{Yd} = 43.6 \cdot F_{GY} \quad (9.1)$$

where σ_{Yd} is in MPa and F_{GY} is in kN (the constant 43.6 must be replaced with 118.2 when using U-notched specimens). It is worth remembering that σ_{Yd} is related to σ_Y by means of Eq. (3.8). For bcc metals, $m \approx 0.05$, whereas for fcc metals, $m \approx 0.005$. Therefore at RT for bcc metal: $\sigma_{Yd} \sim 2\sigma_Y$, and for fcc metals: $\sigma_{Yd} \sim \sigma_Y$.

The load then increases, because of strain hardening, towards a peak value indicated with F_{max} in Fig. 9.2. After F_{max} the load declines. This is because of the nucleation of the fracture crack at the notch root where the strain is maximum. In general, crack nucleation takes place slightly before F_{max} (after the nucleation, the load increases a bit because the effect of strain hardening overwhelms the reduction of the load-bearing section induced by the crack), but the approximation of nucleation at F_{max} is acceptable.

The area below the force-deflection curve represents the fracture energy, E_{CVN} . The area up to F_{max} is the impact energy for crack nucleation, E_n , whereas the area from F_{max} up to fracture is the energy for crack propagation, E_p , such that:

$$E_{CVN} = E_n + E_p \quad (9.2)$$

The impact energy, E_{CVN} , is expected to be proportional to the metal strength and ductility, since these parameters influence both the energy for crack nucleation and that for propagation. However, the energy for crack propagation is also affected by the characteristics of the propagation path that are in turn related to the microstructure. Therefore it is difficult to simply correlate the CVN impact energy, E_{CVN} , with specific mechanical properties. In Fig. 9.3, E_{CVN} is plotted as a function of the yield strength for different metals at RT. Only selected materials were used, and hence a broad indication is simply provided. It is seen that E_{CVN} is comparatively low (below approximately 50 J), when the yield strength is greater than 600/700 MPa. For lower σ_Y values, E_{CVN} generally increases with decreasing the yield strength for every metal. The E_{CVN} data can be compared with the ductility data summarized in Fig. 7.31, and a broad correlation is recognized between E_{CVN} and tensile elongation at fracture. Some exceptions can also be highlighted. One regards C-Mn ferritic-pearlitic steels. They display an intermediate ductility (Fig. 7.31A) and rather high E_{CVN} values at comparatively low yield strength values (approximately below 300 MPa). A similar behavior is also displayed by low-strength Ti alloys, such as α Ti with low interstitial content (not included in the figure).

As a further example, Fig. 9.4 shows the impact energy for three steels containing 0.11, 0.47, and 0.78% C (AISI 1010, 1045, and 1080), obtained using sub-size Charpy V-notch specimens (thickness: 2.5 mm), from RT up to 800°C (Ferreiros et al., 2017). From RT up to 500°C, i.e., up to 0.4–0.5 T_m , the impact energy decreases with temperature. This is mainly due to a decrease in the metal strength with temperature, since in this temperature interval ductility is almost unaffected by

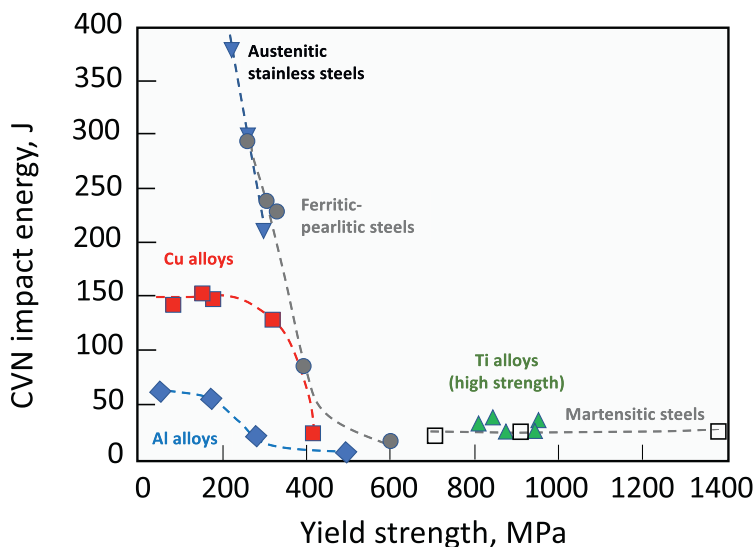


Fig. 9.3 Relationship between E_{CVN} and yield strength for different metals at RT. The amount of data is limited and only a broad indication is provided.

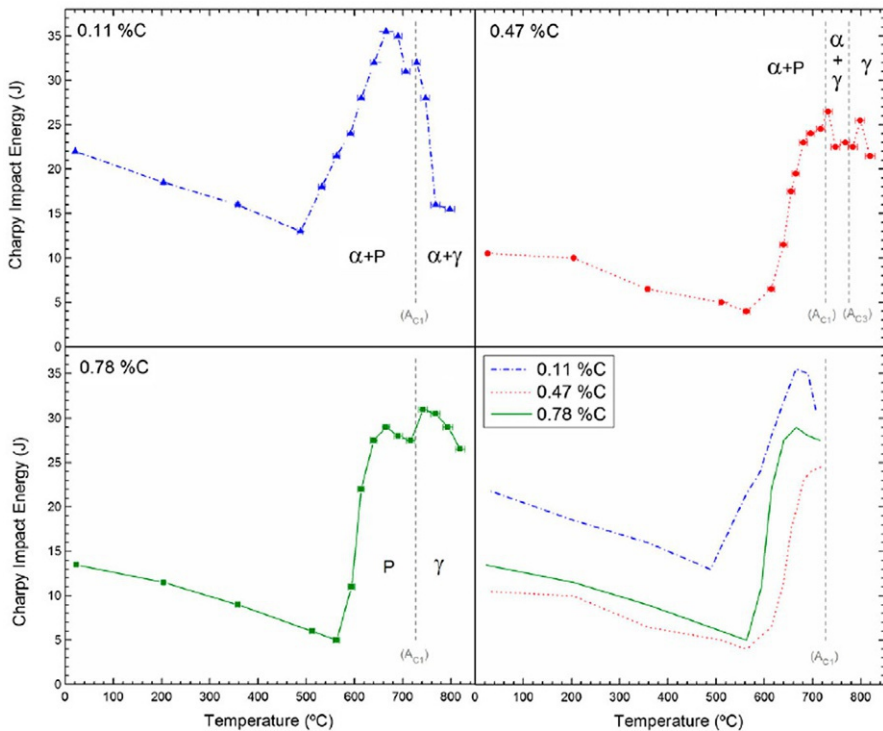


Fig. 9.4 Dependence of the Charpy fracture energy on temperature for three steels containing 0.11, 0.47, and 0.78%C.

From Ferreiros, P.A., Alonso, P.R., Gargano, P.H., Rubiolo, G.H., 2017. High-temperature testing in Charpy impact pendulum using in-situ Joule heating of the specimen. *Fatigue Fract. Eng. Mater. Struct.* 41, 1171–1182, reproduced with permission of the publisher.

temperature. By comparing the behavior of the three steels, it is seen that E_{CVN} decreases with increasing C content, i.e., with increasing the fraction of pearlite (as explained in Section 7.2). For temperatures greater than 500°C (i.e., greater than $0.4\text{--}0.5T_m$), the fracture energy steeply increases with temperature. Since the strength continues to decrease with temperature, this behavior is clearly due to a marked increase in ductility, as explained in Section 8.3. Subsequently, the steel containing 0.11%C shows a drop in fracture energy as temperature is increased above 700°C. This is due to a drop in ductility, since this temperature interval corresponds to the left side of the ductility trough (Fig. 8.10). Such a drop is less evident for the other two steels.

The data in Fig. 9.4 show that impact testing is rather useful to investigate the high-temperature ductility of metals, including the possible presence of ductility troughs, at an average strain rate that is close to the value encountered in forging or other high-temperature forming operations (see Chapter 10).

9.2 Brittle fracture and Davidenkov-Orowan diagram

The ductile-to-brittle transition occurs when the conditions to trigger brittle fracture are attained in the plastic region of a metal, such as at the notch root of a Charpy-V specimen. In this section, we analyze such conditions, discussing the mechanism of brittle fracture. In this way, it is possible to know the parameters that influence the transition, and the role they exert.

We may have two kinds of brittle fracture:

1. *Transgranular cleavage fracture*, when the crack propagates through the grains along specific close packed planes;
2. *Intergranular brittle fracture*, when the crack propagates along the grain boundaries.

Fig. 9.5 shows a schematization of these kinds of fracture, together with two fractographic examples. In both cases, brittle fracture is triggered by the unstable propagation of one or more cracks. This idea originates from the well-known work of Griffith on the brittle fracture of glass. The Griffith theory assumes the pre-existence of a population of cracks in the glass (a material with extremely limited plastic deformability) and predicts the occurrence of brittle fracture when the tensile (opening) principal stress, σ_1 , is sufficiently high to trigger the unstable propagation of one of such

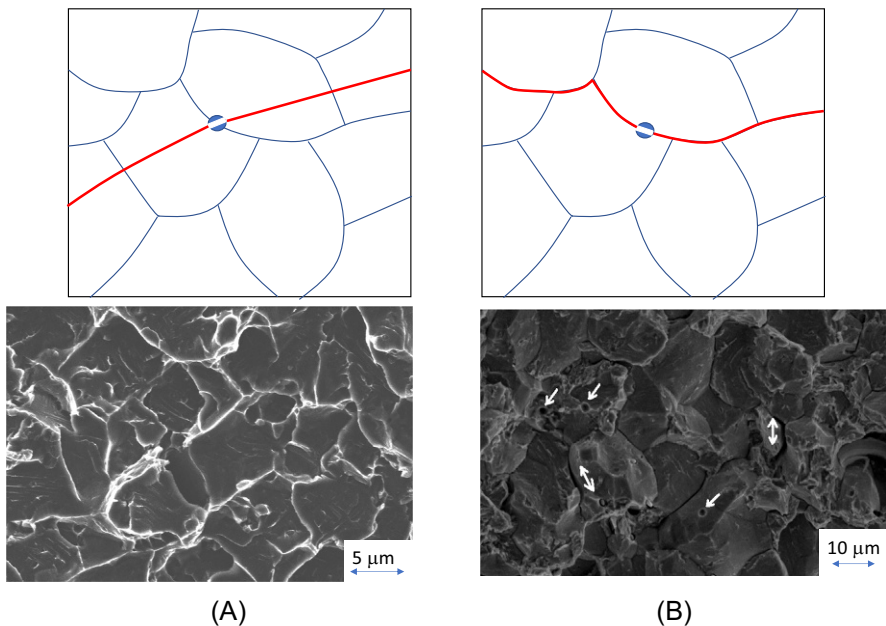


Fig. 9.5 Schematization of microstructure with transgranular brittle (cleavage) crack propagation (A), and intergranular brittle crack propagation (B). In both cases, crack propagation starts from a cracked particle.

Modified from Straffelini, G., Versari, L., 2009. Brittle intergranular fracture of a thread: the role of a carburizing treatment. Eng. Fail. Anal. 16, 1448–1453.

cracks. That is, when $\sigma_1 = \sigma_F$, where σ_F is the *critical brittle fracture stress*, which is given by:

$$\sigma_F = \sqrt{\frac{2 \cdot E \cdot \gamma}{\pi \cdot c}} \quad (9.3)$$

where

- E is the longitudinal elastic modulus;
- γ is the surface energy of the material forming the crack (the only work of fracture considered);
- c is the length of the pre-existing crack that is responsible for fracture (known as the *weakest link*).

The fracture process is statistical in nature, as it is related to the probability to sample a crack with the right length c in a volume of material subjected to a given stress (Herzberg, 1976).

In metals, the Griffith relation must be modified, since these materials contain a sufficiently high density of mobile dislocations that ensure a high plastic deformability. The main consequences of this are two. The first is that pre-existing cracks in the microstructure are blunted. Therefore they cannot act as fracture cracks: brittle fracture cracks must be dynamically nucleated *during* the fracture process. The second relates the required energy for crack propagation: it is greater than the surface energy of the metal. Parameter γ is then substituted with G , which accounts for both the creation of two new surfaces and the dissipation of some plastic deformation during crack propagation. It is identified as the *effective surface energy for crack propagation* and as a measure of the matrix *fracture toughness*.

The brittle fracture mechanism can then be subdivided into three stages (Thompson and Knott, 1993):

1. Formation of a plastic region in the metal (an important situation is the plastic region at a notch root, where the stress σ_1 is intensified by the constraint induced by the notch; see for instance Fig. 5.9);
2. Fracture of a hard particle inside the plastic region (in most cases, it is a hard intermetallic particle);
3. Dynamic propagation of the crack across the particle/matrix interface.

In pure metals, a crack may nucleate and propagate following a mechanism proposed by Zener, which is based on the pile-up of dislocations at the grain boundaries. In engineering metals, the presence of hard particles, such as precipitates, which are able to nucleate brittle cracks, facilitates the brittle fracture process, anticipating the conditions for its onset (note that the hard particles also have a high elastic modulus and the stored elastic energy for a given applied deformation is high; the release of this energy is required to create the new crack surfaces).

The brittle fracture of metals has been extensively investigated in ferritic steels with bcc structure (refer, for example, to Lin et al., 1987). In these steels, the cracks are easily formed by the fracture of grain boundary carbides. In general, the fracture-controlling event is the subsequent crack propagation across the carbide/matrix

interface. This occurs when the local tensile (principal) stress, σ_1 , is greater than the *critical brittle fracture stress* σ_F , which is given by:

$$\sigma_F = \sqrt{\frac{\pi \cdot E \cdot G}{2 \cdot (1 - \nu^2) \cdot D_p}} \quad (9.4)$$

where

- G is the energy required for the propagation of the crack;
- ν is the Poisson ratio of the steel matrix;
- D_p is the diameter of the cracked carbide (the relation assumes local plane strain conditions).

For ferritic steels, G is comparatively low, typically in the range 8–20 J/m² and slightly dependent on temperature. Consequently, σ_F does not depend much on temperature as well. Of course, if $\sigma_1 < \sigma_F$, all cracks are arrested at the carbide/matrix interface, and they can possibly serve as voids for the subsequent ductile fracture (voids nucleation by particle cracking, Section 6.3).

This mechanism can be extended to steels and metals with different microstructures. In any case, a particle that nucleates an unstable crack is responsible for the brittle fracture event. Eq. (9.4) can be generalized in the following way:

$$\sigma_F = \sqrt{\frac{\pi \cdot E \cdot G}{2 \cdot (1 - \nu^2) \cdot x}} \quad (9.5)$$

where x is the size of the particle responsible for the fracture event.

A useful way to rationalize the role of the parameters involved in the ductile-to-brittle fracture transition is the *Davidenkov-Orowan diagram*. In this diagram, shown in Fig. 9.6, the three stresses σ_Y , σ_{1max} , and σ_F are plotted together as a function of temperature.

- σ_Y is the metal yield stress (it is the dynamic yield stress, σ_{Yd} , in the case of impact loading) that must be reached at the notch root to have the onset of plastic deformation (at the notch

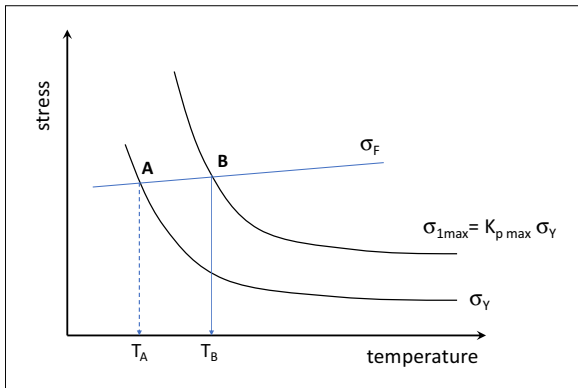


Fig. 9.6 Davidenkov-Orowan diagram.

root, plane strain conditions are present; following the simple Tresca criterion, yielding occurs when: $\sigma_1 = \sigma_Y$).

- σ_{1max} is the maximum stress in the plastic region ahead of the notch root, and it is given by: $K_{pmax} \sigma_Y$ where K_{pmax} is the maximum value of the plastic stress intensification factor (refer to Section 5.5).
- σ_F is the critical brittle fracture stress that is almost unaffected by temperature (only a small increase with temperature is considered).

In the diagram, two points are indicated, *A* and *B*, and correspondingly two temperatures, T_A and T_B , are identified. They are formed by the intersection of σ_F with σ_Y and σ_{1max} , respectively.

For temperatures below T_A , fracture is completely brittle. To have fracture, the applied load must be sufficiently high to induce the onset of yielding at the notch root. A small plastic region is formed (when $\sigma_1 = \sigma_Y$), the extension of which increases with the load. In this region, $\sigma_1 = \sigma_Y > \sigma_F$. Therefore brittle fracture occurs as soon as a suitable particle is sampled. Fracture is then *nucleation-controlled*, since the critical step is the crack nucleation. For temperatures between T_A and T_B , fracture is also completely brittle. When $\sigma_1 = \sigma_Y$ at the notch root, plastic deformation begins. With increasing the applied load, σ_1 also increases because of the stress intensification due to the local constraint (see Fig. 5.9). Brittle fracture occurs when σ_1 in the plastic region becomes equal to σ_F . In this case, brittle fracture is *propagation controlled*. Finally, for temperatures above T_B , brittle fracture does not occur at the notch root since the maximum σ_1 value reached in the plastic region is lower than σ_F .

With reference to the Davidenkov-Orowan diagram, some useful indications can be obtained:

1. An increase in σ_F induces a decrease in T_B (and vice versa). As seen, σ_F may change because of a variation in the size of the particles that nucleate fracture (parameter x in Eq. 9.5), and following a change in G . For example, a weakening of the grain boundary cohesion may induce a decrease in G , favoring the intergranular brittle fracture (Fig. 9.5B) and an increase in T_B .
2. An increase in σ_Y induces an increase in $K_{pmax}\sigma_Y$ and then in T_B (and vice versa). A change in σ_Y can be attained by changing the strain rate, following a change in the impact velocity, or with a microstructural strengthening modification, as outlined in Chapter 1.
3. A change in the notch geometry induces a change in the stress intensification in the plastic region at the notch root, and then in T_B . For example, when using a CVN specimen, K_{pmax} is approximately 2.52 (it is 2.2 from the slip line theory, but the value of 2.52 is most suitable in case of a pendulum tup width in the range of 1.5–2 mm). When using an un-notched tensile specimen K_{pmax} is approximately in the range of 1–1.3, because of necking. A decrease in K_{pmax} induces a decrease in T_B (this is clearly the motivation for using notched specimens in investigating the possible transition towards a brittle behavior of a metal). When using pre-cracked specimens, K_{pmax} is typically in the range of 3.7–4, leading to a shift of T_B towards the right.

9.3 Ductile-to-brittle fracture transition

As mentioned, the transition from a ductile to a brittle behavior is typically observed in ferritic steels with bcc structure. As an example, Fig. 9.7 shows the evolution of the

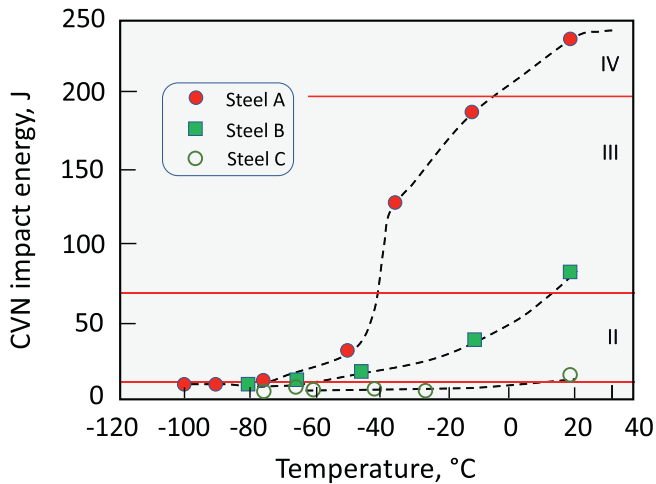


Fig. 9.7 CVN impact transition curves for three steels with different C content. The shape of the impact transition curve (refer to Fig. 7.4) is also indicated.

CVN impact energy with temperature (from RT down to -100°C) for three ferritic-pearlitic steels, labeled with Steel A, B, and C, using standard CVN specimens. The steels have an increasing C content and their main characteristics are summarized in Table 9.1. Four shapes of the load-deflection curves are identified with reference to specific impact energy intervals. They are indicated with I, II, III, and IV and are schematically shown in Fig. 9.8. For simplicity, the corresponding impact energy intervals are also summarized in Fig. 9.9.

Table 9.1 Characteristics of the three steels (A, B, and C) considered in Fig. 9.7.

Steel	Chemical composition	Microstructure	σ_y (MPa)	σ_{yd} (MPa)	T_B (NDT) ($^{\circ}\text{C}$)
A (ASTM A387 GR5)	0.09%C, 0.45% Mn, 0.25%Si, 4.72%Cr, 0.48%Mo, 0.16%Ni	Ferrite (grain size: 4 μm)	310	637	~ -90
B (ISO 30CrMo4)	0.31%C, 0.49% Mn, 0.27%Si, 0.94%Cr, 0.18%Mo	Ferrite—fine pearlite	340	645	~ -80
C (ISO 46Mn7)	0.44%C, 1.65% Mn, 0.28%Si, 0.16%Cr	Fine pearlite (some grain boundary ferrite)	600	922	~ 25 (RT)

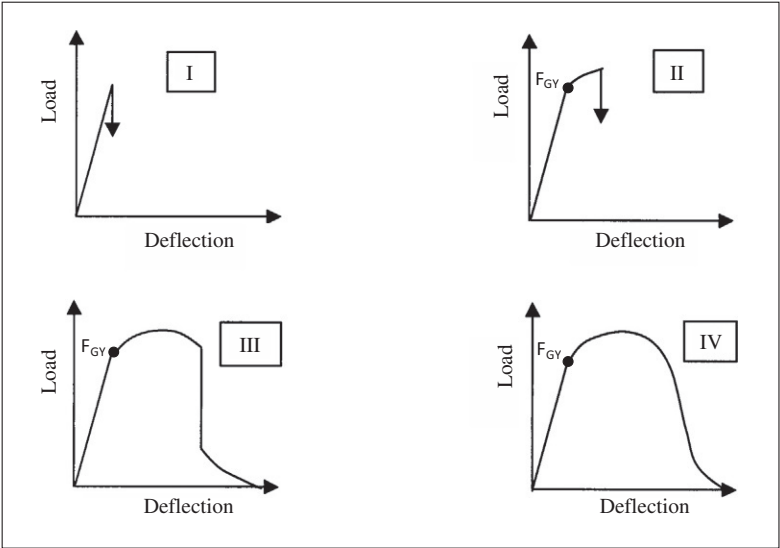


Fig. 9.8 Typical representation of the load-deflection curves in the different regions of the impact transition curve (the typical oscillations due to inertia effects are not indicated).

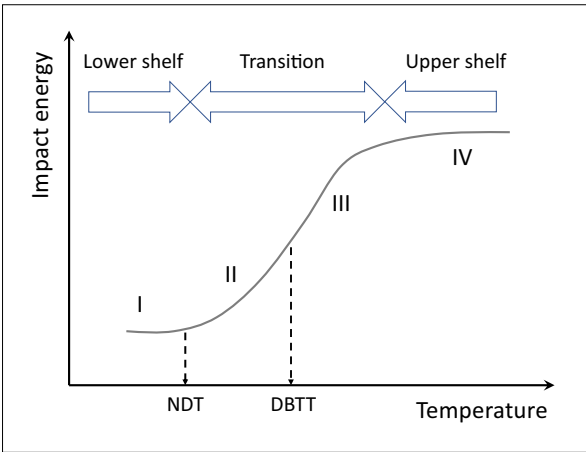


Fig. 9.9 The ductile-to-brittle transition curve with the indication of four distinct regions.

Steel A displays a completely ductile fracture at RT (20°C in this case). The shape of the curve is labeled with IV and is similar to that shown in Fig. 9.2. The high-temperature region, at which this kind of fracture occurs, is denominated the *upper shelf* region. In the upper shelf region, the CVN fracture energy in steels is typically greater than 200J.

As temperature is decreased, the fracture energy is also decreased. An intermediate region, made up of two parts, indicated with II and III, can be recognized. The fracture energy of region III is typically from 200 to $\sim 50\text{--}70\text{J}$ (Tani and Nagumo, 1995). As indicated by the shape of the impact load-deflection curve (Fig. 9.8), strain hardening takes place after general yielding, and at the maximum load the ductile crack is nucleated at the notch root, and then propagates. However, during the propagation, brittle fracture is triggered, leading to a sudden drop in the recorded load. This is due to the increased strain rate at the crack tip, and at the change of the local state of stress that turns out to be close to plane strain. Both these phenomena allow for the attainment of the conditions of brittle fracture at the propagating crack tip (see previous section). Subsequently, the lateral sides of the specimen undergo plastic contraction because of the local decrease in the stress triaxiality; some energy is then absorbed by this contraction before the final separation of the two specimen parts.

With a further decrease in temperature, the fracture energy continues to decrease. When the CVN fracture energy is typically in the range $10\text{--}70\text{J}$ (region II; displayed by all the three steels under consideration), the conditions for the attainment of brittle fracture at the notch root are quickly reached after general yielding, because of the increase in the yield strength. Finally, at low temperature (region I), the impact energy is very low (lower than 10J), since fracture is completely brittle. This is the so-called *lower shelf* region. The load-deflection curve is macroscopically linear up to fracture. Consequently, the fracture energy is low and proportional to the recorded maximum load. The maximum load occurs after the onset of localized yielding at the notch root and before general yielding (Fig. 5.8A).

The transition temperature between regions I and II is indicated with NDT (*nil ductility temperature*) in Fig. 9.9. It corresponds to temperature T_B of the Davidenkov-Orowan diagram. As seen, the transition from ductile to brittle behavior is gradual with the existence of an intermediate temperature region. In addition to the NDT approach, other practical methods are available to define the transition temperature. A simple one considers the temperature at which the fracture energy is the average between the upper shelf and lower shelf fracture energies. This transition temperature is denoted by *ductile-to-brittle transition temperature* (DBTT). The DBTT temperature broadly corresponds to the transition temperature between regions II and III (Fig. 9.9). Another method considers the temperature at which the fracture energy assumes a fixed value, e.g., 27J .

The NDT temperature can be simply estimated assuming that brittle fracture occurs when σ_1 reaches the maximum value in the plastic region, i.e., $\sigma_{1max} = K_{pmax} \sigma_{Yd} = \sigma_F$. The evolution of σ_{Yd} with temperature in the regime of thermally activated slip of dislocations is given by Eq. (3.1). To a first approximation, a linear relationship can be used to relate the yield strength with temperature: $\sigma_{Yd} = a - b \cdot T$, where a and b are two material constants, and T is the temperature (setting T in $^{\circ}\text{C}$, constant a is then the yield strength at 0°C). After making the proper substitutions, the following is obtained:

$$NDT = \frac{1}{b} \cdot \left(a - \frac{\sigma_F}{K_{p,max}} \right) \quad (9.6)$$

This relationship clearly shows that the transition temperature is *not* a material property, since it depends on $K_{p,max}$, i.e., on the notch geometry (a similar reasoning yields also for DBTT).

In Table 9.1, the NDT values for the considered steels are listed. The lowest NDT temperature is shown by Steel A, which is characterized by the lowest yield strength (and then lower “ a ” value). The highest NDT temperature is displayed by Steel C, due to its higher yield strength, caused by the fine pearlitic microstructure. Note that in ferritic-pearlitic steels, the size of the pearlite colonies is the microstructural feature that triggers brittle fracture. In fact, the fracture crack is nucleated by the fracture of the cementite lamellae in a colony, induced by the plastic deformation in the inter-lamellar ferrite.

In high-strength steels, including heat-treated steels, the high σ_Y values (and hence the high “ a ” values) imply a shift of the transition temperatures to the right, i.e., to higher values. However, in quenched and tempered steels the size of the precipitated carbides is quite small (refer to Section 7.1), and therefore the σ_F values for these steels are high (depending on chemical composition and treatment cycle). Values in the range 3000–4000 MPa are typical (in ferritic-pearlitic steels they are lower than 2000 MPa). This means that the transition temperature can be comparatively low, as dictated by Eq. (9.6). For example, it is reported that in a steel containing 0.4%C, DBTT is approximately 10°C, if the steel has a ferritic-pearlitic microstructure, and it decreases to –170°C after heat treatment (Reed-Hill, 1991).

A special family of heat-treated steels is constituted by the tool steels that are used to manufacture the forming tools. Cold work tool steels and high-speed steels have a C content in the range of 0.1%–1.5%. They also have high hardness, in the range of 650–100 HV, depending on the C content and heat treatment. At RT, they display a macroscopically brittle behavior. In Charpy V-notch tests, the shape of the impact curves is of type I (Fig. 9.8), and the fracture energy is then lower than 10 J. Hot work tool steels contain a lower amount of C (approximately 0.35%). Their hardness is lower (in the range of 400–700 HV) and their impact curve at RT is of type II (the fracture energy is between 15 and 30 J). In general, the fracture toughness of these steels is experimentally related to their impact fracture energy, determined with CVN tests. In many manufacturing processes, such as in close die forging operations, tools must be made with a steel that is able to guarantee a high hardness (for improved wear resistance) and sufficient fracture toughness (to resist impact loading during the operations, considering that dies often contain notches). Due to this, the microstructure of tool steels must be optimized with a proper selection of the chemical composition, heat treatment cycle, and a control of the impurities. Impact testing is paramount for a performance control of tool steels to be used in the manufacturing of forming tools.

As already mentioned, the ductile-to-brittle transition is typical of bcc steels and other bcc metals. In several other metals, with different structure, this transition is not observed. This means that $NDT < 0\text{ K} = -273^\circ\text{C}$, and hence, from Eq. (9.6):

$$\sigma_F \geq K_{p,max} \cdot \sigma_{Y0} \quad (9.7)$$

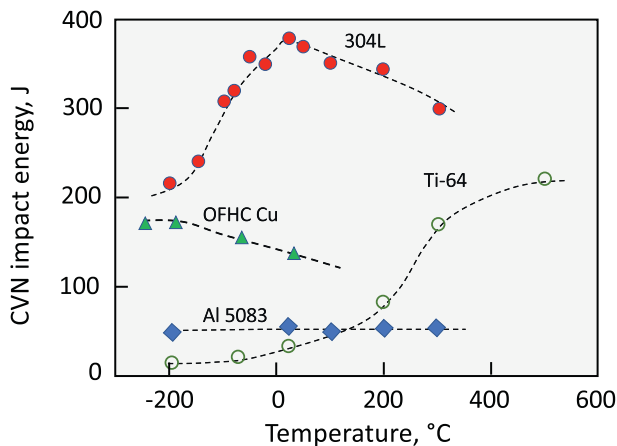


Fig. 9.10 CVN impact energy for selected metals as a function of temperature (data from Byun and Lach (2016) for 304L, Semenova et al. (2019) for Ti-64, Ozden et al. (2007) for Al 5083, and Warren and Reed (1963) for OFHC Cu).

where σ_{Y0} is the predicted yield strength at 0 K. This inequality is satisfied when σ_F is high and/or σ_{Y0} is low. Typical metals that do not display the ductile-to-brittle transition are austenitic stainless steels and Cu and Al alloys.

Fig. 9.10 shows the CVN impact energy for selected alloys as a function of temperature. The 304L austenitic stainless steels is characterized by rather high values of E_{CVN} and fracture is always ductile. This is a typical behavior of austenitic stainless steels. The comparison with Fig. 7.13 confirms that E_{CVN} is proportional to the tensile elongation at fracture. The absence of the transition to brittle fracture is due to the limited σ_Y increase as temperature is decreased (low σ_{Y0}), and the high value of G (Eq. 9.5), due to the ease of plastic deformation of fcc metals, which corresponds to a very high σ_F value.

For OFHC Cu, also with an fcc structure, E_{CVN} is rather high at low temperature. E_{CVN} is then seen to decrease with increasing temperature. Also in this case, the trend is very similar to that displayed by the elongation at fracture, as shown in Fig. 8.1B. The same behavior is displayed by 70/30 brass (not reported in the figure), even if the E_{CVN} values are much lower (on the order of 20 J). The 5083 Al alloy displays a fracture energy broadly independent of temperature, with a ductile fracture behavior at every temperature. For Cu and Al alloys, the yield strength σ_Y does not increase much with decreasing temperature (the σ_{Y0} values are comparatively low) and the matrix fracture toughness is high also at low temperature (the σ_F values are comparatively high). Therefore, the conditions for brittle fracture are never attained.

Fig. 9.10 also shows the CVN impact energy for Ti-64. For this material (with a dual α -hcp/ β -bcc structure) E_{CVN} is rather high at high temperature and then it decreases with decreasing temperature. However, no ductile-to-brittle fracture transition is observed. In fact, fracture remains ductile down to -197°C . The E_{CVN} trend is, also for this metal, rather similar to the evolution with temperature of the elongation at fracture (Fig. 8.6). For this metal, the σ_{Y0} value is rather high (the yield strength increases markedly with decreasing temperature and σ_{Y0} is greater than 1800 MPa), but the absence of hard particles in the microstructure is responsible for the absence of this transition even at very low temperatures.

9.4 Impact testing and embrittling phenomena in steels

The experimental evaluation of the CVN impact energy and of the transition temperature, supplemented with a proper fractographical investigation to evaluate the fracture mode, provide valuable information on the occurrence of undesirable microstructural modifications in a metal. When using an instrumented impact testing, the analysis can be completed with an evaluation of the critical brittle fracture stress, σ_F . For this task, different approaches are proposed in the literature. A very simplified one is based on the evaluation of σ_{Yd} (using Eq. 9.1) at NDT (i.e., at T_B), and then to the estimation of σ_F with the following relationship: $\sigma_F = K_{pmax} \sigma_{Yd}$.

As already mentioned, the ductile-to-brittle transition behavior has been deeply investigated in **ferritic steels**. For these metals, the microstructural feature responsible for the onset of brittle fracture is the size of grain boundaries carbides. In general, the ratio of the carbide size and grain size is constant, since both the parameters depend on the same diffusion phenomena. This means that grain refining is accompanied by refining in the carbide size, and an increase in σ_F . It is reported that σ_F increases from 500 to 1500 MPa in passing from an average grain size of 250 μm to an average grain size of 15 μm (Knott, 1973). This also implies a decrease in NDT (and DBTT), even if grain refining induces an increase in the yield strength (refer to Section 1.6). For example, DBTT was found to increase from -20°C to 30°C with an increase in the carbide size from 0.6 to 1 μm (Petch, 1987). Correspondingly, the upper shelf energy was found to decrease from $\sim 280\text{J}$ to $\sim 200\text{J}$. In ferritic steels, an important role is exerted by Mn. As shown by Eq. (1.16) (Section 1.6), Mn induces an increase in σ_Y due to solid solution hardening. However, it has been found that Mn induces a decrease in the transition temperature. This can be explained by considering that Mn also induces a considerable carbide refinement since it reduces the intensity of the diffusion phenomena (Petch, 1987).

An evaluation of the DBTT (or NDT), together with σ_F and the upper shelf energy, is then useful to detect possible undesired microstructural modifications, such as an excessive grain size accompanied by an excessive thickness of grain boundaries carbides.

An important role in the fracture behavior of steels is exerted by inclusions, such as Mn sulfides. In Chapter 7, it is observed that such particles decrease their ductility. In addition, they also introduce a mechanical anisotropy since they are elongated in the direction of plastic deformation (such as in the rolling direction). In Fig. 9.11, the impact transition curves of an HSLA steel containing 0.03% S are shown (Ghosh et al., 2016). The tests were carried out using specimens with different orientation with respect to the longitudinal, or rolling, direction (the orientations are indicated by the angles). It can be seen that the fracture energy in the upper shelf decreases in passing from the longitudinal direction to the transverse direction. This is in agreement with the corresponding significant reduction in ductility, explained in Section 7.2. It is also seen that the DTT and DBTT temperatures are not much influenced by the specimen orientation (they are approximately -60°C in all cases). This is explained considering that large Mn sulfides do not directly nucleate brittle cracks. When the S content is

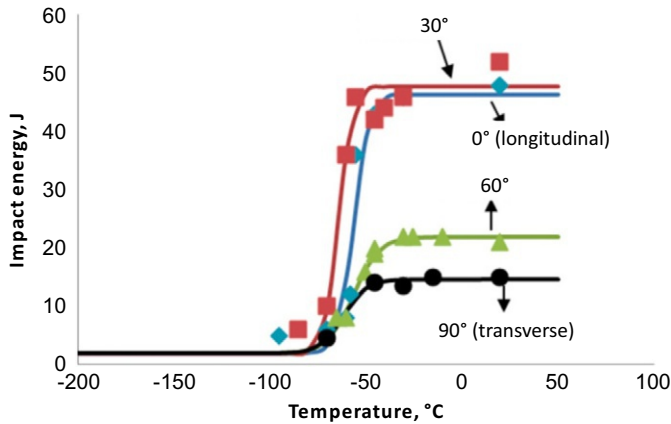


Fig. 9.11 Impact transition curves for an HSLA steel containing 0.03% S. Subsize specimens were used (cross-section: 8 mm $\times$ 10 mm) with different orientation angles with the rolling direction.

Modified from Ghosh, A., Modak, P., Dutta, R., Chakrabati, D., 2016. Effect of MnS inclusion and crystallographic texture on anisotropy in Charly impact toughness of low carbon ferritic steel. *Mater. Sci. Eng. A* 654, 298–308, reproduced with permission of the publisher.

rather low and, correspondingly, the fraction of sulfides is also very low, the impact fracture transition curve is mainly influenced by the matrix properties of the steels.

In heat-treated steels, particular embrittling phenomena may take place during tempering. The *tempered martensite embrittlement* (TME) phenomenon may occur when tempering high-strength steels in the temperature range of 300–400°C. It is due to the precipitation of carbides at the prior austenitic grain boundaries, whose cohesion is reduced by the segregation of impurities (particularly P), during the previous austenitization. Intergranular brittle fracture is then caused by the fracture of the carbides, and the subsequent easy crack propagation along the grain boundaries (Fig. 9.5B). The TME phenomenon is severe in steels containing more than 0.02% of P, since the grain boundary segregation of P reduces G and favors the intergranular crack propagation.

Another embrittling phenomenon is *temper embrittlement* (TE). It may occur in different situations:

- In the case of tempering treatments in the critical range of 450–600°C;
- In the case of slow cooling of large parts, such as hot forged parts, through this critical temperature interval;
- In the case of prolonged service exposure of mechanical parts in this critical temperature interval.

TE is due to the grain boundary segregation of impurities such as P or Sb (even in total contents below 0.01%), which induces a reduction in the grain boundary cohesion and hence in G . As an example, Fig. 9.12 shows the relationship between σ_F and P grain boundary segregation (expressed in terms of fraction of P monolayer) for a low alloy

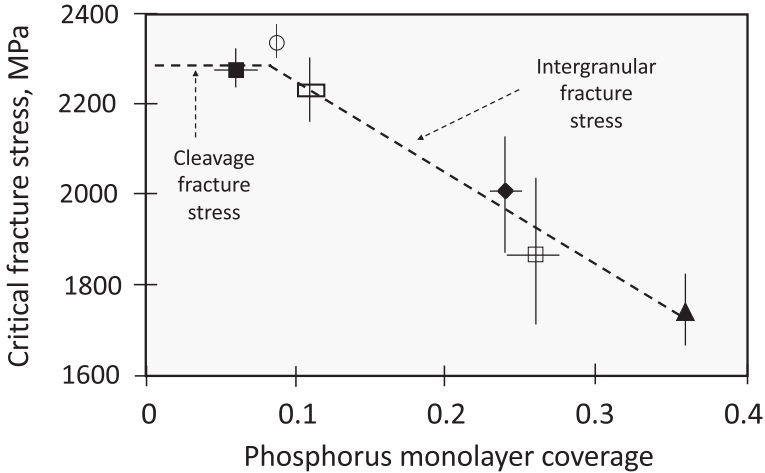


Fig. 9.12 Relationship between σ_F and P grain boundary segregation (in terms of fraction of P monolayer) for a low alloy steel subjected to different treatments.

Redrawn from Naudin, C., Freund, J.M., Pineau, A., 1999. Intergranular fracture stress and phosphorus grain boundary segregation in an alloy-steel. *Scr. Mater.* 40, 1013–1019.

steel, subjected to different treatments (Naudin et al., 1999). The effect of P may be alleviated by the presence of free C. Both elements tend to segregate at the grain boundaries, but C increases their cohesion.

As an exercise, we can consider the following temperature dependency of σ_{Yd} : $\sigma_{Yd} = 510 - 2.5T$ (with σ_Y in MPa and T in °C). From Eq. (9.6) it is obtained:

$$\text{NDT} = \left(510 - \frac{\sigma_F}{2.52} \right) \cdot \frac{1}{2.5}$$

Without segregation, we may set $\sigma_F = 2250$ MPa (from Fig. 9.12), and then we obtain: $\text{NDT} = -153^\circ\text{C}$. But if the P segregation reduces σ_F to 1800 MPa, NDT is correspondingly increased to -82°C .

As seen, austenitic stainless steels do not show the transition to brittle fracture. However, if σ_Y is considerably increased (for example, through a significant addition of N) and/or G is decreased (for example, due to the grain boundary precipitation of embrittling phases, such as specific carbides or other compounds, following aging in critical temperature intervals), a ductile-to-brittle transition can occur in austenitic stainless steels as well. Most austenitic stainless steels are characterized by the precipitation of embrittling σ phase in the temperature interval of 480°C and 815°C . Hot finishing operations should then be carried out at temperatures greater than $\sim 870^\circ\text{C}$, followed by rapid cooling to avoid possible subsequent cracking phenomena.

We consider now in some detail the case of **duplex stainless steel (DSS)** that has a ferritic-austenitic microstructure and displays an impact behavior that is intermediate between that of austenitic and ferritic steels (Nilsson, 1992). A brittle behavior can be

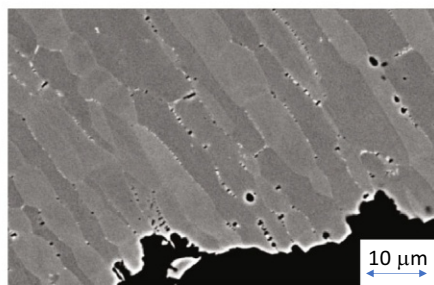
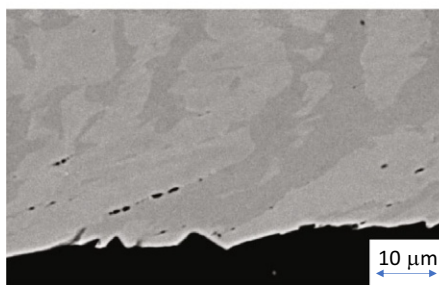
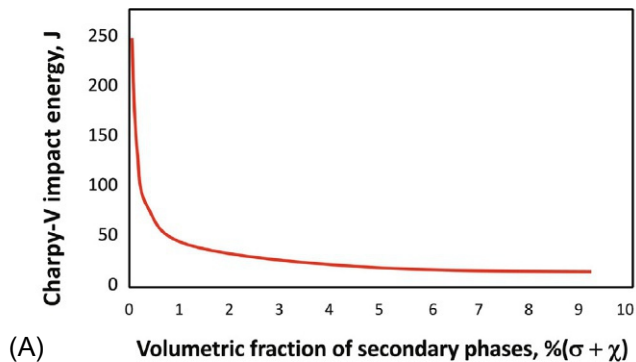


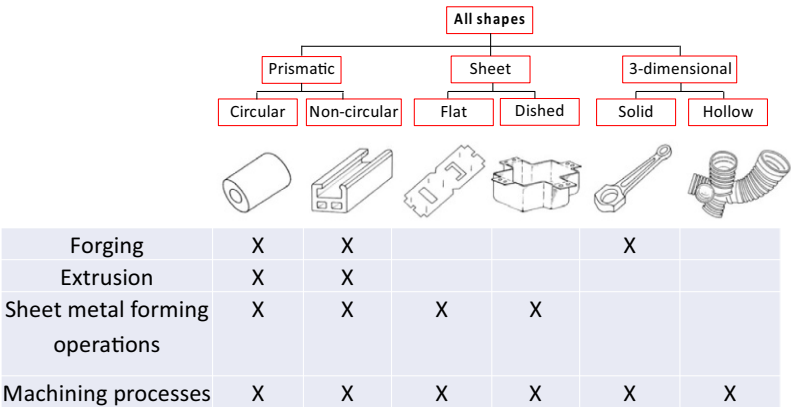
Fig. 9.13 (A) CVN impact energy as a function of content in intermetallic phases, $\chi + \sigma$ in vol%; voids nucleation along the austenitic/ferritic interfaces, in the case of (B) very small precipitation of intermetallics, and (C) in the case of precipitation of 0.5 vol% of intermetallics. From Calliari, I., Straffelini, G., Ramous, E., 2010. Investigation of secondary phase effect on 2205 DSS fracture toughness. *Mater. Sci. Technol.* 26, 81–86, reproduced with permission of the publisher.

observed in DSS even at RT if they are exposed to temperatures over 300°C. In particular, in the interval from 600°C to 1000°C, the kinetic of precipitation of χ - and σ -phases (and other intermetallics, such as nitrides and carbides) is rather high, depending on the chemical composition of steel. As shown in Fig. 9.13A, referring to the very common 2205 DSS, even a small precipitation of these secondary phases induces a marked drop in the impact fracture energy (Calliari et al., 2010). When the content of secondary phases is very small, the fracture energy is greater than 200 J, and the fracture is completely ductile. Fig. 9.13B shows a cross section close to the notch root, showing gross plastic formation with nucleation of voids at the austenite/ferrite interfaces, due to the presence of the precipitates. When the fraction of secondary phases increases to 1 vol%, the fracture energy drops to 50 J. A very large fraction of voids nucleates at the austenite/ferrite interfaces, as seen in Fig. 9.13C. When the fraction of secondary phases is greater than 1 vol%, the fracture energy is rather low, and fracture is almost brittle, by cleavage through the ferritic areas.

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Safe plastic forming of metals: General aspects



Typical shapes of end products that can be obtained with traditional plastic forming processes (modified from [Ashby, 2005](#)). Two other important processes are rolling and wire drawing. With these processes, mainly standard products, to be processed further, are obtained.

Chapter outline

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Ductility is a special property of metals that allows them to be shaped easily, obtaining products with the required (often complex) geometry. The main plastic shaping processes are:

1. Forging;
2. Rolling;
3. Extrusion;
4. Wire drawing;
5. Sheet metal forming.

In every case, a rigid tool plastically deforms a metal workpiece to a certain strain, and at a given temperature and strain rate. The workpiece must have proper ductility to obtain parts free from defects, such as ductile cracks that are the most dangerous defects, or excessive localized (shear) deformation.

Shaping operations also include shearing processes and machining operations by chip removal. In these cases, a low workpiece ductility is taken as an advantage, since ductile fracture favors the process. In this chapter, the general aspects regarding the plastic working processes in safe conditions are outlined, with particular emphasis on the metallurgical characteristics of hot and cold processes. Specific details on the shearing process and machining operations are provided in Section 15.2 and Chapter 16.

The general concept of formability by plastic deformation (and formability window) is introduced. Metal formability is defined as the ease with which a metal is properly shaped through plastic deformation. It is first linked to ductility, i.e., to the necessity to avoid any crack formation by ductile fracture and also to avoid excessive plastic localization. It is also linked to the flow strength of metals, i.e., to the necessity of properly carrying out the process with the capability of the available equipment. Hence, to understand the formability of metals and to design a specific process, the knowledge of the plastic behavior of metals must be coupled with a mechanical analysis of the process, aimed at obtaining the strain and stress distributions (and, in some cases, the involved temperature rise).

In the next chapters, the topics will be presented following the same pattern:

- *First, the main process methods are introduced;*
- *Then, a mechanical analysis is outlined for the evaluation of the involved strains, strain rates, and temperature rise;*
- *At the end, the main types of ductile failures are outlined, with reference to the most important alloys that are used in every specific process.*

For the mechanical calculations, different commercial software packages are currently available, with applications of the finite element method (FEM). However, in the next chapters the basic mechanical aspects of the processes are introduced, since this knowledge is pivotal for using such computer programs with the proper awareness and with the capability of introducing any required improvements.

10.1 Classification of main processes

A metallic *end product* is generally made starting from *standard products* of simple shape, realized in the *production industries*, where the raw materials (minerals or scrap) are processed in different stages, as seen in Fig. 10.1.

The standard products are produced from the intermediate products, which are directly obtained from the solidification of the melt. Plastic shaping processes are involved in the production of the standard products from the intermediate products, and also in the realization of the end products from the standard products.

Table 10.1 summarizes the main geometrical characteristics of the *intermediate* and *standard* products made of **steels**. *Ingots* are generally big intermediate

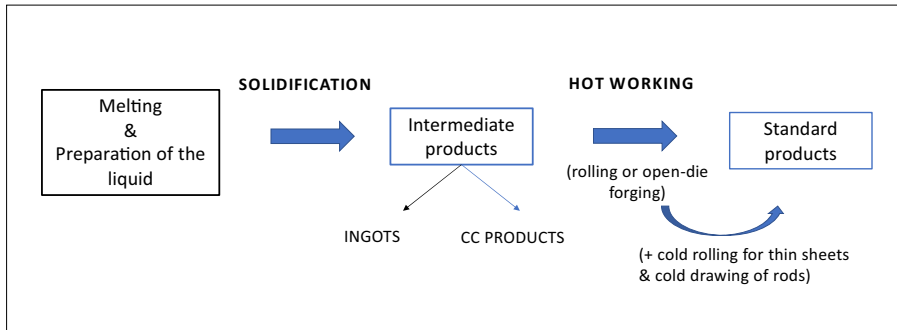


Fig. 10.1 The process from raw materials to standard products realized in the production industries.

Table 10.1 Intermediate and standard steel products.

Intermediate products	Shape	Dimensions
Ingots (by ingot casting)	Square Round Polygonal	Side (or diameter) usually larger than 500 mm
Billets (by continuous casting, CC)	Square Rectangular	Side: 130–160 mm Area < 260 cm ²
Blooms (by CC)	Square	Side: 300–500 mm
Slabs (by CC)	Rectangular	Minor side: 180–250 mm
Rounds (by CC)	Round	Diameter: 140–500 mm
Standard products	Shape	Dimensions
By flat rolling from slabs: Sheets ^a Plates	Rectangular Rectangular	Thickness < 3 mm Thickness: 3–10 mm
By long rolling from blooms or billets: Rods (wires ^b) Shapes	Round Polygonal rectangular	Diameter: 4.5–130 mm

^a Thin sheets are then obtained by cold rolling.

^b Wires are then obtained by cold drawing.

products. They are submitted to preliminary *roughing* hot processes (by rolling or open-die forging) to reduce the section before any subsequent operation to obtain standard products (in specific cases, they may also be directly used to produce large forgings). *Continuous casting* (CC) intermediate products are smaller than ingots and therefore are directly transformed into standard products by hot rolling. Flat laminates, such as sheets and plates, are produced from the slabs. To obtain thin sheets, cold rolling followed by recrystallization annealing is also required after

hot rolling. Long laminates are obtained from blooms and billets, by the means of successive passes in grooved cylinders. In this way, it is possible to obtain bars of various sections, rods, and shapes (T, L, double T beams, rails). Round intermediate products can also be used for extrusion processes. Round rolled bars can be used for subsequent wire drawing processes, carried out in cold conditions (Kalpakjian and Schmid, 2009).

In the case of **Al alloys**, the intermediate products are obtained by *direct chill vertical casting*, a semi-continuous process (Eskin, 2008). The ingots may be round for subsequent extrusion or forging processes, or rectangular for subsequent flat rolling operations. In the case of **Ti** and **Mg alloys**, round ingots are typically obtained by double or triple arc re-melting and solidification. The processes are carried out under vacuum to avoid any contamination of the melt from the surrounding atmosphere. The round ingots are then mainly used for subsequent forging or extrusion operations (Polmear et al., 2017).

The standard products are then supplied to the *manufacturing industries* where the end products are obtained by the means of plastic deformation and/or metal removal processes. In this way, objects of any shape can be obtained: prismatic, flat, and three-dimensional. Very often, additional operations are required such as:

- Welding (when different parts must be assembled to obtain complex or large objects);
- Heat treatment (if particular mechanical properties are to be conferred to the metal);
- Surface treatment (for improved corrosion, fatigue, or wear resistance).

In specific cases, mechanical parts can also be made with other processes, such as foundry and powder metallurgy, which are not considered here.

10.2 Forming principles

During any plastic forming process, the workpiece is first heated to the selected process temperature (if required, i.e., in the hot or warm processes). Subsequently, a properly designed tool gives a *specified plastic deformation* to the workpiece (determined by the geometry of the part to be produced), at a *given strain rate* (determined by the displacement velocity of the tool and the system geometry). Of course, the temperature, strain, and strain rate change in the workpiece from point to point, and in general, reference is made to the average values or to the highest values attained in the most critical positions of the workpiece. The resulting *stress* distribution, and in particular the pressure distribution at the workpiece-tool interface, is what the tool must apply to impart such strains and strain rates. Increasing the contact pressure induces an increase in the required force that the tool must apply (and an increase in the required power). Additionally, it also induces an increase in the die wear, depending on the tool geometry and lubricating conditions.

To clarify further the forming principles, we consider, for example, the case of *flat rolling*, schematized in Fig. 10.2. In this process, the height, H_o , of the workpiece is reduced to H_f by the compressive deformation exerted by two counterrotating rolls

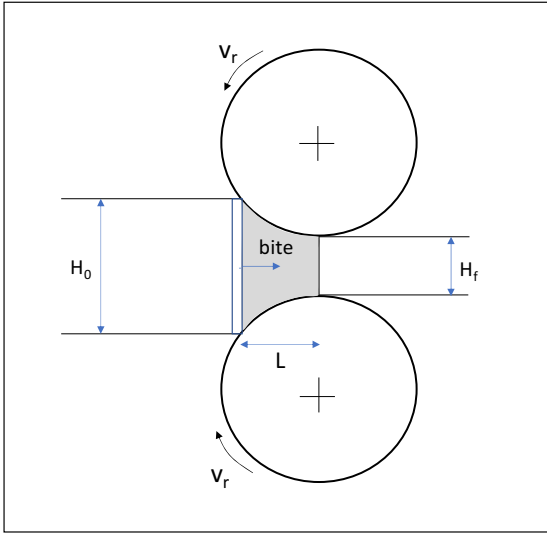


Fig. 10.2 Representation of plastic deformation in a flat rolling operation. Plastic deformation by compression occurs in the bite, i.e., in the gap between the two rotating rolls (further details on the process are given in [Chapter 12](#)).

(that are the tools). The applied true average compressive plastic strain is given by the following relation:

$$\varepsilon = \ln \left(\frac{H_f}{H_0} \right) \quad (10.1)$$

and the total *reduction* is: $\Delta H = H_0 - H_f$. ε represents the strain applied to each metal slice that enters the rolls gap with a thickness H_0 and goes out with a thickness H_f (ε is negative but it is often considered as positive for simplicity). Correspondingly, the average strain rate imposed on the workpiece by the rolls is given by:

$$\dot{\varepsilon} = \frac{|\varepsilon|}{\Delta t} = \frac{\ln \left(\frac{H_0}{H_f} \right)}{\frac{L}{v_r}} = \frac{v_r}{L} \cdot \ln \left(\frac{H_0}{H_f} \right) \quad (10.2)$$

where Δt is the time interval required for the deformation of the metal slice. This is approximately given by the ratio of the length of the contact arc, L (which depends on the total reduction and the radius of the rolls) to the tangential speed of the rolling cylinders, v_r .

Therefore the imposed strain and strain rate depend on the system geometry and the operating parameters. The operating parameters (or process variables) can be directly controlled by the operator. They are:

1. Temperature (such as the workpiece preheating temperature);
2. Applied deformation (such as the rolling reduction);
3. Tool velocity (such as the surface velocity of the rolling cylinders).

Table 10.2 Typical values of strain rate and total plastic strain in the various shaping processes by plastic deformation.

		Typical strain interval (for a single pass)	Typical strain rate interval (s^{-1})
Rolling	Hot, warm, or cold	0.1–0.6	$1-10^3$
Forging	Hot, warm, or cold	0.1–1	$1-10^3$
Extrusion	Hot	2–6	$0.1-10^2$
Wire drawing	Cold	0.05–0.5	$1-2 \times 10^4$
Sheet forming	Cold	0.1–0.5	$1-10^2$

Relationships such as Eqs. (10.1) and (10.2) are obtained for all the other manufacturing processes as well. Table 10.2 summarizes the typical values of plastic strain and strain rates in the various processes by plastic deformation.

The process design procedure develops in two stages. First, the stress distribution at the workpiece-tool interface is calculated, followed by the load to be applied by the tool. Indicating the loading direction with “1,” the applied load is given by: $\sigma_1 A_n$, where σ_1 is the (mean) plastic stress in direction 1 and A_n is the nominal area of contact between the workpiece and the tool (perpendicular to the loading direction). The proper constitutive equation must be used to evaluate σ_{eff} and then σ_1 (considering temperature, strain rate, triaxiality and tool-workpiece friction). The design condition is $\sigma_1 < \sigma_{lim}$, where $\sigma_{lim} = F_{max}/A_n$ and F_{max} is the maximum allowable load given by the capability of the equipment (a similar relationship is obtained for the maximum power).

The second design stage involves the verification of the conditions for the safe production of parts, without strain localizations, ductile damage (such as excessive voids nucleation), and/or ductile cracks. In general, intermediate or standard products contain *solidification defects* (such as surface transversal cracks) that could easily propagate during the forming process. These defects can be easily recognized (often by visual inspection) and can be eliminated by removing the surface layers before proceeding with the deformation process (e.g., partial or complete grinding of the workpiece surface).

However, even if the workpiece does not contain appreciable defects (because they are absent or they were removed), ductile damage may occur during the forming process, depending on the process parameters (such as strain, strain rate, state of stress, expressed by triaxiality factor, T , temperature distribution in the workpiece, tool-workpiece friction), especially when the metal is characterized by a low ductility. For this, the use of a suitable ductile fracture criterion (such as the R-T criterion) is required, and the design condition for the production of sound products is: $\epsilon_{eff} < \epsilon_{efff}$ (in the most critical workpiece position). The overall approach is schematized in Fig. 10.3.

At the end of this section, it is noted that the grinding operations for the removal of surface defects can be rather expensive; in addition, sometimes they are not completely effective in removing the defects. Because of this, effort is being made in several investigations aimed at identifying the process conditions that would allow sound products to be obtained when starting with defective initial workpieces. The topic is important, but it will not be considered in detail in this or the next chapters.

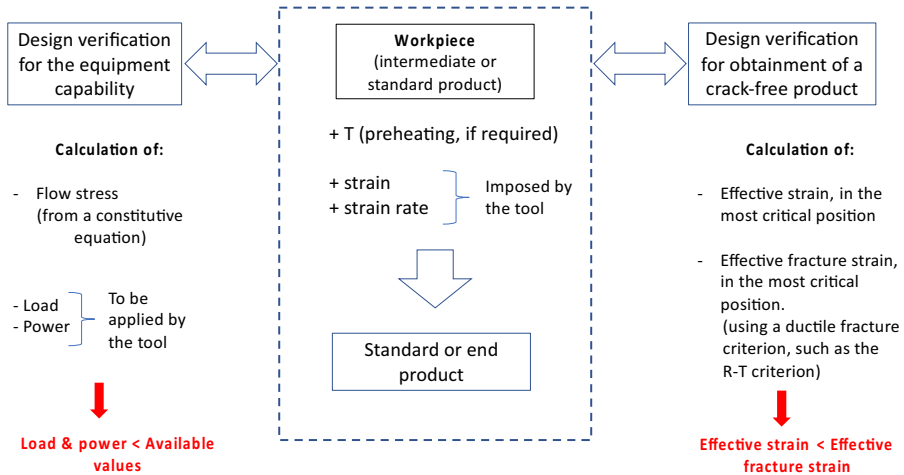


Fig. 10.3 Design flowchart for the safe production of an intermediate or end product.

10.3 Main metallurgical aspects

The plastic forming processes can be carried out in *cold*, *warm*, or *hot* conditions. In the case of hot and warm processes, the metal is heated to a temperature above $0.4\text{--}0.5T_m$ to exploit the softening provided by the creep mechanisms (typically dislocation creep) and the high-temperature ductility (attained closed to the PDT). In general, we properly refer to hot working when dynamic or static recrystallization (DRX or SRX) occurs over the working process. Warm working occurs when creep deformation is active but without any recrystallization. In the case of cold working, the initial temperature of the metal is below $0.4\text{--}0.5T_m$ (in most cases RT), and the metal undergoes strain hardening during the plastic deformation. Hot working processes are adopted to produce complex-shaped products and/or when high-strength alloys with comparatively low ductility have to be shaped. After hot working, the products often do not possess the required strength and must be subjected to additional processes (such as heat treatment). In cold working operations, the involved strains are much lower than in hot working, because strain hardening increases the metal flow strength. After cold working, the obtained part does not need additional strengthening treatments.

Hot working

Hot working operations involve the transformation of intermediate products into standard products (Fig. 10.1), and are also used in specific transformations of a standard product into an end product (or near to an end product).

The intermediate products (ingots and CC products) have a particular cast microstructure, characterized by the presence of:

- Equiaxed and columnar grains;
- Micro-segregations;

- Central porosity and central macro-segregation;
- Primary intermetallic phases in the inter-dendritic/inter-columnar regions;
- Surface cracks (longitudinal and transverse cracks).

Transverse surface cracks are particularly dangerous for subsequent operations (they can be filled with surface oxides that remain entrapped in the surface acting as crack initiation sites). They mainly form during continuous casting because of the strains applied during the unbending of the strand. As said, grinding operations are often required to remove the surface layers containing the cracks, before proceeding with the subsequent operations.

The first working step for intermediate products (realized by open-die forging or rolling) is the **hot roughing process**. The process is aimed at reducing the cross section of the product and, especially, at the *breakdown* of the solidification (as-cast) microstructure. This is achieved by:

- Changing the grain morphology and size;
- Reducing the micro- and (partially) the macro-segregations;
- Closing the internal porosity (completely or partially) and small surface cracks;
- Reducing the content of intermetallics (promoting their dissolution in the matrix) and/or modifying of their shape (especially in Al alloys).

For this, the workpiece is heated to a process temperature that is close to the PDT. This is a *standard process*, selected to deform the metal with the lowest possible flow strength (to minimize the work load and power) and the highest possible ductility, considering that the ductility of metals with a solidification microstructure may be rather low (Sections 8.3 and 8.4). As a broad indication based on experience, metals with a reduction of area at fracture, RA%, in the range of 40%–50% (ϵ_f in the range of 0.5–0.7) display an acceptable ductility in most cases, and metals with RA in the range of 50%–60% (ϵ_f in the range of 0.7–0.9) display a good ductility and can be submitted to forging or rolling operations without problems (Nicolaou et al., 2003).

In addition, the selection of the process temperature must consider the effects of surface oxidation on the workpiece. In steels, surface oxidation involves the formation of a harmful oxide scale, which must be removed before any subsequent processes. Surface decarburization is also associated with oxidation since C reacts with O as well. In the case of Ti alloys, surface oxidation produces an oxide scale and an oxygen-enriched, hard and brittle surface case, which must be removed before the final machining operations. In the case of Al alloys, any intense surface oxidation is absent. These alloys do not develop a surface scale during heating, since they have prior protection from a thin (albeit hard) alumina film. Therefore, if the metal ductility is sufficiently high, the preheating temperature should be reduced to minimize the metal loss due to oxidation. The heating time must be sufficiently long to allow the thermal homogenization of the workpiece. However, a very long heating time must be avoided to reduce excessive costs and surface oxidation.

In steels, preheating induces the transformation of the as-cast structure into new austenitic grains. The average austenite grain size depends on temperature and time, the temperature being the most important parameter. In HSLA steels, the presence of small precipitates (carbonitrides) in the microstructure provides grain boundary

pinning, thus preventing the austenitic grain growth. However, if the preheating temperature is sufficiently high, the precipitates decompose, and they are no longer able to prevent grain growth.

In Al alloys, no phase transformation occurs during heating to the process temperature and during cooling down to RT after the process. However, the as-cast structure is not suitable for direct hot working, and a *homogenization treatment* (typically at 500°C for some hours) is required. In fact, the as-cast microstructure generally contains several intermetallics and low melting point eutectics that significantly reduce the ductility of the alloys (see [Section 8.4](#) and [Fig. 8.16B](#)). Therefore these particles must be partly dissolved in the matrix or reduced in size through the homogenization treatment.

Ti alloys are submitted to rough breakdown processes by alternate forging above and below the β transus. This process has the target of breaking up the cast microstructure, thus obtaining a fine microstructure ([Leyens and Peters, 2003](#)).

[Table 10.3](#) provides an indication of the typical preheating temperatures of steels and other metals in the case of a standard process.

The first (roughing) working step is typically carried out in multiple stages, with possible intermediate reheats, and is completed by cooling the product to RT. When working with big ingots, it might be necessary to make intermediate stops during heating, to avoid the establishment of steep temperature gradients in the ingots and therefore possible failures.

The subsequent **hot finishing process**, aimed at producing the intermediate products, and the **hot or warm processes** involved in the production of the end products, are also typically carried out in a series of multiple passes at successively lower temperatures and, generally, higher strain rates.

In any case, the control of the involved metallurgical phenomena, comprising DRV, DRX or MDRX, and SRX, is very important for minimizing the flow stress and maximizing the workpiece ductility at each pass, and, in the case of the finishing processes, for obtaining a final product in the proper microstructural state.

As seen in [Chapter 4](#), metals with a low SFE easily undergo DRX during plastic deformation at high temperatures. DRX is then primarily observed in steels that are austenitic (fcc) at the process temperature, Cu and Ni alloys, and in hcp metals such as Mg alloys. DRX reduces the required force for plastic deformation and generally induces grain refinement. DRX is generally easier in hot processes that are carried out at higher temperatures. When DRX is not realized, static recrystallization

Table 10.3 Typical preheating temperatures for steels and other metals in the case of standard hot working processes.

Alloy	Temperature (°C)	Alloy	Temperature (°C)
C steels	1050–1280	Al alloys	400–550
Stainless steels	1100–1250 ^a	Cu alloys	700–900
Ti alloys	700–950	Mg alloys	250–350

^a Up to 1290°C in resulfurized steels.

(SRX) may occur in the interpass times between two subsequent passes. In this case, SRX reduces the metal flow strength and the force (therefore, power) to be applied in the subsequent pass (and practically plays the same role as DRX). The occurrence of SRX in the interpass time is rather typical in the finishing processes. DRX is slow in high SFE alloys such as Al alloys, for which it can be further retarded by the possible presence of dispersoids or precipitates (McQueen, 2002). In multistage processes, the plastic strain accumulates, and it may eventually reach the conditions for DRX (during deformation) or SRX (in the interpass period) in the last stages. In some cases, the lowest allowable preheating temperature is the one able to induce recrystallization in the selected working conditions.

A summary of the main metallurgical phenomena occurring in selected metals is provided in Table 10.4.

Table 10.4 Summary of the main metallurgical phenomena occurring in different metals in hot working processes.

Steels	In forging, it is typical to have DRX or MDRX, with subsequent grain growth in the interpass time between two operations. These phenomena may also be present in the hot rolling of long products. In flat rolling, it is typical to have only SRX in the interpass time. In some finishing processing cycles, the last passes are carried out at a low temperature to avoid SRX and obtain a very fine grain size of the transformed microstructure after air cooling to RT (these treatments are called <i>thermomechanical treatments</i> , or <i>treatments at a controlled temperature</i>)
Al alloys	Due to their high SFE, they do not typically undergo DRX during hot working (there are some exceptions). These alloys may undergo SRX with grain growth in the interpass times. The process should be carefully controlled to optimize possible crystallographic textures. In the case of flat hot rolling, SRX is often undesired, since it is preferred to obtain a flattened recovered microstructure, avoiding any formation of coarse grains. Even after extrusion (a one-pass process), SRX with grain growth should be avoided, so as not to obtain very large grain sizes and penalize the material strength. This is done by ensuring rapid cooling immediately after the process. Such rapid cooling is also important to obtain a solution hardening treatment
Ti alloys	DRX is generally not observed in the β phase (with bcc structure), i.e., above T_β , whereas it is rather easy at low temperature, i.e., in the (hcp) α phase. However, DRX is not used to reduce the working stresses. In the hot working operations (after the breakdown forging), emphasis is given to the control of SRX to avoid excessive grain growth, and to the subsequent heat treatment. For example, the widely used α - β Ti-64 alloy is typically forged in the α - β region (30–100°C below T_β to have sufficiently high ductility, Fig. 8.6). Thereafter, it is solution annealed above T_β and then cooled to RT to obtain the desired microstructure (often the typical lamellar microstructure)
Mg alloys	DRX is rather easy during hot extrusion (a common process for these metals). To avoid excessive grain growth after extrusion, in some cases, a water jet is used to force cool the extrudate immediately after its exit from the die

To select the right preheating temperature and control any possible microstructural transformation, the temperature evolution of the workpiece with time should be properly controlled. First, the transfer from the heating station to the working station should be sufficiently quick to limit the heat transfer with the environment. Second, the heat transfer with the tools must also be considered (it is particularly important at low strain rates, i.e., at long process times), together with the temperature rise due to the plastic deformation. As already mentioned, in the case of multiple passes, intermediate heating stages should possibly be carried out.

As an example, Fig. 10.4 shows the results of a finite element analysis to evaluate the temperature evolution in a cylindrical billet made with C-Mn steel subjected to hot forging (Jang et al., 2000). The preheating temperature is 1200°C, and the billet is transferred in 10 s from the furnace to the lower die that is preheated to an initial temperature of 350°C. The dwell time before forging is 4 s. During this period, the billet is cooled through the heat transfer with the environment and with the lower die (refer to Fig. 10.4A). The billet is then deformed with a reduction in height of 40%. It is possible to see that in the regions near the upper and lower dies, the metal temperature is rather low (Fig. 10.4B). The authors observed DRX only at the center of the workpiece, where the temperature remained sufficiently high. The lower formability of

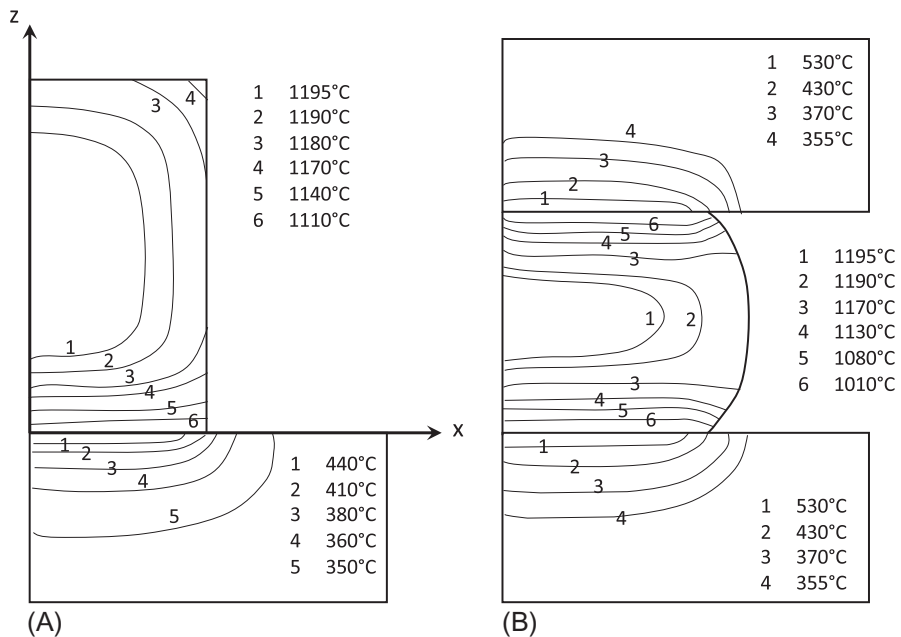


Fig. 10.4 (A) Calculated distribution of temperature in the billet and in the lower die along the axisymmetric plane before deformation, and (B) after a height reduction of 40%, i.e., $\epsilon = 0.33$. Redrawn from Jang, Y.S., Ko, D.C., Kim, B.M., 2000. Application of the finite element method to predict microstructure evolution in the hot forging of steel. *J. Mater. Process. Technol.* 101, 85–94.

the metal close to the die, together with the friction at the workpiece-die interface, reduces the metal radial displacement, inducing the phenomenon of *barreling*, which is explained in detail in the next chapter. To avoid or reduce the metal cooling close to the die, the die can be preheated and maintained at a temperature close to that of the workpiece (*isothermal deformation*).

As seen, metal *formability* is simply defined as the *ease with which a metal is properly shaped through plastic deformation* (Dieter, 1988). It is a technological property that considers the design principles (introduced in the previous section) and the metallurgical properties (introduced in this section) to evaluate the attitude of an alloy towards plastic shaping. In general, the hot formability (or workability) of metal is experimentally determined by hot tensile or compression tests (with on cooling tests), aimed at establishing the metal ductility and the metal flow stress as a function of temperature and strain rate. In some cases, specific technological tests are also carried out. As explained in Chapter 11, a very important role is played by the upsetting (compression) test, which is a standard practice to simply evaluate the hot formability of metals.

Table 10.5 provides a list of metals with a broad classification of their hot formability. Further details are provided in the next chapters, focusing on the alloys with the worst performances.

Table 10.5 Hot formability of some engineering alloys.

Hot ductility	Flow stress		
	Low	Intermediate	High
Good	- Low and medium C steels (included type 400 stainless steels) - Class H tool steels - Low alloyed Al alloys (such as 6061) - Cu alloys	- Type 304 austenitic stainless steel - α-β Ti alloys (such as Ti-64)	- Special purpose stainless steels (e.g., Fe-based superalloys)
Fair	- Mg alloys - High alloyed Al alloys (such as 7076)	- Precipitation hardening stainless steels - Class M tool steels - Type 314 austenitic stainless steels	- Ni alloys - High-strength Ti alloys (such as some β alloys)
Low	- Alloy steels - Resulfurized steels - Lead Cu and Al alloys	- Duplex stainless steel	- Superalloys

Adapted from ASM Handbook, 1991. Vol. 2 – Properties and Selection: Nonferrous Alloys and Special-Purpose Materials, ASM International, and other sources.

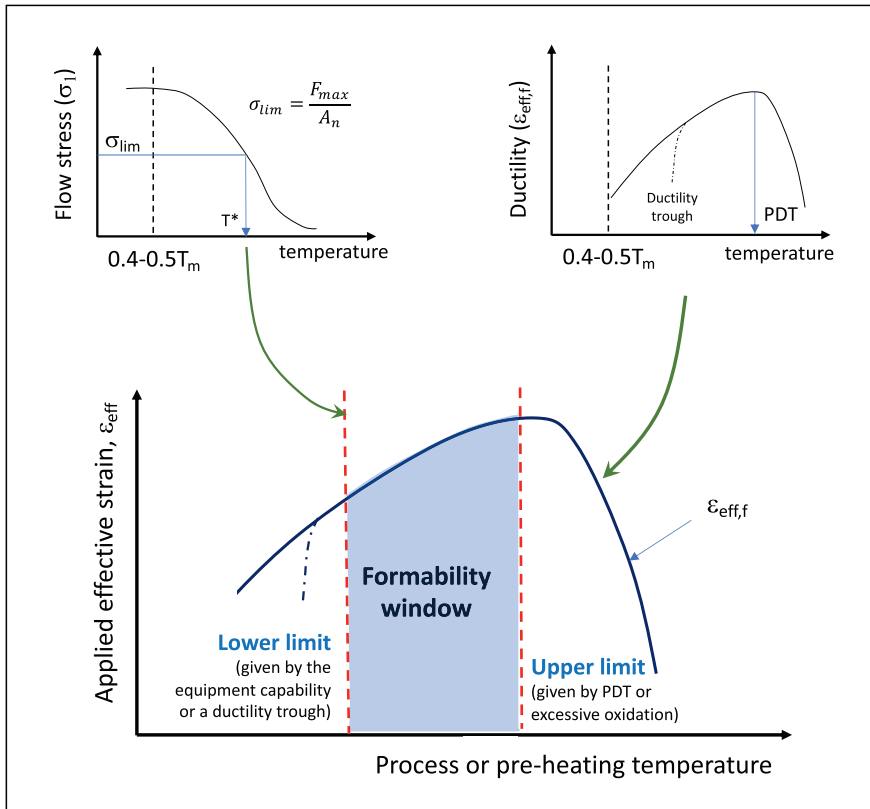


Fig. 10.5 Schematization showing the concept of formability window in hot working. In some cases, the lower temperature limit is given by the lowest temperature for recrystallization. Please note that $\epsilon_{\text{eff},f}$ also depends on strain rate and triaxiality.

Fig. 10.5 shows a schematization of the *process window* in the case of hot working. The diagram indicates the temperature interval for the safe conduction of the process, for a given material and strain rate. The maximum possible effective strain in the process window is also displayed.

Cold working

In cold working processes, the metal grains are flattened and elongated in the deformation direction, with the metal undergoing strain hardening. In the case of homogeneous deformation, the local strain in the microstructure is the same everywhere and equal to the macroscopic strain, ϵ (see next section). It is then expected that the thickness of the grains after deformation is related to ϵ (to be taken with a negative magnitude in the case of compression):

$$d_f = d_0 \cdot e^{\epsilon} \quad (10.3)$$

where

- d_0 is the mean grain size before deformation;
- d_f is the mean thickness of the grains in the direction perpendicular to plastic deformation.

This is an approximate model, especially in the case of heavy plastic deformation, characterized (as seen in [Chapter 2](#)) by intense dislocations multiplication, the formation of small cells and subgrains (including geometrically necessary boundaries), that induce a pronounced refinement of the microstructure.

As an example, [Fig. 10.6A](#) shows the grain boundaries of a pure Cu bar before (left) and after (right) a heavy plastic deformation with $\epsilon = -2.9$ (by compression), obtained

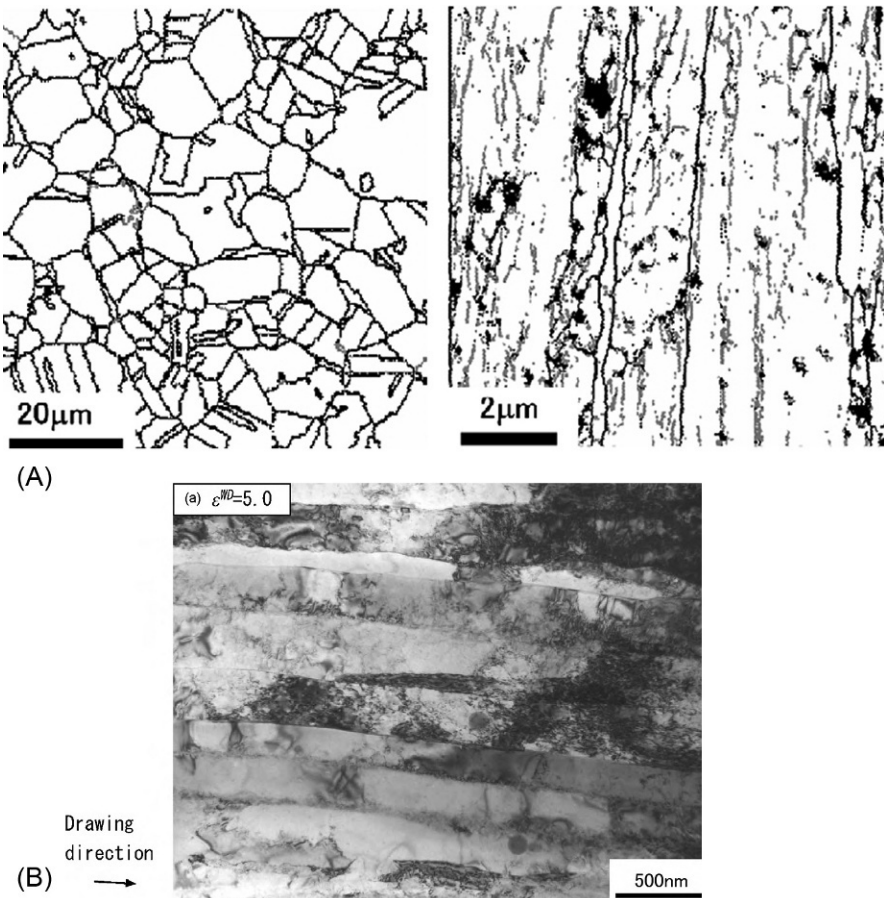


Fig. 10.6 (A) Grain boundary maps (obtained by EBSD) of a pure copper bar before (left) and after (right) a plastic deformation with $\epsilon = 2.9$ (by compression), obtained by cold drawing; (B) microstructure (obtained with TEM) of the copper wire subjected to cold drawing with $\epsilon = 5$. The arrow indicates the drawing direction.

From Hanazaki K., Shigiri N., Tsuji N., 2020, Change in microstructures and mechanical properties during deep wire drawing of copper. *Mat. Sci. Eng. A*, 527, 5699–5707, reprinted with permission of the publisher.

by cold drawing (Hanazaki, 2010). Before drawing, the grains have an average diameter of 9.4 μm . After drawing, the mean thickness of the elongated grains, perpendicularly to the drawing direction, is 2 μm . Fig. 10.6B shows the microstructure of a Cu wire that was drawn with a plastic strain of $\varepsilon = -5$. The grains are elongated in the working direction, and they contain several dislocation cells, in agreement with the fact that Cu has a comparatively high SFE (refer to Table 2.1).

As already explained in Section 2.7, plastic deformation also induces a fiber and crystallographic texture. The alignment of sulfides and other particles in the direction of plastic deformation (see, for example, Fig. 7.1) is an example of fiber texture. Another important example regards *pearlitic steels*, with a fully pearlitic microstructure (Fig. 7.2). During heavy cold plastic deformation, typically by cold drawing, the pearlite colonies rotate and align along the wire axis, as seen in Fig. 10.7. In addition, a reduction of the interlamellar spacing is achieved, accompanied by the formation of fine dislocations cells in the ferritic regions. The final interlamellar spacing depends on the initial spacing and on the reduction, following a relationship as in Eq. (10.3). The refinement of both the interlamellar spacing and the dislocation cells induces a marked increase in the material strength. It is also reported that after heavy cold working, the cementite lamella become unstable and decompose, favoring the diffusion of C into the ferrite. Subsequent annealing at low temperature results in the precipitation of nanocrystalline carbides that contribute to further increasing the strength of the wire (Wei et al., 2020). Values for the ultimate tensile strength up to 7000 MPa are reported in the case of thin drawn wires (Li et al., 2014), which is the highest obtainable strength in metals.

Strain hardening increases the strength of the end product. As seen in Chapters 2 and 3, different equations are available to obtain the flow stress of a metal as a function of the applied strain. The reference equation is the L-H equation (Eq. 2.7). Strain hardening also decreases ductility. If the process is carried out with a series of subsequent passes, the increase in the flow strength and the corresponding decrease in ductility can make the process very difficult. In some cases, intermediate annealing stages (with SRX) are then required.

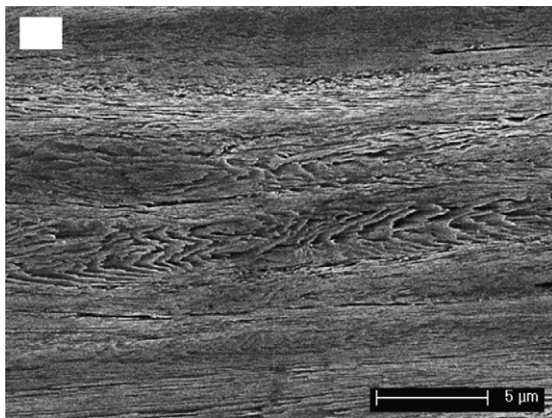


Fig. 10.7 Deformation of pearlite nodules after the cold drawing (plastic strain of 1.6). From Fang, F., Zhao, Y., Zhou, L., Hu, X., Xie, Z., Jiang, J., 2014. Texture inheritance of cold drawn pearlite steel wires after austenitization. Mater. Sci. Eng. A 618, 505–510, reprinted with permission of the publisher.

The analysis of the data in Figs. 7.31 (and in Table 2.3) is very useful to frame the cold formability of metals. Fig. 7.31 shows that the formability of Mg alloys is very limited, as already highlighted. These alloys are typically shaped at temperatures beyond 300°C (see next chapter). Ti alloys also display a rather poor cold formability. These metals are characterized by comparatively high strength and low ductility even in the annealed condition. On the other hand, Cu alloys display excellent cold formability.

In some cases, n changes during plastic deformation. This happens in metastable austenitic stainless steels and in TWIP steels. Therefore the data for these steels deserve a particular analysis. The increase in n during plastic deformation induces asignificant increase in the flow strength, and then in the required shaping load and power. This explains why metastable austenitic stainless steels are not easily cold worked (despite their good ductility).

10.4 Inhomogeneous plastic deformation and secondary tensile stresses

Fig. 10.8 schematizes the tensile plastic deformation of a prismatic sample. The macroscopic plastic strain is: $\varepsilon_t = \ln(l/l_0)$. Small elemental volumes (in the form of squares) are also drawn on the sample in three different positions (indicated with 1, 2, and 3), to show examples of local plastic deformation. The local strains are then indicated with ε_{t1} , ε_{t2} , and ε_{t3} . If $\varepsilon_{t1} = \varepsilon_{t2} = \varepsilon_{t3} = \varepsilon_t$, plastic deformation is *homogeneous*. If, in some

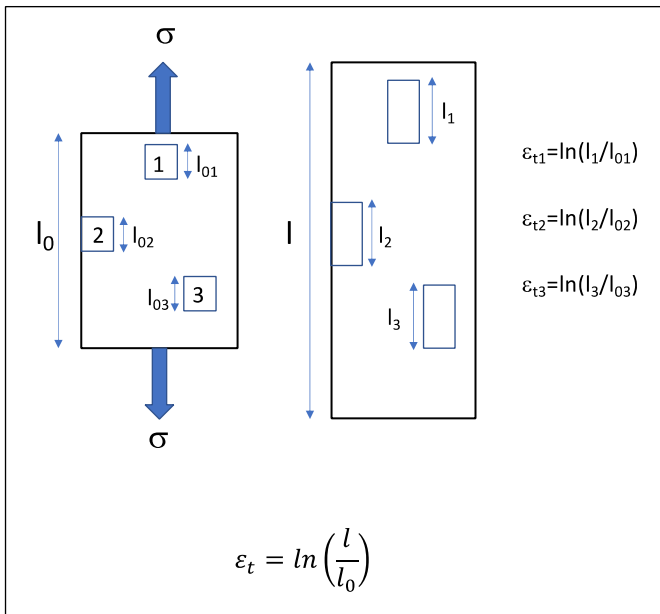


Fig. 10.8 Schematization showing the macroscopic and local deformations in the case of tensile deformation of a prismatic sample.

regions, the local strain is different from the strain observed in other regions, and it is also different from the macroscopic (average) strain, plastic deformation is *inhomogeneous*.

In shaping parts with complex geometry, plastic deformation develops in a rather inhomogeneous way. This is typically investigated using finite element method (FEM) or, experimentally, using the grid method. Some specific examples are seen in the next chapters. In the present section, simple deformation processes are analyzed, with the objective of identifying peculiar situations leading to an inhomogeneous plastic flow. The corresponding and interrelated consequences, concerning the appearance of (unexpected) secondary tensile stress and the establishment of conditions promoting ductile damage and fracture, are then introduced. This information plays a pivotal role in understanding the conditions for safe plastic forming processes, discussed in the next chapters.

The first case of inhomogeneous plastic deformation regards the plastic localization in shear bands, already analyzed in Sections 2.8 and 3.6 for sheet tensile deformation. Inside the shear bands, the plastic strain is greater than in the neighboring regions and is also greater than the macroscopic (average) value. We have seen that when $\varepsilon = n$, diffuse necking starts. In this case, the localization is limited and involves a gauge length that is of the same size as the specimen width. However, the situation changes when the conditions of localized necking are attained. In the presence of imperfections, and with simultaneous negative m values (as well as with low n values), shear deformation localizes spontaneously and rather early in narrow bands. In these regions, very high strains are reached. This condition cannot be generally accepted, since it quickly leads to the formation of ductile damage and *internal ductile cracks* inside the shear bands.

Shear localization can also occur during compressive deformation, even if no necking takes place in this case. A special role is played by barreling, already introduced in Fig. 10.4. This phenomenon is also shown in Fig. 10.9 and is extensively explained in the next chapter. Barreling is due to the presence of friction at the workpiece-die interface that reduces the radial flow of the metal, and, in the case of hot deformation, to the possible workpiece chilling at the contact with the dies (with a consequent reduction in the local deformability). Consequently, plastic deformation is inhomogeneous and displays a particular pattern with bulging at the equatorial region. As schematized in the figure (in the central cross section), the inhomogeneity in the plastic deformation induces the formation of regions of intense *forced shear*. In these regions, internal ductile damage and ductile cracks may form. In addition, mid-height circumferential (hoop) strains are rather high, and this induces the formation of unexpected surface tangential tensile stresses. This latter phenomenon is schematized by the top view of the compressed cylindrical specimen. These tensile stresses may be responsible for the formation of *surface ductile damage and ductile cracks*.

Inhomogeneous plastic deformation can also occur during compression, in the case of comparatively low reductions and depending on the geometry of the region under deformation. This is shown in Fig. 10.10, which shows the compressive deformation of a slab sample in frictionless conditions (and in plane strain for simplicity, i.e., with the assumption of no plastic deformation in the thickness direction). The reduction Δh is given by: $\Delta h = h_0 - h$, and the average height of the workpiece under deformation is:

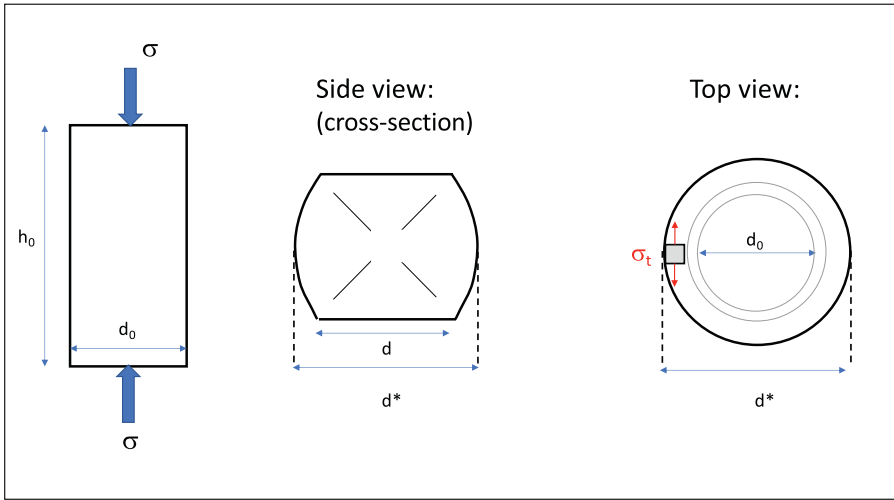


Fig. 10.9 Schematization showing the barreling effect in the compression of a cylindrical specimen. In the side view (cross section), the internal regions with forced shear are shown by the crossing lines; in the top view, the formation of mid-height circumferential (hoop) strains is indicated. Note that d_0 and h_0 are the initial diameter and height of the specimen; d and d^* are the current diameter of the specimen in the contact with the die and in the equatorial region.

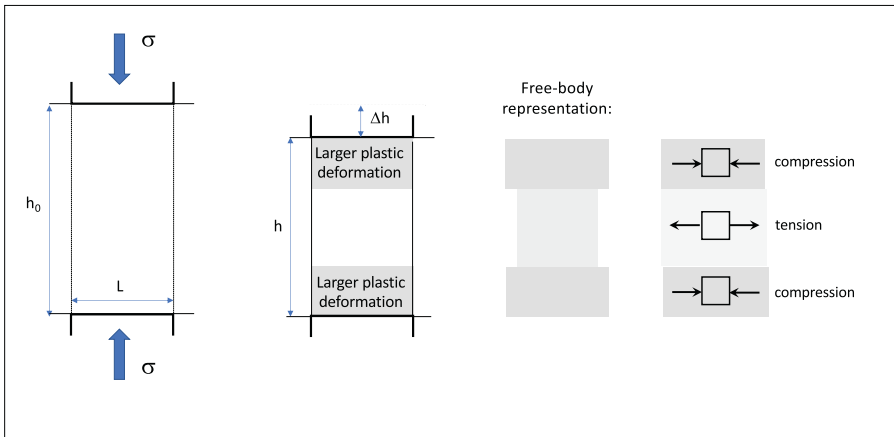


Fig. 10.10 Inhomogeneous plastic deformation by compression of a slab; free-body representation illustrating the formation of center secondary tensile stresses.

$\bar{h} = \frac{h_0 + h_f}{2}$. The length of the workpiece-die contact is constant, and equal to L . The geometry of the deformation zone is represented by the following parameter, Δ (Backofen, 1973):

$$\Delta = \frac{\bar{h}}{L} \quad (10.4)$$

which, for the geometry in Fig. 10.10, becomes:

$$\Delta = \frac{h_0}{2 \cdot L} \cdot (2 - r) \quad (10.5)$$

where r is the *reduction*, given by:

$$r = \frac{\Delta h}{h_0} \quad (10.6)$$

With increasing Δ , which typically occurs when the reduction decreases and the ratio h_0/L increases, plastic deformation is increasingly inhomogeneous and progressively localized in the regions close to the dies. If the material were free to deform, the plastic elongation of the metal close to the dies (and parallel to them) would be greater than in the center, as schematized in the free-body representation of Fig. 10.10. But for continuity reasons, this is not possible, and a strain accommodation occurs. The metal in the center tends to restrain the metal close to the dies, and the metal close to the dies tends to stretch the metal in the center. Consequently, the metal on the surface is submitted to compressive stress, whereas the metal in the center experiences tensile stress. Therefore, also, in this case, the appearance of unexpected tensile stresses occurs *during* plastic deformation, and they can be responsible for internal ductile damage or, at the limit, ductile cracking.

The dependence of the centerline tensile stress, σ_{axial} , and the average or mean stress, σ_m (both normalized to the yield stress) on parameter Δ during plastic deformation, as obtained by a slip-line analysis for a non-strain hardening material (Hill, 1950), is shown in Fig. 10.11. Note that the ratio σ_m/σ_Y is the triaxiality parameter T . The figure clearly shows that with increasing Δ (for $\Delta > 1$ and < 8.74) the tensile stress in the centerline increases (as broadly predicted by the simple analysis of Fig. 10.10), and the triaxiality factor, T , also increases.

This kind of analysis can be applied to several metalworking operations, including forging, rolling, extrusion, and wire drawing. It is very helpful in understanding the possible occurrence of internal ductile damage and cracking due to geometry-induced inhomogeneous deformation.

Finally, conditions of inhomogeneous deformation with the appearance of dangerous secondary tensile stresses can also occur in the case of high friction at the workpiece-die sliding interface under the action of a contact pressure. The situation is simply schematized in Fig. 10.12. Due to friction, the plastic deformation of the material close to the dies and parallel to the sliding direction is more limited than that of the material in the center. If the material were free to deform, the deformation of the metal close to the dies would be lower than in the center due to the braking effect exerted by friction. Because of strain continuity, a state of tensile stress is formed on the external regions during plastic deformation (in contact with the die surfaces), and a state of compressive stress forms in the internal regions. Also, in this case, the unexpected secondary tensile stresses may be responsible for the ductile damage and cracking of the workpiece.

Fig. 10.11 Dependence of the centerline tensile stress, σ_{axial} , and the average or mean stress, σ_m (both normalized to the yield stress) on parameter Δ , as obtained by a slip-line analysis.

From Gouveia, B.P.P.A., Rodrigues, J.M.J., Martins, P.A.F., 2000. Ductile fracture in metalworking: experimental and theoretical research. J. Mater. Process. Technol. 101, 52–63, reprinted with permission of the publisher.

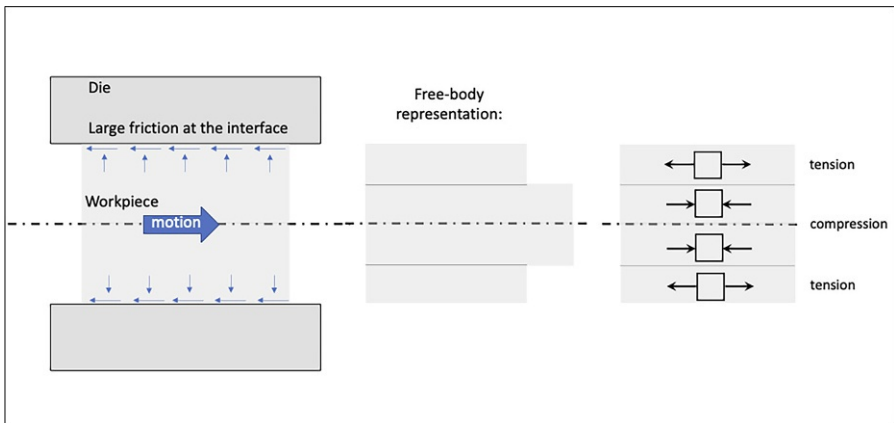
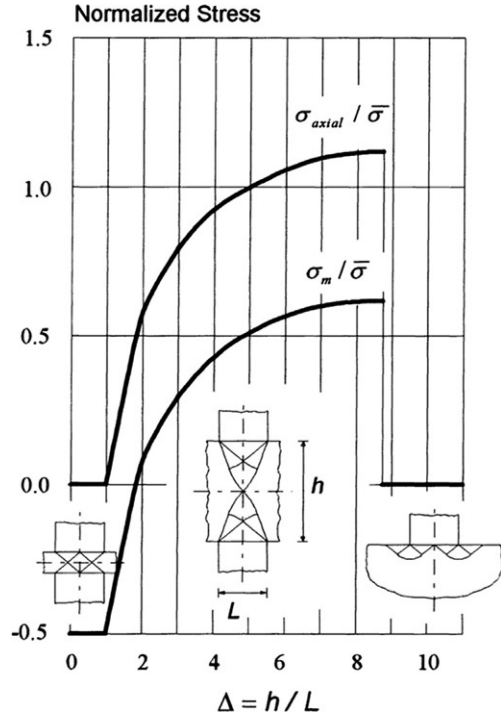


Fig. 10.12 Inhomogeneous deformation with the appearance of secondary tensile stresses in the case of high friction at the workpiece-die sliding interface.

To summarize, different sources of inhomogeneous plastic deformation can be responsible for the appearance of unexpected (secondary) tensile stresses, with a local increase in the triaxiality parameter, T . Of course, this does not necessarily mean that ductile damage and fracture will occur. This depends on the fulfillment of the ductile fracture criterion, such as the R-T criterion (Section 6.6). However,

Table 10.6 Summary of the possible origins of ductile damage and cracking in plastic forming processes following inhomogeneous plastic deformation.

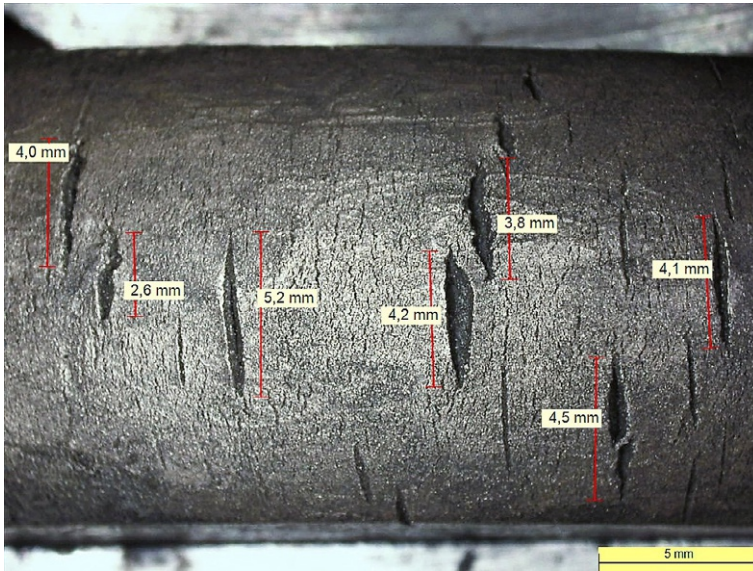
Internal cracks	Caused by spontaneous shear localization (mainly under tensile deformation, when $m < 0$ and n is low), and forced shear localization (mainly under compression, due to barreling)
Surface cracks	Caused by high Δ values (i.e., low reductions under compression) Caused by barreling Caused by high workpiece-die friction

plastic forming operations that typically take place in compression conditions are not intrinsically safe, as could be expected, even when surface or internal defects are not pre-existing in the initial workpiece. Special care is then required in the design of the forming process, considering the consequences of inhomogeneous deformation, and following the guidelines outlined in [Section 10.2](#). [Table 10.6](#) provides a summary of the possible origins of ductile damage and cracking in plastic forming processes.

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Hot upsetting at 1050°C of a cylindrical specimen made with a resulfurized steel containing bismuth, with the formation of surface cracks from barreling (Oliosio, 2021).

Chapter outline

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Forging is an old and very common forming process. It is carried out in compression and most often in hot conditions. Forging can be classified into open-die and impression-/closed-die forging. In the first case, the metal is compressed in the loading direction and is free to spread in the other two directions. In the second case, the lateral plastic flow is free only at the beginning of the process and the metal is then forced to flow into the die cavities to take on a certain shape. Parts of various

geometries and with comparatively good mechanical properties can be obtained by forging. Open-die forging is a very important process for the production industry, especially for transforming intermediate products into standard products. Impression- and closed-die forging processes are pivotal for the manufacturing industry, to produce high-quality near net shape or net shape end products with high mechanical properties.

After a brief introduction to the main forging processes, a mechanical analysis is outlined for the evaluation of the forging strain, strain rate, deformation path and corresponding stress triaxiality. Most of these concepts, including the so-called friction hill and barreling phenomena, are also useful in dealing with the rolling process in the next chapter.

The conditions for the formation of surface and internal cracks are then discussed. Subsequently, the concept of the formability map is introduced. For this, reference is made to the upsetting test. This test is able to realize different deformation paths, and can reveal the cracking attitude of a metal in different process conditions. The analysis is particularly focused on some alloys that are difficult to hot work, comprising Mg alloys, Ti alloys, and duplex stainless steels. The indications obtained from this type of test are useful for all plastic forming processes and, in addition, can serve as guidelines for the development of new alloys for plastic forming.

11.1 Forging processes

Fig. 11.1 shows the ingot *cogging* (or *lengthening*) process, used to obtain standard products such as plates, rounds, or other products from cast ingots. It is a typical hot open-die process, carried out in the production industry, such as in steelworks.

In the manufacturing industry, open-die, impression-die, and closed-die processes are performed, in hot as well as cold conditions. Hot open-die processes, such as cogging and *upsetting* (which is a simple compression process), are used to obtain components by successive operations, which can also include bending, drilling, and

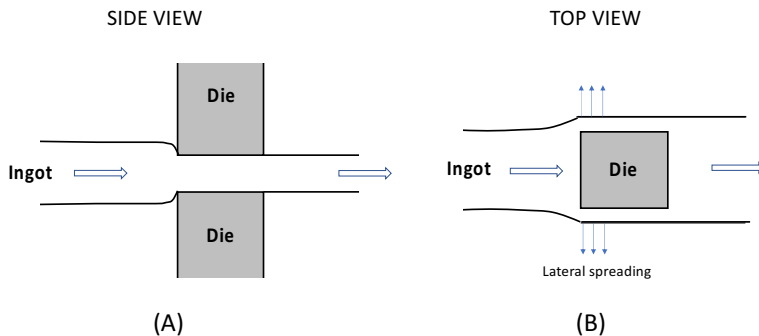


Fig. 11.1 (A) Cogging of an ingot (or a slab) to obtain a plate (a sequence of short bites along the length of the workpiece is carried out); (B) top view of the process showing the phenomenon of lateral spreading.

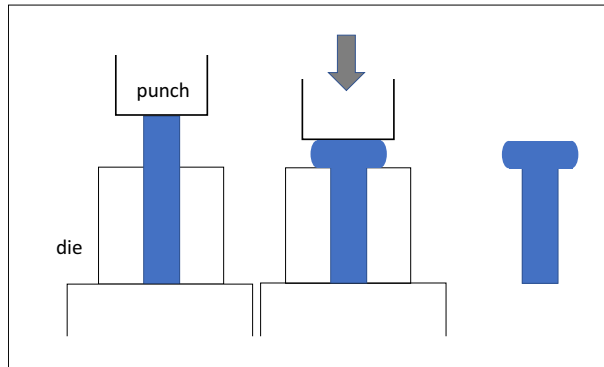


Fig. 11.2 Schematization of the heading process that involves upsetting of the free end of a rod to form a head.

shearing, with possible intermediate reheat stages in the case of hot processes. Generally, these components are large and they are produced in a limited number. Upsetting restricted to one end of a bar is also used to locally increase the diameter of the bar. The process (in hot or cold conditions) is called *heading*, and is schematized in Fig. 11.2. Open-die processes are also used to produce preforms (slugs) for subsequent impression-/closed-die forging operations.

Fig. 11.3A shows a typical sequence of impression-die forging. It depicts the obtaining of a flange in a hot process, carried out in a single step, also showing the flow behavior of the workpiece. In this kind of forging, an upper and a lower die are used, and they contain the flash gutter around the periphery of the lower die cavity. The flash acts as a *safety valve* for the excess metal in the die. Since the thin flash metal increases its flow resistance (due to local intense cooling), the flash also ensures that metal completely fills all die cavities. After cooling to RT, the burrs are removed by shearing (trimming).

In Fig. 11.3B, the impression-die process is compared with the closed-die process. In this latter case, the die is closed by the punch and no flash gutter is present. In this manner, products with higher accuracy are obtained. As an example, Fig. 11.3C compares two connecting rods obtained by impression-die and close-die hot forging.

In most cases, hot forging is conducted in two (or more) steps (after the obtainment of the preform). The first step is the *pre-forging* step, featuring the greatest change in shape, and the second step is the *finish forging* step, aimed at obtaining the final shape. At the end of the process, a heat treatment may be carried out, to provide the metal with the required properties. The pieces are then cleaned by sandblasting to remove the surface oxides and, if required, they are subjected to final machining operations to obtain the required tolerances and surface finishing. To reduce the process cost, the sole pre-forging step may be performed and the part is then submitted to extensive machining operations for the final shape. Alternatively, a preform with optimized geometry can be obtained by casting, and the process is then restricted to the sole finish forging step.

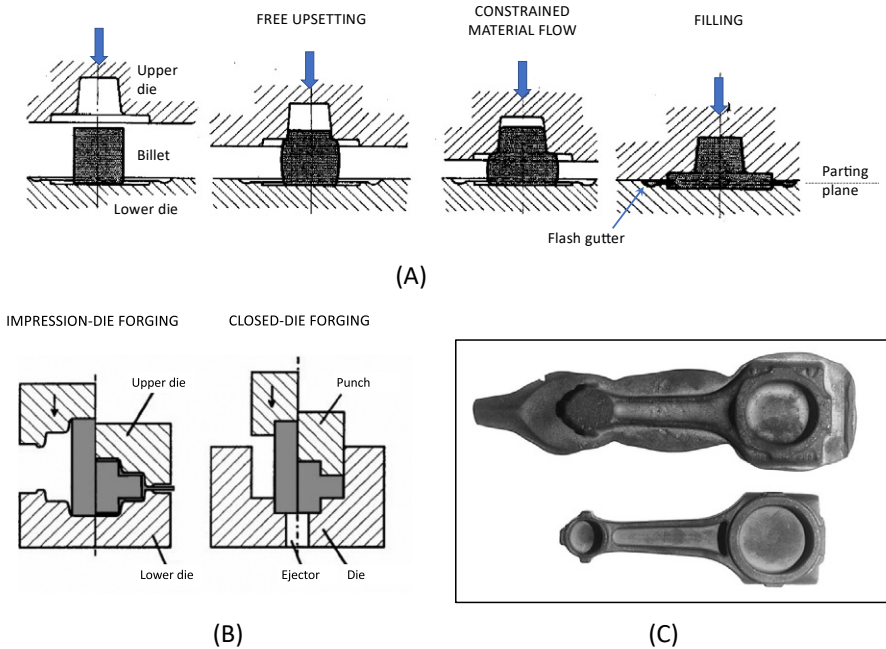


Fig. 11.3 (A) Impression-die forging sequence to obtain an axisymmetric flange; (B) comparison between impression-die and closed-die forging; (C) comparison between impression-die (top) and open-die (bottom) forged connecting rods. B and C: Obtained from Doege, E., Bohnsack, R., 2000. Closed die technologies for hot forging. J. Mater. Process. Technol. 98, 165–170, reprinted with permission of the publisher.

Several technological aspects must be considered in forging, including (Kalpakjian and Schmid, 2009):

Lubrication. It is used to reduce the metal/die friction and favor the metal flow into the die cavities. The reduction in friction also decreases the forging load and die wear. Lubricants for hot forging are: spray of graphite particles, grease with graphite particles, and glass particles. In some cases, such as in the forging of Ti alloys, ceramic glass pre-coats (applied by dipping or spraying processes) are used for lubrication and protection against excessive surface oxidation. Lubricants for cold forging are mineral oils, grease, oils containing graphite, or MoS_2 particles. The average friction coefficient depends on lubrication. It typically varies between 0.05 and 0.1 in the case of cold forging, and between 0.1 and 0.2 in the case of hot forging.

Corner radius and draft allowance. Die corners should not be too sharp to allow an easy flow of the metal in the cavity. Forging dies must be tapered to facilitate the removal of the finished piece. This *draft allowance* is approximately 5 degrees for steel forgings.

Forging machines can be divided into *hammers* and *presses*. Hammers are energy-restricted machines. In *gravity drop hammers*, the load is supplied by a falling weight,

or *ram*. Plastic deformation results from dissipating the kinetic energy of the ram. In *power drop hammers*, the ram is accelerated on the downstroke by steam or air pressure in addition to gravity. In *counterblow hammers*, two opposed rams are used. They strike the workpiece at the same time to result in all energy being absorbed by the workpiece and very little energy being lost as vibration.

Presses are force-restricted machines. In *hydraulic presses*, the oil pressure moves a piston in a cylinder, which generally pushes the upper die. In *screw presses*, the rotary motion of a flywheel is transformed into linear motion by the multiple threads on the large screw and its nut. In mechanical presses, the ram stroke is also restricted. In these presses, an eccentric crank translates rotary motion into reciprocating linear motion of the press slide. Table 11.1 summarizes the main features of hammers and presses.

In the selection of the process equipment for a specific forging process (or in the verification of the ability of existing equipment to carry out a given process), the maximum energy and/or force must be calculated, following the procedures outlined in the next sections. Additionally, it should be verified that the adopted equipment is able to provide the requested strain rate at the selected temperature. The proper strain rate and process temperature are selected based on experimental tests able to individuate the conditions at which metal ductility is maximum or at least sufficient to avoid the formation of ductile cracks, and the flow stress is adequately low (Fig. 10.5).

From a metallurgical viewpoint, in the case of hot forging, it is important to verify the occurrence of recrystallization. In fact, DRX increases the ductility and decreases the flow strength at high temperatures. DRX is favored in the case of large strains applied at high temperatures and at low strain rates. Considering strain rate, it is hence favored when using hydraulic presses. If DRX is not activated, SRX between the forging stages should occur. In general, forging conditions are then selected to promote DRX (or MDRX) or SRX in the interpass time, i.e., between two subsequent passes.

Table 11.1 Main features of hammers and presses.

Hammers	Maximum energy (kJ)	Impact velocity (m/s) (typical strain rate interval)	Presses	Maximum force (MN)	Impact (penetration) velocity (m/s) (typical strain rate interval)
Gravity drop hammers	50–120	3–5 (10–200 s ⁻¹)	Hydraulic presses	2.2–143	0.06–0.3 (10 ⁻² –1 s ⁻¹)
Power drop hammers	1150	5–9 (10–200 s ⁻¹)	Screw presses	1.3–280	0.5–1.2 (0.1–1 s ⁻¹)
Counterblows hammers	1220	5–9 (10–200 s ⁻¹)	Mechanical	2.2–623	0.03–1 (1–30 s ⁻¹)

Adapted from Lefteri, C., 2007. Manufacturing Techniques for Product Design, Laurence King, and other sources.

If the number of steps is such that the workpiece temperature becomes too low to induce recrystallization, it would be necessary to reheat the workpiece during the forging sequence. Only in special cases recrystallization is avoided.

11.2 Mechanics of open-die forging

For the sake of simplicity, we first consider the *upsetting* of a prismatic-shaped workpiece (Fig. 11.4A). The workpiece has an initial height h_0 , an initial half-width a_0 , and an initial depth w_0 . The applied load is F , and the upper die velocity is v . The figure shows an intermediate position when the height of the workpiece is h , and the half-width is a ($h < h_0$ and $a > a_0$). Under the assumption of no friction between the workpiece and dies, and with the hypothesis of plane strain deformation (i.e., the depth w_0 remains unchanged), we have:

$$\begin{cases} \varepsilon_1 = \ln(h/h_0) \\ \varepsilon_2 = 0 \\ \varepsilon_3 = -\varepsilon_1 \text{ (from volume constancy)} \end{cases} \quad \begin{cases} \sigma_1 = F/(2a w_0) \\ \sigma_3 = 0 \text{ (free deformation)} \\ \sigma_2 = \sigma_1/2 \text{ (from Eq. 5.10)} \end{cases}$$

Therefore $\varepsilon_{eff} = 2/\sqrt{3} \cdot |\varepsilon_1|$, and $\sigma_{eff} = \sqrt{3}/2 \cdot |\sigma_1|$, as seen in Section 5.3.

In real conditions, friction is present at each workpiece-die interface. It can be expressed with the usual equation: $\tau = \mu p$, where τ is the friction (shear) stress,

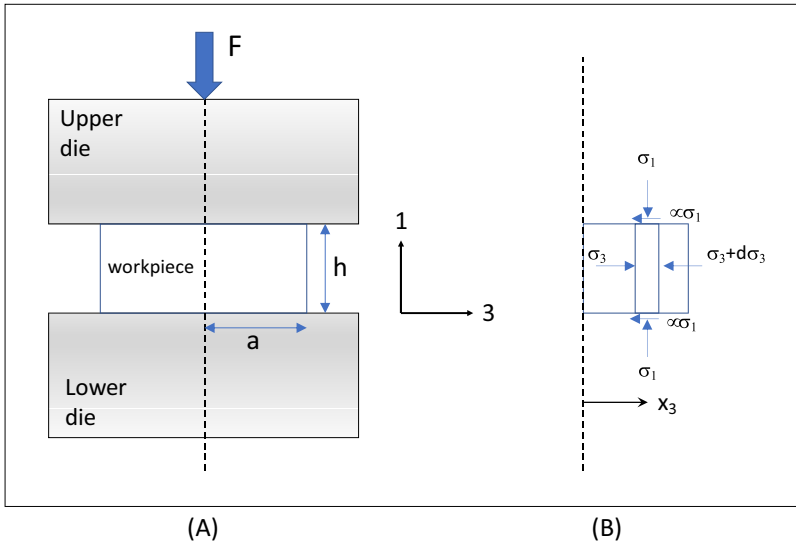


Fig. 11.4 (A) Upsetting test of a prismatic specimen; (B) stresses acting on a slab of material at a distance x_3 from the origin (for symmetry, only the contact region for $x_3 > 0$ is considered).

p is the pressure normal to the sliding surface, and μ is the *coefficient of friction*. Friction induces a change in the stress-strain state in the workpiece. In particular, two aspects must be considered (Brozzo, 1979; Dieter, 1988):

1. The modification of the workpiece-die contact pressure, with the formation of the *friction hill*;
2. The modification of the stress-strain state in the workpiece due to the *barreling* phenomenon.

Contact pressure in plane strain conditions

The workpiece *height reduction* is given by $\Delta h = h_0 - h_f$, where h_f is the height of the workpiece at the end of the process. If the parameter Δ of Eq. (10.4) is high (greater than 1), and the corresponding compression strain, ε_1 , is then low (typically on the order of -0.1), the plastic deformation is inhomogeneous in the workpiece and largely localized in the regions of contact with the dies. If Δ is small (lower than 1) and friction is high, the barreling phenomenon gains in importance (as explained in the next subsection), and also in this case plastic deformation is inhomogeneous, as anticipated in Section 10.4. The analysis presented herewith, based on the *slab method*, is rather simplified and assumes that the reduction is sufficiently large, and the friction is sufficiently low to induce a rather *homogeneous* plastic deformation along with the height of each slab that forms the workpiece. In addition, it is supposed that friction does not affect the principal directions.

The aim of the present analysis is to obtain the evolution of the contact pressure, σ_1 , as a function of x_3 , i.e., of the distance from the origin. In Fig. 11.4B, a slab of metal with height h , thickness dx_3 , and at a distance x_3 from the origin is shown (for symmetry reasons, the contact region for $x_3 > 0$ is only considered). The slab is considered to move to the right (positive 3 direction) under equilibrium conditions, i.e., any dynamic effect is neglected. In the figure, the stresses acting on the faces of the slab in directions 1 and 3 are shown (they are considered to be principal stresses). On the two faces in contact with the dies, the compressive stress σ_1 and the frictional (shear) stress $\mu\sigma_1$ are present. The frictional stress is acting to the left, since the slab is moving to the right. On the lateral faces, the stress σ_3 is acting to the right to maintain equilibrium (in the absence of friction, the deformation in direction 3 is free and σ_3 is 0 everywhere). (Since the thickness of the slab is infinitesimal, the difference between the stresses at the two faces is also infinitesimal.) Hence, the friction changes the stress distribution in the workpiece (in each slab, depending on its position), but it does not change the strain distribution. In other words:

$$\begin{cases} \varepsilon_1 = \ln(h/h_0) \\ \varepsilon_2 = 0 \\ \varepsilon_3 = -\varepsilon_1 \end{cases} \quad \begin{cases} \sigma_1 = F/(2a w_0) \\ \sigma_3 > 0 \\ \sigma_2 = (\sigma_1 + \sigma_3)/2 \end{cases} \quad (\text{from Eq. 5.10})$$

The effective strain does not change ($\epsilon_{eff} = \sqrt{3}/2 \cdot |\epsilon_1|$), and the effective stress, obtained from Eq. (5.3):

$$\sigma_{eff} = \frac{\sqrt{3}}{2} \cdot |\sigma_1 - \sigma_3| \quad (11.1)$$

(note that: $\sigma_1 > \sigma_2 > \sigma_3$), is the same in every slab, since it depends on the relevant constitutive equation.

The equilibrium of the forces in direction 3 can now be written as:

$$\sigma_3 \cdot h \cdot w_0 = (\sigma_3 + d\sigma_3) \cdot h \cdot w_0 + 2 \cdot \mu \cdot \sigma_1 \cdot dx_3 \cdot w_0 \quad (11.2)$$

and then:

$$d\sigma_3 + \frac{2\mu}{h} \cdot \sigma_1 \cdot dx_3 = 0 \quad (11.3)$$

In order to obtain the evolution of the contact pressure, σ_1 , as a function of x_3 , we need to express $d\sigma_3$ as a function of $d\sigma_1$, and then to integrate Eq. (11.3). At each deformation level (i.e., at each value of h , or Δh), σ_{eff} is constant. Therefore, from Eq. (11.1): $\sigma_1 - \sigma_3 = \text{constant}$, and $d\sigma_1 = d\sigma_3$. In Eq. (11.3), we can substitute $d\sigma_3$ with $d\sigma_1$. At the same time, it is convenient to set $p = |\sigma_1|$, where p is the contact pressure:

$$dp + \frac{2\mu}{h} \cdot p \cdot dx_3 = 0 \quad (11.4)$$

We can now integrate Eq. (11.4), setting $p = p_0$ for $x_3 = 0$,

$$\int_{p_0}^p \frac{dp}{p} = -\frac{2\mu}{h} \cdot \int_0^{x_3} dx_3$$

to obtain:

$$p(x_3) = p_0 \cdot e^{-\frac{2\mu}{h} \cdot x_3} \quad (11.5)$$

The contact pressure at the centerline, p_0 , is obtained from the consideration that $\sigma_3 = 0$, for $x_3 = a$ (free surface). Therefore, from Eq. (11.1): $p(a) = \sqrt{3} \cdot \sigma_{eff}$. The contact pressure for $x_3 = 0$ can also be expressed with Eq. (11.5):

$$p(a) = \frac{2}{\sqrt{3}} \cdot \sigma_{eff} = p_0 \cdot e^{-\frac{2\mu}{h} \cdot a}$$

and then:

$$p_0 = \frac{2}{\sqrt{3}} \cdot \sigma_{eff} \cdot e^{\frac{2\mu}{h} \cdot a}.$$

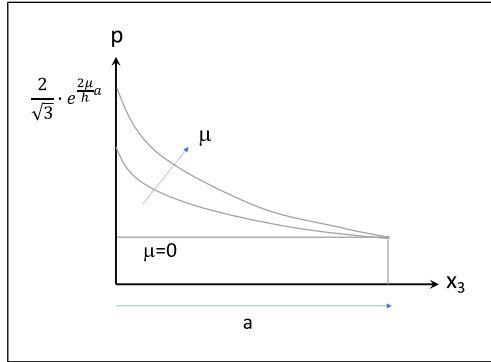


Fig. 11.5 Friction hill in the case of upsetting of a prismatic specimen.

which can be inserted into Eq. (11.5) to finally obtain:

$$p(x_3) = \frac{2}{\sqrt{3}} \cdot \sigma_{eff} \cdot e^{\frac{2\mu}{h}(a-x_3)} \quad (11.6)$$

Eq. (11.6) is graphically shown in Fig. 11.5. The contact pressure is maximum at the centerline (*friction hill*). Measurements of the contact pressure using the embedded pin-transducer technique have confirmed the presence of a maximum of the pressure at the centerline, even if with a smooth trend. The effective stress, σ_{eff} , should be obtained from the effective strain, ϵ_{eff} , using the proper constitutive equation (see Section 3.7 for cold working, and Chapter 4 for hot working).

The applied load, F , can now be estimated as a function of h , by integrating the pressure p (Eq. 11.6) over the resisting section:

$$F(h) = 2 \cdot w_0 \cdot \int_0^a p(x_3) \cdot dx_3$$

to obtain:

$$F(h) = \frac{2}{\sqrt{3}} \cdot \sigma_{eff} \cdot \frac{w_0 \cdot h}{\mu} \cdot \left(e^{\frac{2\mu a}{h}} - 1 \right) \quad (11.7)$$

The mean pressure, $\bar{p}(h) = \frac{F(h)}{2aw_0}$, is simply obtained by considering only the first term of the Taylor expansion of the exponential:

$$\bar{p}(h) \approx \frac{2}{\sqrt{3}} \cdot \sigma_{eff} \cdot \left(1 + \frac{\mu \cdot a}{h} \right) \quad (11.8)$$

Role of lateral spreading

If the metal under compression can spread in direction 2, the conditions of plane strain are no longer present in the deforming metal. An important example is that of cogging, shown in Fig. 11.1. The top view shows the lateral spreading in the width direction, which typically occurs when the width-to-thickness ratio is lower than 8–10 (otherwise, the adjacent workpiece metal under elastic deformation would almost fully restrain the plastically deforming metal). The plastic spread is quantified by the means of the *spreading coefficient*, S , defined as:

$$S = \frac{\varepsilon_2}{\varepsilon_1} \quad (11.9)$$

(Note that $S < 0$, since $\varepsilon_1 < 0$ and $\varepsilon_2 > 0$.) In the case of cogging, the following relationship was proposed for S (Dieter, 1988):

$$S = -\frac{2a/w_0}{1 + 2a/w_0} \quad (11.10)$$

where $2a$ is the width of the die, and w_0 is the depth of the workpiece before plastic deformation. The effective strain can then be calculated with the following relationship, obtained from Eq. (5.7):

$$\varepsilon_{eff} = |\varepsilon_1| \cdot \frac{2}{\sqrt{3}} \cdot \sqrt{1 + S + S^2} \quad (11.11)$$

With increasing spreading, i.e., with S becoming more negative, ε_{eff} decreases with respect to the value under plane strain and, consequently, the contact pressure decreases. Of course, in the case of lateral spreading a friction hill also develops along direction 2, and this complicates the analysis. To calculate the mean pressure, Eq. (11.8) can be rewritten in this way:

$$\bar{p}(h) \approx f \cdot \sigma_{eff} \cdot \left(1 + \frac{\mu \cdot a}{h}\right) \quad (11.12)$$

where f is a factor to allow for plane strain conditions. As seen, $f = 2/\sqrt{3}$ when using the von Mises criterion. In the case of lateral spreading, factor f can be estimated by Eq. (11.11) (Richardson et al., 1985):

$$f = \frac{\varepsilon_{eff}}{|\varepsilon_1|} = \frac{2}{\sqrt{3}} \cdot \sqrt{1 + S + S^2}$$

An important case regards the upsetting of a cylinder. The analysis can be conducted using cylindrical coordinates, obtaining:

$$p(r) = \sigma_{eff} \cdot e^{2\mu \frac{R-r}{h}}$$

where

- r is the distance from the origin;
- R is the instantaneous radius of the deformed cylinder at the height h .

Then, the mean contact pressure is:

$$\bar{p} \approx \sigma_{eff} \cdot \left(1 + \frac{2\mu \cdot r}{h}\right) \quad (11.13)$$

Forging load and power

The average contact pressure, $\bar{p}$, increases with increasing the reduction Δh , i.e., as h is decreased and a (Eq. 11.8) or r (Eq. 11.13) correspondingly increase. In the case of cold process, σ_{eff} also increases with the reduction because of strain hardening. Consequently, the average pressure is maximum at the end of the process and is indicated with: $\bar{p}_{max}$. The maximum forging load, F_{max} , is then given by: $F_{max} = \bar{p}_{max} \cdot A_n$, where A_n is the nominal area of contact at the end of the process. As already mentioned, the calculation of F_{max} is pivotal to verify the ability of a given equipment to perform the requested process (Fig. 10.5).

Eq. (11.13) is also useful in the case of compression tests using cylindrical specimens, as those described in Chapter 4. The flow stress, given by the ratio of the recorded compression load and the effective load bearing section, can be corrected using Eq. (11.13) to account for the effect of friction.

The knowledge of the forging load allows for estimating the *forging energy*, E , required for the process:

$$E = \int_0^c F \cdot dh = \eta \cdot c \cdot F_{max} \quad (11.14)$$

where

- c is the stroke;
- η is a factor typically between 0.15 and 0.25;
- F_{max} is the maximum forging load (attained at the end of the process).

The ratio E/t (where t is the process time, given by c/v , where v is the average die displacement velocity) is the required process power that the forging machine must be able to supply (plus any additional power loss).

Barreling effect and secondary hoop tensile stresses

As mentioned, the workpiece-die friction induces a marked inhomogeneous plastic deformation in the workpiece, which was neglected in the previous analysis. The frictional force at the interface with the die slows down the metal plastic flow in the lateral direction. This phenomenon is accentuated when the die temperature is lower than that

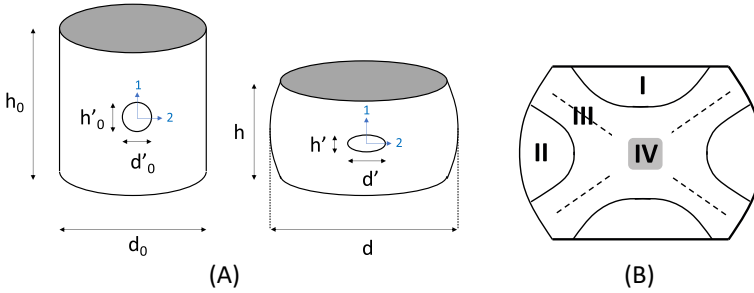


Fig. 11.6 (A) Geometry change of a cylindrical specimen subjected to uniaxial compression (upsetting) with friction at the workpiece-die contact; (B) cross section of a barreled specimen to show the four main deformation regions.

of the workpiece, and the surface cooling of the workpiece induces a decrease in the metal plastic deformability. As a result, the specimen undergoes *barreling* (also called *bulging*), which is shown in Fig. 10.9 and in Fig. 11.6A, for the upsetting of a cylindrical workpiece.

Due to barreling, the plastic deformation in the workpiece is not in pure compression everywhere in the specimen. Rather, it is inhomogeneous and follows the typical pattern shown in the cross section schematized in Fig. 11.6B, in the case of the upsetting of a cylindrical specimen. Four regions can be identified. Region I is close to the die surfaces. Here, friction stress opposes plastic deformation in the radial direction. Therefore the metal remains rather undeformed. This region is called *friction cone*, or *dead-metal zone*. Region II involves the region of barreling. Here, the radial deformation is high (depending on friction). Consequently, the tensile deformation in the hoop (circumferential) direction is also high. Between regions I and II, region III is present, where the material is subjected to large shear deformation. This is the *shear region*. At the center of the specimen, the material is deformed in homogeneous (pure) compression (region IV).

In Fig. 11.6A, a small circle is drawn at the mid-height equatorial region, to mark the evolution of the local strains. Using cylindrical coordinates, we can define the *equatorial surface axial (compressive) strain*, ϵ'_1 , and the *equatorial surface circumferential (hoop) strain*, ϵ'_2 . The two strains are given by the following relationships:

$$\epsilon'_1 = \ln \left(\frac{h'}{h'_0} \right) \quad \epsilon'_2 = \ln \left(\frac{d'}{d'_0} \right) \quad (11.15)$$

The surface hoop strain can also be evaluated through the following relationship:

$$\epsilon'_2 = \ln \left(\frac{d}{d_0} \right) \quad (11.16)$$

where d_0 is the diameter of the cylinder before upsetting, and d is the current diameter at the mid-height. Eq. (11.16) is very useful, as it allows experimentally obtaining the

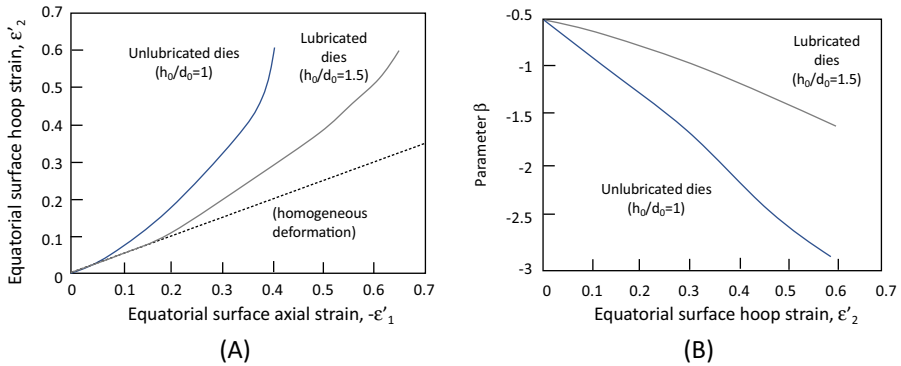


Fig. 11.7 Cylinder upsetting: experimental results showing (A) the relationship between equatorial surface axial strains and equatorial surface hoop strains (in different geometrical and friction conditions; lubrication is realized with MoS_2), and (B) the dependency of parameter β from equatorial surface hoop strain for the two deformation paths shown in (B).

Adapted and redrawn from Thomason, P.F., 1990. Ductile Fracture of Metals, Pergamon Press, Oxford.

surface hoop strain during upsetting (or at the end of the process) by simply measuring the barrel diameter.

In the absence of friction, no barreling develops, and compression is fully homogeneous. Because of the constancy of the volume: $h_0/h = (d/d_0)^2$; thus: $\epsilon_2 = -1/2\epsilon_1$. In the case of barreling, the plastic deformation is inhomogeneous and the surface hoop strain, ϵ'_2 , is greater than $-1/2\epsilon'_1$. The experimental evolution of ϵ'_1 with ϵ'_2 is shown in Fig. 11.7A for two testing conditions. With increasing $-\epsilon'_1$, the deformation path deviates from the homogeneous curve, and the ϵ'_2 values become greater than under pure compression. In general, the barreling effect increases with (Gänsler et al., 2001; Thomason, 1990):

- Increasing the friction coefficient at the workpiece-die interface;
- Decreasing the initial height-to-diameter (*aspect ratio*) of the specimen (h_0/d_0).

At the surface, a state of plane stress is present, since the radial stress is 0 (free surface). As already introduced in Section 5.3, it is convenient to use parameter β in this case. However, Fig. 11.7A clearly shows that the deformation path is not proportional (i.e., the ratio of the strains is not constant during the deformation process). Therefore parameter β is given by:

$$\beta = \frac{d\epsilon'_2}{d\epsilon'_1} \quad (11.17)$$

In Fig. 11.7B, the evolution of β with ϵ'_2 is displayed for the two deformation paths shown in Fig. 11.6A. It is seen that $\beta = -1/2$ (homogeneous deformation) only for small reductions. With increasing the cylinder reduction, β becomes increasingly negative (and this trend is more intense with increasing friction and decreasing the aspect

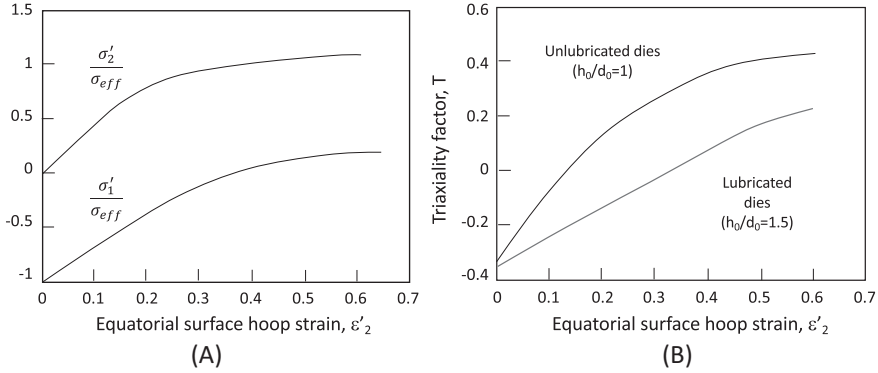


Fig. 11.8 (A) Evolution of stresses in the equatorial region during upsetting of a cylinder for the deformation path $h_0/d_0 = 1.5$ and lubricated dies of Fig. 11.7; (B) evolution of the multiaxiality parameter, T , for both deformation paths of Fig. 11.7.

ratio of the specimen). The evolution of the effective strain in the barrel region can then be obtained from the following relationship (Section 5.3):

$$\epsilon_{eff} = \frac{2}{\sqrt{3}} \cdot \int \sqrt{1 + \beta + \beta^2} \cdot d\epsilon'_1 \quad (11.18)$$

In general, ϵ_{eff} increases with the height reduction (for using this equation, numerical methods, such as FEM, are recommended; an example is shown in Fig. 11.10).

The modification of the strain distribution induced by barreling implies a change in the stress distribution as well. The evolution of the surface circumferential mid-height axial stress, σ'_1 , and the surface circumferential mid-height hoop stress, σ'_2 , with β , can be simply estimated from the flow rules (Eq. 5.10):

$$\begin{aligned} \sigma'_1 &= -\frac{2}{\sqrt{3}} \cdot \sigma_{eff} \cdot \frac{1 + \beta/2}{\sqrt{1 + \beta + \beta^2}} \\ \sigma'_2 &= -\frac{2}{\sqrt{3}} \cdot \sigma_{eff} \cdot \frac{\beta + 1/2}{\sqrt{1 + \beta + \beta^2}} \end{aligned} \quad (11.19)$$

Their evolution is shown in Fig. 11.8A as a function of ϵ'_2 , in the case of the lubricated deformation path of Fig. 11.7A. For very small reductions, $\sigma'_1 = -\sigma_{eff}$ and $\sigma'_2 = 0$, as in homogeneous compression. However, with increasing the reduction, σ'_1 becomes less negative, whereas σ'_2 becomes positive and increases in modulus. Therefore barreling introduces a tensile hoop (circumferential) strain, which is larger than expected and, at the same time, the formation of *unexpected secondary tensile stresses*.

With increasing the height reduction, the stress triaxiality at the mid-height circumferential region also increases. In fact, the parameter T is given by:

$$T = \frac{\sigma_m}{\sigma_{eff}} = \frac{\sigma'_1 + \sigma'_2}{3\sigma_{eff}} = -\frac{\sqrt{3}}{3} \cdot \left(\frac{1 + \beta}{\sqrt{1 + \beta + \beta^2}} \right) \quad (11.20)$$

and the evolution of T with ϵ'_2 is shown in Fig. 11.8B for both deformation paths of Fig. 11.7A. At the beginning, $T = -1/3$. However, T increases with the reduction, and for sufficiently high strains, T tends to be positive. In general, the increase rate of T with ϵ'_2 (and also with the height reduction) increases with increasing friction and decreasing the aspect ratio. It is also noted that, for the unlubricated path, the condition $T \sim 0$ is attained at comparatively low hoop strains (and then at comparatively low reductions), whereas for the lubricated path, the condition $T \sim 0$ is attained at comparatively high hoop strains (high reductions).

11.3 Mechanics of impression-die and closed-die forging

With impression-die and closed-die forging, it is possible to obtain rather complex shapes. Therefore the mathematical analysis of the involved strain, strain rates, and stresses can only be approximated, and for accurate evaluations, it is necessary to use numerical methods.

We initially consider the forging of *axisymmetric* components. An example is shown in Fig. 11.3A for the impression-die forging of a flange. The initial billet is cylindrical, with diameter d_0 and height h_0 , and the process can be considered to occur (at least at the beginning) under uniaxial conditions. Hence, it can be assumed that the forging of an axisymmetric part is like the free upsetting of a disc of diameter L and average height h_m :

$$L = \sqrt{\frac{A}{\pi}} \quad h_m = \frac{V}{A} \quad (11.21)$$

where A is the projected area on the parting plane (including the flash gutter), and V is the volume of the component to be produced (for simplicity, it is equal to the internal volume of the forging die cavity). Therefore the applied uniaxial strain, ϵ_1 , is:

$$\epsilon_1 = \ln \left(\frac{h_m}{h_0} \right) \quad (11.22)$$

The average strain rate is $\dot{\epsilon}_1 = \epsilon_1/t$, where t is the forging duration: $t = c/v$, where c is the stroke and v are the average velocity of the upper die.

As seen in the previous section, the forging load, F , increases with the height reduction. The maximum forging force, F_{max} , is attained at the end of the process and may be expressed with the following relation:

$$F_{max} = \eta \cdot \bar{\sigma}_1 \cdot A \quad (11.23)$$

where $\bar{\sigma}_1$ is the mean flow stress in direction 1 and η is an adjusting parameter. In the case of hot forging, $\bar{\sigma}_1$ can be evaluated using Eq. (4.11): $\bar{\sigma}_1 = C \cdot \epsilon_1$. In case of cold forging, $\bar{\sigma}_1$ can be evaluated from the L-H equation:

$$\bar{\sigma}_1 = \frac{\int_0^{\epsilon_1} \sigma_1 \cdot d\epsilon_1}{\epsilon_1} = \frac{K \cdot \epsilon_1^n}{n+1} \quad (11.24)$$

(or from Eq. 3.18). Parameter η mainly depends on the complexity of the forging, as the state of stress changes from uniaxial compression to plane strain, when the metal starts flowing into the die cavities (constrained flow). For axisymmetric parts, η typically varies between 1.2 and 2.5. The transition in the state of stress is very important in the case of forging of *non-axisymmetric* parts, such as con-rods. For these parts, the transition from plastic deformation under uniaxial conditions into plane strain conditions occurs very early. Consequently, Eq. (11.23) can still be used in evaluating F_{max} , but parameter η must be increased to values typically in the range of 3–12 (Dieter, 1988).

As mentioned previously, the metal plastic flow in the die cavities can be rather complex. An interesting situation is shown in Fig. 11.9. The presence of friction at the workpiece-die interfaces induces a phenomenon like barreling when the metal is flowing into a deep cavity. Therefore, on the external surface of the metal, unexpected secondary tensile deformations and stresses may develop.

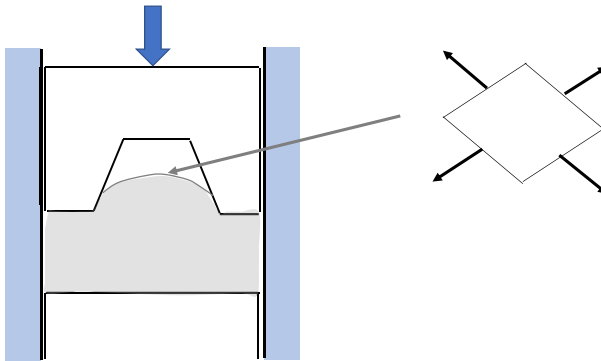


Fig. 11.9 Development of secondary tensile deformations and stresses on the external surface of a metal flowing into die cavities.

Redrawn from Kuhn, H.A., 2003. Workability theory and application in bulk forming processes, in: G.E. Dieter, H.A. Kuhn, S.L. Semiatin (Eds.), Handbook of Workability and Process Design, ASM International, Materials Park, OH.

11.4 Ductile cracks in forgings

Forgings may display different *defects* that must be properly controlled and possibly avoided. In the case of closed-die forging, *geometrical defects* are important. They may originate from an insufficient filling of the cavities (underfills), or from a tiny corner radius that does not allow a proper metal flow. Geometrical defects are also folds (or laps) that occur when a region of the workpiece goes in contact with another one. However, given the comparatively large strains that are involved, the most important defects in forgings (in both closed-die and open-die processes) are *surface* and/or *internal ductile cracks*. In this section the attention is focused on surface and internal cracks that may form due to the inhomogeneous plastic deformation (Table 10.6) together with a possible drop in ductility. The knowledge of the origin of these cracks is paramount for a correct process design.

Of course, the formation of ductile cracks is further favored by possible pre-existing surface defects on the initial billet. An example is given by cracks formed during solidification, such as transverse or longitudinal cracks, in ingots or continuous semi-finished cast products. Another example in hot forging regards the deep entrapment of oxides formed during heating of the billet and not properly removed by sandblasting. In cold forging, cracks can also be present on the initial billet surface due to its production schedule. An example is longitudinal cracks possibly formed on bars obtained by cold drawing. If such defects are too severe, they should be properly removed by turning operations before any process by plastic deformation. The possible presence of these pre-existing defects is not specifically addressed in this section.

Surface cracks in open-die forging

Let us consider the upsetting process. As anticipated in Table 10.6, we may expect the occurrence of surface cracks by barreling, and internal cracks by forced shear localization. We first consider the possible formation of surface cracks.

As seen Section 11.2 during upsetting the local strains in the barrel region (ϵ'_1 and ϵ'_2) increase in modulus with the increasing reduction of Δh , also expressed with the *macroscopic compression strain*, defined by $\epsilon_1 = \ln(h_f/h_0)$. Correspondingly, the local effective strain, ϵ_{eff} , increases. As an example, Fig. 11.10 shows the evolution of ϵ_{eff} with $-\epsilon_1$ (at the mid-height circumferential region), in the case of upsetting of a cylindrical specimen with aspect ratio 1 and a friction coefficient, at the workpiece-die interface, of 0.5 (unlubricated condition). In the same diagram, the evolution of the critical effective strain at fracture, $\epsilon_{eff,f}$, in the barrel region is displayed for three true fracture strains (ϵ_f : 0.2, 0.4, and 0.6). They were obtained using the R-T criterion (Eq. 6.17). For this, the evolution of T in the barrel region was estimated by means of an FEM analysis. Since T increases with the reduction (see Fig. 11.8B), $\epsilon_{eff,f}$ decreases with the reduction. To sum up, during upsetting, the effective strain, ϵ_{eff} , increases in the barrel region and, at the same time, the local effective fracture strain decreases. When: $\epsilon_{eff} = \epsilon_{eff,f}$, a surface ductile crack is formed. The vertical arrows in

Fig. 11.10 FEM prediction of the evolution of the effective (von Mises) strain in the barrel region for a cylindrical specimen under upsetting (aspect ratio: 1; friction coefficient: 0.5). The evolution of the critical effective strain at fracture, $\epsilon_{eff,f}$, obtained using the R-T criterion, is also shown for three tensile fracture strains (Raghavendra and Straffelini, 2022).

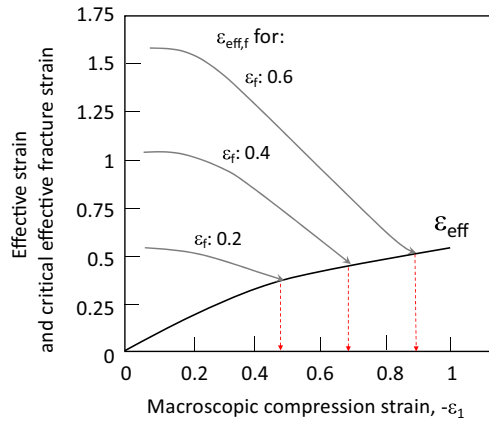


Fig. 11.10 indicate the maximum reductions at fracture in the three conditions. It is shown that the reduction at fracture increases with increasing the metal ductility (in agreement with the seminal experiments of Latham and Cockcroft, 1966).

During the upsetting process, the T values in the barrel region typically vary in the range of -0.3 to 0.4 . In the case of metals that are prone to shear localization, such as some Al and Ti alloys, $\epsilon_{eff,f}$ decreases very significantly in the T range from -0.3 to 0 , as seen in Figs. 6.14 and 6.15. This means that during upsetting, surface ductile cracks are easily formed at comparatively low reductions, since the ϵ_{eff} in the barrel region quickly reaches the critical value. (Note that experimental points in Fig. 6.14 for the Al alloy and pertaining to this T interval were obtained by the means of upsetting tests.) Since fracture takes place when T is close to 0 (i.e., under local pure shear conditions), the surface ductile cracks form an angle of ~ 45 degrees with the equatorial line. This is shown, for example, in Fig. 11.11, referring to unlubricated cold upsetting tests of Al-6061 cylindrical specimens with $h_0/d_0 = 1$ (Li et al., 2011).

In the case of metals that are not particularly prone to shear localization, such as steels, the effective fracture strain does not decline so sharply as T increases in range from -0.3 to 0 . Therefore surface ductile cracks are formed at higher reductions, and at higher T values. When the local T at fracture is rather large (with both σ'_1 and σ'_2 assuming positive values), ductile cracks are *vertical*, i.e., they form an angle of approximately 90 degrees with the equatorial line. An example is shown in Fig. 11.12. It refers to the unlubricated cold upsetting of a cylindrical specimen made with a 1040 carbon steel in the annealed condition (Landre et al., 2003).

In the case of upsetting with a good lubrication and/or using a workpiece with a high aspect ratio, the triaxiality factor, T , increases very slowly with the reduction (refer to Fig. 11.8B), whereas the effective strain, ϵ_{eff} , at the surface barreled region continuously increases. Consequently, when fracture is reached (at comparatively high reductions), T approaches 0 , and the cracks are oblique (Komori, 2020). This has been experimentally shown for steels in the well-known experiments of Kudo and Aoi (1967).

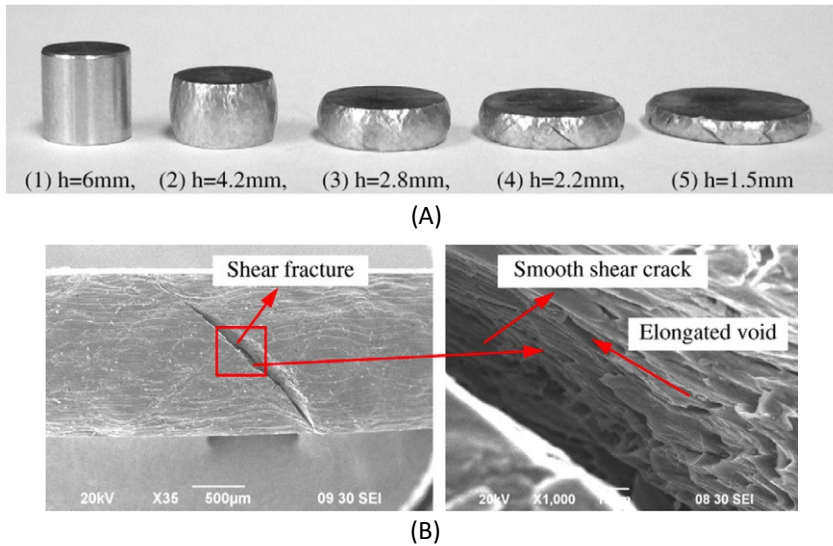


Fig. 11.11 (A) Formation of oblique surface ductile crack in the cold upsetting test of an Al-6061 cylindrical specimen with $h_0/d_0 = 1$; (B) SEM micrographs showing the surface shear ductile fracture.

From Li, H., Fu, M.W., Lu, J., Yang, H., 2011. Ductile fracture: experiments and computations. *Int. J. Plast.* 27, 147–180, reproduced with permission of the publisher.

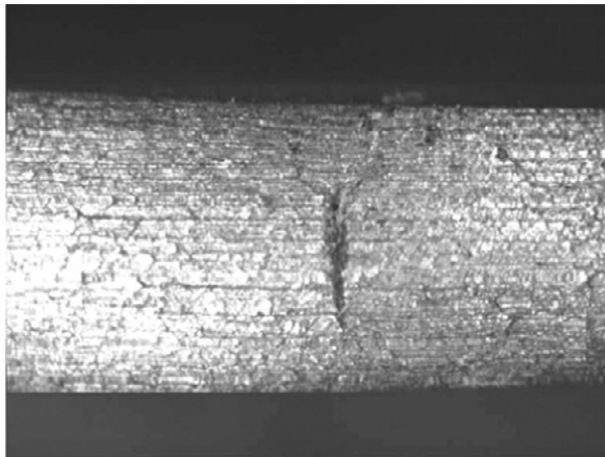


Fig. 11.12 Vertical crack forming at the surface equatorial region in the case of cold upsetting of a cylindrical specimen made with 1040 carbon steel without using a lubricant.

From Landre, J., Pertence, A., Cetlin, P.R., Rodrigues, J.M.C., Martins, P.A.F., 2003. On the utilization of ductile fracture criteria in cold forging. *Finite Elem. Anal. Des.* 39, 175–186, reproduced with permission of the publisher.

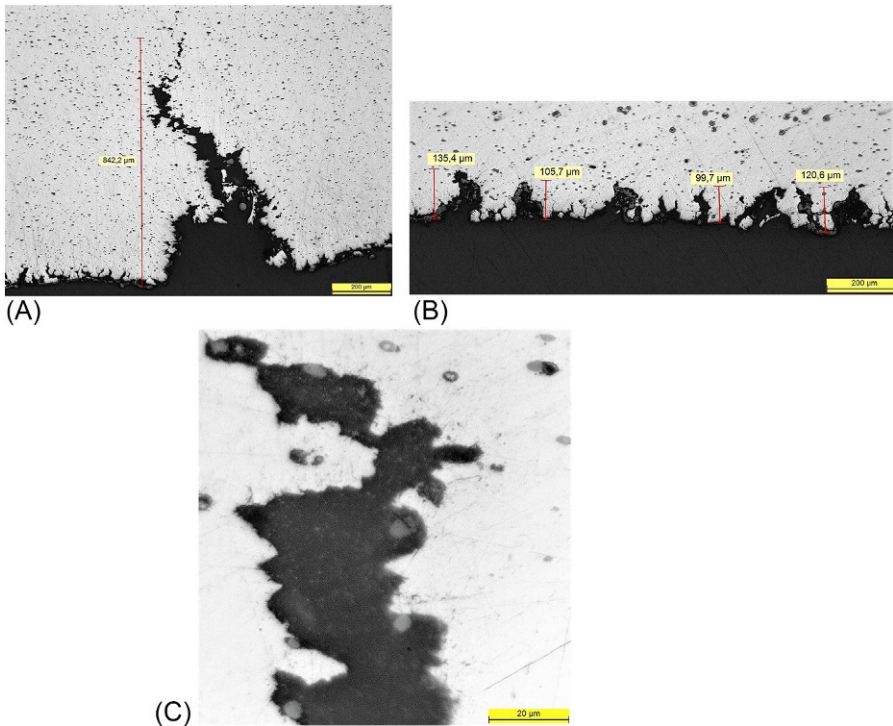


Fig. 11.13 Metallographic cross sections showing surface long (A) and short (B) cracks formed on the surface region in the upsetting test at 1050°C of a resulfurized Bi steel (Oliosio, 2021). In (C) details of the voids nucleation at Bi particles and growth are shown.

Hot upsetting is often carried out in unlubricated or poorly lubricated conditions. The hot metal ductility is generally rather high, as explained in [Chapter 8](#) and the process is then usually conducted in safe conditions. However, critical upsetting conditions may occur in the case of ductility drops, such as, for example, in correspondence to the ductility trough or at very high temperatures, when hot shortness phenomena occur. The first page of this chapter shows numerous vertical surface cracks after upsetting of a (turned) cylindrical specimen at 1050°C, made of a resulfurized steel containing bismuth particles (with Mn/S ratio of ~ 3.9) (Oliosio, 2021). The height reduction was 66%. [Fig. 11.13](#) is a metallographic cross section, showing the characteristics of the surface cracks. The presence of long and short cracks can be recognized. [Fig. 11.13C](#) highlights the crucial role of Bi particles, in nucleating the voids for the subsequent formation, by coalescence, of the surface ductile cracks. [Fig. 11.14](#) shows the results of hot upsetting tests carried out at 1150°C for the same reduction of 66%. In this case, only a surface crack was detected. As a matter of fact, the temperature is sufficiently high to fall at the right side of the ductility trough.



Fig. 11.14 Hot upsetting at 1150°C of a resulfurized steel containing bismuth, with the formation of very few surface cracks by barreling (Oliosio, 2021).

Critical conditions for the hot upsetting of steels are also encountered when the steel contains residual Cu that markedly decreases the hot ductility (Section 8.5).

Surface cracks by barreling can also form during cogging operations (Fig. 11.1), because of lateral spreading that may induce a bulging on the side faces. If friction is high, surface cracks may also form in the contact regions with the die surfaces.

Internal cracks in open-die forgings

During open-die forging, internal cracks may also form. They may form before surface cracks, if their formation is easier. Localized shearing is responsible for the formation of such cracks, and in upsetting, shear cracks can then form in the shear region, i.e., in Region III of Fig. 11.6B. The situation is that of forced shearing, and this kind of fracture mainly involves metals that are rather prone to shear localization. The resulting reductions at fracture may be rather low. An example is shown in the next section, regarding the upsetting of a Mg alloy.

The formation of shear cracks inside the localized band may be facilitated by the formation of adiabatic shear bands, ASB (refer to Section 3.6). The formation of such cracks is favored by:

- High strain rates (they increase the adiabaticity of the process, and the metal strength);
- Low metal thermal conductivity (it reduces the thermal diffusion from the narrow-forced shear band from where it is released);
- High metal strength (it favors the generation of heat in the deformed region);
- Low ambient temperature (it favors a high metal strength).

As seen in Section 8.7, the formation of cracks in ASBs is particularly easy in Ti alloys that match most of the requirements. As an example, Fig. 11.15A shows a cylindrical specimen made with Ti-64 alloys and submitted to an upsetting test, at RT and at

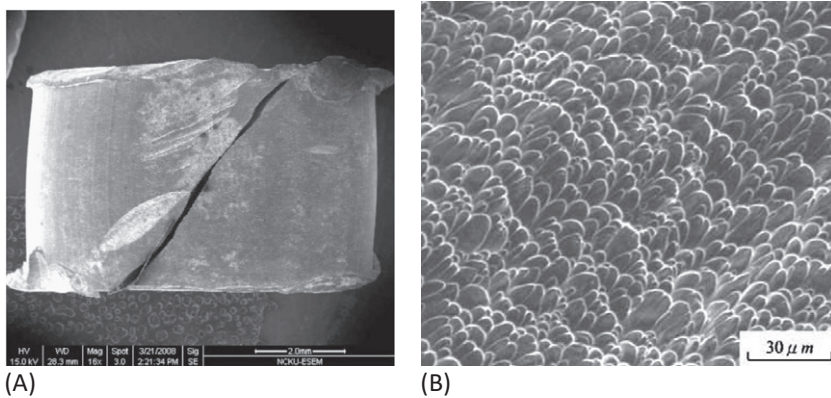


Fig. 11.15 (A) Cylindrical specimen made by Ti-64 fractured under upsetting test through the formation of an ASB (RT, $3 \times 10^3 \text{ s}^{-1}$); (B) fracture morphology of a specimen fractured in similar conditions (RT, $4.3 \times 10^3 \text{ s}^{-1}$) showing the typical shear dimpled ductile fracture morphology.

From Lee, W.S., Chen, T.H., Huang, S.C., 2010. Impact deformation behaviour of Ti-6Al-4V alloy in the low-temperature regime. *J. Nucl. Mater.* 402, 1–7, reproduced with permission of the publisher.

$3 \times 10^3 \text{ s}^{-1}$ (Lee et al., 2010). The specimen fractured because of internal cracking in ASBs. Fig. 11.15B shows morphology of the fracture surface (in a specimen tested in similar conditions), depicting the typical shear ductile dimpled fracture with parabolic dimpling. The formation of internal shear cracks during upsetting at high strain rate may also occur in other alloys if their formation conditions are satisfied (El-Magd and Abouridouane, 2006).

Another important case is that of cold heading, schematized in Fig. 11.2. The process is carried out with high impact velocity. In the region between the head and the stem, a large deformation by shear takes place with strain rates greater than 100 s^{-1} . Adiabatic shear bands (ASBs) easily form and if the metal is characterized by a comparatively low ductility, shear cracks can form within the ASBs, provoking the failure of the part. An example is shown in Fig. 11.16, referring to an industrial bolt made with a 1010 steel (Sabih and Nemes, 2009).

Cracks in impression-die and closed-die forgings

In these cases, the formation of surface cracks is rather difficult, since T generally remains very low and negative. However, in specific situations, such as that depicted in Fig. 11.9, secondary surface tensile stresses are formed, and T becomes greater than $-1/3$ and sometimes positive, following a mechanism very similar to that encountered in the case of barreling. In such cases, surface cracks may form if the ductile criterion is satisfied, especially if the local metal ductility is sufficiently low (Kuhn, 2003). Surface cracks can also form as a consequence of the formation of surface laps (or folds). In fact, the plastic flow can induce formation of surface cracks by localized shearing.

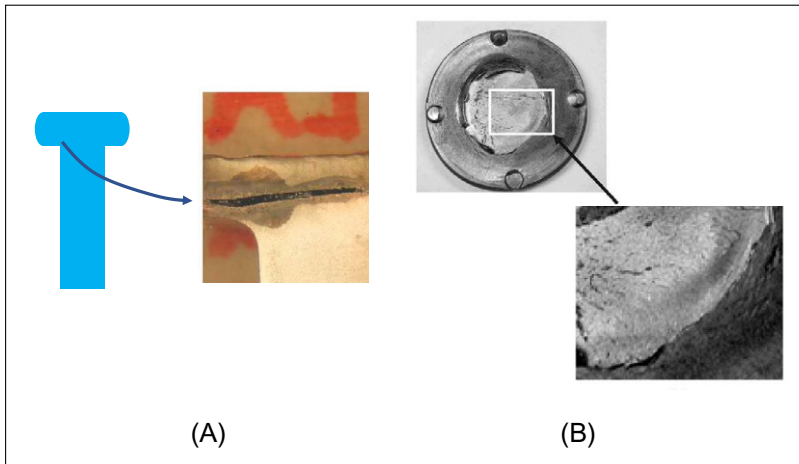


Fig. 11.16 (A) Sectioned bolt revealing an internal shear crack; (B) separated head of an industrial bolt broken by ductile fracture due to internal shear cracking.

From Sabih A., Nemes J.A. (2009) Internal ductile failure mechanisms in steel cold heading process, *Journal of Materials Processing Technology*, Vol. 209, pp. 4292–4311, reproduced with permission of the publisher.

In closed-die forging, internal cracks may form following shear strain localization. Given the geometrical complexity of many forgings, the analysis of the process should be made case by case. In most cases, shear localization occurs at the boundary of a zone with limited plastic deformation (a dead-metal zone) and a zone of well-developed plastic flow (similar to Region III in Fig. 11.6B).

An example is depicted in Fig. 11.17, which shows the three stages adopted to produce a bearing race by hot forging (Moon et al., 2007). The first stage is the upsetting process to obtain the pre-form. The second one is the pre-forging stage, and the last one is the finish forging stage. In the last stage, a region with high shear localization is formed. Simulations with the FE method have shown that in the shear region, the strain rate is as high as 90 s^{-1} and the local temperature is as high as 1230°C (initial pre-form temperature: 1150°C ; initial die and punch temperature: 150°C ; average punch velocity app. 500 mm/s). Upsetting and tensile tests have

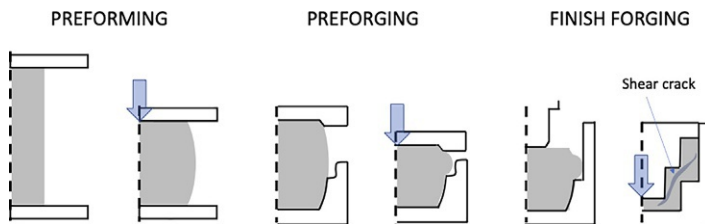


Fig. 11.17 The three stages adopted to produce a bearing race.

Redrawn and modified from Moon, H.K., Lee, J.S., Yoo, S.J., Joun, M.S., Lee, J.K., 2007. Hot deformation behavior of bearing steels. *J. Eng. Mater. Technol.* 129, 349–355.

shown that in these conditions, the material used to produce the forged part (a conventional high C and high Cr steel for bearings) is characterized by the hot shortness phenomenon (Fig. 11.19A). This explains the experimental results that indicate the formation of an internal shear crack. This example highlights the importance of considering strain localization and local temperature rise in the proper design of a forging process.

In impression-die and closed-die forging, it is rather typical to have a significant inhomogeneous distribution of strain, strain rate, and temperature in the material. This may easily give rise to different conditions for the onset of microstructural transformations, such as dynamic recrystallization, and produce a heterogeneous microstructure after cooling to RT. Rather typical is the presence of regions with different grain sizes, which induce the formation of forged parts with mechanical properties that may change from one point to another. In general, such differences (easily revealed through hardness tests) must be kept to a minimum.

11.5 Hot formability maps

The upsetting test using cylindrical specimens is a standard for the evaluation of metal formability for not only forging processes but for all bulk deformation processes. In this section, the use of this test to obtain details on specific hot formability maps will be considered.

The analysis carried out in the previous section clearly shows that the maximum height reduction at fracture, Δh_{max} (where $\Delta h = h_0 - h_f$, h_f being the height at fracture in this case), decreases as the metal hot ductility decreases. In addition, it also depends on the friction coefficient at the workpiece-die interface, and on the specimen aspect ratio (h_0/d_0). As μ increases and/or h_0/d_0 decreases, the circumferential strain and stresses increase, and surface fracture is attained earlier.

To fully determine the process window, the tests should be carried out at different temperatures and at different strain rates, up to ductile fracture (Fig. 10.5). However, in most cases a simplified approach is adopted. In particular, two kinds of laboratory compression tests are generally carried out, obtaining two types of *hot formability maps*:

1. **Type I maps.** The tests are carried out at different temperatures and strain rates up to a given reduction. After each test, the specimens are examined for the possible formation of surface or internal cracks, and the observations are collected in a map (this kind of test is selected when metals are expected to be particularly sensitive to the strain rate);
2. **Type II maps.** The tests are carried out at different temperatures and at a selected strain rate up to the fracture condition, i.e., up to the appearance of a visible ductile crack (a surface crack or an internal crack that produces a visible failure of the specimen).

To increase the barreling effect (and then to allow comparison among materials with a high ductility), different strategies can be adopted:

- Carry out tests in unlubricated conditions or using rough or grooved dies (to increase the interface friction);

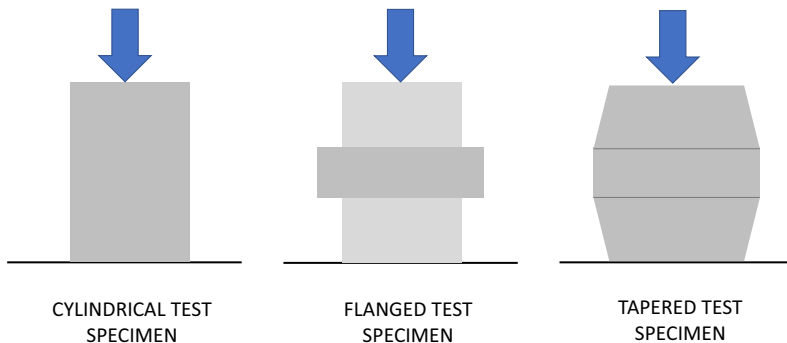


Fig. 11.18 Cylindrical, flanged, and tapered compression test specimens.

- Use of cooled dies (barreling is favored by the decreased deformability of the metal work-piece close to the dies because of cooling);
- Use of flanged or tapered specimens (in place of cylindrical specimens, see Fig. 11.18), i.e., pre-barreled specimens with a larger mid-height diameter that increases the circumferential strains.

Formability of bearing and duplex steels

As seen in Table 10.5, steels generally display a good hot formability. There are some exceptions. One is constituted by the free-machining steels, such as the resulfurized steels, considered in the previous section. Another one is constituted by the duplex stainless steels. In this subsection, the hot formability (forgeability) maps of a bearing steel and of a duplex stainless steel are shown.

In Fig. 11.19A, the Type I map for a **bearing steel**, with chemical composition: 0.17%–0.23%C, 0.15%–0.35%Si, 0.55%–0.9%Mn, 0.85%–1.25%Cr, $S < 0.03\%$, $P < 0.03\%$, is shown (Moon et al., 2007). It was obtained through compression tests of cylindrical specimens (diameter: 10 mm; height: 12 mm), in the range of temperatures from 900°C to 1300°C and strain rates from 1 to 50 s⁻¹. The tests were carried out for a total true strain of -1.38 (height reduction of 75%). The map shows that surface cracks are formed (by barreling) at high strain rates and high temperatures. The authors showed that hot shortness was responsible for cracking, which was related to the grain boundary segregation of elements such as S. As explained in Section 8.5, hot shortness is typically attained at high temperatures. The role of strain rate can be explained considering that as strain rate increases, the steel temperature also increases, since an increased amount of heat is released by plastic deformation (given the increase in the flow strength and in the adiabaticity of the process). Therefore, with increasing strain rate, the hot shortness phenomenon is attained at lower ambient temperatures. In this investigation, the metal behavior close to the PDT is investigated, and for temperatures as low as 900°C, no ductility trough was detected.

Fig. 11.19B is a Type I map for a (lean) **duplex 2101 stainless steel**, with chemical composition: 0.026%C, 0.28%Si, 5.3%Mn, 21.2%Cr, 1.5%Ni, 0.31%Mo, 0.22%N, 0.004%S, 0.017%P (Fang et al., 2011). The data in the map were obtained through

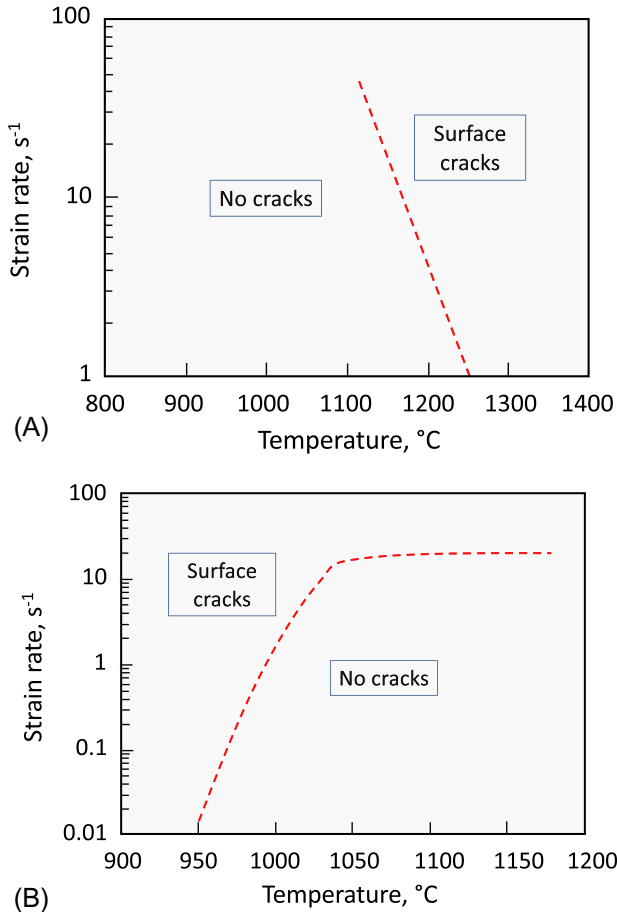


Fig. 11.19 Type I hot formability map for (A) a conventional bearing steel (adapted from [Moon et al., 2007](#)), and (B) a (lean) duplex stainless steel of type 2101 (adapted and redrawn from [Fang et al., 2011](#)).

compression tests of cylindrical specimens (diameter: 8 mm; height: 12 mm), in the range of temperatures from 950°C to 1150°C (after cooling from a preheating temperature of 1200°C), and strain rates from 0.01 to 30 s⁻¹. The tests were carried out for a total true strain of -0.8 (height reduction of 55%). The map shows that at high strain rates and low temperatures, surface cracks are formed (by barreling). Examples of the cracks are shown in [Fig. 11.20](#). The authors showed that the cracks nucleated very close to the equatorial region. It can also be seen that the cracks form an angle of about 45 degrees with the loading axis, showing that they are formed by shear, i.e., at relatively low fracture strains.

The formation of surface cracks in duplex stainless steels at a relatively low temperature (depending on strain rate) is caused by the nucleation and growth of ductile cracks at the ferrite/austenite interfaces. In fact, during hot compression, plastic

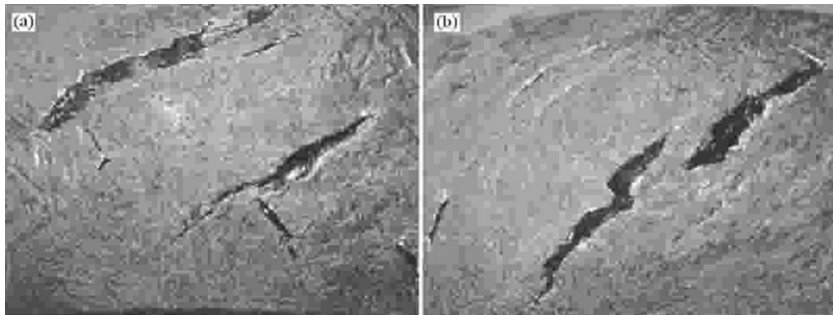


Fig. 11.20 Surface cracks on the specimens deformed at different temperature and strain rate. (A) 950°C, 5 s⁻¹; (B) 950°C, 30 s⁻¹.

From Fang, Y.L., Liu, Z.Y., Wang, G.-D., 2011. Crack properties of lean duplex stainless steel 2101 in hot forming processes. *J. Iron Steel Res. Int.* 18, 58–62, reproduced with permission of the publisher.

deformation localizes in ferrite because of its higher formability at high temperature with respect to austenite; local shear bands can also form, favoring the nucleation of voids (in particular if inclusions, such as sulfides, are present). The fraction of ferrite in the microstructure plays a decisive role. The picture is similar, but opposite, to that encountered in the ductility trough of carbon steels (Section 8.4). In duplex steels, the fraction of ferrite in the microstructure increases with temperature. At high temperatures, it is very high, more than ~70% for temperatures beyond 1170°C, and plastic deformation is rather homogeneous. However, as temperature decreases, the fraction of ferrite also decreases. This, in turn, induces a decrease in ductility that is very pronounced when the ferrite fraction is in the range 20%–30%. In this condition, the constraint exerted by the austenitic regions and the corresponding strain localization in ferrite is maximum. In addition, with increasing strain rate the tendency of local shear localization in the ferritic areas increases, and this explains the transition, at a certain temperature, towards the formation of surface cracks with increasing strain rate (Martin et al., 2012).

The ductility drop at comparatively low temperatures is then a ductility trough, as discussed in Section 8.4. The data in Fig. 8.19, referring to a conventional 2205 duplex stainless steel, show that PDT is around 1200°C. Therefore the map of Fig. 11.19B shows that duplex stainless steels must be forged at rather high temperatures (but not too high, to avoid excessive surface oxidation or hot shortness), and at comparatively low strain rates. In fact, if strain rate is too high, the temperature rise due to the plastic deformation is also high, favoring the attainment of the conditions for hot shortness (also in consideration of the comparatively low thermal diffusivity of the alloy). The optimum temperature window for duplex stainless steels is then in the range 1000–1200°C, depending on the chemical composition and strain rate. The control of the temperature uniformity in the forging is crucial; if the temperature becomes too low at edges or thin sections, the local ductility may decrease, allowing for the formation of ductile cracks.

As seen in [Section 9.4](#), in duplex stainless steels, embrittling phases (such as χ - or σ -phases) may precipitate at the grain boundaries, depending on the steel composition, workpiece size, temperature, and exposure time (typical critical temperature interval: 600–1000°C). Even if plastic deformation may accelerate the precipitation kinetics, the times required for precipitation are generally too long to induce an appreciable embrittlement during hot forging in the right process window. However, the thermal history of these steels must be properly controlled since possible embrittling precipitations may occur if the temperature drops too low. In any case, after hot forming, the steels should be quenched and then possibly subjected to high-temperature solution annealing followed by rapid quenching, to dissolve any precipitate and obtain a microstructure with the right balance of ferrite and austenite (before the use or any subsequent cold forming process). Charpy V-notch impact testing (see [Chapter 9](#)) is a proper method to evaluate the quality of the final products.

Formability map for a Ti-64 alloy

To detect the possible presence of internal cracks and achieve a complete picture of the formability of an alloy, laboratory tests must be supplemented with a careful microstructural analysis of the cross sections of the deformed material. An example is presented here regarding the hot formability of a **Ti-64 alloy**, which is known to be particularly prone to internal cracking.

In the considered example, ingot breakdown was conducted in the β -field (the β -transus of the alloy is 1010°C). After air cooling, the typical lamellar microstructure was obtained. Compression tests were then carried out in a protective atmosphere (to avoid surface oxidation) using cylindrical specimens and with a lubricant to reduce friction. The temperature range was 750–1100°C (recall that: $0.4T_m \sim 500^\circ\text{C}$ for this alloy), and the strain rate interval was 1×10^{-4} – 10 s^{-1} . The tests were carried out for a total height reduction of 55% ([Seshacharyulu et al., 2002](#)).

[Fig. 11.21](#) shows the obtained Type I map, integrated with the observed deformation mechanisms. For temperatures lower than 900°C and strain rates greater than 10^{-1} s^{-1} , ASBs were observed to form in the shear regions; an example is shown in [Fig. 11.22A](#). Their formation is favored by increasing the strain rate. At very high strain rates, beyond 1 s^{-1} , internal ductile cracks were detected in the shear bands (refer to [Fig. 11.22B](#)). Such cracks produced a macroscopic fracture along the shear region like that shown in [Fig. 11.15](#). For temperatures lower than 900°C and strain rates lower than 10^{-1} s^{-1} , voids were detected at the prior β boundaries and in the barreled region (Region II in [Fig. 11.6B](#)). An example is shown in [Fig. 11.22C](#). They formed in the region where circumferential tensile stress developed during upsetting. However, the deformation path was not sufficiently severe to allow the formation of extensive surface ductile cracks (as said, the authors minimized friction at the workpiece-die interfaces). In any case, the temperature of 900°C can be considered as a lower limit (as discussed in [Section 8.4](#), this temperature may be representative of a ductility trough).

With increasing temperature above 900°C, the picture changes. As seen in [Fig. 8.6](#), ductility of the α - β Ti-64 alloy increases significantly with temperature. Therefore no ductile cracks were detected in the barreled region. In addition, the tendency of shear

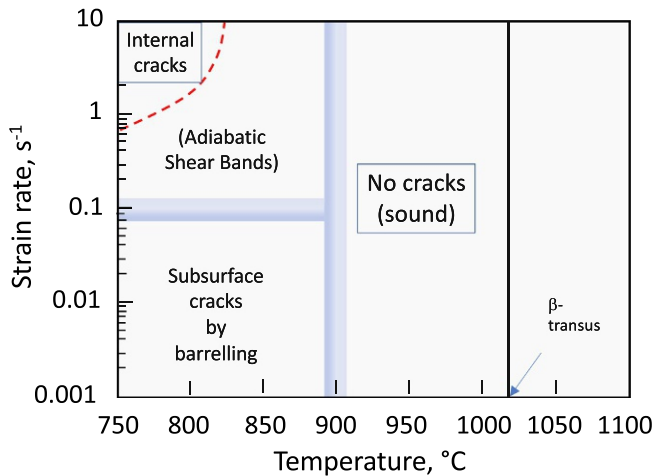


Fig. 11.21 Type I hot formability map for a Ti-64 alloy. The microstructural modifications are also indicated.

Adapted and redrawn from Seshachryuli, T., Medeiros, S.C., Frazier, W.G., Prasad, Y.V.R.K., 2002. Microstructural mechanisms during hot working of commercial grade Ti-6Al-4V with lamella starting microstructure. *Mat. Sci. Eng. A*, 325, 112–125.

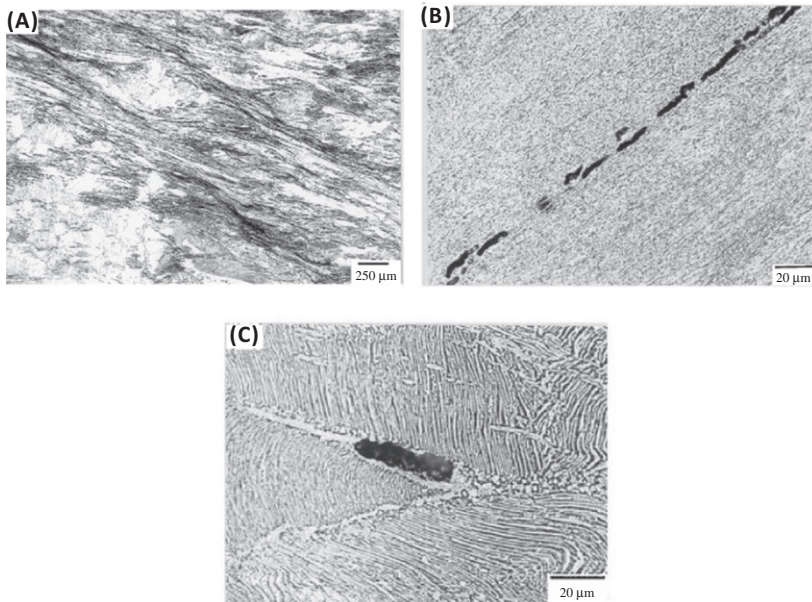


Fig. 11.22 Hot compression testing of a Ti-64 alloy. (A) Example of shear flow localization; (B) formation of a ductile crack with an ASB; (C) Formation of a void in the barreled region at low temperature and low strain rates.

From Seshacharyulu, T., Medeiros, S.C., Frazier, W.G., Prasad, Y.V.R.K., 2002. Microstructural mechanisms during hot working of commercial grade Ti-6Al-4V with lamella starting structure. *Mater. Sci. Eng. A*, 325, 112–125, reproduced with permission of the publisher.

strain localization also decreased (because of the increase in temperature and the corresponding decrease in the specific plastic energy), and hence no shear bands (and, of course, internal cracks) were detected. As a matter of fact, at high strain rates ($>10^{-2} \text{ s}^{-1}$), a kinking of the microstructural lamellae was observed, whereas, at lower strain rates, a globularization of the lamellae occurred, with the formation of a fine microstructure with a high plastic deformability.

As mentioned, Ti alloys are rather difficult to forge. In general, most Ti alloys are hot forged at a temperature below β -transus (α - β forging) and at comparatively low strain rates. For Ti-64, the safe temperature is in the range of 900–980°C. Die heating and control of the die temperature is also crucial (when using a low strain rate, the workpiece cooling could be excessive if the die temperature is not kept sufficiently high, at least 400°C). Forging in the β phase is typically performed to break down the solidification microstructure, but it is subsequently avoided since it leads to excessive grain growth. In the forging of Ti alloys, the involved flow stresses (and loads) are comparatively high (Table 4.4) and this increases the difficulty of the process.

Formability of Mg alloys

As already explained, Mg alloys are characterized by a very poor RT plastic formability. The typical low n values give rise to low uniform elongation values and low strains at fracture, due to the early voids coalescence by localized necking. Consequently, Mg alloys are typically formed in only hot conditions.

As an example, Fig. 11.23 shows a Type II formability map for a **ZK60 Mg alloy** (nominal chemical composition: Mg-6%Zn-0.5%Zr) at different temperatures and at 10 s^{-1} (Ogawa et al., 2002). This strain rate was selected since it is a typical strain rate for mechanical presses (refer to Table 11.1). Some deformed samples are shown in Fig. 11.24. At 100°C, formability is very low (and is even lower at RT) and the

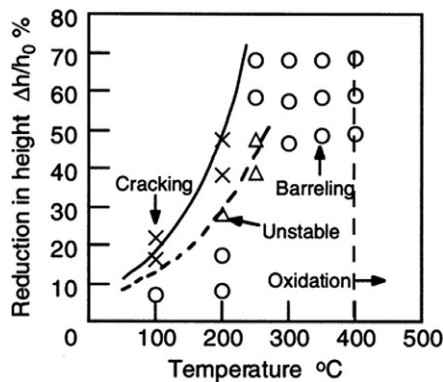


Fig. 11.23 Height reduction-temperature map for the ZK60 Mg alloy, showing the safe and unsafe process regions.

From Ogawa, N., Shiomi, M., Osakada, K., 2002. Forming limit of magnesium alloy at elevated temperatures for precision forging. *Int. J. Mach. Tools Manuf.* 42, 607–614, reproduced with permission of the publisher.

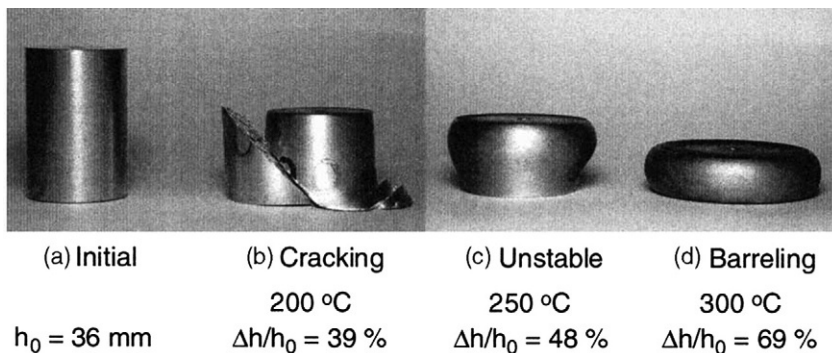


Fig. 11.24 Results of upsetting tests of billets made of the ZK60 Mg alloy.

From Ogawa, N., Shiomi, M., Osakada, K., 2002. Forming limit of magnesium alloy at elevated temperatures for precision forging. *Int. J. Mach. Tools Manuf.* 42, 607–614, reproduced with permission of the publisher.

maximum reduction is approximately 10%. At 200°C, the maximum height reduction is approximately 39%. The specimens display macroscopic cracking, induced by internal shearing. For temperatures higher than 250°C, plastic deformation proceeds without surface cracking, even if at 250°C the deformation is unstable for reductions in the range of 40%–50% (due to compression instability). The comparison with the data in Fig. 8.6 for an AZ31 Mg alloy clearly shows that Δh_{max} is proportional to the tensile ductility, as explained in the previous section.

In conclusion, to carry out safe forging processes at 10s^{-1} and with large reductions, the working temperature should be above 250°C. However, the temperature should be lower than 400°C to avoid extensive surface oxidation.

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Edge cracks in a cold rolled 5754 Al sheet (thickness 1 mm).

Chapter outline

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During the rolling process, the height of a workpiece is reduced by forcing it to pass through the gap between two counter-rotating rolls. As in forging, plastic deformation is achieved by compression. Rolling processes are mainly used in the production industry to obtain standard products from intermediate products. Rolling treats more metals than any other metalworking process, and is mainly used to produce steel products.

Flat rolls are adopted to obtain flat products (such as plates and sheets). On the other hand, grooved rolls are used to obtain long products, such as bars or shapes. The rolling processes are mainly carried out in hot conditions. Cold rolling is employed to obtain thin sheets for subsequent sheet metal forming operations. In any case, the control and minimization of rolling defects, such as edge and internal cracks, are very important to avoid expensive consequences, such as the rejection of rolled material or threats to the safety of the rolling equipment.

Most of the available technology of rolling has been developed for the rolling of steels. However, flat rolling is also important to produce plates and sheets (also with very low thickness) made of non-ferrous alloys, such as Al alloys. In general, the concepts discussed herewith are useful for the development of rolling procedures for specific alloys.

12.1 Rolling processes

Table 12.1 summarizes the main rolling processes and mills used to produce standard **steel** products from intermediate products. Many rolling configurations are possible, and the table provides a simple overview. Each pair of rolls is placed in a frame and each roll is moved by a motor through two spindles, thus forming the so-called *mill stand*. In the mill stand, the rolls can be arranged in different ways. The *two-high* and *four-high mills* use two and four rolls, respectively. As shown in **Fig. 12.1**, in a common four-high mill, the two *working rolls* have a smaller diameter than the two *supporting* (or *backup*) *rolls*, which are in contact with the working rolls along their axes. This configuration is used for decreasing the operating load and power and, at

Table 12.1 Hot and cold rolling processes and mills used to produce standard steel products from semi-finished products.

	Typical mills	Compressive plastic strain and strain rate	Interpass time (s)
Ingot blooming	Two-high reversing mill	0.1–0.4 (for each pass) $1\text{--}30\text{ s}^{-1}$	2–10 (up to 100)
Hot strip rolling			
Roughing	Four-high reversing mill (One or two mill stands)	>0.6 (for each pass) $10\text{--}200\text{ s}^{-1}$	0.4–4
Finishing	Finishing mill (5–10 four-high stands, with a spacing of 5–6 m; see Fig. 10.1)	>0.6 (for each pass) >4 (total) $10\text{--}200\text{ s}^{-1}$	0.4–4
Hot plate rolling			
Roughing	One four-high reversing mill	0.1–0.4 (for each pass) $1\text{--}30\text{ s}^{-1}$	8–10 (up to 100)
Finishing	One four-high reversing mill		

Table 12.1 Hot and cold rolling processes and mills used to produce standard steel products from semi-finished products—cont’d

	Typical mills	Compressive plastic strain and strain rate	Interpass time (s)
<i>Hot rolling to obtain rods and bars</i>			
Roughing	Roughing mill (4–10 four-high stands) Finishing mill (10 or more stands with grooved rolls)	0.2–0.6 (for each pass) 1–8 (total) 0.5–2500 s ^{−1}	0.4–4
Finishing			
<i>Hot rolling to obtain shapes</i>			
Roughing (to obtain the leader pass) Finishing	One two-high reversing mill Finishing train (8 or more stands with grooved rolls)		
<i>Cold rolling to obtain sheets or foils</i>			
	Mill with 4–6 four-high stands		

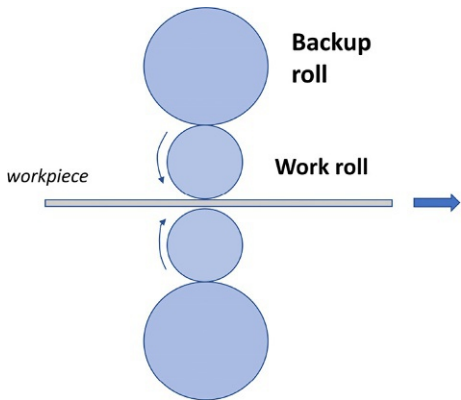


Fig. 12.1 Arrangements of rolls in a four-high mill.

the same time, for increasing the rigidity of the mill, thereby reducing the bending of the working roll. For specific applications, a *cluster* of rolls is used, which can increase the system rigidity and allow the use of working rolls with very small diameters.

Ingot blooming is analogous to ingot forging. A *universal* reversing mill equipped with two vertical rolls (called *edgers*) to avoid lateral spreading is usually utilized. The ingots, with a thickness up to 900mm, are heated to 1150–1200°C (standard hot working). The workpiece is fed into the roll gap from one side. After the rolling deformation and the subsequent roll gap reduction, the workpiece is fed from the

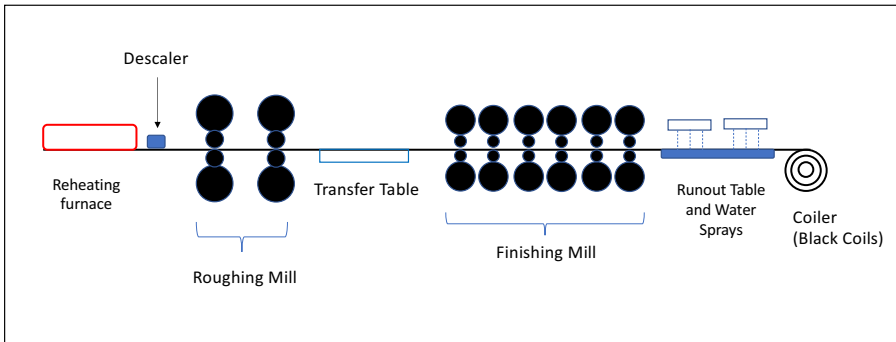


Fig. 12.2 Hot strip rolling process.

opposite side. This process is repeated in subsequent passes. The occurrence of dynamic recrystallization is rather difficult, whereas the occurrence of static recrystallization between two consecutive passes is more likely, if the interpass time is sufficiently long.

The *hot strip (flat) rolling* process is schematized in Fig. 12.2. In this process, the slabs obtained from continuous casting (or ingot blooming) are transformed into strips of thickness ranging between 1 and 10 mm. First, the reheated slabs (1200–1250°C) are descaled by using a high-pressure water jet to remove the primary oxides from the surface. With one or two reversing roughing mills, the thickness is reduced to about 25 mm. Correspondingly, the workpiece temperature decreases to about 1000°C. The workpiece is then transferred to the *finishing rolling mill* to obtain strips that are cooled in the runout table (using water sprays) and coiled to obtain the *black coils*. Depending on the microstructural requirements, the intensity of cooling must be properly designed to obtain *hot coiling* (workpiece temperature above approximately 660°C) or *cold coiling* (workpiece temperature less than approximately 660°C). Static recrystallization typically occurs in the interpass time between each pass. Subsequently, the hot rolled strip is subjected to cold rolling to reduce the thickness down to 1 mm or less. First, the surface oxides are removed by *acid pickling*, followed by cold rolling under lubricated conditions to obtain products with an excellent surface finish. The product is then coiled to obtain *white coils*, which are later cleaned (to remove the lubricant) and annealed. Finally, the product is submitted to *temper rolling* (or *skin passing*) to eliminate the yield-point phenomenon. The production chain is shown in Fig. 12.3. Thin sheets are typically made by very low C steels (<0.1%) to ensure high plastic deformability after annealing and skin passing. The sheets are mainly used for stamping operations, including stretch forming and deep drawing operations. Recently, advanced high-strength steel sheets

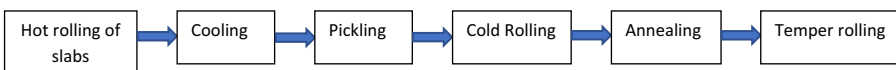


Fig. 12.3 Production chain to obtain steel sheets for subsequent stamping operations, including stretch forming and deep drawing.

have been proposed, mainly by the automobile industry. These steels comprise BA steels, dual phase steels, and TRIP and TWIP steels. They are typically characterized by a lower plastic deformability compared to mild (low C) steels, but display a higher strength, which is useful for reducing the thickness and weight of the produced parts.

With *hot plate rolling*, the slabs are transformed into plates with a thickness of a few centimeters. Generally, C-Mn steels are used, and the process is carried out using one or two four-high reversing mills. Hot rolled plates are typically used for pressure vessels, pipelines, or ships.

Long products, such as *rods and bars*, are obtained from billets. A rather high number of reductions in grooved rolls is adopted, comprising roughing, intermediate, and finishing rolling passes. For example, wire rods with a typical diameter of 5.5 mm, used for subsequent cold drawing operations, are rolled in 4–6 roughing stands, 10–12 intermediate stands, and 14–16 finishing passes carried out in a compact rolling machine. These wire rods are often made with pearlitic (high C) steels. In this case, after hot rolling, they are subjected to controlled cooling to obtain full pearlitic microstructure (Fig. 7.2). Steels for mechanical applications (with a wide range of chemical composition) are typically produced as long products. They are obtained as standard products for subsequent forging or drawing processes.

Hot rolled long products such as beams, profiles, channels, and rods for reinforced concrete (generally made of C-Mn steels; for specific applications austenitic stainless steels or duplex steels are also employed) are end products for buildings, bridges, and other applications. Shapes like beams are obtained from blooms or *beam blanks* that are produced by continuous casting and have a shape like that of the beam to be produced. Roughing is carried out using one two-high reversing mill, and the rolls have different channels that modify the shape of the workpiece. The final product is obtained in the finishing mill. Universal mills with vertical rolls can also be used to profitably roll external surfaces (for example, refer to Fig. 12.4, for the production of an H-beam).

As mentioned, flat rolling is also important in the production of plates and sheets (even with very low thickness) made of **Al alloys** (and, to a lesser extent, Cu, Ti, and Mg alloys). For this purpose, different mill configurations are available. Hot roughing

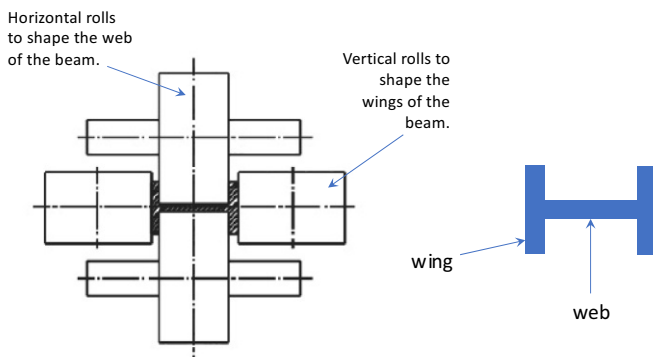


Fig. 12.4 Rolls configuration in a finishing mill stand to produce an H-beam.

operations are typically carried out using single four-high reversing mills. Slabs obtained by direct chill casting are used after a homogenization treatment. Hot finishing operations are then carried out using single four-high reversing mills, or a finishing mill with 2–4 four-high mills. Static recrystallization typically occurs in the interpass times and grain growth is limited by the presence of the microstructural dispersoids. Cold rolling is then carried out to obtain thin sheets. With proper lubrication, a mill of two or three four-high mills are used. Cold rolling is generally followed by static annealing of the final product.

12.2 Mechanics of rolling

Definitions and entry conditions

Fig. 12.5 shows the geometry of the workpiece in initial contact with rotating rolls (for symmetry, only the upper roll is shown). In the contact region, indicated with ΔA , a normal repulsion force F is established, and its horizontal component is $F \cdot \sin \Phi$, where Φ is the *angle of bite*. Due to the sliding in the contact region, a frictional force, T , is also established, whose horizontal component is $T \cdot \cos \Phi$. The rotating rolls are then able to start the process, i.e., drag the workpiece into the throat of the roll, if: $T \cdot \cos \Phi \geq F \cdot \sin \Phi$. The entry condition is then given by: $\mu \geq \tan \Phi$, indicating that the coefficient of friction at the workpiece-roll contact must be sufficiently high to start the process. This explains the well-known slogan: “without friction no rolling” (Vergne et al., 2006).

A more useful equation is obtained by replacing $\tan \Phi$ with a simplified relationship. The length of the *arc of contact*, L , is approximately given by its projection on the horizontal axis (refer to Fig. 12.6). Since the reduction, ΔH , is given by: $\Delta H = H_0 - H_f$, we have:

$$L = \sqrt{R^2 - \left(R^2 + \frac{\Delta H^2}{4} - R \cdot \Delta H \right)} \cong \sqrt{R \cdot \Delta H} \quad (12.1)$$

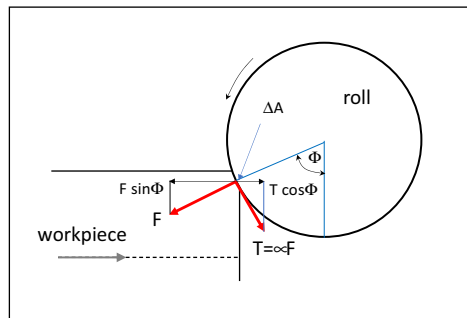


Fig. 12.5 Geometry of the workpiece in initial contact with rotating rolls (for symmetry, only the upper roll is shown). If the angular velocity is ω , the surface velocity of the rotating roll is: $v_r = \omega R$, where R is the roll radius.

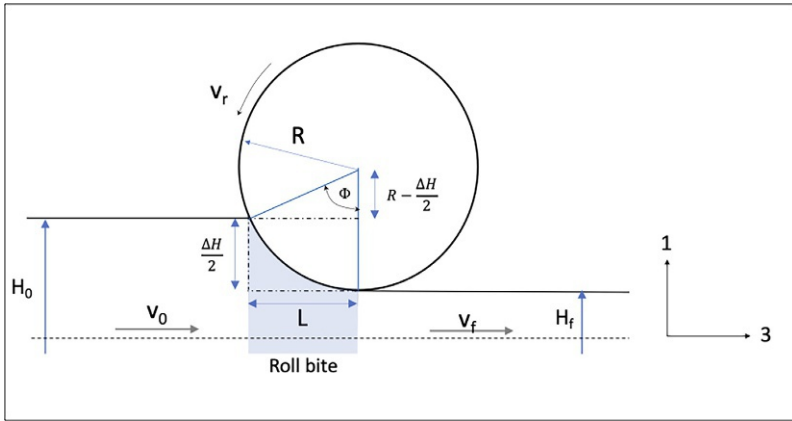


Fig. 12.6 Rolling gap and definition of the length of the arc of contact (note that the reduction in the picture is exaggerated; in general, it is much smaller and L approximates the length of the arc of contact in a reasonable manner). Direction 2 (not indicated) is parallel to the roll's axis.

where R is the roll radius. Therefore:

$$\tan \Phi = \frac{L}{R - \frac{\Delta H}{2}} \cong \sqrt{\frac{\Delta H}{R}}$$

The entry condition then becomes:

$$\mu \geq \sqrt{\frac{\Delta H}{R}} \quad (12.2)$$

and the maximum possible reduction is:

$$\Delta H_{max} = \mu^2 \cdot R \quad (12.3)$$

The friction coefficient is typically in the range of 0.3 and 0.6 for steels and Al alloys (Pellizzari et al., 2005). To reduce the friction coefficient (broadly to 0.15–0.4), specific lubricants (typically emulsions) with proprietary compositions are used in the case of rolling Al and Cu alloys. In cold working, lubrication is always used, and the friction coefficient is approximately 0.1 or lower. A reduction in the friction coefficient implies a decrease in the rolling load and power and in the probability of obtaining cracks in the rolled products (see next section), but makes the rolling process more difficult. This means that a higher number of passes are required to reach the desired reduction.

As indicated in Fig. 12.6, the workpiece is dragged into the rolling gap with a velocity v_0 , due to the action of the friction force. Due to compressive deformation, each material volume is elongated in direction 3 in the roll gap, causing an increase in its velocity. At a specific point along the arc of contact, the metal velocity becomes equal

to the surface roll velocity, v_r , and no sliding at the workpiece-rolling interface occurs. This is known as the *neutral point*. In the region between the neutral point and the exit, the friction force is directed in the opposite direction with respect to the workpiece motion. Therefore it exerts a resisting action at the workpiece motion. In the case of plane strain conditions ($\epsilon_2 = 0$), the relationship between v_0 and v_f can be simply obtained by the constant metal flow at the entry and the exit:

$$v_f = v_0 \cdot \frac{H_0}{H_f} \quad (12.4)$$

Different relationships were proposed to relate v_f and v_r (which is an input parameter), through which both v_0 and v_f can be estimated (Ford and Alexander, 1963).

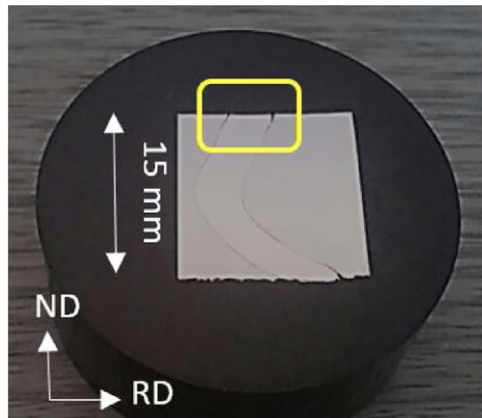
In the case of flat rolling, the parameter Δ of Eq. (10.4) is given by the following relationship (Hosford and Caddell, 1993):

$$\Delta = \frac{2-r}{2} \cdot \sqrt{\frac{H_0}{r \cdot R}} \quad (12.5)$$

where r is given by $\Delta H/H_0$. When the rolling workpiece is thick with respect to the roll radius, Δ is greater than 1, and plastic deformation is inhomogeneous, being mainly concentrated in the regions close to the rolls. For $\Delta < 1$ plastic deformation is expected to be homogeneous (by compression). However, friction at the workpiece-rolls interface may play an important role, in particular in case of hot rolling. Friction induces an increase in the longitudinal plastic deformation in the workpiece regions close to the rolls because of the rolls notation. Fig. 12.7 shows a cross section of a steel slab hot rolled from $H_o = 25$ mm to $H_f = 15$ mm ($\Delta H = 10$ mm, and $r = 0.4$) where a stainless steel pin was inserted perpendicular to the rolling plane (Nioi et al., 2019). The roll radius, R , is 111.5 mm, and therefore $\Delta = 0.6$. Plastic deformation is then expected to be homogeneous. However, Fig. 12.7 shows that deformation significantly changes through the thickness since the friction at the

Fig. 12.7 Cross section of a steel slab containing a pin perpendicular to the rolling plane. The deformed pin after rolling shows an increased deformation in the rolling direction of the metal close to the rolls.

From Nioi, M., Pinna, C., Celotto, S., Swart, E., Ferrugia, D., Husain, Z., Ghadbeigi, H., 2019. Finite element modelling of surface defect evolution during hot rolling of Silicon steel. J. Mater. Process. Technol. 268, 181–191, reproduced with permission of the publisher.



interface with the roll (estimated friction coefficient: 0.25) induces a distortion of the pin, showing an increased plastic flow in the rolling (longitudinal) direction close to the roll surfaces.

Contact pressure and rolling (or roll separating) load

The contact pressure at the roll-workpiece interface depends on the intensity of the plastic deformation and friction. The case of flat rolling in plane strain conditions, i.e., without plastic spreading along the rolls' axes ($\varepsilon_2 = 0$), will be considered. The situation is similar to cogging (Fig. 11.1). This assumption is acceptable when the width-to-thickness ratio (w_0/H_0) of the workpiece is larger than 8–10.

A simplified analysis is presented, starting from the assumption that plastic deformation is homogeneous in the bite. The compressive plastic strain in direction 1 is then given by:

$$\varepsilon_1 = \ln \frac{H}{H_0} \quad (12.6)$$

where H is the thickness of each slab moving along the arc of contact (see also Fig. 10.2). The corresponding effective strain is: $\varepsilon_{eff} = \frac{2}{\sqrt{3}} \cdot |\varepsilon_1|$. The total applied strain is obtained by setting $H = H_f$. Neglecting friction, the local pressure along the arc of contact is:

$$p = |\sigma_1| = \frac{2}{\sqrt{3}} \cdot \sigma_{eff} \quad (12.7)$$

As in the case of forging, the presence of friction induces a change in the evolution of p along the arc of contact, with the formation of a friction hill, reaching the maximum at the neutral point (Ginzburg and Ballas, 2000). The rolling (or roll separating) load, F , can then be calculated by the integral of the pressure, p , along the arc of contact. A simple approach is to use the following relationship:

$$F = A_n \cdot \bar{\sigma}_1 \cdot G = w_0 \cdot L \cdot \bar{\sigma}_1 \cdot G \quad (12.8)$$

where

- A_n is the nominal (projected) area of contact;
- w_0 is the width of the workpiece (that remains unchanged because of the assumption of plane strain conditions);
- L is the length of the arc of contact (Eq. 12.1);
- $\bar{\sigma}_1$ is the mean flow stress in plane strain conditions;
- G is a parameter that accounts for the friction hill.

For the estimation of $\bar{\sigma}_1$, the L-H equation may be used in the case of cold rolling, considering plane strain conditions. In the case of hot rolling, the role of primary creep should be considered, since the strains applied in each rolling pass are generally low.

Therefore the Voce or other similar equation (Chapter 4) should be used. During the analysis, the role of recrystallization should also be taken into account, and the task might be rather difficult. In general, the applied strains are too low for DRX, but SRX may occur in the interpass time (and this increases ductility and decreases the flow strength for the subsequent pass). In some cases, such as in the hot finishing rolling to obtain rods, bars, or strips, SRX may not occur between two subsequent rolling passes, especially in steels that contain elements that retard it (such as Nb in HSLA steels), since the interpass time is too short. In this case, $\bar{\sigma}_1$ must be calculated considering the accumulated plastic strain. However, it is also possible that at a certain mill stand, the accumulated strain becomes sufficiently high to promote DRX (ϵ accumulated $> \epsilon_c$). In this case, the softening induced by DRX must be considered (Section 4.3; see also Siciliano et al., 1996). It must also be considered that rolls are cooled and hence the workpiece is chilled in the contact region with the rolls. This may influence the plastic deformation of the workpiece and its recrystallization behavior.

For the estimation of parameter G , different relationships are proposed in the literature. A simple one is the following (Roberts, 1988):

$$G = \exp\left(\frac{\mu \cdot L}{H_0 + H_f}\right) \cong 1 + \frac{\mu \cdot L}{H_0 + H_f} \quad (12.9)$$

If the initial width-to-thickness ratio of the workpiece (w_0/H_0) is lower than 8–10, plastic spreading along the roll's axis must be considered and deformation is no longer under plane strain. The spreading coefficient, S , is used (Eq. 11.9): $S = \epsilon_2/\epsilon_1$. Different empirical equations are obtained. If w_0/H_0 is between 1 and 2, the following equation was proposed (Richardson et al., 1985):

$$S = -0.61 \cdot \left(\frac{H_0}{w_0}\right)^{1.3} \cdot \exp\left[-0.32 \cdot \left(\frac{H_0}{L}\right)\right] \quad (12.10)$$

If w_0/H_0 is between 2 and 6, the equation obtained by Helmi and Alexander can be used (Helmi and Alexander, 1969).

In the hot rolling of long products, the calculation of the rolling load is rather complex. Different models are proposed to simplify the analysis, such as the method based on the equivalent rectangular cross section with the maximum width (Lee et al., 2002).

In the case of **cold rolling**, the roll separating load can be rather high due to the accumulated strain hardening. This causes an elastic flattening of the rolls and the actual radius of curvature of the roll in the contact region, R' , increases with respect to the nominal roll radius, R . It has been shown that the roll force increases drastically when the workpiece thickness is below 1 mm. This is attributed to the severe roll flattening. Specific models are under continuous development for the correct evaluation of the roll forces in these very particular conditions.

The high roll separating loads in cold rolling can be alleviated by applying a front and back tension, σ_f and σ_b , respectively, that facilitates the compressive plastic deformation. This can be done by properly adjusting the rolling velocities of two

neighboring rolling mills. With reference to the Tresca criterion, the front and back tension reduces the rolling force by decreasing $\bar{\sigma}_1$ to approximately $(\bar{\sigma}_1 - \sigma')$, where σ' is the average between σ_t and σ_b . In flat cold rolling, the metal spreading along the roll's axis can be generally neglected given the comparatively high w_0/H_0 ratio.

From the roll load, F , the required *roll torque*, M , can be calculated as: $M = a \cdot F$, where a is the lever arm that is a fraction of L , and is close to 0.5 (see [Ginzburg and Ballas, 2000](#), for different data and models for the lever arm). Therefore the *roll power*, W , for the plastic deformation is: $W = 2 \cdot M \cdot \omega$, where ω is the roll's angular velocity. To obtain the power requirement for the driving motor, the mechanical losses in the drive mill must also be considered. As explained in Section 10.2, the evaluation of the roll separating load, F , and the corresponding roll power, W , is pivotal in determining the correct process design. In fact, if the calculated values result over the limit for the equipment the involved reductions should be decreased. In case of hot rolling, the preheating temperature should be possibly increased.

12.3 Temper rolling

Temper rolling, also known as *skin-passing*, is applied to cold rolled sheet coils after annealing ([Fig. 12.3](#)). Temper rolling is the last process of the sequence.

The applied reduction is low, in the range of 0.5%–6%, and it induces a small surface plastic deformation by compression. The internal part of the workpiece remains in the elastic deformation and, after rolling, the surface layer is characterized by compressive residual stresses, whereas the inner part of the sheet displays tensile residual stresses. The process is generally carried out without a lubricant, in order to maintain the surfaces clean ([Roberts, 1988](#)). Temper rolling improves the sheet flatness and the surface roughness. However, it is mostly aimed at preventing the formation of the Lüders bands during the subsequent stamping operations ([Section 15.4](#)). With increasing skin passing reduction, the yield point phenomenon gradually disappears, and the metal yield strength correspondingly decreases. This decrease is induced by the presence of internal residual tensile stresses. However, with further increase in the reduction, the metal yield strength starts to increase, as schematized in [Fig. 12.8](#). This phenomenon must be minimized, as an increase in the yield strength is accompanied by a decrease in the strain-hardening coefficient, n , and, in turn, in the ductility ([Chapter 7](#)). The *optimum degree of temper rolling* depends on the metal properties and is approximately 1%–2%.

For the correct design of the temper rolling process, two additional factors should be considered. The first one is the possible formation of Lüders bands during the production process. Such defects are called *edge breaks* or *edge strains* since they are initiated at the edges of the sheets, from local imperfections. They form during the uncoiling processes, after annealing and before skin passing. In fact, uncoiling produces a bending of the sheet and, as seen in [Section 15.3](#), tensile stresses and strains on the sheet surfaces that may produce the Lüders bands due to local stress/strain intensifications ([Tan and Liew, 2021](#)). Some examples are shown in [Fig. 12.9](#) ([Zhou et al., 2019](#)). These defects can be eliminated by the subsequent temper rolling

Fig. 12.8 The influence of temper rolling reduction on the stress-strain behavior of an annealed steel.

Adapted and redrawn from Stelzer, R., Putz, P.-D., Diegelmann, V., Gorgels, F., Piesack, D., 2005. Optimum temper rolling degree: pre-set and influencing effects of bending deformations. *Steel Res. Int.* 76, 105–110.

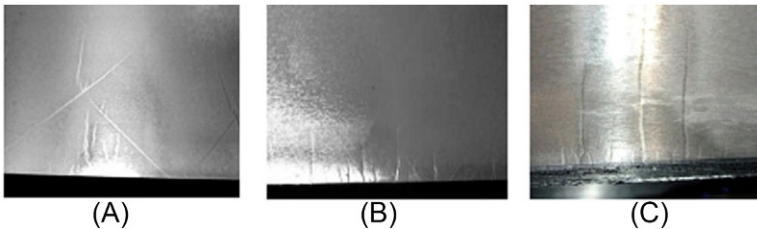
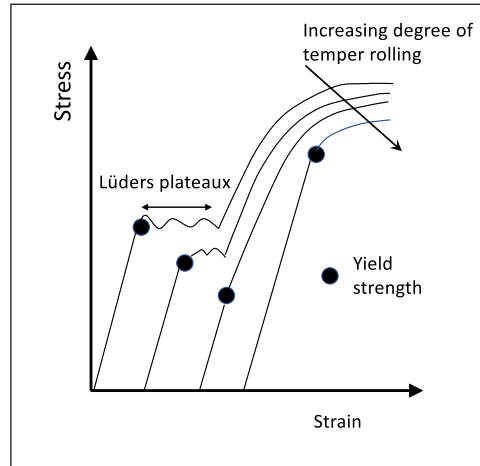


Fig. 12.9 Examples of typical edge breaks.

From Zhou, T., Zurob, H., Zhang, P., Kuuskman, K., Cho, S.H., Burella, D., 2019. Control of edge breaks during cold mill processing of commercial and drawing quality low-carbon steels. *Ironmak. Steelmak.* 46, 656–662, reproduced with permission of the publisher.

treatment, if the corresponding reduction is greater than the Lüders plateau of the edge breaks. As mentioned, the reductions are on the order of 1%. Any edge break with a Lüders plateau lower than 1% elongation can be easily eliminated.

The second factor is the *shelf life* of the material. In fact, after temper rolling, the coils are generally stored before use. Therefore it is important to avoid strain aging, i.e., the reappearance of the upper and lower yield point phenomenon (Section 1.4), if the storage, even at RT, is too long. This would induce the formation of the surface stretcher markings during the subsequent stamping operations. As an example, in an investigation carried out with a DC01 mild steel sheet (C content: 0.036%), subjected to temper rolling with a reduction of 1%, signs of aging were visible after 3 months, and the reappearance of the upper and lower yield point phenomenon was evident after 4 months (Colak, 2021). In general, 1% of reduction is recommended for this kind of

steel to ensure a shelf life of 3 months. Increasing the reduction degree induces an increase in the shelf life.

12.4 Ductile cracks in rolled products

Rolled products may contain different defects, such as geometrical, surface, and internal defects (Dieter, 1988). In hot rolled parts, surface oxides are typically present on the surface and must be thoroughly removed. Surface oxidation can also interact with surface defects in ingots or semi-finished continuous casting products. Depending on the casting process, products may contain defects such as cracks or grooves, which are oxidized and may grow during the hot rolling process if not removed previously.

In cold rolling, an inadequate lubrication may induce adhesive or abrasive wear on the surfaces of the rolled products. Consequently, surface defects may appear, enhanced by the possible presence of surface inclusions. The occurrence of defects like *scratches*, *flakes* (peelings), *dents*, or *holes* is observed. Such defects are detrimental to the surface appearance of the sheets and must be kept to a minimum. This could be done by optimizing lubrication and reducing the fraction of inclusions, such as Si and Al oxides, in the starting slabs (Yu et al., 2013). Inclusions are particularly damaging, as they may induce a local inhomogeneous plastic flow with the formation of another type of defect, i.e., *slivers*. They are regions with overlapping material along the inclusions (Dhua, 2019). Edge strains are other possible surface defects in cold rolled sheets that can be eliminated by a proper temper rolling process, which has been seen in the previous section.

The most detrimental defects in rolled products are ductile cracks, and in particular the *edge* and *internal cracks*. Edge cracks are surface cracks that form during rolling at the edges of the workpiece. When present, they must be removed before further processing. This could lead to a considerable increase in the manufacturing costs.

Edge cracks may form in flat products, both under hot and cold conditions, and in rolled shapes. The occurrence of edge cracks is a consequence of an inhomogeneous plastic deformation, with the corresponding appearance of secondary tensile stresses, possibly coupled with a reduction in the metal ductility; they will be considered first. At the end of the section, the internal crack formation leading to fracture by *central bursting* and *alligatoring* is also considered.

Surface cracks in hot rolled products

Edge cracks form during rolling at the edges of the workpiece when the local effective strain becomes equal to the local effective fracture strain: $\epsilon_{eff} = \epsilon_{eff.f}$. They are a special case of surface cracks.

A pivotal role is played by the inhomogeneous plastic flow that induces the formation of secondary tensile stress with the local increase in the triaxiality factor, T . In the case of a series of subsequent rolling passes, the ductile crack can also propagate towards the internal part of the workpiece by a low-cycle fatigue mechanism.

Different phenomena may be responsible for the appearance of such an inhomogeneous flow at the workpiece edges. First, let's consider the case of hot flat rolling, with a comparatively low initial width-thickness ratio (such as hot strip/plate rough rolling), for which lateral spread along the roll axis is important. Due to the presence of friction at the workpiece-roll interface, plastic deformation is inhomogeneous. In fact, the deformation along the roll axis is lower at the center, where the action of friction in opposing such deformation is high due to the presence of the friction hill (see Fig. 12.10, in the top part). As a consequence, there is a tendency to have a higher plastic deformation in the rolling direction at the center, in comparison to the regions close to the edges (consider that the deformation in the thickness direction is not affected by the friction hill). However, there must be a continuity in the geometrical deformation. This infers that in the edge regions, a tensile stress is developed in the bite (it is a longitudinal stress, oriented along the rolling direction), and then the stress triaxiality factor, T , locally increases. In addition, barreling also plays an important role, as depicted by Fig. 12.10, in the bottom part. As explained in Section 11.2, the state of stress and strain changes in the barrel region. In particular, the tensile (longitudinal) stress increases, whereas the vertical (along direction 1) compressive stress decreases in magnitude. Therefore in the barrel region the triaxiality parameter significantly increases and the local effective fracture strain decreases (to summarize, both spread and barreling contribute to the local tensile longitudinal stress, and barreling also reduces the local compressive stress). Therefore the conditions for the formation of ductile cracks can be attained if the metal ductility is sufficiently low.

In this respect, it must also be considered that, close to the edges, the cooling rate is faster than in the center of the workpiece, and this may induce a local reduction in the

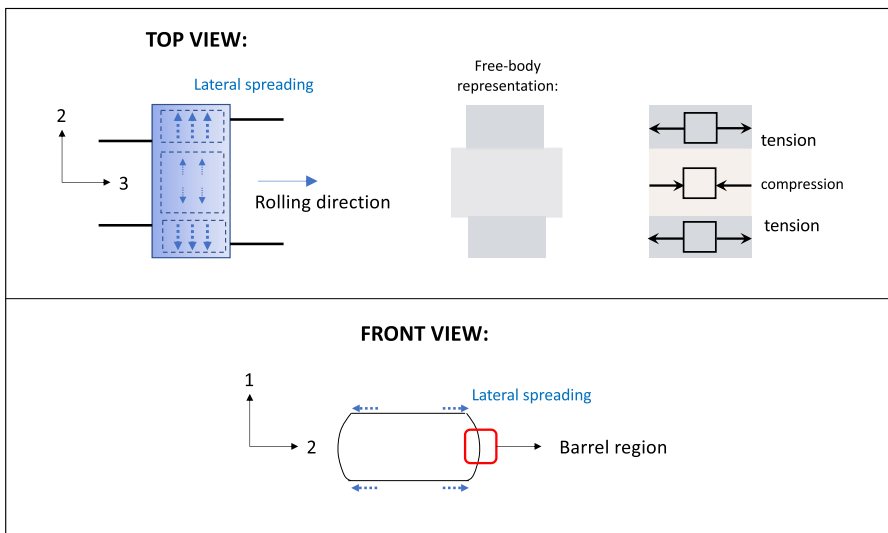


Fig. 12.10 Flat rolling with spreading and friction: origin of secondary longitudinal tensile stresses by inhomogeneous longitudinal flow and barreling.

metal ductility, in particular if the expected recrystallization does not occur (DRX and/or SRX in the interpass time). This contribution gains importance in the finishing operations, during which the workpiece temperature is comparatively low.

The following strategies can be implemented to prevent the formation of edge cracks in hot flat rolling:

- Improvement of lubrication: this reduces friction and the inhomogeneity of plastic deformation;
- Use of edge (vertical) rolls: they will reduce the lateral spread and barreling (and improve the quality of the edge surface);
- Use of edge heaters: this will avoid the temperature drop and possibly increase the temperature at the edge regions, inducing recrystallization and an increase in ductility.

Edge cracking in hot rolling is rather uncommon in plain carbon steels because of their high hot ductility. However, it is possible, and often rather common, in steels that display poor hot ductility, i.e., a low PDT due to the extensive hot shortness phenomena, or hot ductility trough in a specific temperature interval. With reference to the results exposed in Section 8.5, it can be stated that steels with an excessive amount of residual Cu and/or with low Mn/S ratio (particularly the free cutting steels), are rather prone to edge cracking.

Another critical case is constituted by duplex stainless steels. As seen in Section 11.5, when considering hot forgeability, these steels present low ductility values in the temperature range of 700–950°C (Bao, 2015). The formation of edge cracks is then rather common in this type of steel, during the roughing or finishing mill operations, as shown in Fig. 12.11 (Martin, 2011). The edge cracks shown in Fig. 12.11A and B are oblique, i.e., forming an angle of ~ 45 degrees with the plate plane, similarly to the observation with hot forging.

Another type of steel that may suffer from edge cracking in hot rolling are the Si-containing steels, such as the *electrical steels* (with typically 1.5% Si even if higher values are possible, and low C content: below 0.01%) and the *spring steels*, containing up to 1.8% Si. During reheating (that may be carried out at high temperature, above 1350°C, when the C content is low), the presence of an oxidizing atmosphere promotes formation of a scale containing Fe and Si oxides. A separate phase, fayalite (Fe_2SiO_4), may form (Ginzburg and Ballas, 2000). It melts at 1171°C and this may help in increasing the internal grain-boundary oxidation of the steels. As a consequence, the metal hot ductility in the surface regions decreases, promoting the formation of edge cracks. An example is shown in Fig. 12.12, which shows numerous cracks close to the barreled region of an electrical Si-containing steel (Han et al., 1999).

In hot rolling, surface cracks perpendicular to the rolling direction can also form on the surface of the rolled product, far from the edges. An example is shown in Fig. 12.13. In this case, friction is responsible for cracking at the workpiece-roll interface (Table 10.6). Friction increases the plastic deformation near to the contact region due to the rotation of the roll (see also Fig. 12.7) and induces formation of surface tensile stresses. As a result, the local effective strain, ε_{eff} , increases and the local critical effective fracture strain, $\varepsilon_{eff,f}$, decreases. If $\varepsilon_{eff} = \varepsilon_{eff,f}$, surface fracture occurs.

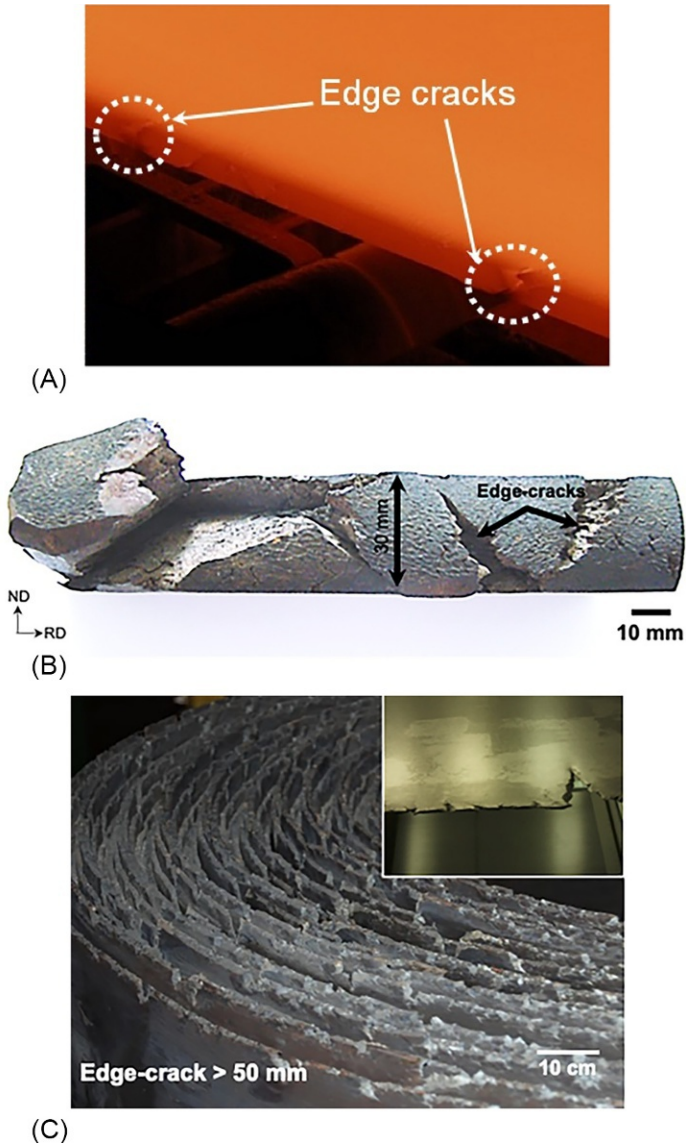


Fig. 12.11 (A, B) Edge cracks in a duplex stainless steel after roughing-mill operations; (C) severe edge cracking in a duplex stainless steel at the end of the hot rolling process. Martin, G., 2011. Hot Workability of Duplex Stainless Steel. Autre. Université de Grenoble. Français. NNT: 2011GRENI078. tel00721764, reproduced with permission of the Author.

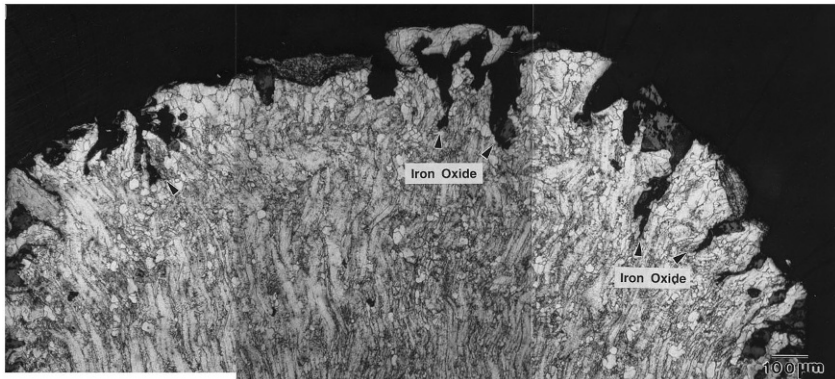


Fig. 12.12 Edge cracks in the barreled region of a hot rolled electrical steel.

From Han, M.H., Lee, S., Kim, N.J., Lee, K.J., Chung, T., Byun, G., 1999. Analysis and prevention of edge cracking phenomenon during hot rolling of non-oriented electrical steels sheets. *Mater. Sci. Eng. A* 264, 47–59, reproduced with permission of the publisher.



Fig. 12.13 Surface transversal cracks in a strip after hot rolling. (Magnification is not known.)

From Yu, H.L., Tieu, K., Lu, C., Deng, G.Y., Liu, X.H., 2013. Occurrence of surface defects on strips during hot rolling process by FEM. *Int. J. Adv. Manuf. Technol.* 67, 1161–1170, reproduced with permission of the publisher.

Surface cracks, and especially edge cracks, may also form during hot rolling of shapes. Fig. 12.4 represents a finishing stage to produce an H-beam. The rolling sequence must be optimized to achieve the desired shape and avoid inhomogeneous plastic deformation during each pass. In particular, the web and wing areas must be properly designed in order to have the same metal flow in the rolling direction. If the

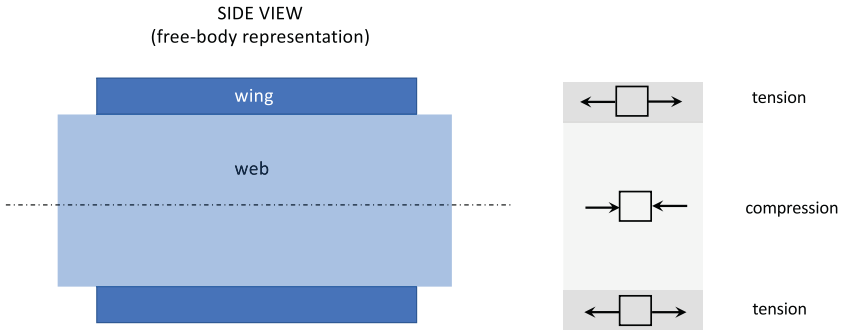


Fig. 12.14 Schematization of the formation of secondary tensile stresses in the wings of an H-beam, during finishing rolling (refer to Fig. 12.4).

flow in the web is higher than that in the wings, tensile stresses are induced in the wings, which can lead to ductile edge cracks (whereas the compressive stresses in the web may induce buckling phenomena due to instability under compression). Fig. 12.14 shows a schematization of the phenomenon.

Cracking phenomena may also occur in the initial roughing stages, during passage through the grooved rolls. As seen in Fig. 12.15A (where two passes are schematized), the roll shaping induces local surface tensile strains and stresses that increase the tri-axiality factor. Since the surface regions also experience a temperature drop (due to heat exchange with the surrounding atmosphere and the rolling roll), the metal may locally undergo a ductility decrease. All these phenomena may then lead to the formation of transverse ductile cracks in the wings. An example is shown in Fig. 12.15B.

Edge cracks in cold rolled parts

In cold rolling the initial width-to-thickness ratio is rather high, and intense lateral spreading can be neglected. However, the effect of barreling is still present. In fact, at the edges, no plane strain condition is present, and the metal is free to deform laterally. Since the metal deformation is hindered by the friction in the regions in contact with the rolls, some barreling may occur (Hosford and Caddell, 1993). In addition, other mechanisms may play an important role in inducing an inhomogeneous plastic deformation with the corresponding formation of local tensile stresses. It is, for example, the case of the phenomenon defined as *over-crowning* (or *over-cambering*) (Dodd and Boddington, 1980). Due to the high separating loads, rolls undergo elastic bending (Fig. 12.16A), leading to the introduction of a geometrical defect (uneven thickness) in the rolled products. To avoid this, rolls are often ground into a crown or camber shape, so that a parallel gap is produced during rolling (Fig. 12.16B). However, if *over-crowned* rolls are used, a higher reduction in the center may be induced (Fig. 12.16C). Hence, a lower longitudinal plastic deformation is observed at the edge regions, resulting in the secondary tensile stresses (Fig. 12.16D). Such stresses can cause the formation of edge cracks. The occurrence of edge cracks is favored by the accumulation of plastic deformation in the subsequent passes of cold rolling,

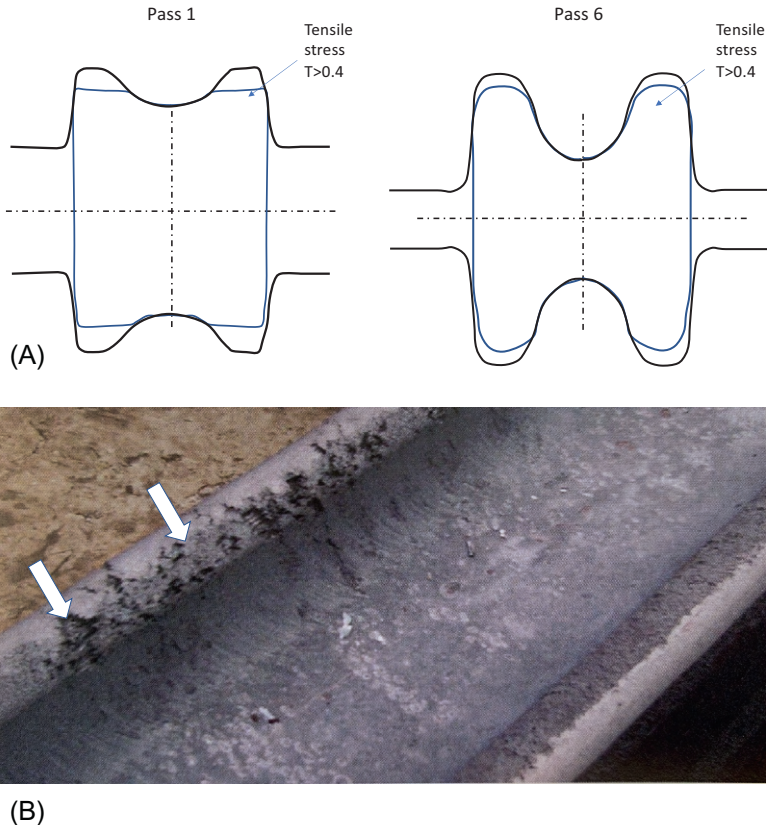


Fig. 12.15 (A) Formation of secondary tensile stresses in the wings of an H-beam during initial rough rolling. (B) Examples of wind cracks (see arrows) in rough rolled products (Pozzobon, 2011).

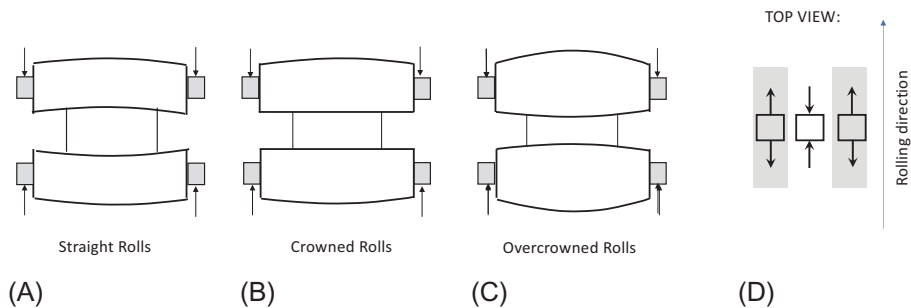
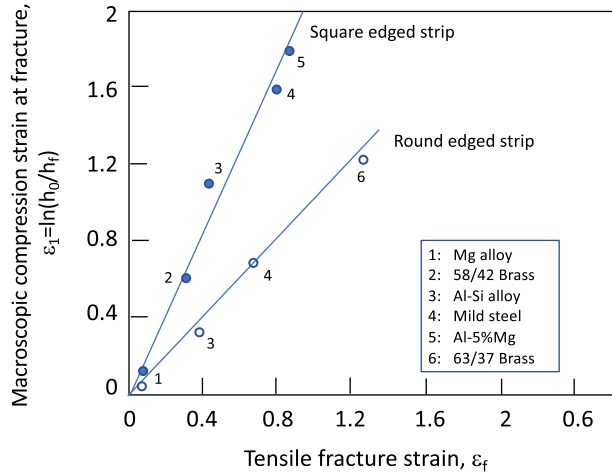


Fig. 12.16 Schematization showing the formation of secondary longitudinal tensile stresses when using over-crowned rolls.

Fig. 12.17 Experimental relationship between the compression strain at fracture (cracking reduction) and material ductility expressed by the true fracture strain (data from Latham and Cockcroft, 1966, reported in Dodd and Bai, 1987).



along with the corresponding sharp decrease in metal ductility. Of course, in case of *under-crowning* secondary tensile stresses are induced in the center that can induce formation of transverse cracks. Today, different methods are often used to produce flat products with even thickness without roll crowning to avoid the problems of over- or under-crowning.

The formation of edge cracks in cold rolling is rather common in steels, in Al, Mg, and Cu alloys, and in particular in metals with a low ductility (Dodd and Boddington, 1980). This is shown in Fig. 12.17, indicating an almost linear relationship between the macroscopic compression strain at edge cracking and the tensile ductility of the materials, expressed by the true fracture strain (Latham and Cockcroft, 1966). The experimental data in the figure were obtained using square-edged strips and round-edged strips. In the second case the cracking reductions were comparatively lower, since the edge geometry increased the barreling effect (the situation is similar to the use of tapered or flanged specimens in the upsetting tests, as discussed in Section 11.5).

Fig. 12.18A shows typical edge cracks in cold rolled 1050 Al alloy. In some cases, the cracks form a zigzag pattern, i.e., they form an angle of ~ 45 degrees with the sheet plane (Riedel et al., 2007). As said, edge trimming is then required to remove the cracks. It is reported that in the cold rolling of Al alloys, about 50% of the metal is lost during the process, and edge cracking is responsible for a large portion of this loss (Ghosh et al., 2004). Another important case is that of duplex stainless steels. If submitted to an insufficient annealing after hot working, they do not recover the required ductility for the subsequent cold rolling process, and then display formation of edge cracks and/or alligatoring fracture (see next subsection). An example is shown in Fig. 12.18B that shows the edge cracks formed in a cold rolled sheet made with a 2205 duplex stainless steel previously annealed (after hot rolling) at 950°C (Fargas et al., 2009). This temperature is below the minimum recommended temperature of 1000°C , required to dissolve in the matrix any precipitated intermetallic phase. As explained in Section 9.4, by means of specific Charpy V impact test it is possible

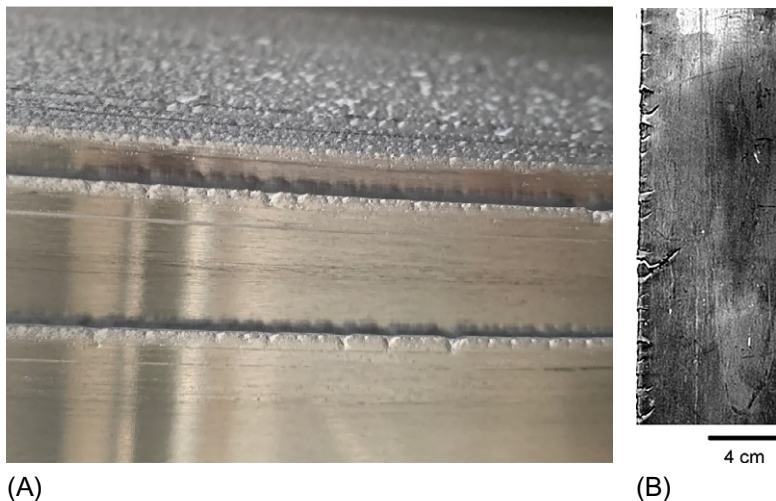


Fig. 12.18 Edge cracks in cold rolled (A) 1050 Al sheet (thickness 1 mm), and (B) duplex stainless steel.

Part B is from Fargas, G., Anglada, M., Mateo, A., 2009. Effect of the annealing temperature on the mechanical properties, formability and corrosion resistance of hot-rolled duplex stainless steel. *J. Mater. Process. Technol.* 209, 1770–1782.

to reveal the effectiveness of a certain annealing process, i.e., if it is able or not to recover the required ductility.

Internal cracks in flat rolled products

Fig. 10.11 shows that when $\Delta > 1$, a positive tensile stress forms along the center axis of the workpiece due to the inhomogeneous plastic deformation. Consequently, conditions for local ductile fracture may be attained with the formation of internal cracks called *central bursts*, schematized in **Fig. 12.19A**. A detailed analysis has shown that the critical conditions for their formation are attained when $\Delta > 0.75$ – 0.79 (Turczyn, 1996). Of course, a positive stress triaxiality factor, T , is only an indication of the

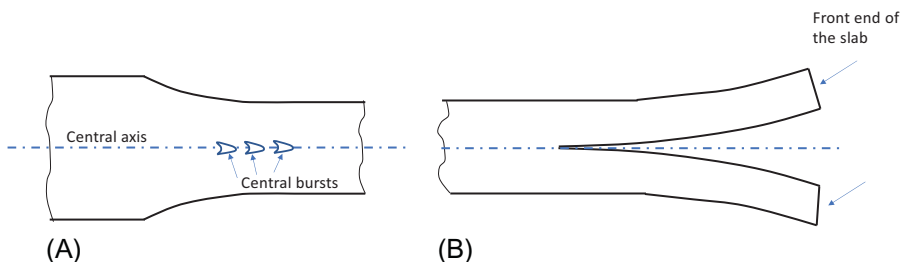


Fig. 12.19 Schematizations showing (A) central bursts and (B) a split end (alligatoring).

possibility of internal cracking. In general, flat rolling (both in hot or cold conditions) develops through a series of subsequent passes that may induce an accumulation of strain and damage along the centerline. Therefore, after a certain number of passes, voids can first nucleate and then grow to coalesce into a macroscopic internal crack, depending on the ductility of the metal. This kind of crack may also form in the case of other metalworking processes, such as extrusion and wire drawing, and will be reconsidered in more detail in the next chapters.

Another kind of rolling defect is the so-called *split end* or *alligatoring* phenomenon. It is characterized by the splitting of a rolled flat product into two halves along the center plane, parallel to the rolling plane. It is schematized in Fig. 12.19B and shown in Fig. 12.20 in the case of the hot rolling of a rod made with a resulfurized steel.

Experimental observations have shown that alligatoring also originates when Δ values are greater than 1, i.e., when plastic deformation is inhomogeneous with the formation of a positive triaxiality along the central axis. In addition, this phenomenon has been mainly observed in materials with a low ductility. Therefore alligatoring also occurs due to the development of ductile damage in a material under rolling and it is often triggered by the formation of central bursts.

Fig. 12.21 shows a possible sequence of events leading to alligatoring (based on Xu and Dahen, 1994). Since $\Delta > 1$, plastic deformation is largely localized close to the rolls, and it develops in the longitudinal (rolling) direction (A). At the front end of the slab (where plastic deformation is free), it induces the formation of a concave shape (B). It is reported that a slight bending of the workpiece occurs when it enters the roll gap; this induces the formation of a stress and strain concentration at the concave shape (see arrow in the figure), which in turn induces the formation of a ductile crack (C). Due to the repeated contacts during rolling (in the subsequent passes), the crack grows in length (D). This phenomenon particularly occurs if the metal is damaged along the central axis (due to the previous formation of central bursts or the presence of preexisting central porosity or inclusions due to the solidification process). The final splitting is caused by the action of an opening normal stress (E), originating from friction with the rolls.

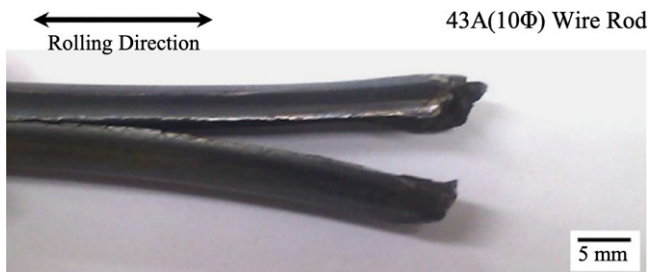


Fig. 12.20 Alligatoring phenomenon in a hot rolled rod made with a resulfurized steel. From Kim, H., Kang, M., Shin, S.Y., Lee, S., 2013. Alligatoring phenomenon occurring during hot rolling of free-machining steel wire rods. Mater. Sci. Eng. A 568, 8–19, reproduced with permission of the publisher.

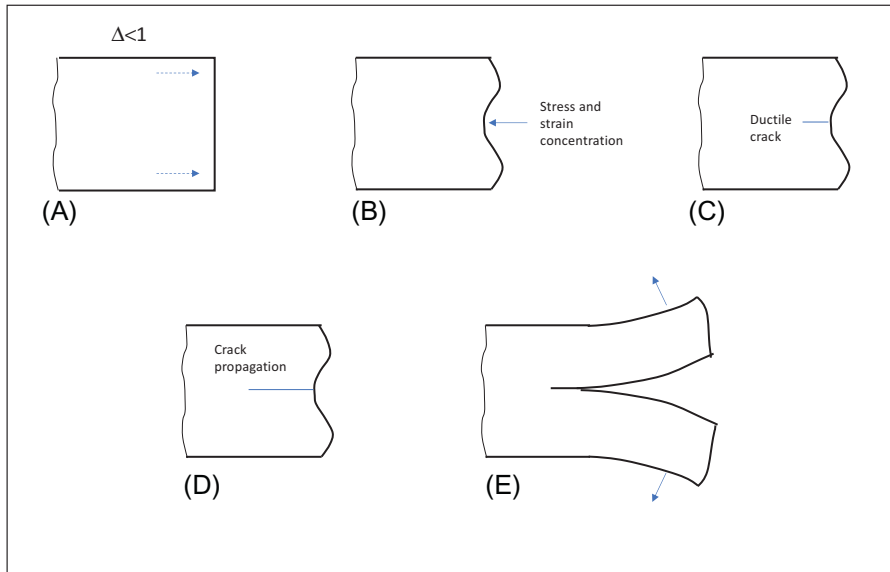


Fig. 12.21 Possible sequence of the events leading to alligatoring.

Based on Xu, L., Dahen, G.S., 1994. Alligatoring and damage in the cold rolling of spheroidized steels. *Metall. Mater. Trans. A* 25, 589–598.

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Edge cracks in an Al 6060 extruded profile (Bernard, 2018).

Chapter outline

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Extrusion process involves forcing a metal billet to pass through a die, obtaining long and straight products of constant cross-section. It is used to produce bars, tubes, wires, solid and hollow sections, and strips. The process can be carried out in hot or cold conditions. In general, cold extrusion is used to produce small parts with simple shape, whereas complex shapes are obtained by hot extrusion.

Hot extrusion is a very important production process for Al alloys, and the alloys of the 6000 class are widely used for the production of extrudates.

This chapter presents the main extrusion processes. A simple method for evaluating the extrusion load and the temperature rise during plastic deformation is presented. The calculation of the temperature rise is useful for estimating the process temperature and the extrudate temperature at the die exit. In addition, the inhomogeneous plastic deformation is discussed, highlighting the possible formation of secondary (unexpected) tensile stresses in the extrusion die, both at the extrudate surface (due to friction at the metal/die interface) and along the extrudate axis (due to high Δ values).

The first kind of stress forms especially in hot extrusion and may be responsible for the formation of surface cracks. The second kind of tensile stress develops mainly in cold extrusion and may cause the formation of central (burst) cracks.

13.1 Extrusion processes

In Fig. 13.1A the common *direct extrusion* process is schematized (Misiolek and Kelly, 2005). The cylindrical billet (preheated in a hot process) is loaded into a heavy walled container (also cylindrical) and then pushed by a ram by means of a hydraulic press. The metal initially undergoes upsetting to assume the same diameter of the container, and then flows through the die in the same direction as the ram travels. Friction takes place at the billet-container wall interface and in the die, between the metal and the bearing channels. Friction increases the extrusion load. Therefore in cold extrusion (always) and in hot extrusion (in the case of extruding high-strength metals, like steels and Ti alloys), various lubricants are used, such as graphitic grease, graphite with MoS₂ powder, phosphate-conversion coatings together with soap or wax coatings (for cold extrusion), and glass particles (for hot extrusion). A dummy block with a diameter that is lower than the internal diameter of the container (about 1–2 mm) is used when no lubricant is adopted, to avoid the cortical layer of the billet (rich in oxides and other impurities) being extruded. To avoid the problems associated with friction, the process of *indirect extrusion* (Fig. 13.1B) was developed. In indirect extrusion, the die is pushed inside the container, forcing the metal to flow backwards. Other methods were developed to avoid the friction problems, such as *isostatic extrusion* or *active friction extrusion* (Li et al., 2008).

In the hot extrusion of Al alloys, cylindrical billets produced by direct chill vertical casting are typically used. They are preheated to 420–520°C in an induction or

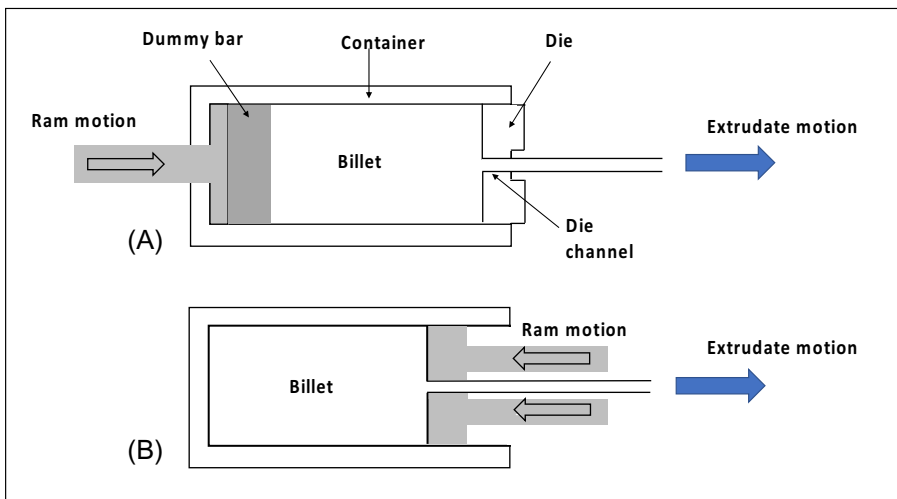


Fig. 13.1 Schematization of direct extrusion (A) and indirect extrusion (B). In both cases a square (or flat) die is used, as is typical in the hot extrusion of Al alloys.

Redrawn and modified from Misiolek, W.Z., Kelly, R.M., 2005. Extrusion of aluminum alloys. In: ASM Metals Handbook; vol. 14A, Metalworking: Bulk Forming, pp. 532–527.

gas furnace and extruded without a lubricant (only a small amount of graphite is used), since the trapped lubricant would favor the formation of surface defects, such as *blisters* (Laue and Stenger, 1981). In addition, some metal transfer onto the container wall surfaces takes place in unlubricated conditions, preventing impurity pick-up from the container walls (Flitta and Sheppard, 2003). The flow stress at high temperature of Al alloys is comparatively low and friction does not increase the extrusion force to an unacceptable level. In heat-treatable Al alloys, such as 6000 class alloys, the extrudate is subsequently forced cooled in air or water to RT to obtain a supersaturated solid solution. The extrudates are then stretched (2%–3%), to eliminate any geometric distortion, and finally submitted to an aging heat treatment to achieve the requested mechanical properties (by precipitation hardening).

An important parameter in extrusion is the *extrusion ratio*, R , given by the ratio of the area of the billet cross section before extrusion, A_0 , and the area of the cross section of the final product, A_f ($R = A_0/A_f$). For Al alloys, R may be up to 400, whereas for steels the maximum R values are approximately 40. The plastic deformation in the extrusion direction is therefore:

$$\varepsilon = \ln \frac{L_f}{L_0} = \ln \frac{A_0}{A_f} = \ln R \quad (13.1)$$

where L_0 is the billet length before extrusion and L_f is the length of the extrudate. Extrusion plastic strains, ε , are rather high, up to 6 in Al alloys and up to 3.5 in steels. The two most important process parameters are the *ram* (or *extrusion*) *velocity*, v_0 , and the billet *pre-heating temperature*, T_0 . They can be controlled by the process operator. The ram velocity determines the productivity of the extrusion equipment. It is commonly intended to adopt the highest possible extrusion velocity to increase the productivity. It is typically in the range 0.1–1 m/s and the corresponding strain rates are in the range of 10^{-1} – 10^2 s $^{-1}$.

To obtain simple extruded profiles, such as bars, flat dies are used. Instead, to produce hollow shapes, it is necessary to use *porthole dies*. They are composed of two pieces: the *mandrel* and the *die cap*. The *mandrel* forms the internal shape of the hollow section. It is located inside the die cap and has two or more ports to control the flow of the metal. During the extrusion, the billet separates into each port and then joins again in the welding chamber prior to reaching the bearing area. The ports are divided by webs, which have the function of supporting the *mandrel*. The die cap forms the external shape of the hollow section. It is evident that with this type of die it is not possible to use a lubricant, which would prevent the re-welding of the metal.

With cold extrusion, comparatively small parts are produced. Conical dies are typically used together with proper lubrication to reduce the extrusion force required. In some cases, the end product is obtained with a combination of processes, including extrusion and forging, and the strength of the product is quite high because of strain hardening.

13.2 Mechanics of extrusion

Plastic flow

Fig. 13.2 schematically shows two common plastic flow patterns in extrusion. In Fig. 13.2A a flat (or square) die is used (see also Fig. 13.1), whereas in Fig. 13.2B, a *conical die* is used. The die half-opening angle is indicated with α ($\alpha = 90^\circ$ for a flat die). The plastic deformation of the metal in the container is limited, as shown by the rather undeformed grids. In fact, the metal is not free to deform radially, being impeded in this by the container walls. The metal is here submitted to a high compressive hydrostatic stress.

Most of the plastic deformation occurs in the die, as shown by the large distortion of the grids. In the flat die the flowing metal accumulates at the die corners, forming a *dead zone*. The extrusion angle, α_s , is then formed spontaneously. With good lubrication (e.g., when using glass lubrication in a hot process), $\alpha_s \approx 70^\circ$, while in the absence of lubrication (in hot extrusion of non-ferrous metals flat dies are typically used without lubricant), α_s can be estimated with the following experimental relationship: $\alpha_s = 54 + 3.45 \cdot \ln R$ (Kalpakjian and Schmid, 2009). As already mentioned, a conical die is typically used in cold extrusion with proper lubrication; in this case $\alpha_s = \alpha$.

The deformation of the grids shows that plastic flow in the die region is very inhomogeneous, presenting a combination of compression in the radial direction, which induces the elongation of the metal, and shear plastic deformation, which induces a distortion of the metal and requires an additional (*redundant*) deformation work.

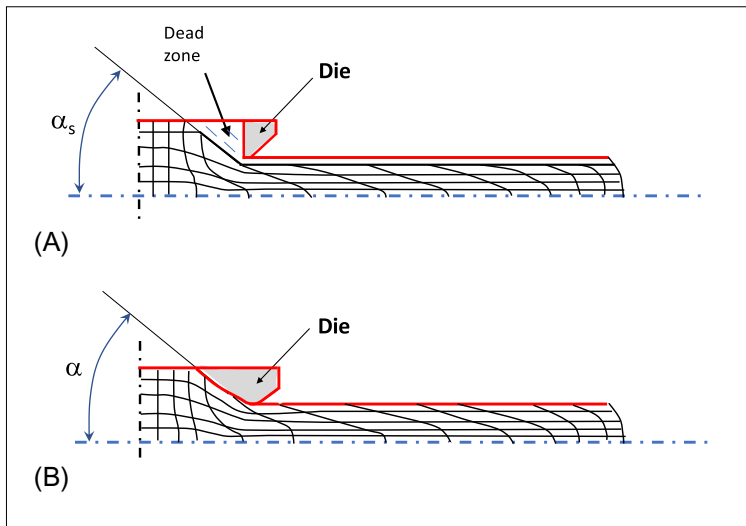


Fig. 13.2 Typical plastic flow pattern in a flat (A) and a conical (B) die.

As seen, the inhomogeneity of plastic deformation is represented by the parameter Δ (Eq. 10.4), which is given by Richardson et al. (1985):

$$\Delta = \frac{\bar{D}}{d} \cong \frac{4 \cdot \alpha}{\varepsilon} \quad (13.2)$$

where:

- $\bar{D}$ is the average of D_0 and D_f ;
- D_0 is the diameter of the billet before extrusion;
- D_f is the diameter of the extrudate;
- d is the traveled distance of a metal slice in the extrusion die (shown in Fig. 13.4).

The inhomogeneity of plastic deformation increases as Δ increases (for $\Delta > 1$).

Extrusion load

The extrusion load, F_{ext} , is given by: $p_{\text{ext}} A_0$, where p_{ext} is the extrusion pressure, i.e., the pressure that the ram must exert for the extrusion of a billet. The extrusion power is: $F_{\text{ext}} v_0$.

To calculate p_{ext} , we initially neglect the contribution of friction due to sliding at the container wall and in the die bearing channels. We also start to consider obtaining a cylindrical bar from a cylindrical billet, through a simple uniaxial deformation process.

Indicating with $\bar{\sigma}$ the mean flow stress of the metal (obtained from the corresponding constitutive equation, such as Eq. (3.18) or (3.19) for a cold process, and Eq. (4.12) or (4.13) for a hot process), the work for plastic deformation is given by: $\bar{\sigma} \cdot \varepsilon \cdot L_0 \cdot A_0$ (where $L_0 \cdot A_0$ is the volume of the metal to be deformed). The external work exerted by the press is: $p_0 \cdot L_0 \cdot A_0$ (where $p_0 \cdot A_0$ is the extrusion force in this case). By equating these two relations we obtain:

$$p_0 = \bar{\sigma} \cdot \varepsilon = \bar{\sigma} \cdot \ln R \quad (13.3)$$

i.e., the pressure for plastic deformation is equal to the specific plastic work. However, plastic deformation is rather inhomogeneous during extrusion and the following relationship should be used:

$$p_0 = \bar{\sigma}_{\text{eff}} \cdot \varepsilon_{\text{eff}} \quad (13.4)$$

which is rather difficult to manage. A simple and widely used equation is the following one:

$$p_0 = \bar{\sigma} \cdot (a + b \cdot \ln R) \quad (13.5)$$

where the constants a and b account for the inhomogeneity of the plastic deformation (and the associated change in the stress triaxiality), as well as for friction in the die bearing channels (die lands). Typical data for a and b are listed in Table 13.1 (Richardson et al., 1985). Note that in the absence of lubrication, the friction

Table 13.1 Typical values of the parameters in Eq. (13.5).

	Very good lubrication	No lubrication	Intermediate friction
<i>a</i>	0.88	1.06	0.9
<i>b</i>	1.3	1.55	1.5

Data from Richardson, G.J., Hawkins, D.N., Sellars, C.M., 1985. Worked Examples in Metalworking. The Institute of Metals, London.

coefficient in the die lands typically ranges between 0.3 and 0.4 (Laue and Stenger, 1981). Eq. (13.5) is valid for the production of cylindrical rods. The extrusion ratio should be modified to consider the influence of the different section shape on the extrusion pressure (Sheppard, 1993). One approach is to replace R with R' , defined by (Richardson et al., 1985):

$$R' = R \cdot \sqrt{\frac{p_e}{p_r}} \quad (13.6)$$

where p_e is the sum of the external and internal periphery of the shape and p_r is the periphery of a rod of an equivalent cross section. Of course, a complicated shape will require a higher ram pressure because of the increased deformation strains.

In direct extrusion, the role of friction at the metal-container interface must also be considered. The extrusion pressure, p_{ext} , is then given by: $p_0 + p_\mu$, where p_μ is the contribution made by friction in the container. Indicating with μ the average friction coefficient at the metal-container interface, we have (Dieter, 1988):

$$p_{\text{ext}} = p_0 + p_\mu = p_0 + p_0 \cdot \frac{4 \cdot \mu \cdot L}{D_0} \quad (13.7)$$

where L is the remaining length of the billet in the container, and D_0 is the internal diameter of the container. In Fig. 13.3, the evolution of the extrusion force with the ram travel is schematically illustrated for both direct and indirect extrusion. Three

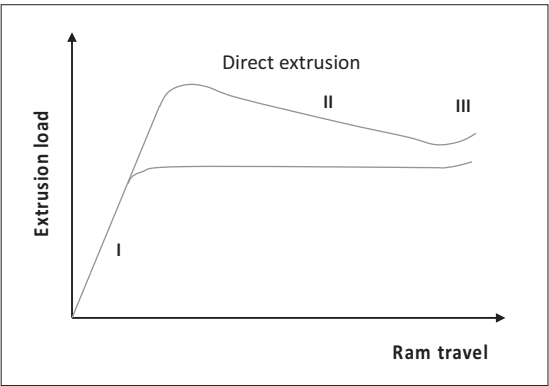


Fig. 13.3 Extrusion load (pressure) vs ram travel for direct and indirect extrusion.

stages are indicated. During stage I, the force increases (in both cases) for the billet upsetting to fill the container. In the case of direct extrusion, the peak force at the end of Stage I is rather high because of the contribution of the container's friction to p_{ext} . This contribution decreases with the ram travel and the extrusion force correspondingly decreases. In indirect extrusion, this contribution is absent. At the end of the stroke, a pressure is exerted on the metal remaining in the container (including the metal in the dead region, where present) and the force slightly increases (Stage III).

In general, the friction coefficient, μ , is estimated indirectly, i.e., with the experimental determination of the extrusion pressure and the use of Eq. (13.7). In the case of the hot extrusion of Al alloys (without lubricant), a material transfer onto the container wall occurs, and the friction coefficient is typically lower than 0.1 (Flitta and Sheppard, 2003), because of the fully plastic contact and the rather high contact pressure in the container (Straffelini, 2015).

For the estimation of the mean flow stress in hot extrusion, both the average strain rate, $\dot{\epsilon}$, and the metal process temperature, T^* , must be estimated. The average strain rate is given by:

$$\dot{\epsilon} = \frac{\epsilon}{t} = \epsilon \cdot \frac{v_0}{d} \quad (13.8)$$

where:

- t is the time required for a metal slice to travel through the region of the die where plastic deformation occurs;
- d is the corresponding traveled distance (Fig. 13.4);
- v_0 is the extrusion velocity.

Considering, for the sake of simplicity, a two-dimensional die, we have:

$$d = \frac{D_0 - D_f}{2 \tan \alpha}$$

and therefore:

$$\dot{\epsilon} = \frac{2 \cdot v_0 \cdot \tan \alpha}{D_0 - D_f} \cdot \ln R$$

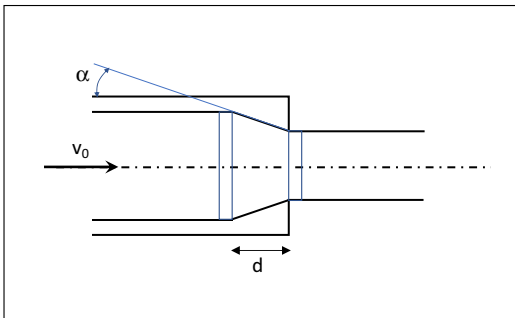


Fig. 13.4 Schema of the traveled distance, d , in the die during direct extrusion.

The approach can be extended to a three-dimensional conical die, obtaining the following relationship (Dieter, 1988):

$$\dot{\epsilon} = \frac{6 \cdot v_0 \cdot D_0^2 \cdot \tan \alpha}{D_0^3 - D_f^3} \cdot \ln R \quad (13.9)$$

It is clear that Eq. (13.9) broadly estimates the average strain rate during extrusion since the local strains change from point to point in the extrusion die, and the corresponding strain rates change as well. However, experimental data have shown that it can be satisfactorily used to a first approximation, even in the case of complex shapes, by replacing D_f with the equivalent diameter of the extrudate, i.e., the diameter of the rod having the same cross section as the shape (Richardson et al., 1985).

Temperature rise in extrusion

To obtain the mean flow stress in hot extrusion, the process temperature T^* , i.e., the average metal temperature in the die during plastic deformation, must be estimated. Also, estimating the product *exit temperature* (i.e., the temperature at which the extrudate emerges from the die) is very important for controlling the process, as shown in Section 13.3. The estimation of these temperatures is a rather difficult task, and a very approximate approach is presented here.

The temperature rise during plastic deformation in the die can be evaluated using Eqs. (3.12) and (13.5), and assuming an adiabatic process:

$$\Delta \bar{T}_{\text{def}} = \beta \cdot \frac{p_0}{\rho \cdot c} \quad (13.10)$$

where β is generally set to 0.9. Parameters ρ and c are, respectively, the density and heat capacity of the metal under deformation. To a first approximation, the average temperature rise in the metal under plastic deformation is: $\frac{1}{2} \Delta \bar{T}_{\text{def}}$, assuming a linear increase of the temperature in the die. Neglecting any thermal loss, the process temperature, T^* , is then given by:

$$T^* = T_0 + \Delta \bar{T}_\mu + \frac{1}{2} \cdot \Delta \bar{T}_{\text{def}} \quad (13.11)$$

where T_0 is the billet preheating temperature and $\Delta \bar{T}_\mu$ is the temperature rise in a metal slice due to friction in the container. To a first approximation, it can be evaluated with the following equation:

$$\Delta \bar{T}_\mu = \frac{1}{2} \cdot \frac{p_\mu}{\rho \cdot c} = \frac{1}{2} \cdot \frac{p_{\text{ext}} - p_0}{\rho \cdot c} \quad (13.12)$$

where $\frac{1}{2}$ takes into account that the frictional heat equally flows into the metal and into the die walls.

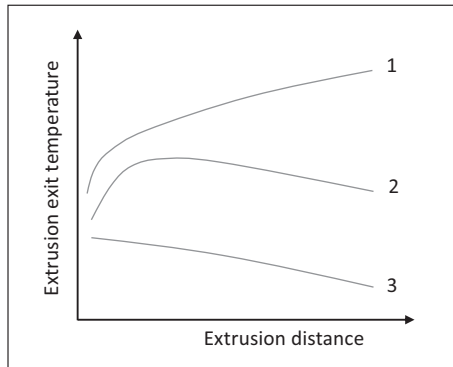


Fig. 13.5 Schematization of the exit temperature profiles of products obtained by hot extrusion. Curve 1 represents the case when the container temperature is similar to the billet temperature. Curves 2 and 3 are typical when the container is colder than the billet.

Finally, the extrudate's exit temperature (T_{exit}) can be estimated using the following equation:

$$T_{\text{exit}} = T_0 + \Delta \bar{T}_{\mu} + \Delta \bar{T}_{\text{def}} \quad (13.13)$$

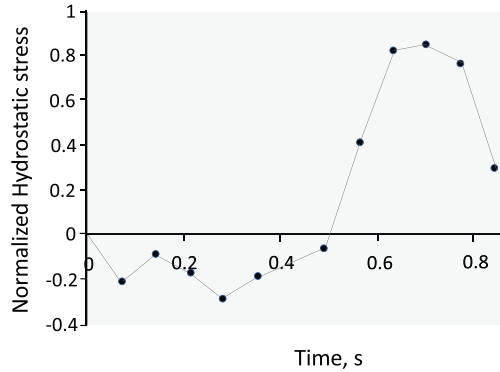
In hot extrusion, the container is preheated to reduce thermal loss. In the extrusion of Al and Mg alloys, the container temperature is on the order of the alloy temperature and any thermal loss can be approximately neglected. Curve 1 of Fig. 13.5 shows a schematization of the *exit temperature profile* for the hot extrusion of an Al or a Mg alloy. It is maximum at the end of the process. If the container is colder than the billet, as in the case of hot extrusion of steel, heat loss by conduction occurs, and the exit temperature profile may assume the shape of curves 2 or 3 in the figure. Its evaluation is more difficult and it may require numerical methods.

Stress triaxiality in extrusion

As already highlighted, the plastic flow in the die region is very inhomogeneous, with a combination of compression and shear plastic deformation. In case of high values of the die semi-angle, α , and low reductions, $\Delta > 1$ and *secondary (unexpected) tensile stresses* develop along the central extrusion axis, as explained in Section 10.3. Plastic deformation (by compression) tends to localize in the radial regions close to the die walls. In these regions, the metal tends to flow in the longitudinal (extrusion) direction markedly, and the metal in the center of the extrudate is then forced to deform plastically under tensile deformation (and, at the same time, large shear stresses develop at the compressive-tensile interface). Consequently, the hydrostatic stress (and then the triaxiality factor, T) becomes tensile upon exit from the die (Fig. 10.11).

Fig. 13.6 Hydrostatic stress (normalized with the metal yield strength) along the central axis in the die versus time, during cold extrusion of an Al alloy.

Redrawn from McVeigh, C., Liu, W.K., 2006. Prediction of central bursting during axisymmetric cold extrusion of a metal alloy containing particles. *Int. J. Solids Struct.* 43, 3087–3105.



As an example, Fig. 13.6 shows the evolution of the hydrostatic stress (normalized with the initial yield strength; this parameter is $\sim T$) along the central axis, for an Al alloy passing through the die during cold extrusion (McVeigh and Liu, 2006). The diagram was obtained using a numerical analysis, with reference to a conical die with a high extrusion angle, 48° , and a small extrusion ratio, $R = 1.11$ (and then: $\Delta = 16$). It is seen that initially, i.e., when the metal is in the container, the hydrostatic stress is negative, as mentioned before. However, hydrostatic stress becomes very positive as the metal passes through the die. In general, a positive tensile stress along the centerline occurs when Δ is approximately greater than 1–2.

The development of tensile stresses along the central axis of the extrudate is common when $R < 4$ and then typically in the case of cold drawing processes. In hot extrusion, the R values are much higher than in cold extrusion and, correspondingly, Δ values are rather low. As a consequence, tensile stresses along the extrusion axis do not develop. However, in these cases the effect of the comparatively large friction with the bearing channels plays a marked role, particularly when the process is carried out without lubrication. As schematized in Fig. 10.12, the friction in the die bearing channels reduces the longitudinal plastic deformation of the metal at the surface in comparison with the metal in the center of the extrudate. Because of material continuity, the metal in the outer regions is forced to deform plastically under tensile deformation. The development of a surface tensile stress in the extrusion direction induces the formation of a positive value of the triaxiality factor, T . All these aspects play an important role in the formation of surface and internal ductile cracks in the extrudates. They are summarized in Table 13.2.

Because of the inhomogeneous plastic deformation in the extrusion die, after unloading, the cold extruded parts may display residual stresses that are tensile on the surface and compressive in the interior of the extrudate. The origin of these residual stresses is explained in the next chapter, when considering the cold wire drawing

Table 13.2 Formation of tensile stresses during hot or cold extrusion.

HOT EXTRUSION	COLD EXTRUSION
Dry or poor lubrication (high friction at the metal/die interface)	Good lubrication (low friction at the metal die interface)
	
High surface tensile stresses	
Comparatively high R ($\Delta < 1$)	Comparatively low R ($\Delta > 1$)
	
	Large inhomogeneous deformation with $T > 0$ at the centerline

process. The intensity of residual stresses increases with an increase in the inhomogeneity of plastic deformation, i.e., with Δ .

13.3 Ductile cracks in extrudates

Extrusion products may contain different defects, such as geometrical and surface defects (Dieter, 1988). Surface defects are often due to the extrudate wear during sliding in the die bearing channels, which is particularly intense in the absence of lubrication (hot processes) and when the die lands are worn. The surface roughness is characterized by comparatively high R_a , often in the range of 2–5 μm . Abrasive scratches (also called *die lines*) may also form on the surface of the extrudates and compromise the quality of the product (Clode and Sheppard, 1990). This may occur when oxide particles accumulate in the bearing area or when the die lands wear is excessive. Comparatively high local temperatures may also intensify the adhesive wear of the extrudate with transfer of particles from the alloy onto the die surfaces, and then back transfer onto the surface of the product (also called *pick-ups*). If the intensity of these phenomena is too high, the surface quality of the products can be unacceptably deteriorated (Menapace et al., 2022).

However, in extrusion products the most important defects are ductile cracks, which may either be:

1. Surface cracks (mainly in hot extrusion products);
2. Internal cracks (mainly in cold extrusion products).

The formation of the cracks is a consequence of inhomogeneous plastic deformation and the secondary tensile stresses it causes, possibly coupled with a reduction in the metal’s ductility. Such cracks are examined in the present section.

Surface cracks in hot extrudates

The concomitance of two factors may form surface cracks in hot extrudates:

1. The development of surface tensile stresses due to friction in the die bearing area;
2. The high surface temperature the extrudate reaches at the exit (especially at the end of the process, in the case of Al or Mg alloys, as seen in Fig. 13.5), which can induce a ductility drop in the metal (hot shortness or, in some cases, a ductility trough; see Chapter 8).

For example, Fig. 13.7 shows two bars of Mg alloy (AZ 31) extruded at 420°C at two velocities (Lapovok et al., 2004). The formation of surface cracks, perpendicular to the extrusion direction, can be clearly appreciated at the highest extrusion velocity. In fact, an excessive increase in the surface temperature was induced at the highest velocity, and the conditions for hot shortness were reached. It is recalled that, in general, the PDT temperature is close to the melting temperature of the alloys.

Therefore the hot extrusion process must be controlled to avoid the formation of surface cracks. In general, it needs to be verified that the exit temperature does not become excessively high and, in particular, it does not exceed a maximum limit determined by the metallurgical characteristics of the material. In Al alloys, the maximum allowable exit temperature is close to 570–580°C, and it is determined by ductility drop phenomena. In particular, it is close to the solidus temperature in low solute alloys, such as 6060, and it is close to the comparatively low eutectic temperature in high solute alloys, such as 6082 (ductility drop by eutectic melting). Therefore, in low solute alloys, the critical temperature is usually rather high and it is unlikely to be reached; die lines and pick-up phenomena can be a serious factor in determining the integrity of the extrudates. On the contrary, surface cracking is the biggest limitation to extrudability due to the low melting point eutectics in high solute alloys. In this respect, tuning the chemical composition of the alloy is rather important. In the case of the widely used 6060 alloy, a decrease in the Mg and Si content (the two principal alloying elements, complying, of course, with

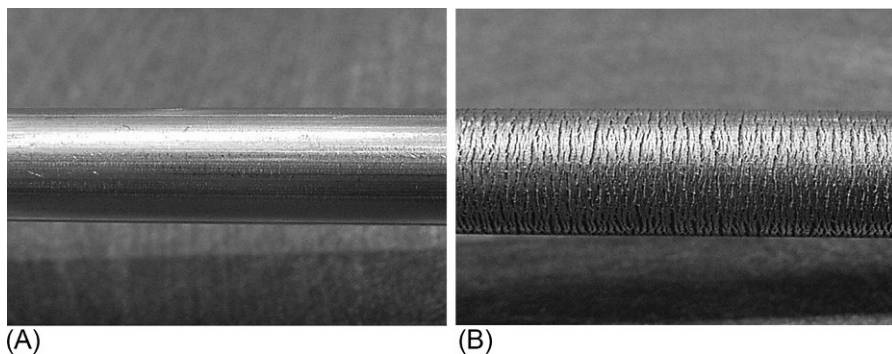


Fig. 13.7 Surface appearance of a Mg alloy bar extruded at 420°C with $R = 36$. (A) Ram velocity of 10 mm/s; (B) Ram velocity of 22 mm/s.

Reproduced from Lapovok, R.Ye., Barnett, M.R., Davies, C.H.J., 2004. Construction of extrusion limit diagram for AZ31 magnesium alloy by FE simulation. *J. Mater. Process. Technol.* 146, 408–414, with permission of the Publisher.

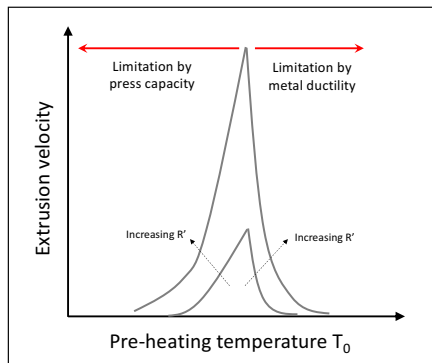


Fig. 13.8 Schematization of extrusion limit diagrams.

the compositional limits) induces a decrease in the flow stress and an increase in the solidus (melting) temperature, with beneficial effects on the extrudability (i.e., with the possibility of increasing the extrusion velocity).

As shown by Eq. (13.13), the exit temperature mainly depends on the billet preheating temperature, T_0 , and the temperature rise due to plastic deformation, which, in turn, depends on the average strain rate (and therefore on the extrusion velocity) and the total applied strain (and then R or R'). In general, to produce a part with a given extrusion ratio R (or R') and using a given alloy, the two operating parameters, T_0 and v_0 , must be properly selected. To achieve this aim, reference to the *extrusion limit diagram* is paramount. This diagram is typically obtained by combining numerical evaluations and experimental data. A schematization is shown in Fig. 13.8. The press capacity limits the maximum extrusion velocity at low billet preheating temperature, when the extrusion force is too high. To increase v_0 is then necessary to increase T_0 . However, a maximum is attained at a certain preheating temperature. With a further increase in T_0 , the maximum possible extrusion velocity is decreased to avoid the formation of surface cracks. In fact, in these conditions the exit temperature reaches the critical value for surface cracking. Of course, it is also possible to obtain limit diagrams in which the maximum allowable R' values are plotted as a function of the preheating temperature for different extrusion velocities.

Suppose the maximum temperature is reached toward the end of the extrusion (as in curve 1 of Fig. 13.5). In that case, it is possible to preheat the billet, conferring a temperature profile in which temperature is maximum at the entrance and then decreases, becoming minimum for the last metal slice being extruded, so as to maintain the exit temperature of the extrudate as constant as possible during the process (and lower than the critical value).

In hot extrusion it is also possible to observe the formation of surface cracks at thin sections or in correspondence of the corners. Fig. 13.9 shows two typical examples. Also in these cases, the cracks form because of the local formation of tensile stresses and possibly a drop in ductility due to an excessive temperature rise. The appearance of tensile stresses is due to an inhomogeneous plastic flow induced by the friction with the die channels. Fig. 11.9A refers to cracks formed in a thin section that is close to

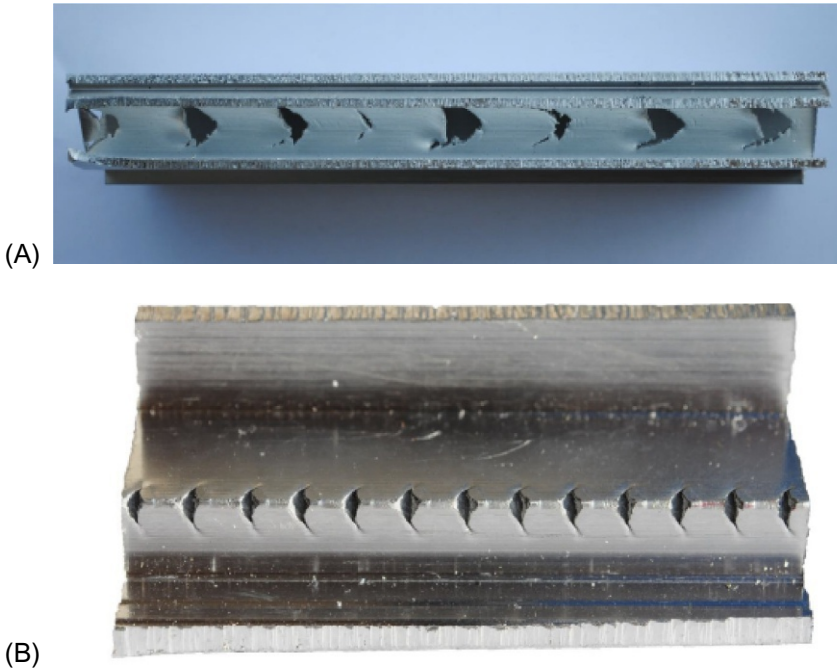


Fig. 13.9 Ductile cracks in thin sections (A) and close to the corners (B) (Bernard, 2018).

wider sections. In the thin section, the plastic flow in the extrusion direction is slowed down by the friction with the two surfaces of the bearing channels. On the contrary, the metal flow in the neighboring wider sections is less limited and it then exerts a tensile stress in the metal in the thin section that may produce ductile fractures. A similar phenomenon may occur in regions close to a corner (Fig. 13.9B), where the effect of friction slows down the metal flow and a tensile stress is induced by the flow of the neighboring metal that does not feel the effect of friction.

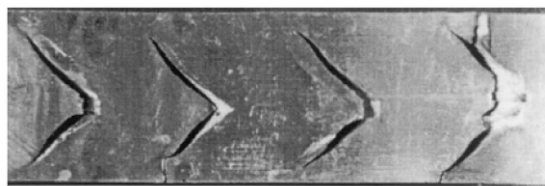
To avoid the formation of such cracks, two strategies may be followed. One is to increase the die bearing channel length for the metal flowing in the wider channel, in order to reduce its local flow velocity. Another one is to reduce the exit temperature (reducing, e.g., the extrusion velocity), in order to reduce the metal ductility drop.

Internal cracks in cold extrudates

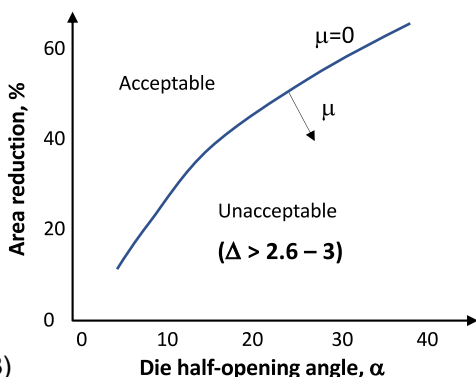
In extrusion parts obtained by cold processes, the friction at the metal-die interface is low because of lubrication, and the surface cracks generally do not form. On the contrary, in the cold extruded parts *internal central burst cracks* (often indicated as *chevron cracks*) may form, if Δ is sufficiently high and, at the same time, the metal ductility is sufficiently low to satisfy the ductile fracture criterion.

Fig. 13.10A shows an example of internal cracks and Fig. 13.10B shows a map obtained combining analytical and numerical analyses, and verified by several experimental datasets. It shows the critical operating region for the formation of internal cracks. The formation of such cracks is a possible consequence of the development of tensile stresses in the central region of the extrudate as the metal enters the die, as shown in Fig. 13.6. The ductile cracks nucleate along the central axis and then propagate along the maximum shear bands, forming a crack with the typical chevron morphology (the chevron cracks point in the extrusion direction). The experimental data show that such cracks generally form when $\Delta > 2.6-3$ (Fig. 13.10B).

Friction at the metal/die interface plays a crucial role. As said, it is generally very low. But, if lubrication is poor and friction increases a bit (up to, say, $\mu = 0.15$), the plastic flow in the extrusion direction close to the die interfaces becomes reduced (Saanouni et al., 2004). Consequently, the intensity of the tensile stresses (and then of T) along the centerline also decreases. This reduces the possibility of formation of internal cracks. But if μ becomes too high, i.e., greater than 0.15/0.2, surface cracks may be formed. This highlights the importance of a careful control of friction, i.e., of the lubrication degree to avoid formation of ductile cracks.



(A)

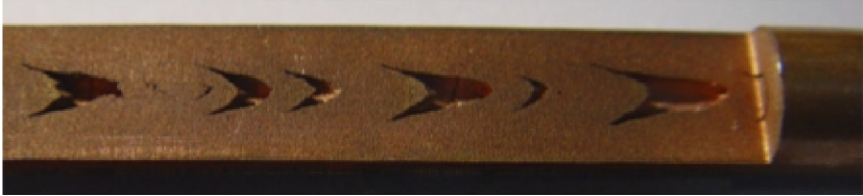


(B)

Fig. 13.10 (A) Typical experimental internal (chevron) cracks (reproduced from Ko, D.C., Kim, B.M., 2000. The prediction of central burst defects in extrusion and wire drawing. *J. Mater. Process. Technol.* 102, 19–24, with permission of the Publisher); (B) map showing the role of the extrusion reduction and semi-die angle on the possible formation of internal cracks (obtained considering experimental data and analytical/numerical models, from: Avitzur, 1968; Zimmermann et al., 1971; and McVeigh and Liu, 2006).

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Internal chevron cracks on the cross section of a drawn Cu wire.

Reproduced from Norasethasopon, S., Yoshida, K., 2008. Prediction of chevron crack initiation in inclusion copper shaped-wire drawing. Eng. Fail. Anal. 15, 378–393, with permission of the Publisher.

Chapter outline

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Wire drawing is a cold plastic deformation process by which the diameter of a rod or a wire is reduced and its length is increased by pulling the workpiece through a calibrated hole of a conical die. The products are characterized by high strength because of strain hardening.

Thin wires with a diameter in the range of 50 μm to 3 mm are typically used for making ropes, cords for tire reinforcement, prestressed concrete structures, springs, cables, and welding electrodes. For this, they may be submitted to bending and side-pressing. Thick wires with a diameter greater than 3 mm are commonly used for subsequent cold forging operations, for example in the production of bolts and fasteners (cold heading, upsetting), or cold extrusion. Steel cold drawn wires are mainly made of high-carbon (eutectoid) pearlitic steel, that permits very high tensile strength values to be obtained.

The mechanics of wire drawing presents many similarities with extrusion, even if similar concepts are often treated differently or with a different notation.

The process must be optimized to avoid formation of surface and internal defects, such as internal voids or chevron cracks. The defects may induce the failure of the workpiece during the drawing process (which greatly increases the manufacturing costs), or may impair the mechanical properties of the products.

Ductility and Formability of Metals. <https://doi.org/10.1016/B978-0-323-99203-9.00004-1>

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14.1 Wire drawing processes

Fig. 14.1 shows a schematization of a wire (or rod) drawing process. Dies are typically made of hard metals, such as WC-Co, because of their high sliding wear resistance. They are inserted into a metallic casing. The die geometry is composed of an inlet zone, indicated with 1 in the figure; a conical zone with a semi-die angle α , indicated with 2, where the plastic deformation takes place; a cylindrical bearing zone, indicated with 3; and the outlet cone, indicated with 4. The incoming wire (possibly pickled or mechanically bent to remove surface oxides due to the previous hot rolling) is tapered to facilitate the initial entry into the die and is then pulled by a *capstan*, moved by an electric motor. Shape drawing processes are also available, but they are not considered in this text.

The wire is lubricated before entering the die. Lubrication reduces friction in the workpiece-die sliding, and therefore reduces:

- The drawing force;
- The die wear;
- The damage to the wire surface;
- The heating of the wire.

Different lubricants are used. The main ones are *soap powders* (e.g., Zn or Ca stea-
rates), used in the so-called *dry drawing process* with wire diameters in excess of 1–1.5 mm, and water-in-oil emulsions with EP additives, used in the so-called *wet drawing process* with smaller wire diameters. Steel wires can be previously phosphated to create a surface layer capable of better anchoring the lubricant, or coated with metal deposits, capable of reducing the coefficient of friction. In the production of thick wires, the phosphate layer is maintained for the subsequent forging operations.

The reduction in a drawing process is indicated by r , which is expressed by the following relation:

$$r = \frac{A_0 - A_f}{A_0} \quad (14.1)$$

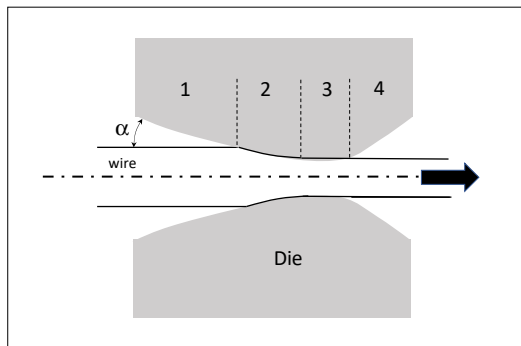


Fig. 14.1 Schematic representation of the wire drawing process.

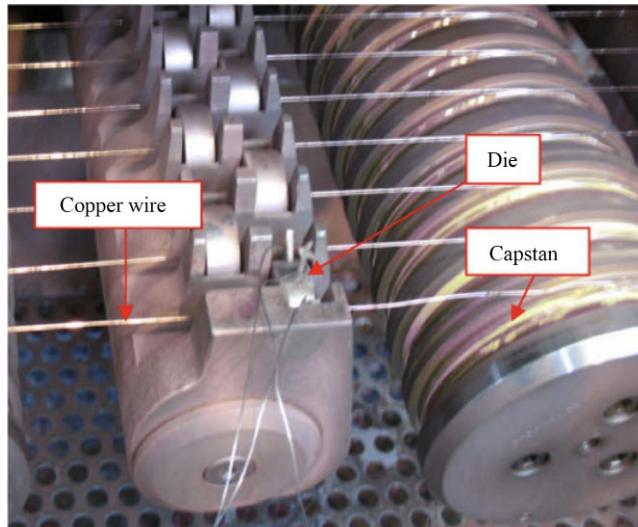


Fig. 14.2 Multi-pass drawing machine for Cu wire. The whole system is fully immersed in the lubricant.

Reproduced from Vega, G., Haddi, A., Imad, A., 2009. Investigation of process parameters effect on the copper-wire drawing. *Mater. Des.* 30, 3308–3312, with permission of the Publisher.

where A_0 and A_f are the areas of the cross sections of the wire (or rod) before and after the reduction. In general, the r values are between 0.1 and 0.4 per pass and for steel drawing they are typically limited to the range of 0.2–0.3. To achieve the required total reduction, several steps are then required (in a *multi-pass* process). For example, in the drawing of a hot rolled wire with a typical diameter of 5.5 mm to obtain a wire of 2 mm, approximately 10 passes are required. This process is carried out in a dry *rough drawing machine* with different dies in series. To obtain diameters of 0.2–0.5 mm, another 20 passes are required, in a *wet drawing machine*; an example is shown in Fig. 14.2. To obtain diameters around 0.1 mm, a further 30 passes are typically required in a *wet drawing machine for thin wires*. There are also drawing machines for producing capillary wires (0.05 mm) and microwires (diameters on the order of tens of micrometers).

The drawing velocity varies from 0.15 to 50 m/s (the most used values range from 20 to 50 m/s), depending on the diameters, the reductions per pass, and the power of the motor. Of course, the drawing velocity continuously increases in the case of a multi-pass process, because of the reduction of the cross section. Generally, the highest values of the drawing velocity are obtained in the production of the smaller diameters. As velocity increases, productivity increases and the degree of lubrication improves, so friction is reduced. At high drawing speeds, the lubrication can be plasto-hydrodynamic and the coefficient of friction close to 0.01 (Lenard, 2002). An increase in the drawing velocity typically induces an increase in the strain rates; the typical values in practice range from 10^3 to 10^4 s⁻¹ (see also Table 10.2).

14.2 Mechanics of wire drawing

Wire drawing stress and load

The average plastic deformation in the drawing direction is given by:

$$\varepsilon = \ln \frac{L_f}{L_0} = \ln \frac{A_0}{A_f} = \ln \left(\frac{1}{1-r} \right) \quad (14.2)$$

where L_0 and L_f are the length of the wire before and after drawing. A metal slice that enters the die undergoes a continuous increase in plastic deformation, which reaches the maximum at the exit from the die. However, plastic deformation is rather inhomogeneous, as in extrusion, and parameter Δ is expressed by Eq. (13.2) ($\Delta \cong 4\alpha/\varepsilon$).

The *drawing force*, T , is given by:

$$T = \sigma_d \cdot A_f \quad (14.3)$$

where σ_d is the *drawing stress*, i.e., the applied tensile stress at the metal slice that is exiting the die. For the sake of simplicity, we start neglecting the friction at the sliding wire-die interface. Hence, the relationship that expresses the equality of the external work done on the system and the work for plastic deformation is given by:

$$T \cdot L_f = \bar{\sigma}_{\text{eff}} \cdot \varepsilon_{\text{eff}} \cdot V$$

where:

- $\bar{\sigma}_{\text{eff}}$ is the mean average effective plastic stress;
- ε_{eff} is the total effective strain;
- V is the volume of the plastically deformed metal ($A_f L_f$).

From Eq. (14.3) we then obtain:

$$\sigma'_d = \bar{\sigma}_{\text{eff}} \cdot \varepsilon_{\text{eff}} = \Phi_1 \cdot \bar{\sigma}_1 \cdot \varepsilon \quad (14.4)$$

where the apex indicates that friction is neglected, and Φ_1 accounts for the so-called *redundant work*, i.e., the work (due to the inhomogeneity of plastic deformation) that is needed in addition to that required for a simple (tensile) uniaxial deformation. This aspect is schematized in Fig. 14.3.

As seen in the case of forging and rolling, the mean effective strain, ε_{eff} , is also expressed by: $\varepsilon_{\text{eff}} = f \cdot \varepsilon$. The parameter Φ_1 is then related to f through the following relationship:

$$\Phi_1 = \frac{\bar{\sigma}_{\text{eff}} \cdot \varepsilon_{\text{eff}}}{\bar{\sigma}_1 \cdot \varepsilon} = f^{n+1} \quad (14.5)$$

Parameters Φ_1 and f are related to the inhomogeneity of plastic deformation and are thus expected to be proportional to Δ . From different experimental data, the following rather general relationship has been proposed (Wright, 2011):

$$\phi_1 = 0.8 + 0.22 \cdot \Delta \quad (14.6)$$

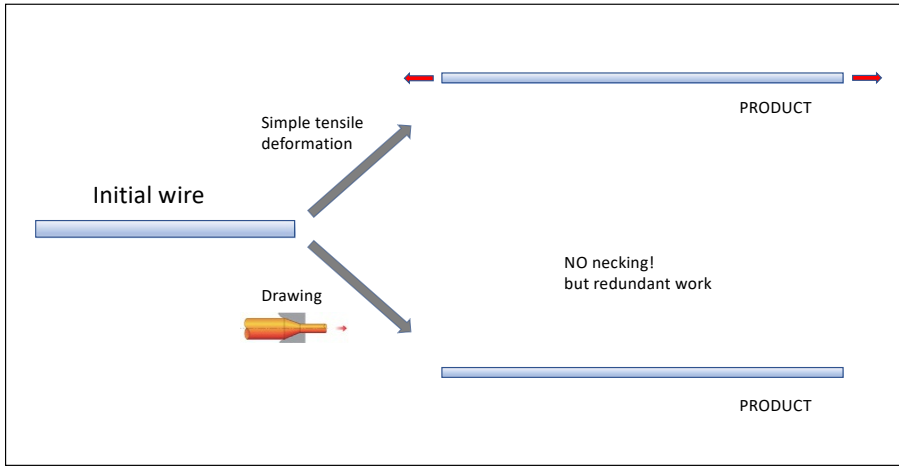


Fig. 14.3 Schematization showing the concept of redundant work in wire drawing. Please note that in the case of wire drawing the deformation limit due to necking does not exist.

The influence of friction at the workpiece/die interface (expressed by the friction coefficient μ) on the drawing force can finally be accounted for with parameter Φ_2 (Dieter, 1988):

$$\Phi_2 = 1 + \mu \cdot \cot g\alpha \quad (14.7)$$

so that Eq. (14.4) becomes:

$$\sigma_d = \Phi_1 \cdot \Phi_2 \cdot \bar{\sigma}_1 \cdot \varepsilon \quad (14.8)$$

In the case of a multi-pass process, the wire strain hardening must be considered, and therefore $\bar{\sigma}_1$ for the i th pass is given by:

$$\bar{\sigma}_1 = \frac{\int_{\varepsilon_i}^{\varepsilon_{i+1}} K \cdot \varepsilon^n \cdot d\varepsilon}{\varepsilon_{i+1} - \varepsilon_i} = \frac{K}{n+1} \cdot \frac{\varepsilon_{i+1}^{n+1} - \varepsilon_i^{n+1}}{\varepsilon_{i+1} - \varepsilon_i} \quad (14.9)$$

when the L-H equation is used. The effect of strain rate (which can be calculated using the equation proposed for extrusion) and of the temperature rise should be properly taken into account, and therefore the extended L-H equation or the J-K equation should be used (see Section 3.7), even if the two effects tend to cancel each other. In some metals, such as in steels, the analysis is further complicated by the DSA phenomenon, which introduces an anomalous dependency of the flow stress on strain rate and temperature (Section 3.4).

During drawing, it is necessary that the wire does not deform plastically outside the die, i.e., $\sigma_d < \sigma_Y$ (where the yield strength must be considered after drawing and the

corresponding strain hardening). This introduces an upper limit for the drawing reduction r , which explains the necessity of conducting a number of passes to obtain the desired wire reduction.

As already mentioned, the plastic deformation in the conical die is rather inhomogeneous. In particular, intense compressive plastic deformation takes place close to the die walls and this localization increases with Δ (see the schematization reported in Fig. 10.11). Because of this, the region along the central axis is subjected to secondary (unexpected) tensile stress.

The curves in Fig. 10.11 can be used to have a broad estimation of the centerline tensile stress and the triaxiality factor T . Positive T values in wire drawing are generally observed for $\Delta > 1.3$. More precise information, able to consider the role of strain hardening, can be achieved by numerical methods (see, e.g., Gonzales et al., 2018).

Temperature rise

The evaluation of the wire temperature rise during drawing is very important for estimating the correct value of $\bar{\sigma}_1$ and, most of all, to check if it remains below an upper limit, to avoid:

- An excessive reduction of the lubricant viscosity (that would impair the lubrication regime);
- Damage to the metal microstructure (e.g., this may occur in steels because of the phenomenon of blue brittleness connected with DSA; see the next section).

The wire temperature at the exit of the drawing die, T_{exit} , is given by:

$$T_{\text{exit}} = T_0 + \Delta\bar{T}_{\text{def}} + \Delta\bar{T}_{\mu} \quad (14.10)$$

where T_0 is the temperature of the wire at the entry with the die, and the other two terms account for the temperature rise due to plastic deformation and friction (see also Eq. 13.13). Following the same approach used for extrusion, we can write:

$$\Delta\bar{T}_{\text{def}} = \frac{\sigma'_d}{\rho \cdot c} \quad (14.11)$$

$$\Delta\bar{T}_{\mu} = \frac{\sigma_d - \sigma'_d}{\rho \cdot c} \quad (14.12)$$

In the calculation, it can be considered that 80%–90% of the generated heat is transferred into the wire (Haddi et al., 2009). As an example, in Fig. 14.4 the results of the calculation are shown for some metal alloys and different reductions, r . In practice, the maximum temperature rise is attained at the wire surface because of the contribution of friction, and Eqs. (14.11) and (14.12) underestimate the real surface temperature achieved during drawing. Thermal equilibrium is reached after the exit from the die, at a distance of three to five times the wire diameter (Wright, 2011).

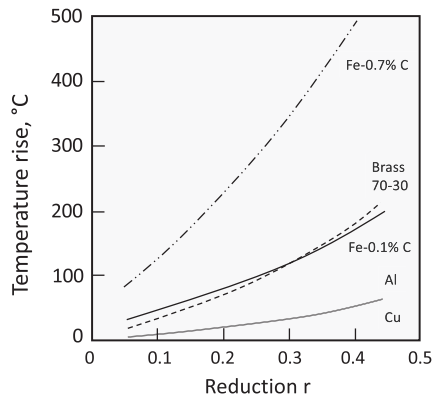


Fig. 14.4 Calculated temperature rise during cold drawing of different metals as a function of the reduction, r .

In the case of a multi-pass process, the temperature rise can become very intense, because of the material strain hardening and the continuous increase in drawing velocity. The wire must then be properly cooled in the interpass time by convection with the surrounding air and, most of all, through conduction with the capstans. The die and the capstans can also be water cooled to reduce the temperature rise in the wire.

Residual stresses

Cold drawn wires are generally characterized by the presence of residual stresses (as this is rather common in cold worked products). They are tensile on the surface and compressive in the interior of the wire. As an example, Fig. 14.5 shows the evolution of experimentally measured and computed longitudinal residual stresses at different depths for a cold drawn ferritic steel (Ellices, 2004). It can be noted that at the surface a tensile residual stress of approximately 250 MPa is present. In the case of cold drawing of a pearlitic steel from an initial diameter of 12 mm to a final diameter of 5.2 mm, a surface tensile longitudinal stress of approximately 600 MPa was measured (Ellices, 2004).

Strain hardening during wire drawing induces a considerable increase in the material strength but residual stresses modify the stress-strain behavior of the wire. In particular, they induce an early deviation from linearity, i.e., the phenomenon of continuous yielding introduced in Section 1.3. This is shown, for example, in Fig. 14.6, which compares the profiles of computed and measured residual stresses (Fig. 14.6A) with experimental stress-strain curves (Fig. 14.6B), in the case of a pearlitic wire submitted to: drawing; drawing with a subsequent small reduction (1%); and drawing with a subsequent thermomechanical treatment consisting of tensile loading at 400°C (Ellices, 2004). The last two treatments were carried out with the aim of reducing the residual stresses. The stress-strain curves show that the cold drawn wire

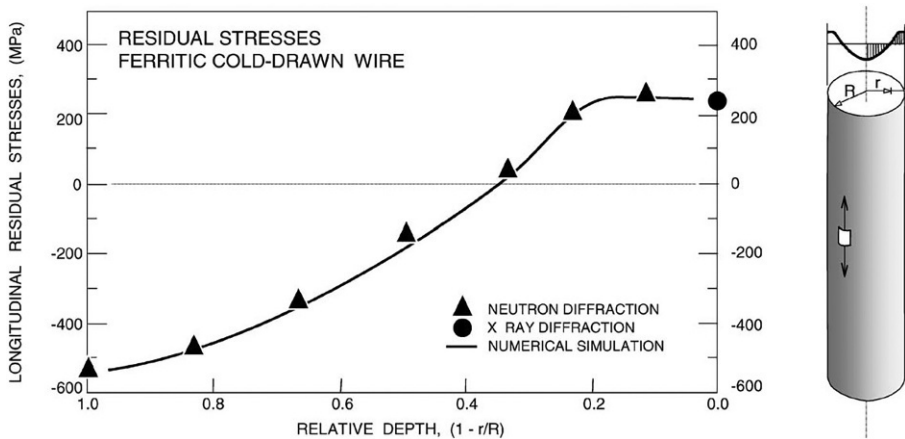


Fig. 14.5 Measured and computed residual stress profile in a drawn ferritic steel. Reproduced from Elices, M., 2004. Influence of residual stresses in the performance of cold-drawn pearlitic wire. *J. Mater. Sci.* 39, 3889–3899, with permission of the Publisher.

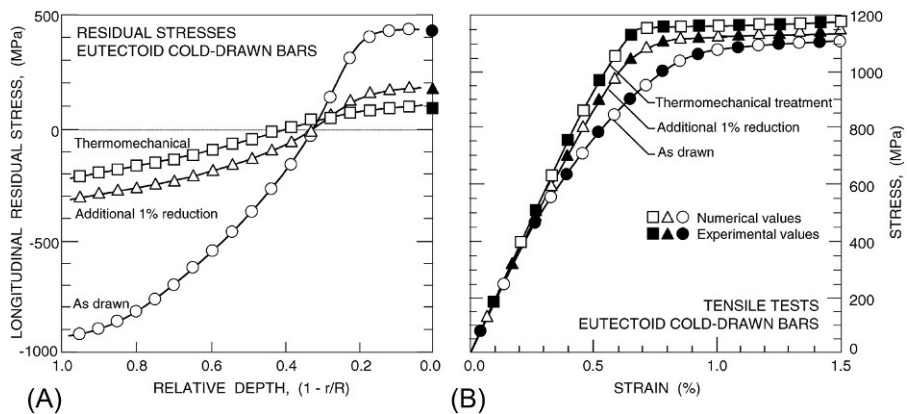


Fig. 14.6 (A) Profiles of computed residual stresses in a pearlitic wire, in the as-drawn condition, after an additional small plastic deformation and after a thermomechanical treatment; (B) corresponding stress-strain curves.

Reproduced from Elices, M., 2004. Influence of residual stresses in the performance of cold-drawn pearlitic wire. *J. Mater. Sci.* 39, 3889–3899, with permission of the Publisher.

with comparatively high residual stresses shows an early deviation from linearity, whereas the two treatments proved to be effective in reducing residual stresses and avoiding the localized yielding phenomenon, with beneficial consequences on the strength, in particular the fatigue resistance, of the wires.

Residual stresses also favor different damaging phenomena that may occur in service, such as fatigue cracking or hydrogen embrittlement (Borchers and Kirchheim, 2016).

Residual stresses develop as the cold drawn wire emerges from the die exit. Two explanations have been proposed. The first is related to the inhomogeneity of plastic deformation of the metal in the die. The outer metal regions undergo plastic deformation under a prevailing compressive triaxiality, whereas the metal along the centerline undergoes plastic deformation under tensile triaxiality. As explained in [Chapter 2](#), strain hardening under compression is in general more intense than under tension. As confirmed by hardness measurements on the product cross sections ([Hosford and Caddell, 1993](#)), a greater hardening is reached in the zones deformed in compression than in those deformed in tension. Therefore, after unloading (i.e., after the exit from the die), the outer metal would recover a higher elastic strain than the interior metal. For continuity reasons, the outer metal is then stretched in tension (and the underlying metal in compression). It is clear that the intensity of residual stresses increases with increasing in the localization of plastic deformation, i.e., with Δ . This has actually been verified for the extrusion of steel rods. Such an increase is of particular intensity when $\Delta > 2$ ([Hosford and Caddell, 1993](#)).

The second explanation is related to the temperature rise in the drawn wire. As seen, temperature is higher in the surface regions and lower in the interior. As the wire emerges from the die, the temperature differences are reduced. During this equilibration stage, the surface regions of the wire tend to contract because of cooling, whereas the internal regions tend to elongate because of the heat diffusing from the surface zones. Therefore, this phenomenon contributes to the formation of tensile residual stresses at the wire surface and compressive stresses in the interior ([Wright, 2011](#)).

14.3 Ductile damage and cracks in drawn wires

An excessive temperature rise may induce the deterioration of the lubricant and the lubrication regime with the formation of surface defects, following the adhesive wear between the sliding wire and die surfaces ([Byon et al., 2011](#)). An increase in friction may also lead to the formation of surface cracks that are perpendicular to the drawing direction ([Table 10.6](#)). An example is shown in [Fig. 14.7](#). Ductile cracking is favored

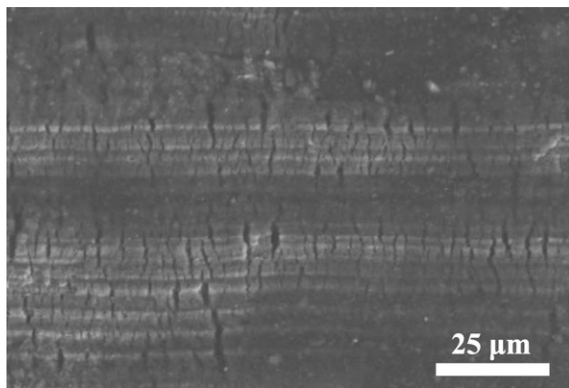


Fig. 14.7 Surface cracks in a drawn carbon steel perpendicular to the sliding direction ([Larsson et al., 2019](#)).

by formation of surface tensile stresses induced by friction and the decrease in ductility due to strain hardening (and the corresponding decrease in the strain-hardening coefficient, n). In some cases, intermediate annealing stages are then carried out to increase n and ductility.

In addition, surface martensitic regions may form in the case of steel drawing, following the austenitization and subsequent rapid cooling of surface overheated regions. Such regions then become brittle (because of the typically high C content), and this can favor the formation of surface brittle cracks and possibly the wire fracture in the subsequent passes (Wei et al., 2020). As already mentioned, the maximum temperature rise must be properly controlled to avoid all these problems.

Drawing may also induce the formation of internal ductile damage, which is difficult to reveal and can be very dangerous. Fig. 14.8 shows the formation of internal voids in a pearlitic wire after seven passes (Lee et al., 2009). Voids nucleation is the initial stage of the ductile fracture process, as explained in Chapter 6. It took place because the total (cumulative) effective strain, together with the increased stress triaxiality, was sufficient to achieve the voids nucleation condition, i.e., $\epsilon_{\text{eff}} > \epsilon_{\text{eff},n}$. Such internal damage may induce the complete ductile fracture under subsequent drawing passes, or lower the strength of the product. If submitted to torsional deformation, internally damaged wires display the typical *delamination* morphology, characterized by an early longitudinal splitting at the wire surface that propagates along the wire length with a helical geometry.

In steels, the occurrence of the internal damage during drawing is greatly facilitated by the possible occurrence of the DSA phenomenon. As seen in Sections 3.4 and 8.2, DSA is related to the interaction of dislocations with solute atoms, in particular N in the case of steels, and it induces an increase in the strength and a decrease in ductility. With RT tensile tests, specific DSA maps are obtained, and two examples are shown in Fig. 3.23. The maximum temperature rise in a drawing process is typically up to 200°C

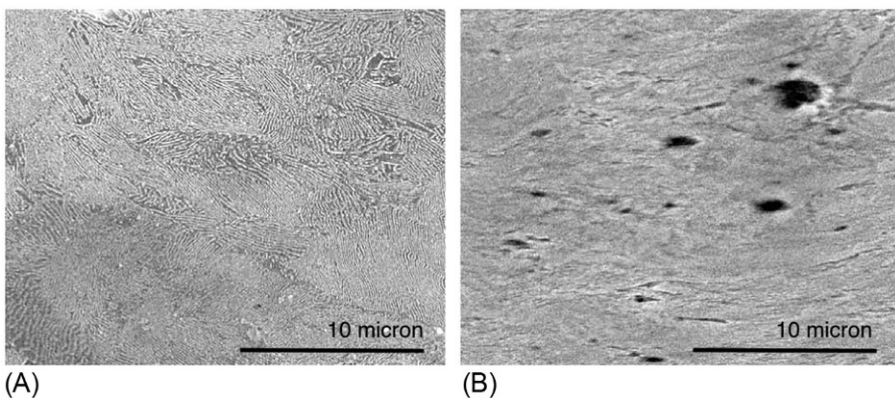


Fig. 14.8 Microstructure of a pearlitic wire before (A) and after (B) a drawing process (reproduced from Lee, S.K., Ko, D.C., Kim, B.M., 2009. Pass schedule of wire drawing process to prevent delamination for high strength steel cord wire. Mater. Des. 30, 2919–2927, with permission of the Publisher). In (B) the presence of ductile voids can be appreciated.

and 300°C. A comparison with the maps of Fig. 3.23 shows that the upper boundary for DSA to occur is well below 10^{-1} s^{-1} at these temperatures, depending on the chemical composition of the steel (in particular, the content of N). In cold drawing, strain rates in the range 10^2 – 10^3 s^{-1} are rather common, that are larger than the upper limit for DSA obtained by tensile testing. However, we have to consider that the limiting strain rate for DSA, $\dot{\epsilon}_{\max}$, depends on the diffusivity of the interstitial atoms and on the density of mobile dislocations. Making reference to Eq. (3.10), the following relationship has been proposed (McCormick, 1973):

$$\dot{\epsilon}_{\max} \propto \frac{\rho}{T} \cdot \exp\left(-\frac{Q}{R \cdot T}\right) \quad (14.13)$$

where:

- ρ is the density of dislocations;
- Q is the activation energy for DSA (that in steels corresponds to the activation energy for interstitial diffusion of N in ferrite);
- T is the absolute temperature.

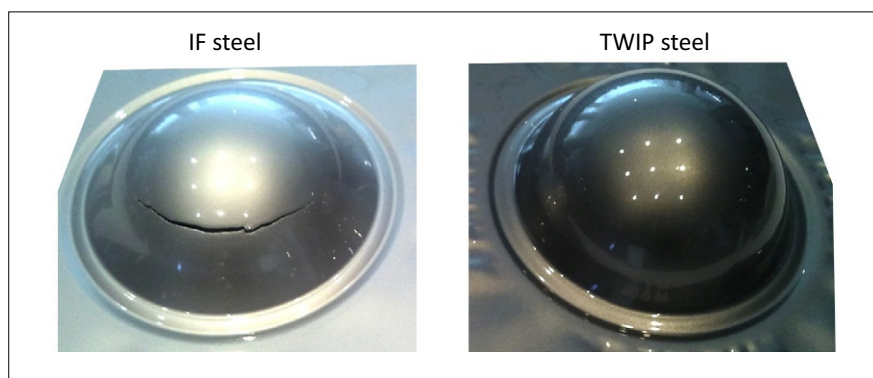
In a drawing process, the total strain is much larger than in a tensile test, and therefore the corresponding density of dislocation is also much higher, typically two orders of magnitude (Karimi Taheri et al., 1995). This means that the upper boundary for DSA obtained by tensile testing can be shifted upwards approximately two orders of magnitude to obtain a map valid for a drawing process. From the new map and Eq. (14.13), it is possible to obtain the temperature conditions to have DSA in a given drawing process, and therefore to design the system, i.e., the drawing speed and the cooling systems, to assure that the wire temperature achieved during drawing is always below the calculated limit (Karimi Taheri et al., 1995).

With increasing drawing semi-angle, α , and decreasing reduction r (and then ϵ), parameter Δ increases, and therefore the triaxiality parameter increases considerably along the wire centerline. This may pose the conditions for the formation of central bursts along the drawing axis (also called *chevron cracks*); the mechanism is the same as that introduced for extrusion. An example is displayed in the first page of this chapter, which shows internal chevron cracks on the cross section of a drawn Cu wire (Norasethasopon and Yoshida, 2008). It is clear that chevron cracks represent an internal damage that is definitively more severe than internal voids. To avoid their formation, the indications of the diagram in Fig. 13.10B should be followed.

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Bulge test on an IF and a TWIP steel; the TWIP steel is undamaged at a dome height that is larger than that of the IF steel at failure.

Reproduced from De Cooman, B.C., Estrin, Y., Kim, S.K., 2018. Twinning-induced plasticity (TWIP) steels. *Acta Mater.* 142, 283–362, with permission of the Publisher.

Chapter outline

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Sheet metal forming operations, typically carried out in cold conditions, allow simple or complex flat and prismatic shapes to be obtained. These operations are carried out using a mechanical or hydraulic press and involve flat rolled products: plates or sheets. Plates, obtained by hot rolling, are supplied with a thickness greater than 3 mm. Steel plates have a C content generally around 0.15% to ensure high weldability; in fact, welding is often an important process in the production of parts obtained from plates, such as pipelines, hulls for ships, or metallic buildings. To increase the strength characteristics without compromising the weldability, HSLA steels are also used. The main processes for hot-rolled plates are shearing, bending, and, in specific cases, deep drawing.

Sheets with a thickness typically lower than 1 mm are produced by hot and cold rolling, followed by annealing and possibly skin-passing (and also galvanizing or pre-painting, to protect the product from corrosion). Steel sheets typically contain very low C content, below 0.1%, to have a very high formability (low flow strength combined with high ductility). Recently, special high-strength steels have been developed, mainly for the car industry, comprising high-strength IF steels, bake-hardening steels, dual phase steels and TRIP/TWIP steels. Rolled and annealed flat products made with non-ferrous alloys, such as Al alloys, are also formed by sheet metal working operations. Thin sheets are formed by shearing, bending, and stamping operations. Stamping operations comprise deep drawing, stretch forming, and some special processes, such as hydroforming, superplastic, or high-velocity forming.

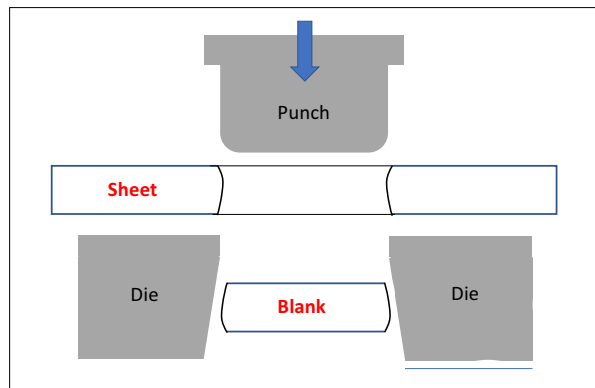
The metal ductility should be sufficiently high to avoid formation of ductile cracks during forming (an exception is shearing, in which ductile fracture is taken as an advantage). The use of metals with high values of the strain-hardening coefficient, n , and of the strain rate sensitivity coefficient, m , are then preferred. Metals with these properties are also resistant to excessive localized necking that generally anticipates ductile fracture and in most cases cannot be accepted.

15.1 Main processes

With *shearing* it is possible to obtain planar parts with complex shapes, by means of an appropriate choice of the die geometry and operation sequence. The principle of the process is schematized in Fig. 15.1. When shearing a close contour and the internal part is the required one, the process is called *blanking*. This process is also used to prepare semi-finished products (called *blanks*) for subsequent processing: for example, to obtain blanks with very complex geometry for subsequent bending operations, or simply circular blanks for subsequent stamping operations.

With *bending*, the starting sheet or plate is curved. Two common bending processes, free and press bending, are schematized in Fig. 15.2. By starting with a blank with a proper geometry and carrying out operations in a sequence (even in an automated way), it is possible to obtain three-dimensional objects of very complex shape.

Fig. 15.1 Principle of shearing in the case of the blanking process (if the desired part is outside the contour, the process is called punching).



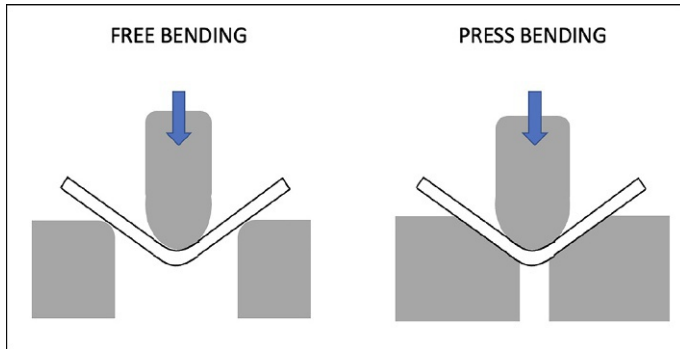


Fig. 15.2 Common bending processes.

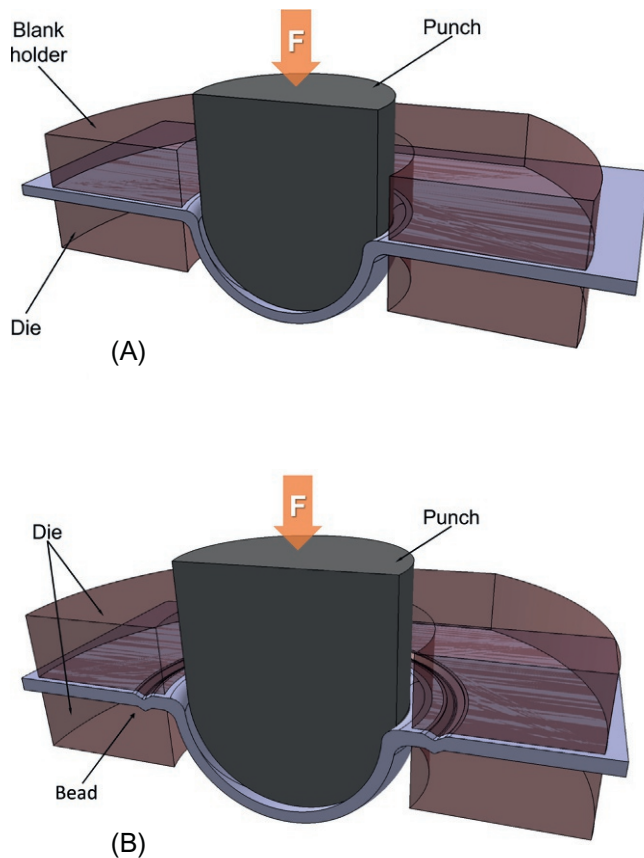
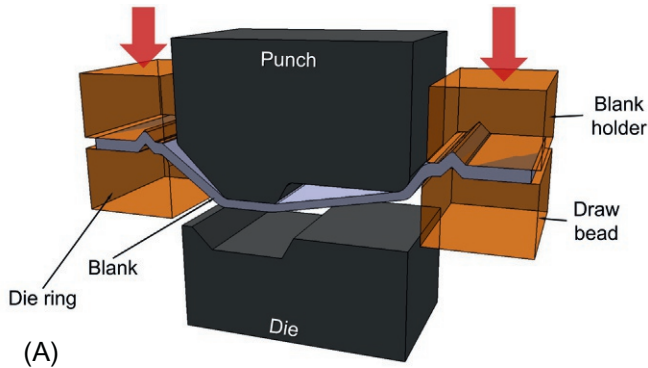
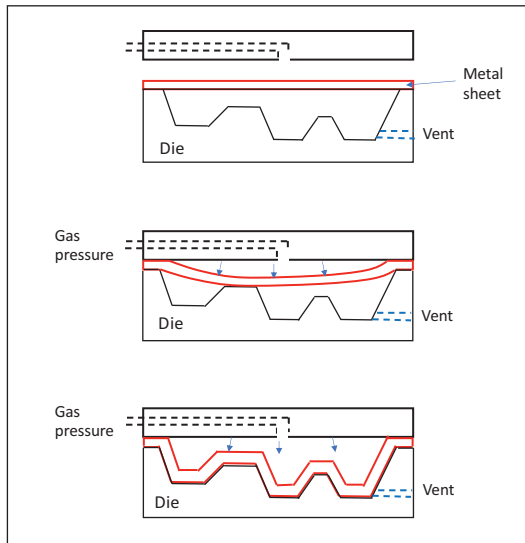


Fig. 15.3 Basic stamping operations: (A) deep drawing; (B) stretch forming.

Stamping operations typically involve thin sheets, and can be subdivided into *deep drawing* and *stretch forming* operations. The basic operations are schematized in Fig. 15.3. By means of these processes, a sheet is transformed into a dished body. We refer to deep drawing when the blank holder does not prevent the radial sliding of the sheet. We refer instead to stretch forming when the blank holder (with or



(A)



(B)

Fig. 15.4 (A) Simple stretch forming operation using two forming dies, and (B) using a single forming die and the action of a fluid pressure.

without a bead) blocks the sheet and prevents it from sliding on the die surfaces. The different operations can therefore be classified into deep drawing or stretch forming types. With deep drawing operations, axial symmetrical parts, such as containers for drinks and pots, and non-axisymmetrical parts, such as panels for cars, can be obtained. Stretch forming operations include simple stretch forming (Fig. 15.3B), press stretch forming using an upper and lower die (Fig. 15.4A), and stretch forming under the action of a fluid pressure that forces the sheet metal to deform against a forming die (Fig. 15.4B).

Among the advantages offered by sheet metal forming, the so-called *tailored blanks* deserve to be mentioned. These blanks are formed by two or more sheets with different characteristics (thickness and chemical composition) welded together. In this way, it is possible to produce parts with different mechanical properties with respect to the position.

15.2 Shearing

The most important parameters in shearing operations are the *clearance*, c ; the punch and die radii, R_p and R_d ; and the punch speed, v . They are defined in Fig. 15.5. Typical values of c are in the range of 1%–15% of the sheet thickness, t , being the lowest values used in the case of *precision blanking*. The punch displacement is typically provided by a hydraulic press and the maximum industrial punch speed is around 50 mm/s.

The shearing process develops through three main stages:

1. The bending stage. The punch motion first induces a bending of the sheet, forming the so-called roll-over region;
2. The shearing stage. The metal starts to deform plastically by forced shear ($T \cong 0$), in a band with a thickness close to the clearance, c , between the punch and the die. The metal spreads along the punch and forms a smooth shear zone almost parallel to the direction of the punch motion. The length of the shear zone increases with the metal ductility. The mean shear strain rate is approximately given by: v/c . For example, with $v = 50 \text{ mm/s}$, $t = 1 \text{ mm}$ and $c = 10\%$, the average strain rate is about $5 \times 10^2 \text{ s}^{-1}$;
3. The fracture stage. When the local effective strain, ϵ_{eff} , reaches the critical value for fracture, $\epsilon_{\text{eff},f}$, a ductile crack nucleates and grows along the shear band. A fracture zone is then formed. Depending on c , the local plastic deformation can be rather large, with the consequent formation of a *burr* that reduces the surface quality of the manufactured part and must be therefore minimized.

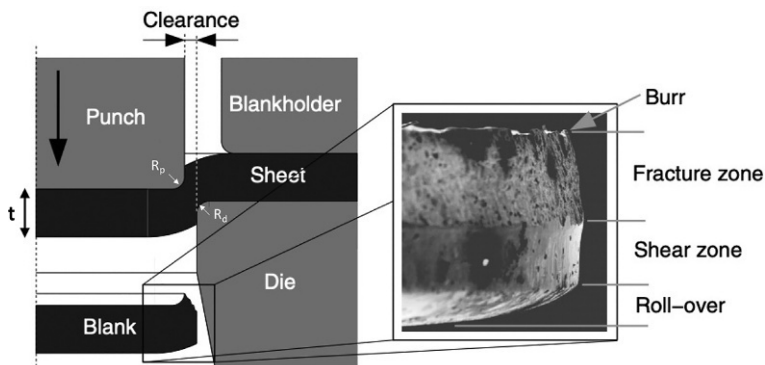


Fig. 15.5 Schematization of the shearing (blanking) process with indication of the regions determining the shape of the blank edge.

Reproduced from Goijaerts, A.M., Govaert, L.E., Baaijens, F.P.T., 2001. Evaluation of ductile fracture models for different metals in blanking. *J. Mater. Process. Technol.* 110, 312–323, with permission of the Publisher.

It is apparent that ductile fracture in shearing is taken as an advantage. The R-T fracture criterion is then used to evaluate the fracture load, F_f , that is close to the maximum load, $F_{\max}$. The approach is of course simplified, but is useful to grasp the main aspects that are involved.

From the R-T criterion (i.e., Eq. 6.17 with $T=0$), we obtain that: $\epsilon_{\text{eff},f} = 1.89 \epsilon_f$. Considering the L-H constitutive equation, the effective stress at fracture is: $\sigma_{\text{eff},f} = K \cdot (1.89 \cdot \epsilon_f)^n$. The shear stress at fracture, τ , is then given by (see Section 5.3):

$$\tau = \sigma_1 = \frac{\sigma_{\text{eff},f}}{\sqrt{3}} = \frac{K}{\sqrt{3}} \cdot (1.89 \cdot \epsilon_f)^n$$

The applied load increases with the punch stroke because of strain hardening, and the load at fracture is then given by: $F_{\max} = \tau b t^*$, where b is the total length of the shear edge and t^* is the thickness of the sheet at the onset of fracture (please note that the fracture load may actually be lower than the maximum load, since the decrease in the cross section may overwhelm the effect of strain hardening). As seen in Fig. 15.5, in general $t^* < t$ because of the penetration of the upper die into the workpiece. In general, we can set: $t^* = \delta t$. Parameter δ depends on the sheet metal and the sheet thickness. If the sheet thickness is lower than 1 mm, δ typically ranges between 0.5 (high alloy steel) and 0.8 (Al alloys) (Omar, 2011).

Considering that: $K = s_{\text{UTS}} \cdot e^n / n^n$ (see Section 1.1), we obtain:

$$F_{\max} = \frac{e^n}{n^n} \cdot \frac{(1.89 \cdot \epsilon_f)^n}{\sqrt{3}} \cdot s_{\text{UTS}} \cdot b \cdot t^* = C \cdot s_{\text{UTS}} \cdot b \cdot t \quad (15.1)$$

where:

$$C = \delta \cdot \frac{e^n}{n^n} \cdot \frac{(1.89 \cdot \epsilon_f)^n}{\sqrt{3}} \quad (15.2)$$

As seen in Chapter 7, the true fracture strain, ϵ_f , is proportional to the strain-hardening coefficient n . This means that parameter C depends mainly on n only. From the data in Fig. 7.11 (and converting the EL values into RA values using the experimental data for steels reported in: Warren et al., 1963), parameter C can be estimated as a function of n . This is graphically shown in Fig. 15.6 for $\delta = 0.65$. It is seen that C increases with n .

In general, an average value of 0.7 for C is recommended (Dieter, 1988). Some experimental values for C are listed in Table 15.1 for different metals (based on data from: Goijaerts et al., 2001). In the table, the main properties of the metals are also listed, together with typical n values taken from Chapter 3. The experimental C values are in reasonable agreement with the predictions of Fig. 15.6, and in particular they increase with increasing coefficient n .

Of course, the picture presented here is simplified, since C is influenced by other parameters in addition to n . One is the friction at the punch-sheet interface and at the

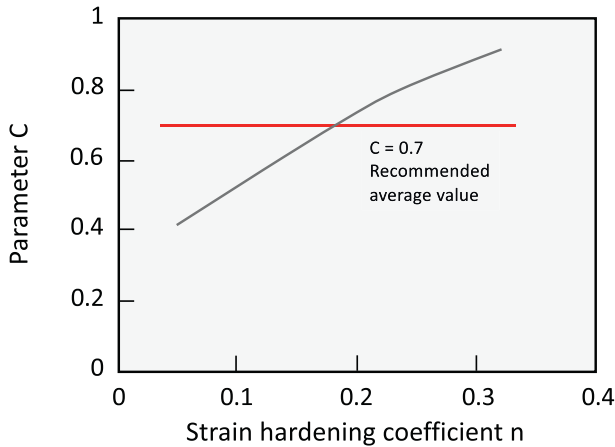


Fig. 15.6 Graphical representation of Eq. (15.2), showing the predicted dependency of C on n (for $\delta = 0.65$).

Table 15.1 Results of shearing tests conducted on sheets with a thickness of 1 mm and made of different alloys (data reduced from: Goijaerts et al., 2001).

Metal		s_{UTS} (MPa)	Thickness at fracture (mm)	n (*)	C
X30Cr13	Ferritic stainless steel	628	0.448	0.28	0.8
X10CrNiMoTi 18 10	Austenitic stainless steel	628	0.320	0.4	0.91
CuZn36	Brass	364	0.342	0.4	0.87
DC 04	Low C steel	285	0.132	0.3	0.95
Al MgSi 1 Mg	Al alloy	343	0.642	0.25	0.74

*The strain-hardening coefficient n is estimated using the data reported in Chapter 3.

die-sheet interface, even if it is generally reduced to a minimum through proper lubrication. Another parameter is the punch velocity, which determines the local strain rate, as mentioned before. In general, as the punch velocity is increased, $F_{\max}$ also increases, but only up to a limiting value. For example, this value is close to 1 mm/s for the ferritic stainless steel of Table 15.1 (Goijaerts, 1999). Above this value, $F_{\max}$ becomes independent of v . In fact, from one side an increase in v induces an increase in the local strain rate and hence in the metal strength, but from the other side, it also induces a metal softening due to the increased contribution of local heating (the process tends to be adiabatic). At a certain velocity, the two phenomena almost counterbalance and $F_{\max}$ is no longer affected by v .

For a precise modeling of the shearing process aimed at the determination of $F_{\max}$ and other related parameters, such as the punch displacement at fracture, a numerical approach (such as FEM) should be employed, together with a proper constitutive equation (Section 3.7).

15.3 Bending

Fig. 15.7 schematizes the bending process. The thickness of the sheet (or plate) is indicated with t , whereas b indicates its width. The bend radius is indicated with R . The plastic deformation concentrates in the bent region, and tensile engineering deformation in direction 1, in correspondence to the outer surface, is:

$$e_1 = \frac{1}{\frac{2R}{t} + 1} \quad (15.3)$$

As the R/t ratio decreases, the tensile deformation at the outer surface increases. The *bend limit* is therefore given by the minimum value of the R/t ratio that can be reached without surface ductile fracture. If the b/t ratio is below, say, 2–4, a state of uniaxial tensile stress is present on the outer surface since the material is rather free to deform in direction 2. See the schematization in Fig. 15.8A.

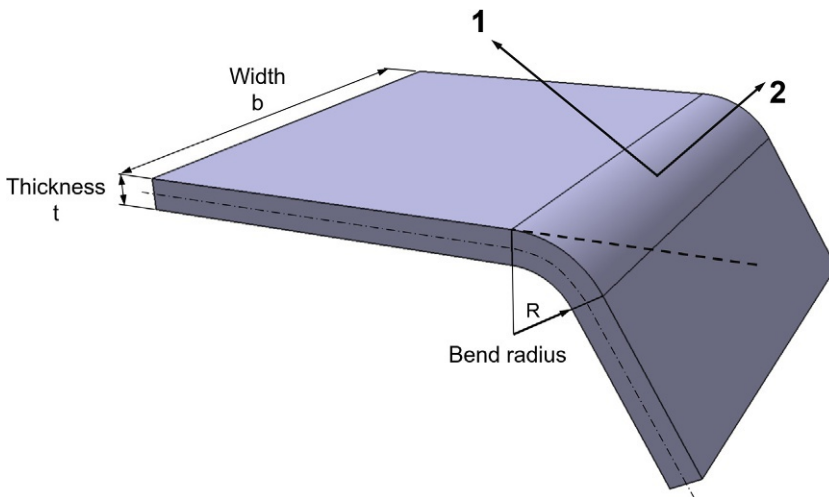


Fig. 15.7 Schematization of the bending process.

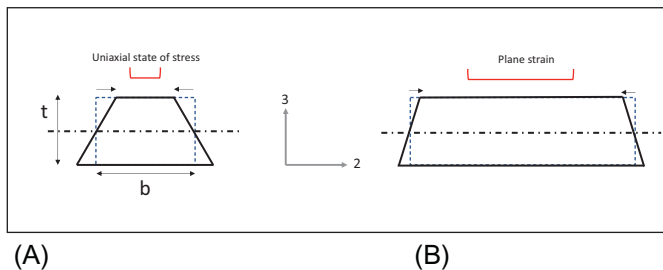


Fig. 15.8 Cross section of a plate in the bend region. (A) b/t ratio lower than 2–4; (B) b/h ratio greater than ≈ 6 .

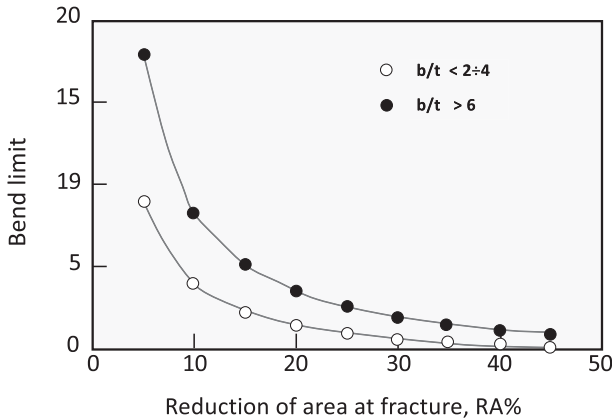


Fig. 15.9 Evolution of the bend limit as a function of metal ductility and the deformation state.

In this case, ductile fracture takes place when $\varepsilon_1 = \varepsilon_f$. The true plastic strain is: $\varepsilon_1 = \ln(1 + e_1)$, where e_1 is given by Eq. (15.3). The metal ductility can be simply expressed with the tensile reduction of area at fracture, AR, introduced in Section 1.1. In this way: $\varepsilon_f = \ln(100/100 - AR)$. By setting $\varepsilon_1 = \varepsilon_f$, we obtain:

$$\left. \frac{R}{h} \right|_{\min} = \frac{50}{AR} - 1 \quad (15.4)$$

Eq. (15.4) is plotted in Fig. 15.9. The bend limit is seen to decrease, and then bendability to increase, with increase in the metal ductility. Eq. (15.4) has been validated by data from several experimental studies (Kalpakjian and Schmid, 2009). Table 15.2 shows typical values of the bend limit for various metals. As expected, the lowest values are attained by metals in the annealed state, i.e., with a highest ductility. However, metals such as Mg and Ti alloys also display a low bendability in the annealed state, due to their low ductility, as explained in Chapter 7.

Table 15.2 Typical values of the bend limit for various metals.

	Annealed state	Work-hardened state
Low-C steels	0.5–0.6	4
Austenitic stainless steel	0.5–0.8	6
Al alloys	0.5–2	6
Ti	0.7	3
Ti alloy	2–4	4
Mg	5	13
Brass		2
Cu-Be alloys		4

Data from Kalpakjian, S., Schmid, S.R., 2009. Manufacturing Engineering and Technology. Prentice Hall, and other sources.

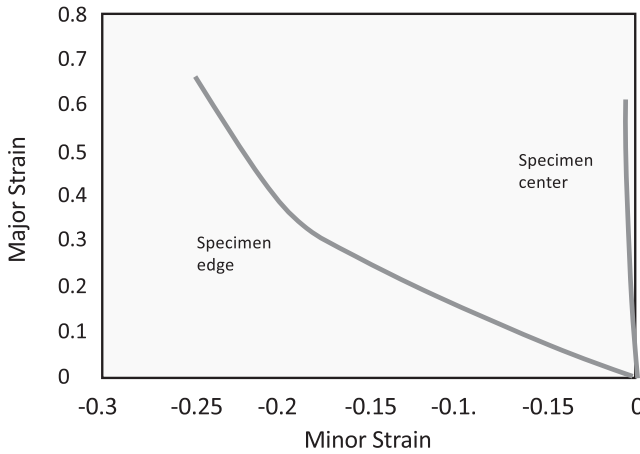


Fig. 15.10 Deformation paths in the center of a specimen and at its edge, in case of an Al sheet with a width-to-thickness ratio of 6.

Data from Kupke, A., Barnett, M., Luckey, G., Weiss, M., 2018. Determination of the bendability of ductile materials. IOP Conf. Ser. Mater. Sci. Eng. 418, 012077.

The bend limit does not depend only on the metal ductility but also on the stress state that is present on the outer stretched surface, and ultimately on the geometry of the sheet being bent. In fact, as the b/t ratio increases, the stress/strain state on the outer surface changes from uniaxial stress to plane strain. As an example, Fig. 15.10 shows the deformation path on the tensile surface of an Al sheet (with $b/t = 6$) submitted to bending, in the center of the specimen and at the edge. The indicated major strain is ε_1 , whereas the minor strain is ε_2 (they were measured with an optical technique). It is seen that at the edges $\varepsilon_2 \approx -1/2\varepsilon_1$, i.e., the deformation path is close to uniaxial deformation. In the specimen center $\varepsilon_2 \approx 0$, i.e., the deformation path is close to plane strain. In general, when b/t is in excess of 6, the deformation path in the surface center of the specimen is close to plane strain (Fig. 15.8B). Therefore the conditions for ductile fracture are attained for smaller values of the local effective strain than for uniaxial deformation. Using the R-T criterion, it is found that ductile fracture occurs when $\varepsilon_1 = 0.52 \varepsilon_f$. In Fig. 15.9, the predicted evolution of the bend limit with tensile ductility in the case of plane strain on the outer bent surface is also shown.

As seen, plastic deformation in bending is largely concentrated in the bend regions. After unloading, the elastic portion of the deformation is recovered and this induces the phenomenon of *springback*, given by the difference of the sheet angle after forming and that after the load removal. This phenomenon increases in intensity with decreasing the ratio R/t , and with increasing the intensity of the elastic strain to be recovered, which is given, to a first approximation, by the ratio σ_Y/E , where σ_Y is the yield strength of the metal and E is its elastic modulus. For example, springback is more severe in high-strength steels than in conventional mild steels and this should be considered in the design practice.

15.4 Sheet stamping

General aspects

In the case of stamping operations, sheets with a small thickness with respect to length and width are subjected to forming. Considering that the sheet surfaces are free of stresses, the through thickness stress in the sheet is negligible, and therefore the deformation process takes place in *plane stress conditions* (see also Fig. 5.3).

Fig. 15.11A schematizes a metal sheet with a printed small circle (with diameter d_0) on the surface. After a certain deformation, the circle transforms into an ellipse (Fig. 15.11B), whose geometry provides several pieces of information. First, the directions of the *major* and *minor deformations* are obtained, right in correspondence with the major and minor axes of the ellipse. They are indicated with 1 and 2 in Fig. 15.11B, and are principal directions. The corresponding strains and stresses are principal strains and principal stresses, respectively. The principal strains, ε_1 and ε_2 , can be obtained by the usual relationships:

$$\varepsilon_1 = \ln \frac{d_1}{d_0} \quad \varepsilon_2 = \ln \frac{d_2}{d_0}$$

The trough thickness strain, ε_3 , is given by the following relationship:

$$\varepsilon_3 = \ln \frac{t}{t_0} \quad (15.5)$$

where t and t_0 indicate the sheet thickness after and before deformation; ε_3 is also given by: $\varepsilon_3 = -(\varepsilon_1 + \varepsilon_2)$, and therefore:

$$t = t_0 \cdot e^{\varepsilon_3} = e^{-(1+\beta) \cdot \varepsilon_1} \quad (15.6)$$

where $\beta = \varepsilon_2/\varepsilon_1$ (see Section 5.3). If $\beta > -1$, the sheet undergoes thinning ($t < t_0$), whereas if $\beta < -1$ it undergoes thickening ($t > t_0$); of course, if $\beta = -1$, the sheet thickness is unchanged.

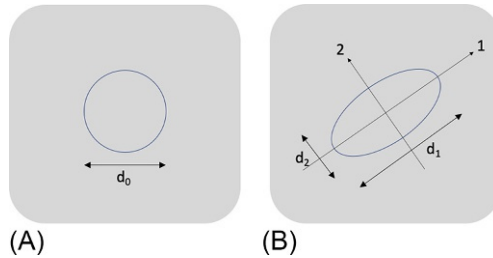


Fig. 15.11 Schematization showing a sheet before stamping with a printed circle with diameter d_0 (A), and the same sheet after plastic deformation, with the transformation of the circle into an ellipse (B).

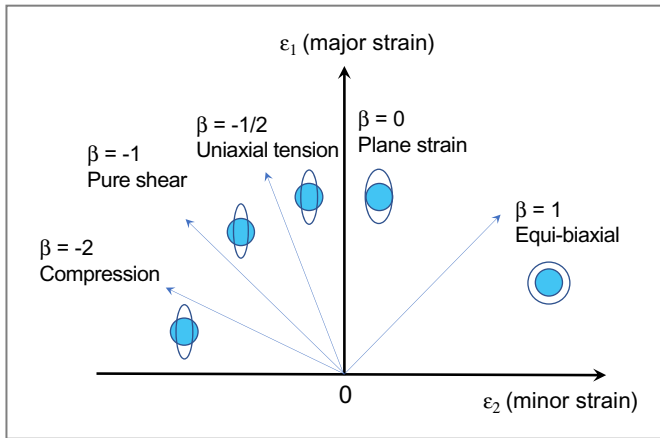


Fig. 15.12 Deformation diagram.

The deformation of a sheet sample may follow different paths that can be properly represented in the *deformation diagram* schematized in Fig. 15.12. Of course, at the beginning, the sample is undeformed: $\epsilon_1 = \epsilon_2 = 0$. Assuming a proportional (or linear) deformation behavior, each possible deformation path is represented by a straight line radiating from the origin, and it is then determined by parameter β (the slope of the line). In the first quadrant, both ϵ_1 and ϵ_2 are positive. The deformation path indicated with $\beta = 1$ occurs in equi-biaxial conditions (it is also called *equi-biaxial stretching*). Of course, the region where $\epsilon_2 > \epsilon_1$ simply mirrors the region with $\epsilon_1 > \epsilon_2$. Each deformation path with β between 0 and 1 takes place in *unequal biaxial stretching*. The deformation path indicated with $\beta = 0$ takes place under plane strain conditions ($\epsilon_2 = 0$). The deformation paths in the second quadrant are characterized by negative values of a principal strain component, namely ϵ_2 , that by definition is the minor one. In the case of $\beta = -1/2$, deformation takes place in uniaxial conditions, whereas it is under pure shear for $\beta = -1$. As seen, where β is greater than $-1/2$, sheet thinning takes place. The deformation path with $\beta = -2$ takes place under compression, and all the paths with $\beta < -2$ proceed with sheet thickening.

The principal stresses, σ_1 and σ_2 , along each deformation path can be calculated simply using the equations for plane strain conditions reported in Section 5.3, after the estimation of α from β using Eq. (5.13), and using the corresponding constitutive equation (the relations are of course valid in the case of proportional deformations and isotropic materials).

Mechanics of deep drawing and sheet stretching

The basic deep drawing process is schematized in Fig. 15.1. It can be conveniently described with reference to the schematization of Fig. 15.13A, where the blank is subdivided into three regions, indicated with x , y , and z (Beddoes and Bibby, 1999).

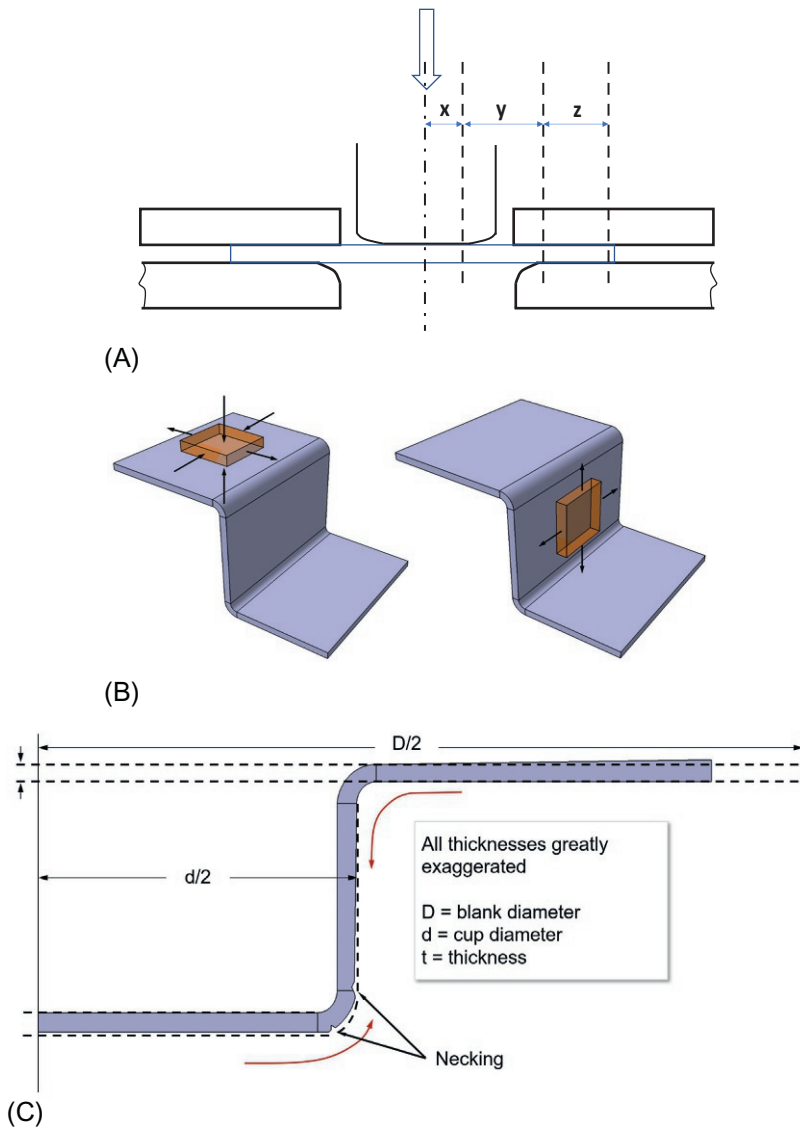


Fig. 15.13 Schematization of the deep drawing process: (A) initial condition with definition of the regions x , y , and z (redrawn and modified from Beddoes and Bibby, 1999); (B) pie-shaped segments showing the stresses in the flange, A, and in the sidewall, B; (C) thickness variation of the drawn part (redrawn from Colgan, M., Monaghan, J., 2003. Deep drawing process: analysis and experiment. *J. Mater. Process. Technol.* 132, 35–41). The arrows indicate the regions where ductile cracking is more likely.

Region x is in contact with the punch bottom and will form the *bottom* of the cup. The metal is submitted to equi-biaxial stretching ($\beta = 1$) and, as schematized in Fig. 15.13C, undergoes thinning (although rather limited).

Region y will form the cup *bottom radius*. The part close to region x undergoes bending around the punch radius (with a state of stress close to plane strain, as seen in Section 15.3), and then unbending; the part close to region z undergoes bending around the die radius and then unbending.

The metal in region z close to region y will form the *sidewall* of the cap after bending around the die radius and stretching. The part remaining in the region between the die and the blank holder will form the *flange*. In the sidewall cup, the metal is submitted to the punch pulling in plane strain condition ($\beta = 0$), since the circumferential strain is 0 (the sheet assumes the circumference of the punch). The sheet in the flange is submitted to radial pulling and then to a tensile radial strain and a tensile radial stress. In addition, the flange is submitted to a compressive hoop (circumferential) strain (and stress). To explain this, consider a volume element at a distance r_1 from the punch axis. As the process progresses, this distance is reduced, becoming $r_2 < r_1$. The corresponding compressive hoop strain is then: $\epsilon_2 = \ln \left(\frac{2\pi r_2}{2\pi r_1} \right) = \ln \left(\frac{r_2}{r_1} \right)$. Therefore in the flange β is negative (it may also be close to -2 , i.e., in pure shear), and the sheet undergoes thickening. As schematized in Fig. 15.13B, in the flange a compressive stress normal to the sheet is also present, which is caused by the blank-holder force. Its presence is important to prevent sheet *wrinkling* that could be caused by the circumferential compressive stress (because of instability under compression).

It is very difficult to obtain a relationship for the maximum punch force, $F_{\max}$, in deep drawing. In general, empirical relationships are used. One is very similar to that proposed for shearing:

$$F_{\max} = C \cdot s_{\text{UTS}} \cdot b \cdot t \quad (15.7)$$

where t is the original sheet thickness, b is the die diameter, and C is a constant that depends on many parameters, including the metal with which the blank is made (in particular its strain-hardening ability), the D/d ratio (D is the diameter of the blank, and d is the diameter of the punch), the blank holder force, the punch velocity, and the friction between the contact surfaces (that is usually reduced to a minimum adopting a suitable lubricant). For example, for low C (ferritic) steels, C is close to unity, and for Al alloys C is around 0.7.

In the case of stretch forming to obtain a simple spherical cup (see Fig. 15.3B), plastic deformation develops in an equi-biaxial condition, i.e., $\beta = 1$. In general, the deformation paths in stretch forming operations take place in the first quadrant with β values in the range 0–1. To calculate the forming forces in the two principal directions, reference is made to the forces for unit sheet length, defined as:

$$\begin{aligned} T_1 &= \sigma_1 \cdot t \\ T_2 &= \sigma_2 \cdot t \end{aligned} \quad (15.8)$$

where t is the sheet thickness at the end of the deformation process (see Eq. 15.6). To estimate σ_1 and σ_2 , it is first necessary to evaluate ε_1 and ε_2 , considering the geometry of the process. Parameters β and α are then calculated (see Eq. 5.13), and σ_{eff} is obtained from the proper constitutive equation. Finally, σ_1 is obtained using Eq. (5.14) (and, of course: $\sigma_2 = \alpha \cdot \sigma_1$).

Forming limit diagrams for localized necking

Stamped parts may suffer from different defects that must be avoided or properly controlled. In deep drawing, geometrical defects include the *wrinkles* in the flange that are due to compressive hoop stress and may occur especially if the blank holder force pressure is insufficient (an example is shown in Fig. 15.24). Geometrical effects are also the *ears*, that may occur in the outer part of the flange, if the sheet metal is characterized by a large anisotropy, and then tends to deform more in some directions and less in other directions. However, the most important defects that strongly limit the formability of metal sheets are the appearance of regions with excessive thinning due to localized plastic deformation, or the ductile cracks. Localized necking is considered first, whereas the formation of ductile cracks is treated in the next subsection.

As seen in Section 2.8, during tensile deformation of a thin sheet (deformation path with $\beta = -1/2$), uniform necking first occurs (when $\varepsilon_1 = n$). This kind of strain localization can be accepted since it does not impair the surface quality and the appearance of the sheet. However, with increasing plastic strain, the conditions for localized necking are attained. Plastic deformation is then localized and it develops, inside the neck, under plane strain conditions. Therefore localized necking sets a limit to the process since the surface quality of the sheet is impaired and the conditions for ductile fracture are quickly attained. Under tensile deformation, this forming limit is given by $\varepsilon_1 = 2n$, even if in some cases it is anticipated due to the presence of imperfections (Section 2.8). If plastic deformation takes place in plane strain ($\beta = 0$), once the conditions for diffusional necking are attained ($\varepsilon_1 = n$), localized necking immediately starts, since the deformed material is already in plane strain conditions (Marciniak et al., 2002). In this case, the deformation limit is given by $\varepsilon_1 = n$. On this basis, it has been proposed that the forming limit for deformation paths of the second quadrant is given by the following relationship:

$$\varepsilon_1 = \frac{n}{1 + \beta} \quad (15.9)$$

This is schematically shown in Fig. 15.14 for the case $n = 0.2$. The scenario becomes more complex in the case of deformation paths pertaining to the first quadrant. A well-established model indicates that localized necking should not take place for $\beta > 0$. Nevertheless, experimental observations actually show that excessive localized thinning can be reached. It is reported that, at large plastic strains, the accumulated internal damage, due to the nucleation and growth of voids, increases so much as to induce intense localized deformation when approaching the conditions for ductile fracture

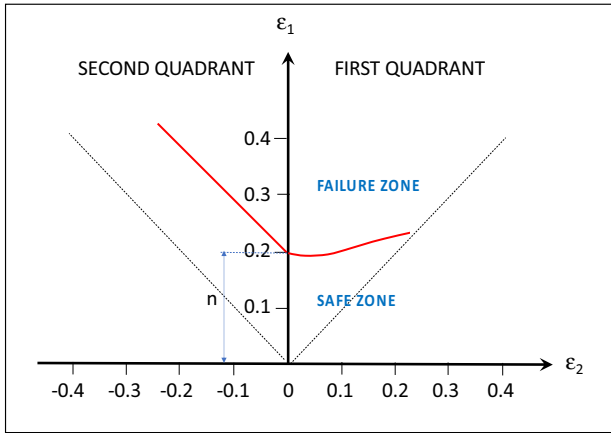


Fig. 15.14 FL curve obtained using Eqs. (15.8) and (15.9), for $n = 0.2$.

(Marciniak et al., 2002). A widely used model to predict the conditions for such localized deformation is the *S-R model* (Storen and Rice, 1975):

$$\varepsilon_1 = \frac{3\beta^2 + n \cdot (2 + \beta)^2}{2(2 + \beta) \cdot (1 + \beta + \beta^2)} \quad (15.10)$$

Eq. (15.10) is also graphically shown in Fig. 15.14, for the case $n = 0.2$. The reported curve is then a model representation of a *forming limit diagram* (FLD). It allows the identification of the *safe zone*, where the deformation paths take place under safe deformation (below the curve), and the *failure zone*, where the deformation paths lead to unacceptable localized deformation (above the curve). This theoretical analysis is a basic framework within which to understand the experimental data.

From an experimental point of view, the FL diagrams can be simply obtained through three tests on sheet specimens: a tensile test in uniaxial conditions (with $\beta = -1/2$); a tensile test in plane strain conditions ($\beta = 0$, Fig. 5.2); and a hydraulic bulge test (with $\beta = 0$, Section 5.7). The FL diagram is then obtained connecting the three experimental points, using the indications of the theoretical model outlined previously. In general, however, a testing procedure allowing for obtaining a greater amount of experimental data is used. In particular, the *Nakazima test* is often adopted. The test is similar to a stretching operation to form cup-shaped parts (see Fig. 15.3B), but different specimens are used with varying widths. The system is well lubricated, and circular grids are printed on the external surface of the specimens. Each test is carried out up to localized necking and the corresponding local minor and major strains are evaluated with the approach displayed in Fig. 15.11 (optical methods coupled with computer-based image analysis are used). Each specimen with different width provides experimental data for different deformation paths. For example, when using large circular or square specimens to produce a full cup, the deformation path is close to equi-biaxial stretching ($\beta = 1$). When using stripes with small width, the

deformation path is closed to uniaxial tension ($\beta = -1$). By changing the width in between, deformation paths between $\beta = -1$ and $\beta = 0$ can be obtained. When testing metals with low strain-hardening ability or when adopting a sub-size punch, the measured strains should be corrected to account for the contribution of the specimen bending around the punch.

As an example, Fig. 15.15 shows the experimental FL diagrams for three alloys, whose representative properties are listed in Table 15.3. The limiting major strain (ϵ_1) for $\beta = 0$ (plane strain path) is often indicated with FLD_0 , and it is expected to be equal to n . However, comparing with the data listed in Table 15.3, it is seen that FLD_0 depends on n as well as on m . Let us first consider the two steels. In passing from the dual phase steel to the IF steel, n increases by 10% whereas FLD_0 increases by 30%; it is clear that the large increase in m also contributed to the increase in FLD_0 . In fact, an increase in m makes strain localization more difficult, as explained in Section 3.3 and schematized in Fig. 3.19. For the Al alloy, coefficient n is much greater than for the two steels, but the FLD_0 value is significantly lower. This can be attributed to the negative m value (due to the DSA phenomenon) and also to the

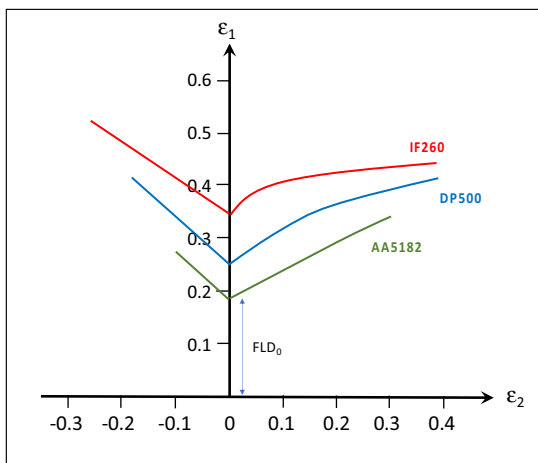


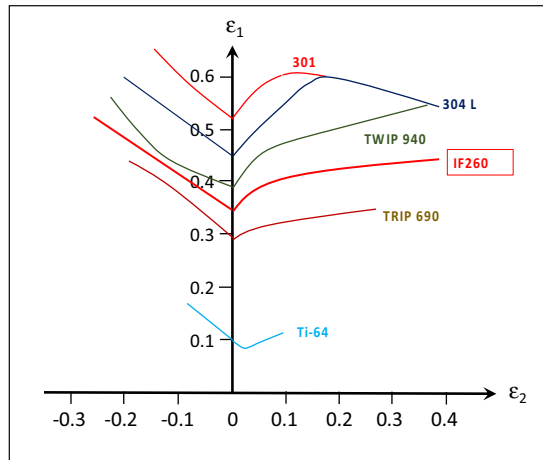
Fig. 15.15 Experimental FL diagrams for three alloys obtained using the Nakazima test (see also Table 15.2). Redrawn from Bortolotti, M., 2002. Curve limite di formabilità di leghe strutturali per impiego nel settore automobilistico, tutor: A. Molinari. Master Thesis. University of Trento.

Table 15.3 Representative properties of the alloys considered in Fig. 15.15.

Metal	n	m	EL%	FLD_0
IF 260 (IF steel)	0.22	0.017	35	0.34
DP500 (dual phase steel)	0.2	0.006	26	0.26
AA 5182 (Al alloy)	0.32	-0.006	32	0.19

Data from Bortolotti, M., 2002. Curve limite di formabilità di leghe strutturali per impiego nel settore automobilistico, tutor: A. Molinari. Master Thesis. University of Trento.

Fig. 15.16 Experimental FL diagrams for different alloys (data taken from: Talyan et al., 1998, for the stainless steels; Chung et al., 2011, for the TWIP steel; Li et al., 2010, for the TRIP steel; Badr et al., 2015, for Ti-64).



softening experienced by the alloy during plastic deformation even at RT, which induces a decrease in n with strain.

Parameter FLD_0 and the FL diagram in the second quadrant mainly depend on the matrix properties of the alloys, such as n and m . The FL diagram in the first quadrant also depends on the presence of particles (such as inclusions or precipitates) that nucleate the voids, i.e., the imperfections that are responsible for the strain localization. For example, it has been shown that an increase in the S content in a steel from 0.001% to 0.29% (with a corresponding increase of the volume fraction of sulfides, from 0.03% to 1.5%) anticipates the conditions for localized necking in biaxial tension ($\beta = 0$, first quadrant), with the minor strain (broadly equal to the major one) decreasing from ~ 0.4 to ~ 0.27 (Yuki et al., 1998).

FLD_0 also depends on experimental parameters; it generally decreases with increasing friction at the sheet-die interfaces, and increases with increasing the sheet thickness (Schey, 1992).

To summarize, the FL diagrams shift upwards, and then the safe zone is extended, when the coefficients n and m increase, and (at least for the first quadrant) the content of engineering particles decreases. Fig. 15.16 displays some additional FL diagrams for different steels and a Ti-64 alloy, and their characteristics can be explained considering these concepts. The curve pertaining to the IF260 steel is included (in red), since it can be considered as a reference curve (of typical low C steels). Better performances are then shown by the reported TWIP and austenitic stainless steels, in particular type 301 stainless steels. For the two stainless steels, the TRIP effect is responsible for the high strain-hardening rates and the corresponding higher forming limits. Note that the more stable 316 L stainless steel displays an FL diagram that is generally below that of 304 L. Of course, the use of these steels in sheet metal forming applications is restricted by their higher strength (and then the higher loads that are involved), compared with common mild or advanced high-strength sheet steels. For the TWIP steel of Fig. 15.16, a similar behavior is observed but, in this case, it

is due to the TWIP effect. The performance of the TWIP steel is lower than that of the stainless steels, despite the very high strain-hardening rate that characterizes these steels. This is due to the fact that these steels typically display negative m values and are particularly sensitive to geometrical imperfections.

In general, advanced high-strength steels (such as dual phase, TRIP, BA, and HSLA steels) display a lower formability than mild steels. Metals with very low n and m values, such as Ti and Mg alloys (see [Chapters 2 and 3](#)), display a very low sheet formability. As an example, in [Fig. 15.16](#) a typical FL diagram for a Ti-64 alloy is included.

Ductile cracks in stamped parts

In most cases, the limit to a stamping operation is given by localized thinning. However, in some circumstances ductile fracture may anticipate strain localization. This may occur during stretching parts with complex geometry ($\beta = 1$), or when using metals that contain a too-high fraction of engineering particles, which significantly reduce ductility. In some cases, ductile fracture takes place immediately after localization (such as in metals with negative m values), and it is not possible to distinguish the two stages (localization/fracture).

As an example, [Fig. 15.17](#) shows the FL diagram for localized necking and ductile fracture, for a 5143 Al alloy ([LeRoy and Embury, 1978](#); reported in: [Hosford and Duncan, 1999](#)). The experimental points referring to the fracture were obtained making reference to imprinted surface circles (or grids) in the necked region using a special procedure (the major strain cannot be measured because of the crack; it is then evaluated from the measured minor strain and the measured change in the sheet thickness, with the usual assumption of the constancy of the volume during plastic deformation). It is seen that fracture generally occurs after localized necking but a critical region exists in the first quadrant, i.e., during typical stretching operations. If the metal ductility is low, ductile fracture may clearly anticipate the occurrence of

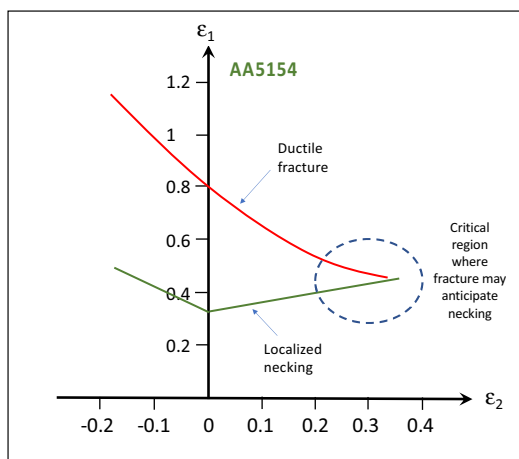


Fig. 15.17 FL diagrams for localized necking and fracture for a 5154 Al alloy ([LeRoy and Embury, 1978](#)). Redrawn from Hosford, W.P., Duncan, J.L., 1999. Sheet metal forming: a review. JOM, November.

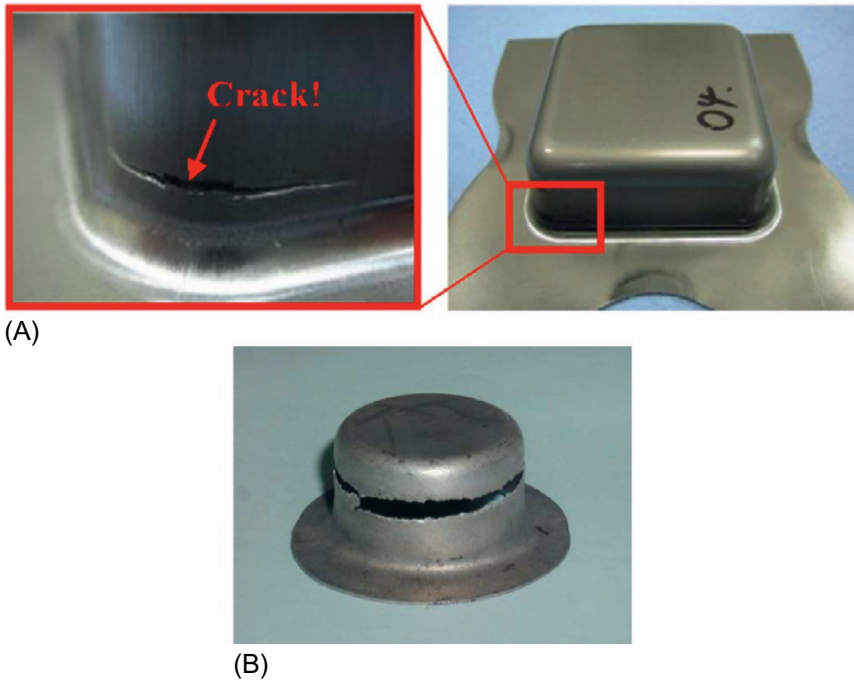


Fig. 15.18 Examples of ductile cracks in drawn parts. (A) Cup made with a TRIP steel and drawn using a square punch (Li et al., 2010); (B) cup made with a Mg alloy and deformed at 300°C (reproduced from Yoshihara, S., Manabe, K., Nishimura, H., 2005. Effect of blank holder force control in deep-drawing process of magnesium alloy sheet. *J. Mater. Process. Technol.* 170, 579–585, with permission of the Publisher).

localized necking. Another critical region exists in the second quadrant (it is not indicated in the figure), in particular near the deformation paths close to pure shear conditions ($\beta = -1$). As seen in Chapter 6, some metals (including specific Al, Ti, and Mg alloys—characterized by low n values, low or negative m values, and containing imperfections) are prone to anticipated fracture by localized shearing and are then characterized by low values of the effective fracture strain (see Fig. 6.15B). Therefore, during plastic deformation, localized necking is not attained (the deformation path with $\beta = -1$ is almost parallel to the necking locus) and fracture by shear may then take place at critical conditions (i.e., when $\epsilon_{\text{eff}} = \epsilon_{\text{eff},f}$).

Two examples of ductile cracks in deep drawn parts are shown in Fig. 15.18. In Fig. 15.18A a ductile crack in a cup obtained using a square punch is shown (Li et al., 2010). The sheet metal is made with a TRIP 690 steel (an advanced high-strength steel for the automotive industry; see also Fig. 15.16). The crack is located close to the die region, where sheet bending plays an important role, and then propagates circumferentially. In general, this is the most critical region for localized necking and cracking in deep drawing parts and therefore the punch radius plays an

overwhelming role, since it determines the local strains (see Eq. 15.3). Fig. 15.18B shows a ductile crack in a drawn cup made with a Mg alloy (Yoshihara et al., 2005). The process was carried out at 300°C (i.e., in hot conditions), since at RT the metal ductility is negligible. The ductile crack nucleated at the bottom of the side-wall region, where the plastic deformation in plane strain conditions is maximum. In both cases, ductile fracture took place rather early after localized necking.

Ductile cracks are nucleated where the local effective strain reaches the critical effective plastic strain for ductile fracture. The regions where the onset of ductile fracture is more likely in deep drawing are displayed by the arrows in Fig. 15.13C. In stretch forming, ductile cracks mainly form near the dome region (see also the cover page of this chapter).

The data in Fig. 15.17 show that the major strain at fracture decreases with increasing β , i.e., with deformation paths characterized by increasing triaxiality, T (Chapter 5). In Fig. 15.19A, the experimental data on 11 different alloys were collected (Wierbicki et al., 2005). The plot is normalized with respect to the ductile fracture strain in plane strain conditions, i.e.:

$$\sim \varepsilon_1 = \varepsilon_1/\varepsilon_f^* \quad \text{and} \quad \sim \varepsilon_2 = \varepsilon_2/\varepsilon_f^*$$

where ε_f^* is the ductile fracture strain in plane strain conditions given by: $\sqrt{3}/2 \cdot \varepsilon_f$ (where ε_f is the tensile true strain at fracture).

The experimental data display the same trend as the fracture data in Fig. 15.17. Using the R-T criterion (Eq. 6.17), it is easy to obtain the fracture curve to plot in the deformation diagram, with reference to the normalized fracture strains. This is shown in Fig. 15.19B. It can be clearly seen that the experimental data (A) follow the same trend predicted by the R-T analysis (B). The observed experimental scatter can be attributed to different factors, such as variable materials properties and testing parameters. It should also be noted that the deformation path is not truly proportional up to fracture, as assumed in the model. In fact, the deformation paths during sheet forming operations are proportional up to localized necking. But after localized necking, plastic deformation proceeds (in the neck) under plane strain conditions. As a consequence, the major strain continuously increases, whereas the minor strain does not increase any more.

The obtention of the FL diagrams, for localized necking or fracture, is often a tedious and long procedure. Because of this, simplified tests are often carried out, including the *Swift cup test* (to simulate deep drawing) and the *Limiting Dome Height test* (LDH, to simulate stretch forming operations) (Bruschi et al., 2014).

The Swift cup test simply involves the drawing of circular blanks with different diameters under the action of a punch with a flat bottom (Swift, 1952). In the test, the geometrical conditions are fixed: the punch diameter is 50 mm, the die hole diameter is 52.5 mm, the radius of both punch and die is 6.36 mm. The formability of the sheet metal is then given by the ratio of the maximum blank diameter drawn without sheet cracking and the punch diameter. Such a ratio is called the *Limiting Drawing Ratio* (LDR). Considering first the role of material properties, the experimental LDR values are found to increase with the metal ductility and also with the Lankford

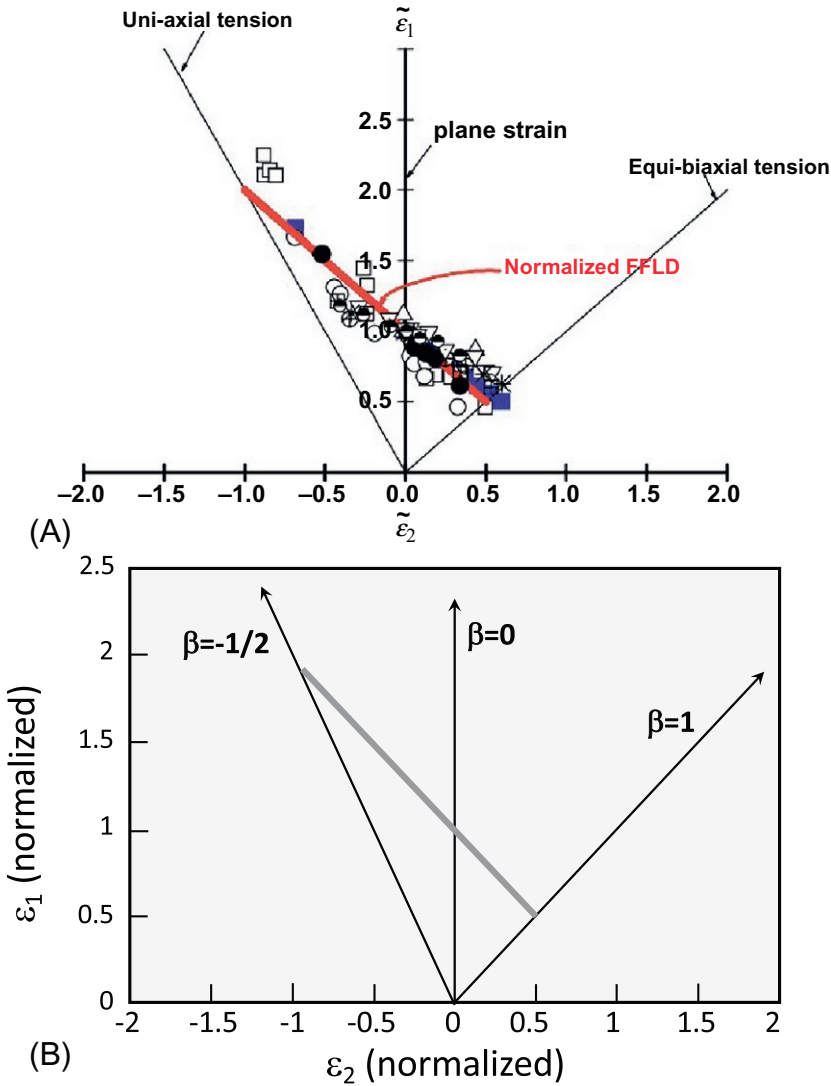


Fig. 15.19 Normalized deformation diagram showing (A) experimental data for 11 sheet metals (reproduced from Wierbicki, T., Bao, Y. Lee, Y.W., Bai, Y., 2005. Calibration and evaluation of seven fracture models. *Int. J. Mech. Sci.* 47, 719–743, with permission of the Publisher), and (B) the fracture locus predicted by the R-T criterion.

parameter, R , introduced in Section 2.6. As already mentioned, with $R > 1$ the thickness reduction during plastic deformation is less pronounced and this is very beneficial since the critical conditions for strain localization (and then for ductile fracture) are mostly attained under plane strain conditions and therefore after large thinning. Both ductility and anisotropy then play an important role and must be

considered together. Rather typical is the case of Mg alloys that are characterized by high R values at RT, typically in the range of 4–5 as seen in [Section 2.6](#), but their drawability is very low, because of their very low ductility. The same holds true for Ti alloys that also have an hcp structure like Mg alloys, and display high R values, but their drawability is markedly limited by their low ductility (it is high for pure Ti only).

The limiting dome test simply involves a stretch test ([Fig. 15.3B](#)) using a hemispherical punch with a large diameter (100 mm) ([Ayres et al., 1979](#)). Specimens with different widths are used and the height of the dome at fracture is measured. The obtained dome heights are then plotted as a function of the specimen width, and a minimum is generally recorded, which corresponds to a deformation path close to plane strain. This minimum value is the limiting dome height (LDH₀).

Sheet formability improvements

As seen, some alloys, including Al, Ti, and Mg alloys, display a rather poor sheet formability. Much research is being devoted to finding the best forming conditions to produce parts, to exploit the engineering advantages of these alloys. Specific efforts are also aimed at further increasing the sheet formability of metals such as steels, for producing parts with complex geometry without any failures.

Different strategies have been investigated and are here summarized, considering the relevant concepts introduced in the first chapters of this text:

- Cryogenic forming;
- High-velocity forming;
- Hot stamping;
- Superplastic forming (SPF).

As seen in [Chapters 3](#) and [7](#), fcc alloys with high-moderate SFE, including Al and Cu alloys, display an increase in n and in ductility as temperature is decreased below RT, and this phenomenon can be used in stamping as an advantage. Testing and stamping operations at cryogenic temperature require the design of a proper cooling system able to maintain a uniform temperature distribution in the sheet, and the use of suitable lubricants. [Fig. 15.20](#) shows the FL diagrams for a 5182 Al alloy ([Sotirov et al., 2015](#)). In [Fig. 15.20A](#) experimental curves are shown, obtained at RT, -50°C and -100°C . It is seen that the safe zone shifts upwards with the decrease in temperature. In [Fig. 15.20B](#), simulated curves (obtained using a numerical method and employing experimental data obtained with low-temperature tensile tests) are shown, for temperatures down to -196°C . The results are rather promising. Specific deep drawing tests at cryogenic temperatures were also carried out, aimed at obtaining the Limiting Drawing Ratio (LDR) for a 2219 Al alloy ([Yuan et al., 2020](#)). The results are shown in [Fig. 15.21](#). The LDR was found to increase from 1.8 at RT to 2.08 at -160°C (the tensile fracture strain at -160°C was found to be 33.3% higher than at RT).

In [Chapter 3](#) we have also seen that in passing to very high strain rates (beyond 10^3 s^{-1}), both coefficients n and m increase considerably. Consequently, strain localization is made more difficult and ductility increases markedly ([Section 8.2](#)). At the

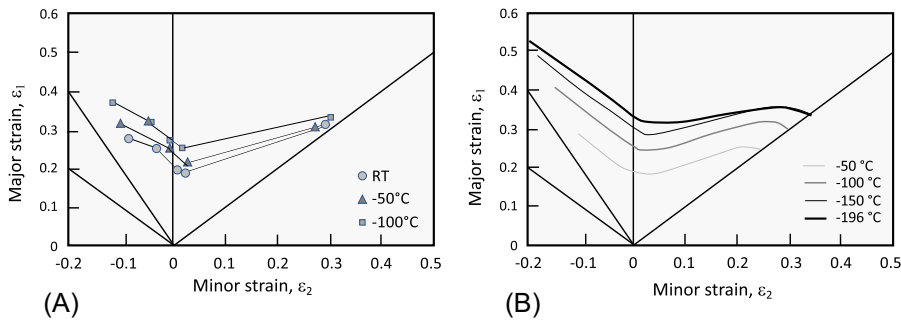


Fig. 15.20 FL diagrams for a 5182 Al alloy. (A) Experimental tests at RT, -50°C and -100°C ; (B) numerical estimations. Redrawn from Sotirov, N., Falkinger, G., Grabner, F., Schmid, G., Schneider, R., Grant, R., Kelsh, R., Radlmayr, K., Scheerer, M. Reichl, C., Sehrschoon, H., Loipetsberger, M., 2015. Improved formability of AA5182 aluminum alloy sheet at cryogenic temperatures. Mater. Today: Proc. 2S, S113–S118.

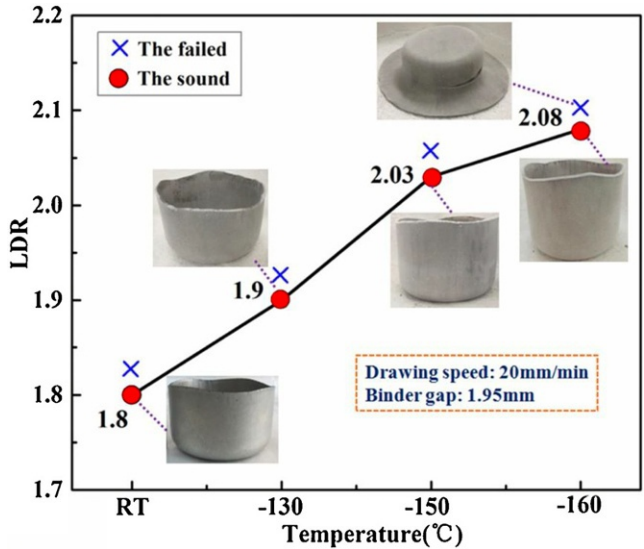


Fig. 15.21 Experimental LDL values for a full annealed AA2219 alloy between RT and -160°C . Reproduced from Yuan, S., Cheng, W., Liu, W., Xu, Y., 2020. A novel deep drawing process for aluminum alloy sheets at cryogenic temperatures. J. Mater. Process. Technol. 284, 116743, with permission of the Publisher.

same time, inertial effects increase the metal ductility since they help in spreading deformation throughout the sample, postponing the conditions for strain localization. On this basis, the process of *high-velocity metal forming* has been proposed. It is similar to that described in Fig. 15.4B, and the high-velocity deformation may be obtained through a shock wave produced by explosives, or through the energy delivered by

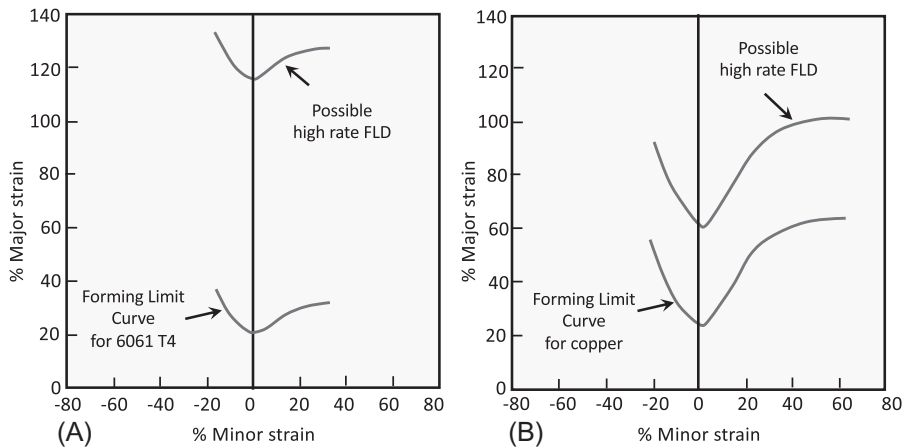


Fig. 15.22 Forming limit curves for a 6061-T4 Al alloy (A) and pure Cu (B). The experimental curves were obtained both in static conditions and at high-velocity deformation. The reported strains are engineering strains.

Redrawn from Balenethirman, V.S., Daehn, G.S., 1994. Hyperplasticity: increased forming limits at high workpiece velocity. *Scr. Metall. Mater.* 30, 515–520.

electromagnetic or electrohydraulic equipment (Daehn et al., 1995). The improvement in formability is not restricted to Al alloys, as demonstrated by the FL curves shown in Fig. 15.22 and relevant to a 6061 Al alloy and an annealed and oxygen-free Cu (Balenethirman and Daehn, 1994).

Finally, in Chapter 4, we have seen that for temperatures greater than $0.4\text{--}0.5T_m$, the strain rate sensitivity coefficient, m , markedly increases with temperature (whereas the strain-hardening coefficient, n , decreases, and it is equal to 0 at the steady-state, i.e., in the secondary creep stage). The strong increase in m brings about a noticeable increase in hot ductility. As an example, Fig. 15.23 shows the FL diagrams at high temperatures for a 5083 Al alloy (Bariani et al., 2013), and a 316L stainless steel (Dharavath et al., 2020). In both cases, the safe zones shift upwards with increasing temperature.

Additional advantages of hot stamping operations are a marked reduction of the forming loads and of the springback effect, due to the decrease in the flow strength. However, hot stamping also has some disadvantages:

- The process is more difficult (because of the heating system) and surface oxidation processes must be controlled;
- The permanence at high temperature may induce damaging microstructural transformations; for example, in Al alloys, undesired transformations, such as recrystallization, precipitation, or excessive grain growth, can easily occur;
- Excessive thinning may occur during the diffusion necking (because of the low n values).

Some disadvantages of hot forming are relieved using warm forming processes, in the temperature range of $0.35\text{--}0.55T_m$. Today, hot stamping is mainly used to form steel sheets (mainly boron steels), taking advantage of the metallurgical possibility of

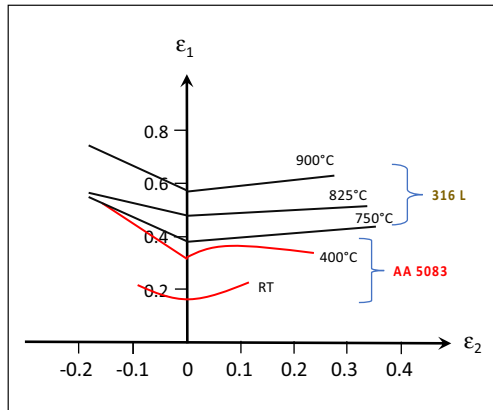


Fig. 15.23 FL diagrams at high temperatures for a 5083 Al alloy (data from Bariani, P.F., Bruschi, S., Ghiotti, A., Michieletto, F., 2013. Hot stamping of AA5083 aluminium alloy sheets. CIRP Ann. Manuf. Technol. 62, 251–254), and a 316L stainless steel (data from Dharavath, B., Morchhale, A., Singh, S.K., Kotkunde, N., Naik, M.T., 2020. Experimental determination and theoretical prediction of limiting strains for ASS 316L at hot forming conditions. J. Mater. Eng. Perform. 29, 4766–4778). For the 316L steel, the tests were carried out using a sub-sized punch (with diameter: 50 mm) and the data were then corrected for bending.

directly heat treating the part after high-temperature shaping (boron significantly increases steel hardenability) (Merklein et al., 2016).

A special case of hot stamping is *superplastic forming* (SPF), in which the superplastic behavior of some metals, such as Ti and also Al alloys, is exploited (Section 8.6). SPS is typically used in stretch forming operations, like that shown in Fig. 15.4B. The preheated sheet is placed on a preheated forming die with the desired geometry and fixed to guarantee deformation by stretching. Under the action of a fluid pressure, the sheet is plastically deformed against the forming die. The process is carried out at a fixed temperature and also at a given average strain rate by setting the proper pressure of the fluid. Considering the comparatively low strain rates that are required (in the range of 10^{-4} and 10^{-3} s^{-1}), the process can be quite long (several minutes) and also expensive. In addition, it requires the preliminary production of metal sheets with a sufficiently small grain size.

The sheet formability can also be improved by optimizing specific processing conditions. For example, deep drawability can be improved by adopting a multi-stage process (see, e.g., Kalpakjian and Schmid, 2009), or by controlling the blank holder force (BHF). As seen, a proper blank holder pressure is required to avoid wrinkling. However, if BHF is too high, tensile stresses in the sample are increased (stretching is favored) and LDR is decreased. But if BHF is variable and in particular it is lower at the beginning (when the compressive hoop stresses are still comparatively low), hence favoring the drawing process (i.e., more material is drawn into the die cavity), and it is increased in the second half of the process to better counteract wrinkling, the drawability can be markedly increased. As an example, Fig. 15.24 shows the results of a deep drawing test using rectangular blanks (in AA574-0 Al alloy) and a shaped

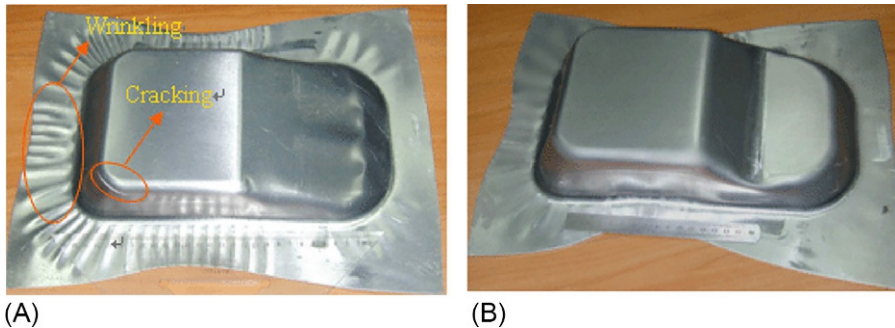


Fig. 15.24 Comparison of deep drawn parts obtained with (A) constant BHF and (B) optimized variable BHF.

Reproduced from Zhing-quin, L., Wu-ring, W., Guan-long, C., 2007. A new strategy to optimize variable blank holder force towards improving the forming limits of aluminum sheet metal forming. *J. Mater. Process. Technol.* 183, 339–346, with permission of the Publisher.

punch (Zhing-quin et al., 2007). In (A) the process was carried out with a constant BHF, whereas in (B) an optimal variable BHF was adopted.

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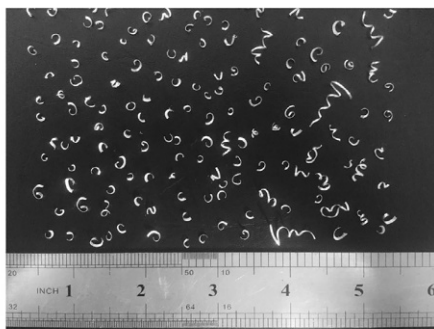
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Machining operations by chip removal

16



(A) Tool without chip breaker



(B) Tool with chip breaking dot

Continuous and discontinuous broken chips obtained by turning using different HSS tools. Reproduced from Shi, C., Yu, A., Wu, J., Niu, W., He, Y., Hong, X., Shang, Q., 2017. Study on position of laser cladded chip breaking dot on rake face of HSS turning tool. *Int. J. Mach. Tools Manuf.* 122, 132–148, with permission of the Publisher.

Chapter outline

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Machining operations by gradual metal removal with a wedge-shape tool allow simple or complex flat and prismatic shapes to be obtained with high precision and dimensional tolerances. The most important operations are:

- *Turning on a lathe, to produce parts mainly with axi-symmetric geometry;*
- *Milling, to produce flat surfaces and surfaces of complex geometry;*
- *Drilling (boring, reaming), to produce holes in parts.*

Grinding operations are finishing operations similar to milling. They use an abrasive wheel containing hard particles that operate like a multi-edge cutting tool. These operations are not considered in this chapter.

The machinability of a metal is defined by the ease with which it is machined. From an engineering viewpoint, this is typically given by the maximum productivity at the optimized cutting velocity. It is then mainly determined by two parameters:

1. The cutting force, F_c , and the required cutting power (given by $F_c v$, where v is the cutting velocity);
2. The tool life, t , which is strongly related to the cutting velocity; typically, $t = C/v^m$, where C and m are two experimental constants, depending on the workpiece and tool materials, and the operating parameters.

In the case of machining metals with sufficiently high ductility, a continuous chip is formed. However, when the workpiece ductility is low, or in special machining conditions, a discontinuous chip may form, due to the ductile fracture of the chip under formation. The formation of a broken chip has several benefits. First, it is easily removed from the process equipment, whereas a discontinuous chip is not. This positively affects the operator safety and avoids a deterioration of the workpiece surface quality that may occur when the continuous chip wraps around the workpiece. Secondly, with a discontinuous chip there is a reduction of the chip-tool contact length, a reduction of the average cutting force, and the cutting fluid is allowed to enter the cutting zone. All this is very beneficial with regard to the tool wear, which is much reduced, and determines an increase in the machinability.

In this chapter the main processes are first introduced and the basic concepts of metal cutting are presented with reference to orthogonal cutting. Subsequently, the characteristics of the chip formation are considered, with particular emphasis on the role of the workpiece ductility.

16.1 Main processes

Fig. 16.1A shows a schematization of a turning operation that is the most widespread machining process. In turning, the *primary (cutting) motion* is provided by the rotating workpiece, whereas the *feed motion* is given by the tool. The average cutting velocity, v , is given by the following relationship:

$$v = n \cdot \pi \cdot \bar{D} \quad (16.1)$$

where:

- n is the rotational velocity of the workpiece (usually in revolutions per minute);
- $\bar{D}$ is the average between the diameter of the workpiece and of the turned bar.

Note that v is typically in the range 10–1000 m/min. The average cutting velocity directly determines the productivity, expressed by the *volume material removal rate*, MRR:

$$\text{MRR} = v \cdot b \cdot a_1 \quad (16.2)$$

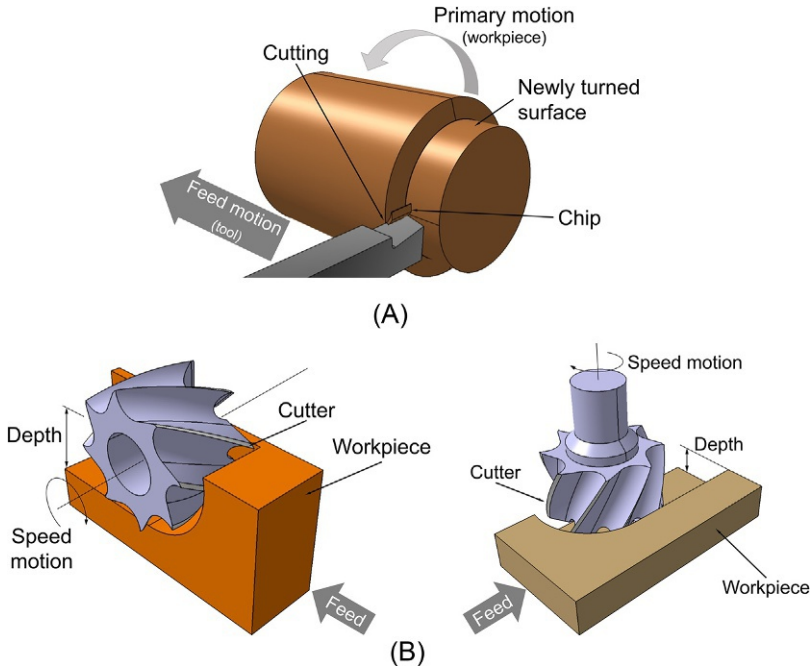


Fig. 16.1 (A) Scheme of a turning on a lathe operation; (B) schematizations showing the process of peripheral and face milling.

where:

- b is the width of the chip;
- a_1 is the thickness of the chip before cut (see also the next section).

In milling operations (Fig. 16.1B), a multi-tooth cutter is used, which produces several chips in one revolution. In this case, the primary motion is provided by the rotating tool, whereas the feed motion is given by the workpiece. Indicating the workpiece velocity (feed motion) with v_a , the MRR in peripheral milling is simply given by:

$$\text{MRR} = v_a \cdot B \cdot a_1 \quad (16.3)$$

where B is the width of the cut.

In drilling, a rotating tool called a twist drill provides both the cutting (rotating) and feed (vertical) motion.

Because of the sliding at the tool/chip and tool/fresh surface interfaces, the tool life is limited by its wear. When wear is too much, the tool must be replaced. The frequency of this replacement affects the cost of the process. To reduce the tool wear and increase its life, specific materials are employed. High-speed steels and hard metals are commonly used, even after coating with ceramic thin films. Ceramics, such as alumina and silicon nitrides, are used in turning operations because of their high wear resistance even at high tool temperature. Cutting fluids (emulsions of oils and

water) are commonly adopted, mainly to reduce the local temperature rise in the tool/chip interface, due to the heat dissipated by the plastic deformation (which occurs in adiabatic conditions because of the high local strain rates), and by the friction at the sliding interfaces. Typical temperature increases are in the range of 200–1100°C.

The adhesive interaction at the tool/chip interface may also produce some material transfer from the chip onto the rake face of the tool. This transfer is favored by a low sliding velocity, which favors the adhesive interactions. A built-up edge (BUE) is then formed, which increases the tool wear and reduces the finish of the newly formed surface. This phenomenon is avoided in different ways, for example by increasing the cutting velocity, reducing friction and favoring the formation of a discontinuous broken chip (Section 16.4).

Productivity is related to the MRR and increases with v (or v_a). However, if velocity is too high, productivity decreases, because of excessive tool wear that requires a too-frequent change of the cutting tool. Therefore v is a key process parameter that must be carefully optimized.

Recent developments in machining operations involve (Kalpakjian and Schmid, 2009):

- Cutting in dry condition or with a limited coolant supply (to avoid environmental problems connected with the disposal of the cutting fluid);
- Cutting of metals in their hardened state (such as *hard turning*) using tools made with advanced ceramics such as cBN (to avoid subsequent finishing processes by grinding and to reduce process cost);
- High-speed machining, with cutting velocities in excess of 1500 m/min (to increase the productivity).

16.2 Orthogonal cutting

The analysis of the cutting process, even for experimental research purposes, is usually carried out in the two-dimensional or *orthogonal cutting* condition. An example of orthogonal cutting is shown in Fig. 16.2. In this process, the cutting edge producing the chip is orthogonal to the cutting direction. In most cases the cutting process is

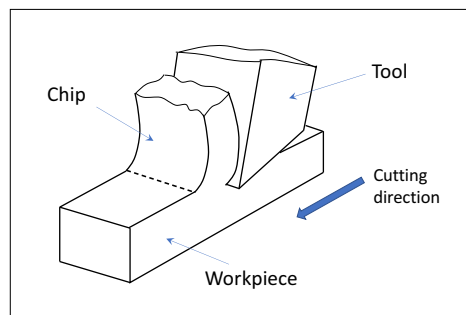


Fig. 16.2 Example of orthogonal cutting in which the cutting edge producing the chip is orthogonal to the cutting velocity vector.

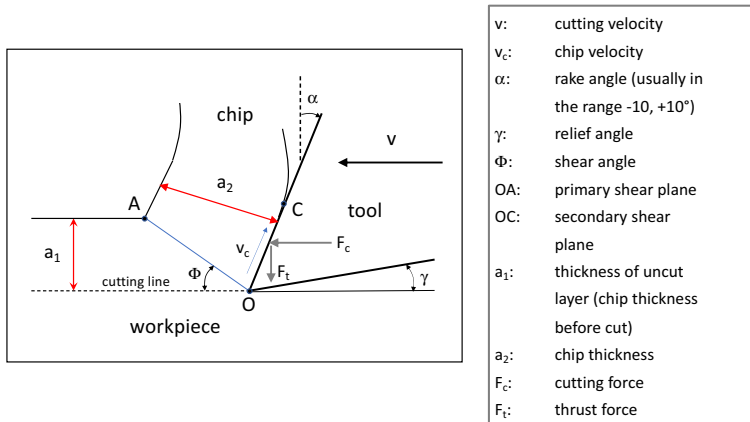


Fig. 16.3 Schematic diagram of the tool-chip region with reference to orthogonal cutting conditions. O, tool tip; A, chip root.

three-dimensional (*oblique cutting*). However, the information obtained from simple orthogonal machining investigations has a rather general validity and is very useful in understanding the cutting process by metal removal (Shaw, 1984).

Fig. 16.3 indicates the main parameters involved. Figs. 16.2 and 16.3 show a *continuous chip* formation, but the chip can also have a *discontinuous* geometry, as is seen in the next sections.

The formation of the chip is basically a two-step process, comprising:

- Ductile fracture along the *cutting line* (see Fig. 16.3) with the formation of two new surfaces: the surface of the chip initially in contact with the rake face of the tool, and the fresh surface of the workpiece;
- Plastic deformation and curvature of the chip for its proper evacuation from the cutting region.

The fracture along the cutting line is induced by the continuous formation, by ductile fracture, of a short crack in the workpiece, right in front of the tool tip (Dodd and Bai, 1987). The short crack is formed rather easily, due to the very large tensile strain occurring very close to the tool tip caused by the wedge (opening) action of the advancing tool, and also the local stretching induced by friction, as seen later in this section. However, most of the energy required for the process (and, correspondingly, the largest contribution to the cutting force) is not required by the formation of the two new surfaces but by the large plastic deformation involved in the curvature of the chip, and the sliding friction at the chip-tool (rake face) interface.

As an example, Fig. 16.4 shows the sequence of the formation of a continuous chip in the case of orthogonal cutting of an annealed OFHC Cu, with $\alpha = 0$, $a_1 = 50 \mu\text{m}$, and $v = 1 \text{ mm/s}$ (Viswanathan et al., 2017). It was obtained by in situ observations using a high-speed camera/optical microscope system. A single grain on the workpiece surface is colored in green, and the sequence shows its deformation behavior (the grain is representative of a surface volume material). When approaching the tool,

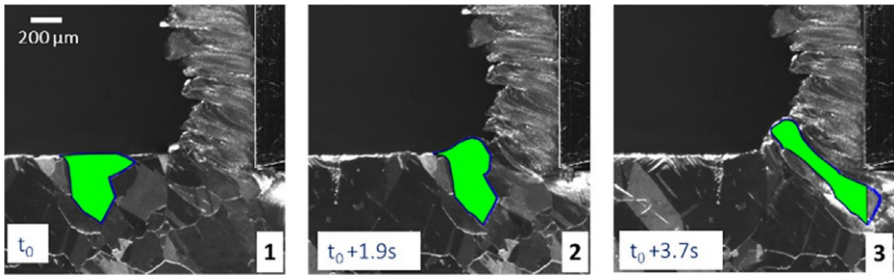


Fig. 16.4 Time sequence showing the formation of a continuous chip in the case of annealed OFHC Cu, with $\alpha = 0$, $a_1 = 50 \mu\text{m}$, $v = 1 \text{ mm/s}$. Reproduced from Viswanathan, K., Udupa, A., Yeung, H., Sagapuram, D., Mann, J.B., Saei, M., Chandrasekar, S., 2017. On the stability of plastic flow in cutting of metals. CIRP Ann. Manuf. Technol. 66, 69–72, with permission of the Publisher.

it is submitted to compression and then forms a bump on the free surface. Subsequently the grain rotates, and is deformed mainly by shear. The sequence is repeated, and the chip formation is then a cyclical process; only in particular conditions, the chip forms by uniform plastic flow.

Another example is shown in Fig. 16.5. In this figure, the microstructure of a cross section through a chip of a 6061 Al alloy is shown (Zhang and Alpas, 2002). It was obtained with an interrupted test, carried out in orthogonal cutting ($\alpha = -5$ degrees, $a_1 = 200 \mu\text{m}$, $v = 0.6 \text{ m/s}$; feed rate: 0.3 mm/rev ; dry conditions, i.e., no lubricant was employed). The figure shows that the plastic deformation in the region between the tool tip and the chip root is not characterized by a single value of Φ , as simply schematized in Fig. 16.3. This can be appreciated by considering that the undeformed parent metal displays vertical microstructural markers resulting from the previous extrusion process, and that such lines are differently inclined in the direction of plastic flow, depending on the position between the tool tip and the chip root. In particular, for metal close to the tool tip $\Phi \sim 32^\circ$, whereas at the chip root $\Phi \sim 14^\circ$. The estimated average effective strains (by shear), were approximately 2.5 close to the tool tip, and were found to decrease, moving along the *primary shear region*, to approximately 0.8 close to the chip root. This is actually a general trend.

A close inspection of the chip metal close to the tool in Fig. 16.5 also reveals the significant effect of friction that introduces an additional shear deformation region along the chip/tool interface. To sum up, plastic deformation involved in the formation of the chip is caused by three main contributions:

1. Plastic shear along the *primary shear region*, extending from the tool tip to the chip root (with triaxiality factor, T , close to 0);
2. Plastic shear due to friction at the chip/rake face interface (*secondary shear region*, with triaxiality factor, T , close to 0);
3. Plastic compression at the entrance of the primary shear region (with rather negative triaxiality factor, because of the constrained deformation).

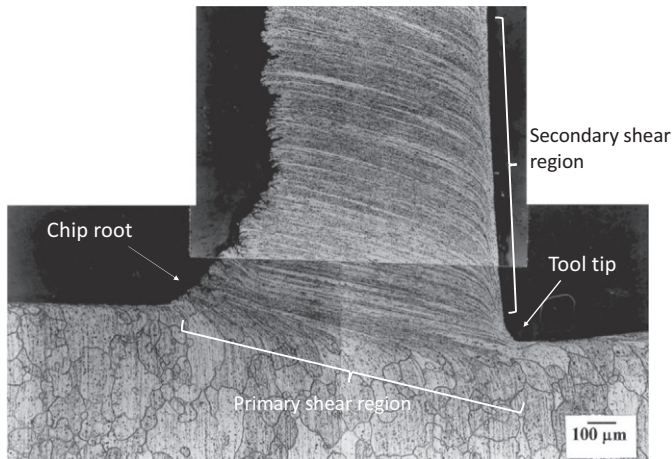


Fig. 16.5 Microstructure of a cross section through a cut chip of 6061 Al alloy, showing its formation. Note that $a_1 \sim 200 \mu\text{m}$, and $a_2 \sim 800 \mu\text{m}$. Reproduced from Zhang, H., Alpas, A.T., 2002. Quantitative evaluation of plastic strain gradients generated during orthogonal cutting of an aluminum alloy. *Mater. Sci. Eng. A* 332, 249–254, with permission of the Publisher.

In addition, a small region with tensile deformation (with triaxiality factor, T , greater than ~ 0.3) is present very close to the tool tip. It is responsible for the formation of the short crack leading to the generation of the chip, as already mentioned. A thin region with positive T values is also present close to the free surface, at a certain distance from the chip root. In Fig. 16.6 the main characteristics of the process zone are shown, and the typical values of the corresponding triaxiality factors, T , are indicated (Abushawashi, 2013; Mamalis et al., 2001).

In this respect, we have to note that the stress and strain distribution in the process zone is very complex and significantly influenced by the tool geometry, the workpiece material properties, and the cutting parameters. A very important role is played by the rake angle, α . In fact, an increase in the rake angle increases the extension of the deformation by compression (with $T < 0$) in the region between the tool tip and the chip root

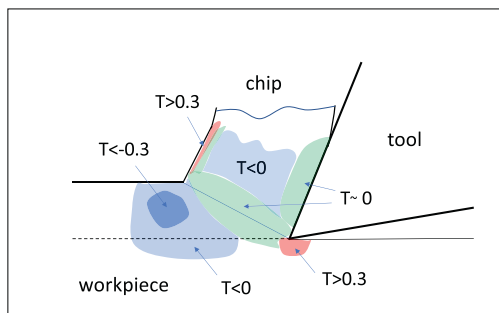


Fig. 16.6 Schematization showing the process zone; in particular, the different regions characterized by different T values are highlighted.

at the expense of the shear. At the same time, the tensile region (with $T > 0.3$) close to the tool tip increases in size, moving into the interior of the workpiece.

An important parameter that is used to characterize the intensity of plastic deformation in the process zone is the so-called *cutting ratio*, r , defined by the ratio of the chip thickness before and after the cut: a_1/a_2 . Parameter r can be easily obtained experimentally. It decreases as the tool rake angle is decreased, and the friction with the rake face of the tool increases; in fact, as α is decreased and friction is increased, an increased metal pile up against the tool rake face is observed.

Under the assumption of chip formation without lateral spreading perpendicularly to the cutting direction, parameter r can be defined in the following way:

$$r = \frac{a_1}{a_2} = \frac{L_2}{L_1} = \frac{v_c}{v} \quad (16.4)$$

where:

- L_1 and L_2 are the chip lengths before and after cut;
- v and v_c are the cutting and chip velocity.

Parameter r typically varies in the range of 0.3–0.4. Small r values mean large chip thickening and hence increased plastic deformations in the process zone. They are typically encountered in soft annealed metals: for example, in pure Al, Cu, and α -brass, r ranges between 0.05 and 0.07 (Viswanathan et al., 2017).

Using a very simplified model that considers plastic deformation in the process zone occurring by shear only (with the presence of a single Φ value, as shown in Fig. 16.3), the effective shear strain can be expressed by the following relationship (Childs et al., 2000):

$$\epsilon_{\text{eff}} \approx \frac{0.5 \div 0.75}{r} \quad (16.5)$$

16.3 Mechanics of orthogonal cutting

Fig. 16.3 shows the two main forces acting on the tool and that the tool must provide for the cutting process. The *cutting force*, F_c , is applied (by an external motor) in the cutting direction. The motor must then supply a power given by: $F_c v$. It is mainly due to the deformation and fracture processes responsible for the formation of the chip. The *thrust force*, F_t , must be applied to maintain the tool in its position, and it is mainly due to the sliding friction at the chip-tool interface.

The corresponding *specific cutting energy* required for the formation of the chip is usually indicated with u . It is defined by the ratio of the energy spent in the formation of the chip and the volume of involved material. Therefore it is given by the ratio of the cutting power and the volume rate of material removal (MRR):

$$u = \frac{F_c \cdot v}{v \cdot b \cdot a_1} = \frac{F_c}{b \cdot a_1} \quad (16.6)$$

Table 16.1 Representative data of the specific cutting energy (in MJ/m³).

Metal	Specific cutting energy (MJ/m ³)	Metal	Specific cutting energy (MJ/m ³)
Al alloys	400–1000	Ti alloys	2500–5000
Cu alloys	1400–3300	Steels	1500–9000
Mg alloys	300–600	Stainless steels	2000–5000

Data from Kalpakjian, S., Schmid, S.R., 2009. Manufacturing Engineering and Technology. Prentice Hall, and other sources.

Typical experimental values of u (obtained by measuring the relevant parameters in specific tests) are listed in Table 16.1.

As already mentioned, most of the energy required for the formation of the chip is spent in the plastic deformation in the process zone and therefore u is proportional to the metal hardness, H (to appreciate this, consider that Vickers, or Brinell, hardness is given in kg/mm² that is approximately 0.1 MPa). Since hardness values are obtained at RT and in quasi-static conditions (and also in a constrained compressive deformation), this observation shows that the local temperature rise above RT, which induces a decrease in u , is roughly counterbalanced by the increase in the local strain rate, which induces an increase in u . The contribution of friction at the chip/rake face interface is approximately one-fourth of the total (Shaw, 1984), even if in dry cutting of rather soft metals the contribution of friction may be prevailing (Elmadagli and Alpas, 2003). Of course, a specific energy requirement is also due to the formation of the two new fresh surfaces, even if it is in general a small fraction of the total (Atkins, 2003).

In general, u experimentally depends on the rake angle, α , and it has been found to decrease by $\sim 1\%$ per degree increase. In addition, it has also been reported that u is size dependent. In particular the u values increase significantly with approaching rather low uncut chip thickness values (a_1). In fact, when a_1 becomes very small, the volume of material involved in the process also becomes very small and u is then mainly given by the energy required for the formation of the two fresh surfaces by ductile fracture. This energy is linked to the metal fracture toughness and does not depend on a_1 . Therefore, when this lower limit is approached, a further reduction of a_1 (i.e., of the volume of material involved in the process zone) is accompanied by an increase in u .

Eq. (16.6) is very useful for estimating the cutting force:

$$F_c = u \cdot b \cdot a_1 \quad (16.7)$$

Therefore using experimental u values it is possible to verify the capability of the adopted equipment to carry out the process. In this respect, metals with low u values are characterized by a better machinability. However, other parameters play an important role, as discussed in the next sections. As mentioned, the thrust force, F_t , is mainly determined by the sliding friction at the chip/tool interface. Considering that the friction coefficient is typically close to 0.5, we have: $F_t \sim 0.5 F_c$ (Shaw, 1984).

The average temperature rise in the chip under formation can be evaluated using Eq. (3.12) considering that the specific cutting energy, u , is dissipated into heat in an almost adiabatic manner (the local strain rates typically range from 10^2 to 10^5 s^{-1}). A comparison with experimental data revealed that Eq. (3.12) should be better modified in the following way (Trent, 1991):

$$\Delta T = 0.4 \cdot \frac{u}{\rho \cdot c} \cdot \left(\frac{v \cdot a_1}{a} \right)^{0.33} \quad (16.8)$$

where v is expressed in meters per second, a_1 in meters, and the thermal diffusivity of the workpiece, a , is expressed in square meters per second. The tool wear is influenced by the cutting force F_c , and therefore by u , and also by the local temperature rise ΔT . This explains why the cutting velocity plays a leading role in determining the tool life. If the temperature rise is comparatively low, the process is carried out in cold conditions and the chip is correspondingly strain hardened.

16.4 Chip formation

Continuous chip

Let us consider the continuous chip formation shown in Fig. 16.5, referring to the orthogonal cutting of a 6061 Al alloy. The authors of the investigation carried out a metallographic analysis of the chip in the process zone to estimate the possible occurrence of ductile damage. Close to the tool tip, where the effective strain is comparatively larger, only few voids were detected, as shown in Fig. 16.7A. On the contrary, several voids that merged into a small ductile crack were found at a certain distance from the chip root, where the effective strain was lower (even if the small crack remained confined at the chip root).

To explain these results, the R-T criterion for ductile fracture (Eq. 6.17) must be considered. Therefore reference to the local intensity of the effective strain and the distribution of the stress triaxiality values, T , must be made. As schematized in Fig. 16.6, ahead of the tool tip, plastic deformation is mainly by shear, and T is close to 0. At a certain distance from the chip root, T is positive (>0.3). Following the R-T criterion, we can then state that close to the tool tip only a small voids nucleation occurred because of the comparatively high local effective strain to fracture ($\epsilon_{\text{eff},f}$), due to the comparatively low T values. On the contrary, a higher ductile damage was observed at a certain distance from the chip root, despite the local lower effective strain, because of the much higher value of T , which induced a marked decrease in $\epsilon_{\text{eff},f}$.

In general, we can say that a continuous chip typically forms in the case of cutting alloys with sufficiently high ductility, such as Al alloys, ferritic- and ferritic-pearlitic steels. In these cases, the plastic strains involved in the formation of the chip are not sufficient to induce the onset of ductile fracture and are unable to break the chip in the process region during its evacuation from the cutting region.

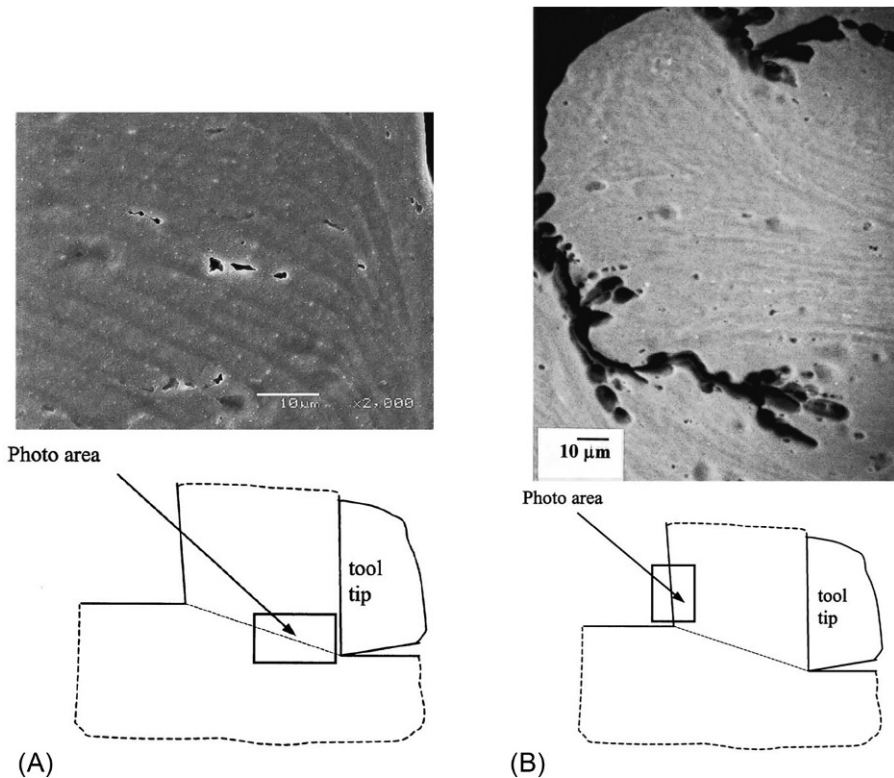


Fig. 16.7 Metallographic analysis at high magnification of the chip in Al alloy shown in Fig. 16.3, in two regions: close to the tool tip (A) and close to the chip root (B). Reproduced from Zhang, H., Alpas, A.T., 2002. Quantitative evaluation of plastic strain gradients generated during orthogonal cutting of an aluminum alloy. *Mater. Sci. Eng. A* 332, 249–254, with permission of the Publisher.

Discontinuous chip

Discontinuous chips form in the case of cutting metals with low ductility, following the attainment of ductile fracture conditions in the process zone. Fig. 7.31 shows that martensitic steels and Mg alloys are characterized by low ductility: these alloys typically display the formation of a discontinuous (broken) chip during machining. For these metals, the extension of the process zone is very limited, since the critical effective strain for fracture is small, and small local effective strains are then required for the nucleation of a ductile crack.

As an example, Fig. 16.8 reports a sequence of frames showing the chip formation in the case of orthogonal cutting of a Mg AZ31B alloy (Saei, 2021). The arrow in the frame 2 shows the nucleation of a ductile crack ahead of the tool tip (in the primary shear region). The arrow close to the chip root in the frame 3 shows the nucleation of a second ductile crack (at a certain distance from the chip root where T is positive). The

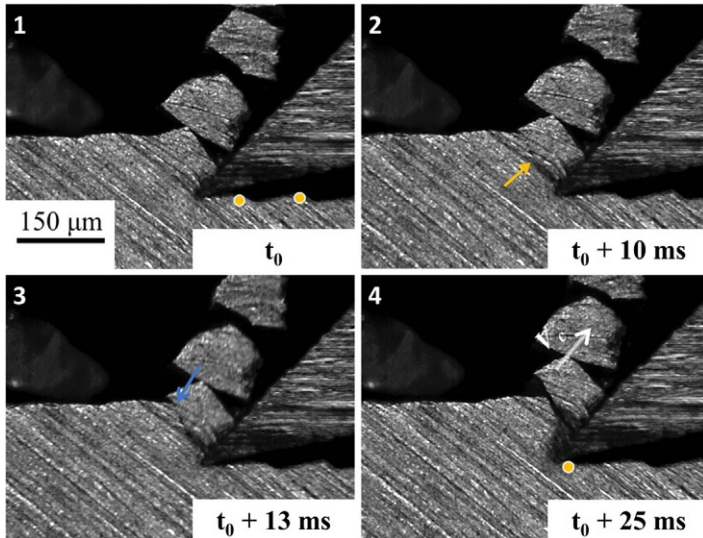


Fig. 16.8 Sequence of frames showing the sequence of chip formation in the case of cutting of Mg AZ31B alloy (Saei, 2021; reproduced with permission of the Author). Rake angle $\alpha = 40^\circ$; $v = 3 \text{ mm/s}$; $a_1 = 100 \text{ mm}$.

two cracks propagate and then merge forming a broken fragment. The process is cyclic. After the breaking of the chip, new material approaches the tool and is submitted to compression and then to shear, and is finally fractured. The distance between two successive cracks (and hence the length of the broken chips) decreases with increasing the cutting speed, and increases with increasing the feed rate.

The described sequence responsible for the formation of discontinuous chips is supported by the results of an FEM modeling focused on the orthogonal cutting of a 4340 steel, quenched and tempered to have a hardness of 33 HRC (Guo and Yen, 2004). The corresponding tensile elongation at fracture is comparatively low, approximately 12%. The onset of ductile fracture is modeled using Eq. (8.1), i.e., considering the R-T criterion modified to account for the influence of temperature and strain rate on ductility. The FEM analysis has shown that plastic deformation first develops in front of the rake face of the tool, and then at the chip root free surface. Subsequently, two ductile cracks nucleate, the first in front of the tool face and the second at the chip root, as shown in Fig. 16.9A. The two cracks propagate and then merge, hence forming a broken chip (Fig. 16.9B).

As already mentioned, mechanical parts manufactured with martensitic steels are increasingly finished using turning operations after heat treatment (in the so-called *hard turning* process). This procedure increases the productivity with respect to a conventional process consisting of (rough) turning after annealing and finishing by grinding after heat treatment. Of course, in hard turning the thickness, a_1 must be carefully selected to avoid excessively high cutting forces and power (see Eq. 16.7). Because of the high hardness of these metals, the specific cutting energy, u , is also

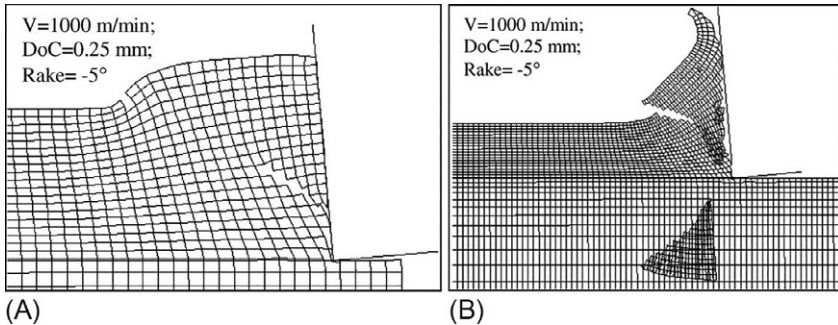


Fig. 16.9 (A) FEM modeling showing the nucleation and growth of two ductile cracks forming close to the tool tip and the chip root, in the case of cutting of a martensitic steel. (B) Formation of a broken chip when the two cracks meet.

Reproduced from Guo, Y.B., Yen, D.W., 2004. A FEM study on mechanisms of discontinuous chip formation in hard machining. *J. Mater. Process. Technol.* 155–156, 1350–1356, with permission of the Publisher.

high, and this may induce a large temperature rise in the primary shear region with the possible formation of adiabatic shear bands (Mabrouki et al., 2016). As shown in the next subsection, this mechanism may further contribute to the formation of a discontinuous chip.

In the formation of a discontinuous chip, an important role is played by the rake angle, α . As explained in Section 16.2, as the rake angle increases, the extension of the deformation by compression in the region between the tool tip and the chip root also increases, and this makes attainment of the condition for ductile fracture (R-T criterion) more difficult. Therefore, in metals with a moderate/low ductility, a discontinuous chip may form adopting negative values of the rake angle that induce intense shear in the process zone ($T \sim 0$, and not negative).

An important case of discontinuous broken chip formation is observed in the cutting of the so-called *free machining alloys*, such as, for example, resulfurized steels, steels containing Pb or Bi, and Cu alloys containing Pb (all introduced in Chapter 7). These metals contain particles (MnS sulfides, Pb or Bi particles) added with the objective of reducing the ductility, i.e., favor the ductile fracture of the chip under formation. As an example, Fig. 16.10 shows a cross section of the chip in the case of a resulfurized steel with composition: 0.16%C, 1.18%Mn, and 0.27%S (Sidjanin and Kovac, 1997). It is characterized by the presence of several ductile cracks (prone to completely break the chip in different regions), originating from the tool tip and the chip root. It should also be noted that, during cutting, some particles also spread on the rake surface of the cutting tool, exerting a solid lubrication that reduces the chip-tool friction. This reduces both the tool wear and the cutting force, with positive additional consequences for machinability.

If the metal ductility is not sufficiently low, the formation of a discontinuous chip can be obtained by forcing the chip to bend, and therefore by submitting it to large tensile strains (see Section 15.3). This aim can be achieved by using a *chip breaker*,

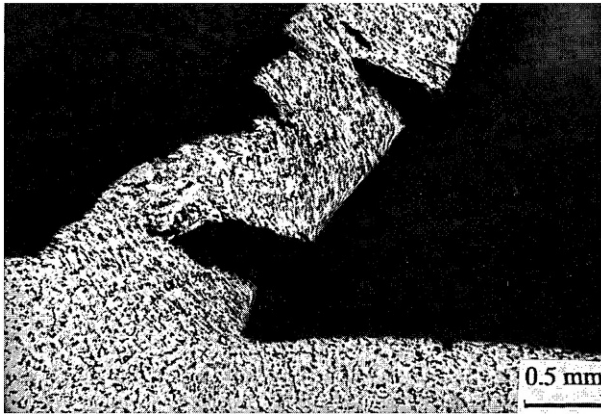
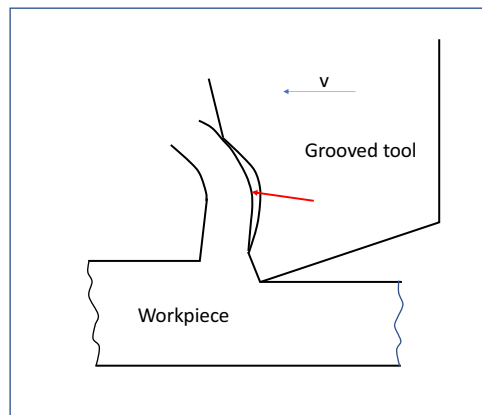


Fig. 16.10 Cross section of the chip under formation in case of a resulfurized steel with the following composition: 0.16%C, 1.18%Mn, and 0.27%S.

Reproduced from Sidjan, L., Kovac, P., 1997. Fracture mechanisms in chip formation processes. *Mater. Sci. Technol.* 13, 439–444, with permission of the Publisher.

Fig. 16.11 Schematization showing the principle of chip breaking using a grooved tool. The *arrow* indicates the region of chip bending with the formation of tensile strains that can be responsible for ductile fracture.



i.e., a cutting tool with a grooved geometry (see schematization in Fig. 16.11), or a tool with an obstruction-type device (such as a dot) attached to the rake face (Yilmaz et al., 2020). The cutting parameters must be carefully selected to properly force the chip into the groove or against the obstruction, in order to obtain the right bending for its ductile fracture (Lofti et al., 2015).

Segmented (or serrated) chip

The intense plastic shear deformation and the high strain rate in the primary shear region may induce the formation of adiabatic shear bands (ASBs), if the local temperature rise is sufficiently high, as explained in Section 3.6. This will give rise to the

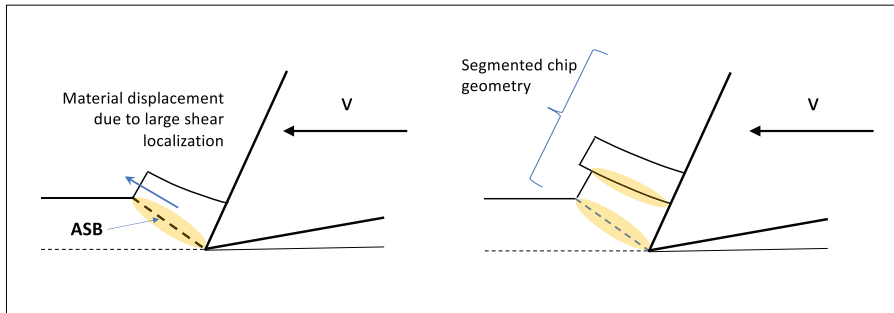


Fig. 16.12 Schematization showing the formation of a segmented chip geometry (saw-tooth geometry) following ASB formation.

formation of segmented (or serrated) chips, with a typical *saw-tooth* geometry. This is determined by the intense plastic localization in the primary shear region and the corresponding large displacement of material close to the shear band and parallel to it, as schematized in Fig. 16.12. ASBs may be accompanied by the nucleation of ductile cracks. In this case, a discontinuous broken chip is formed. However, the ASB does not generally produce the breaking of the chip, with the associated advantages already mentioned, and a continuous segmented chip is typically formed. The thickness variation of the segmented chip introduces variations in the cutting force and also in the chip evacuation velocity that must be properly considered in order to obtain the desired surface quality after machining.

Adiabatic shear banding in machining is favored by different factors that may also act together (Section 3.6):

1. A low thermal diffusivity of the workpiece material that favors a high temperature rise in the shear region;
2. A high hardness of the workpiece material, i.e., a high u value that gives rise to an intense heat generation and therefore a high local temperature rise;
3. A high cutting velocity that induces an increase in the plastic resistance of the workpiece material and then a comparatively high local temperature rise.

Austenitic stainless steels and Ti alloys are characterized by low thermal conductivity, and they form ASBs rather easily, even at comparatively low cutting velocities and feed rates. As an example, Fig. 16.13A shows the chip formation in the case of orthogonal cutting of an AISI 316L austenitic stainless steel, with $v = 150$ m/min, feed = 0.154 rev/min, and $a_1 = 3$ mm (M'Saoubi and Ryde, 2005). The chip displays a segmented geometry with intense shear localizations, which were found to be characterized by the presence of mechanical twins because of the low SFE of the alloy. Fig. 16.13B shows details of ASBs for a Ti-64 alloy submitted to orthogonal cutting with 60 m/min and $a_1 = 0.1$ mm (Calamaz et al., 2008). In this case the nucleation of a ductile crack in the shear band can be appreciated. It is close to the chip root where plastic deformation by shear is prevailing. It is reported that for the Ti-64 alloy, segmented chips form at a cutting velocity as low as 9 m/min (Hou and Komanduri, 2002).

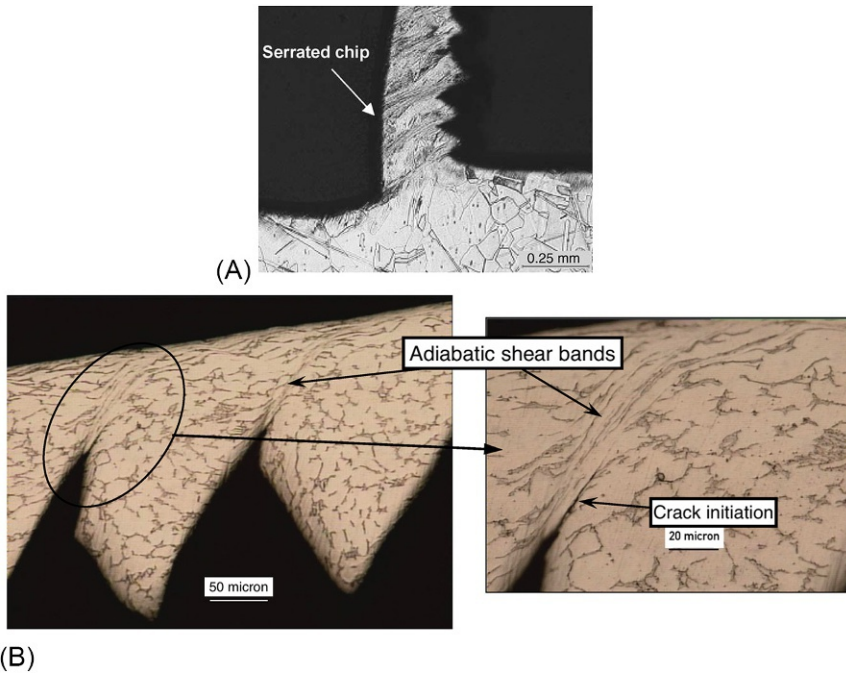


Fig. 16.13 Serrated (or segmented) chip formation in (A) an austenitic stainless steel (reproduced from M'Saoubi, R., Ryde, L., 2005. Application of the EBSD technique for the characterization of deformation zones in metal cutting. *Mater. Sci. Eng. A* 405, 339–349, with permission of the Publisher), and (B) a Ti-64 alloy (reproduced from Calamaz, M., Coupard, D., Girot, F., 2008. A new material model for 2D numerical simulation of serrated chip formation when machining titanium alloy Ti-6Al-4V. *Int. J. Mach. Tools Manuf.* 48, 275–288, with permission of the Publisher).

The low thermal conductivity (and then low diffusivity) of alloys prone to give ASB also favors a marked temperature rise in the tool, and this strongly favors the diffusional wear of the same. As a consequence, materials such as austenitic stainless steels and Ti alloys are characterized by a poor machinability. For specific applications, resulfurized austenitic stainless steels are commonly used (if the presence of the sulfides does not impair the mechanical behavior of the produced parts).

The formation of segmented chips due to ASB phenomena can also be observed in the case of high-speed machining (HSM). HSM increases the productivity and induces a fast evacuation of the chip that avoids a massive transfer of heat into the tool, with a corresponding reduction in the tool wear. The high cutting velocity also induces a high strain rate deformation in the chip, in particular along the primary shear region, and therefore ASB at critical velocity-feed conditions (Duan and Zhang, 2012). Also in this case, the cutting process must be carefully controlled in order to fully exploit the advantages of HSM.

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DUCTILITY AND FORMABILITY OF METALS

A Metallurgical Engineering Perspective

Giovanni Straffelini

Ductility and Formability of Metals: A Metallurgical Engineering Perspective uses metallurgical, mechanical, and physical principles and concepts to explain ductility while emphasizing the influence of material microstructure on ductile fracture mechanisms. Focusing on steel, aluminum, copper, titanium, and magnesium alloys, the book examines the strain hardening behaviors of these metals and alloys, the influence of strain rate and temperature, and ductile fracture, including a ductile fracture criterion. Hot plastic deformation is covered, with special consideration given to its interplay with recrystallization phenomena. Other phenomena such as dynamic strain ageing (DSA) and adiabatic shear banding (ASB) are discussed. Metal working applications such as forging, extrusion, and machining are included throughout, with particular emphasis on the control of ductile cracks in parts obtained with such forming operations.

Key Features

- Provides an overview of the plastic deformation behavior and ductile fracture of steel, aluminum, copper, titanium, and magnesium alloys
- Illustrates the influence of microstructure on yield behavior and strain hardening of metals, and the influence of strain rate and temperature
- Covers the role of the strain hardening coefficient (n), strain rate index (m), dynamic strain aging (DSA), and adiabatic shear banding (ASB) on the ductile damage mechanisms
- Metalworking applications are provided throughout, including forging, rolling, extrusion, wire drawing, sheet metal forming, and machining

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Giovanni Straffelini graduated in materials engineering at the University of Trento in 1989 and received his PhD in metallurgical engineering from the University of Padova in 1993. He currently teaches courses on metallic materials engineering and steelmaking and foundry technologies at the University of Trento. His research activity mainly concerns the fields of mechanical metallurgy, plastic working of alloys, powder metallurgy, and tribology, topics on which he has published more than 200 scientific works, mostly in international journals. He is also included in the World Ranking of Scientists, developed by Stanford University.



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