Chapter 3
Plastic Deformation Mechanisms

Abstract This chapter begins with an introduction of the crystalline structure and crystalline defects of solid materials and the definition of hot and cold work. Then, a full description of the dislocation slip mechanism is presented, including an explanation of the direct observation of dislocations by transmission electron microscopy. Based on the preceding ideas, a description of the slip modes and their associated dislocation substructures is given. The chapter concludes with a brief description of the plastic deformation mechanism by twinning.

3.1 Crystalline Defects and Deformation Mechanisms

Thanks to the works of the physicists Max Von Laue from Germany and the British W. H. Bragg and W. L. Bragg, in 1913, it is known that most engineering metals and alloys are *crystals*. A crystal is a solid whose atoms are arranged in an order formed by the repetition of a unitary geometry that extends throughout the solid. This idea was proposed by the self-taught English scientist William Barlow in 1883, based on the ideas of the French physicist Auguste Bravais, who introduced in 1848 the field of crystallography, describing the *unitary cell*, a term given to the geometric unit that repeats in a crystalline solid, being also the symmetry-unit that defines the type of crystalline structure. The most important crystalline structures, called *Bravais lattices*, are shown in Table 3.1

The ordered repetition of a group of unitary cells in a solid is called *lattice*. Lattices are tridimensional, so each plane and direction in the cell unit is defined by a vector, described by the *Miller indexes*, (Miller's notation can be found in textbooks of Materials Science or Crystallography). Figure 3.1 illustrates these concepts in two dimensions for further simplicity.

Soon after discovering the crystalline structure, it was found that the crystalline lattices in natural and engineering materials are not perfect; but rather contain flaws called *crystalline defects*. The most common types of crystalline defects are listed in Table 3.2 and a schematic representation of them is shown in Fig. 3.2.

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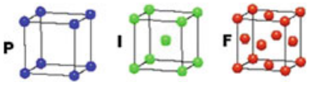
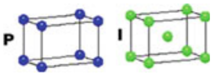
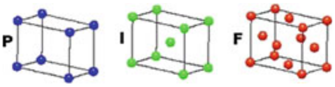
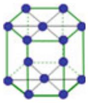
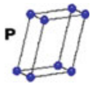
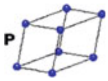
Table 3.1 Common crystalline structures in solid materials

Structure P = Primitive, I = Body centered, F = Face centered	Material
Cubic 	Face centered cubic (*fcc*): Cu, Ni, Al, Fe-γ, Au, Ag. Body centered cubic (*bcc*): Fe-α, Mo, W.
Tetragonal 	Fe_3C, Martensite (α')
Orthorombic 	Marcasite, Olivine, Aragonite
Hexagonal 	Compact hexagonal (*hc*): Cd, Mg, Ti, Zn
Monoclinic 	Gypsium, Orthoclase
Triclinic 	Tantite, Wollastonite

Evidently, the plastic deformation of solid materials is closely related to the crystalline structure and its crystalline defects. The sequence of events that produce plastic deformation in solids is called *deformation mechanism*. At this point it is necessary make clear that the elastic deformation mechanisms is just the result of the stretching or shortening of atomic links, therefore it is quite straight forward,

Fig. 3.1 Schematic two dimensions representation of a single cubic crystal lattice

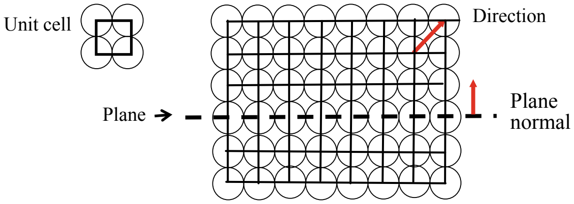


Table 3.2 Classification of crystalline defects

Type	Defect
Punctual	• Vacancies • Substitutional impurity atom • Interstitial impurity atom
Lineal	• Dislocations–perfect [edge, screw and mix] • Partial dislocations
Planar	• Stacking faults: Intrinsic y extrinsic • Twin boundary • Grain boundary
Volume	• Precipitates

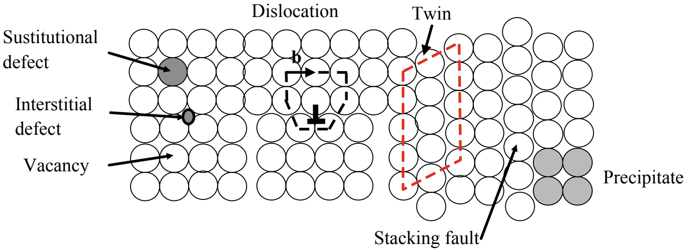


Fig. 3.2 Schematic representation of main crystalline defects

and requires no further study, whereas plastic deformation mechanisms involve the rearrangement of particles within the solid, including defects movement, all of which make them very complex.

3.2 Definition of Cold and Hot Work

Before studying the deformation mechanisms, it is convenient to bear in mind that the mechanical behavior of engineering materials is strongly influenced by the temperature, and metallic materials may feature two types of behavior with respect to it, referred as: low temperature or *cold work*, and high temperature or *hot work*. The main difference is that at high temperature, the deformation mechanisms involve time-dependent thermally activated processes such as diffusion and viscous

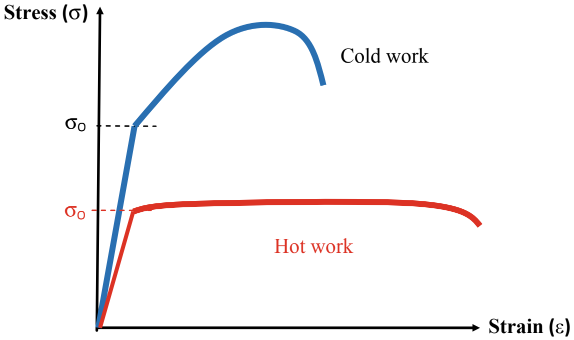


Fig. 3.3 Tension Stress–Strain curve of hot and cold work

flow, whereas at low temperature, deformation occurs by processes that are independent of time. The difference between hot and cold work is exhibited in the tension stress-strain curve, as the absence of strain hardening in hot deformation, in addition to reduction of the elastic modulus, yield strength and ultimate tensile strength, as schematically shown in Fig. 3.3.

The parameter to determine whether it is hot or cold work is the *homologous temperature*, which defined as:

$$\Theta = T/T_f$$

where Θ is the homologous temperature, T is the work temperature in absolute degrees and T_f is the material's melting point absolute temperature (in Kelvin). The most widely accepted criterion is:

$$\begin{aligned} &If\ \Theta > 0.4, \quad it\ is\ hot\ work \\ &If\ \Theta < 0.4, \quad it\ is\ cold\ work \end{aligned}$$

However, the value of Θ that determines the transition from cold work to hot work is not a fixed one, but it is rather a range of values. Furthermore, since Θ depends on the melting point, the condition of hot work does not necessarily indicate that the material is at a temperature much higher than room temperature, as shown in the following example:

Example Determine the Θ value steel and a Pb–Sn alloy at room temperature (25 °C, 298 K)

For steel: T_f = 1600 °C = 1873 K, thus: Θ = 298/1873 = 0.16

For the Pb–Sn alloy: T_f = 200 °C = 473 K, thus: Θ = 298/473 = 0.63

Notice that at room temperature, steel is in cold work whereas the Pb–Sn alloy is hot work condition.

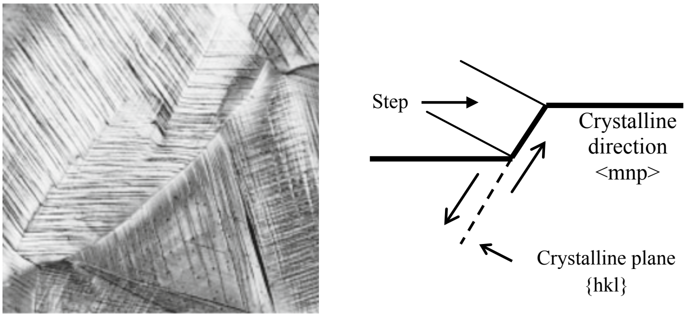


Fig. 3.4 Slip lines on a polished surface as seen in the metallographic microscope and schematic representation of the formation of a surface step by slip

3.3 Deformation by Dislocation Slip

In March, 1899, Ewing and Rosenhain, from St. John's College, Cambridge, published a short paper, where they reported observations through the metallographic microscope, of the polished surface of a metal strip stretched up to reaching plastic deformation. They observed fine parallel dark lines within the grains, that increased in number and new line systems appeared in different directions, as the specimen was stretched further. After performing a crystallographic analysis, they concluded that the lines were actually tiny surface steps produced by the slip of crystalline planes, as shown in Fig. 3.4. Such lines were called *slip lines* and it was concluded that plastic deformation in metals is caused by the slip of crystalline planes.

Later on, it was demonstrated that slip occurs in specific crystallographic planes and directions for each crystalline structure, where the slip direction is always parallel to the slip plane. The combination of plane and direction of slip is called *slip system*. The slip systems of the main crystalline structures of engineering metals and are shown in Table 3.3.

Notice that the *bcc* structure has a greater number of slip systems than the *fcc*, but metals with *fcc* structure are usually more ductile. This is because the slip planes in *fcc* metals are more compact than those in *bcc* metals, thus facilitating slip. Once experimentally demonstrated that plastic deformation occurs by the slip of

Table 3.3 Main slip systems of common crystalline structures

Structure	Planes	Directions	No. of systems
fcc	{111}	<110>	12
bcc	{111} {112} {123}	<111>	48
hc	{0001}	<1120>	3

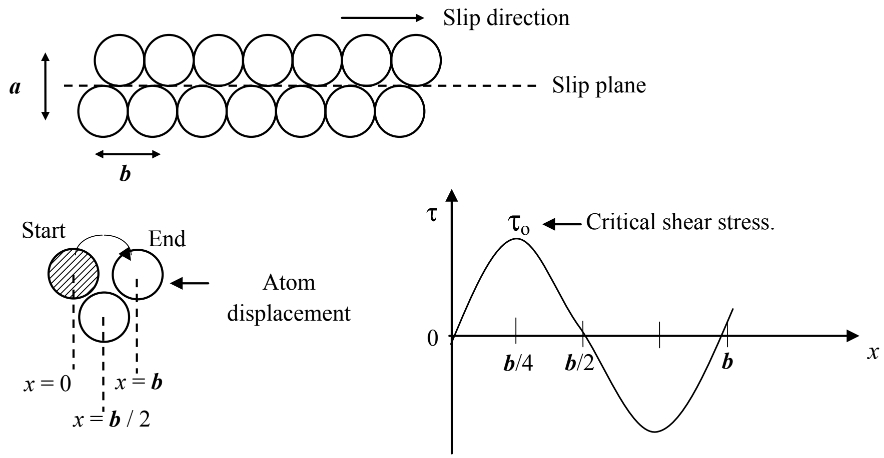


Fig. 3.5 Slip model in a perfect lattice to determine the stress necessary to produce slip

crystalline planes, the next step was to calculate the stress necessary to produce the slip of an atomic plane over another one. Such estimation was done by Frenkel in 1926 by the following procedure:

Consider a model of two parallel rows of atoms as shown in Fig. 3.5, when the upper row of atoms moves over the one below, the shear stress varies according to the following equation:

$$\tau = \tau_0 \sin\left(\frac{2\pi x}{\boldsymbol{b}}\right)$$

For small values of the argument $(2\pi x/b)$, the approximation $\sin\theta = \theta$ is valid, therefore the previous equation can be simplified as:

$$\tau = \tau_0\left(\frac{2\pi x}{b}\right)$$

On the other hand, the shear deformation may be calculated as:

$$\gamma = \frac{x}{a}$$

Before the upper row of atoms reach the position $x = \boldsymbol{b}/4$, it can be assumed that the deformation is elastic, thus the shear elastic deformation is:

$$\tau = G\gamma = \frac{Gx}{a}$$

Therefore:

$$\frac{Gx}{a} = \tau_0\left(\frac{2\pi x}{b}\right)$$

Taking $a = b$ (as typical in cubic crystals) and solving for τ_0, the theoretical shear stress for slip in the perfect lattice is obtained:

$$\tau_0 = \frac{G}{2\pi}$$

The typical values of G for metals range from 6000 to 70,000 MPa (10^6 to 10^7 psi), thus calculating τ_0 with the previous equation, the resulting values are, from 100 up to 10,000 times higher than those obtained experimentally, which are in the range of 7–2000 MPa (10^3 to 3×10^5 psi). The discrepancy between the theoretical and experimental slip stress values does not disregard slip as the plastic deformation mechanism, since it is an experimental observation, but instead it implies there must be "something" that helps slip to occur at a much lower stress.

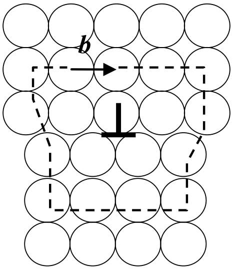
In 1905, the Italian scientist Vito Volterra developed the dislocation theory, from which, later in 1934, Orowan, Taylor and Polanyi, each one separately, proposed that slip could be the result of dislocation movement. Orowan is the most recognized dislocation scientist in the world, but it is Taylor's model that more clearly illustrates how the dislocation movement can be responsible of the slip. The Taylor's model of slip is based on an edge dislocation, which is characterized by a Burger's vector (***b***) parallel to the dislocation plane, as shown in Fig. 3.6.

To demonstrate that the slip mechanism is caused by dislocation movement, the following conditions must be meet:

1. Dislocation movement must produce plastic deformation.
2. Dislocation movement must occur in directions over a plane and produce surface steps.
3. Dislocation movement must take place at a stress lower than the theoretical one.

The following mechanism demonstrate conditions 1 and 2: By applying a shear stress parallel to the Burger's vector, the atomic bonds unbalance, as shown in Fig. 3.7a, this makes the dislocation move forward one atomic position in the

Fig. 3.6 Edge dislocation with Burgers vector in equilibrium position



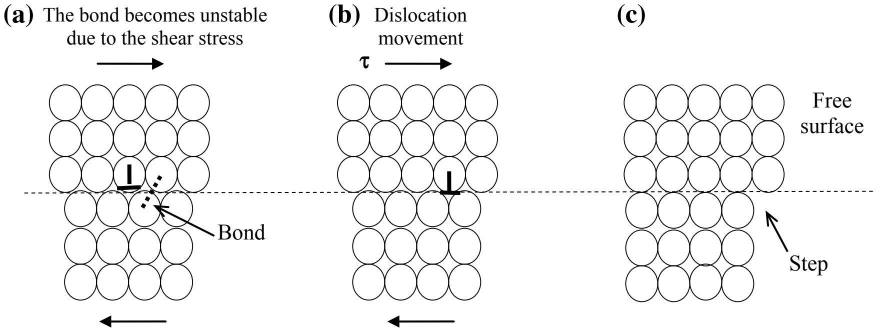


Fig. 3.7 Taylor's model showing that slip can be produced by the movement of an edge dislocation

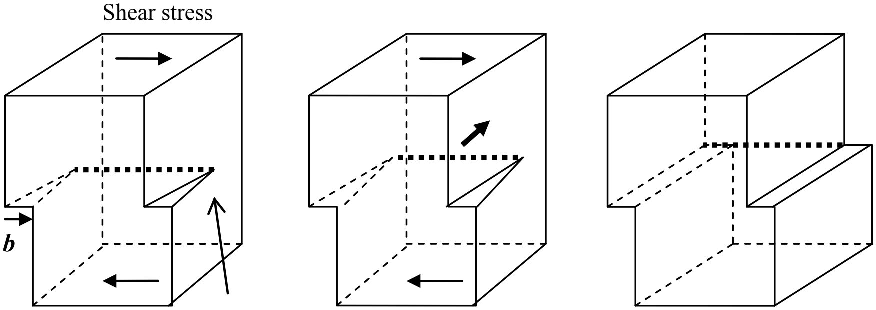


Fig. 3.8 Crystal slip produced by the movement of a screw dislocation

lattice, Fig. 3.7b. By holding the stress, the dislocation keeps moving and shears the crystal until it reaches a free surface, where a step is formed, as seen in Fig. 3.7c. An analog behavior is produced by a shear stressed screw dislocation, as shown in Fig. 3.8. The models depicted in the previous figures demonstrate that indeed the dislocation movement is consistent with the characteristics of plastic deformation by slip observed experimentally.

The shear stress to move a dislocation was calculated by Pierls and corrected by Nabarro in 1947, and is given by the equation:

$$\tau_{dis} = \left\{ \frac{2G}{(1-v)} \right\} \exp\left(\frac{-2\pi w}{\boldsymbol{b}} \right)$$

where: v is Poisson's ratio, $\boldsymbol{b}$ is Burgers vector, w is the dislocation width, defined as the distance up to which the dislocation distorts the crystal and G is the shear module. This equation combines the elastic stiffness, given by G, while the exponential term is related to an energetic barrier that has to be surpassed in order to

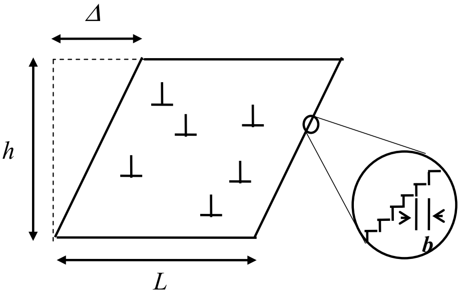


Fig. 3.9 Two dimensional model to estimate the number of dislocations necessary to produce a significant shear strain

reach an equilibrium state of lower energy. It has been estimated that $w = 10\boldsymbol{b}$ in ductile metals and $w = 2\boldsymbol{b}$ in brittle metals, thus if: $G = 11 \times 106$ psi, $v = 0.35$ and $w = 5$b:

$$\tau_{\text{dis.}} = \{2[11 \times 10^{6}]/(1 - 0.35)\} \exp(-10\pi) = 7.68 \times 10^{-7}\,\text{psi}$$

This calculation shows that an extremely low stress is required to move a dislocation, but now it is necessary to explain how the dislocation movement reach the experimentally observed yield strength. What actually happens is that a very large number of dislocations has to be moved in order to achieve a significant amount of plastic strain, in addition to the fact that dislocations have to overcome obstacles, which altogether, increases the stress necessary to produce the slip.

The estimation of the number of dislocations necessary to plastically strain a crystal by slip can be done by the following reasoning: Consider an originally rectangular crystal, of h height and L length, which is deformed by a group of dislocations moving in parallel planes, as depicted in Fig. 3.9.

If the total shear displacement (Δ) produced by the group of dislocations is the sum of the individual displacements produced by each single dislocation, then:

$$\Delta = \boldsymbol{b}\left(\frac{\sum x_i}{L}\right)$$

where x is the slip distance of each individual dislocation, L is the crystal width and $\boldsymbol{b}$ is Burgers vector. On the other hand, the shear strain is equal to:

$$\gamma = \Delta/h$$

Substituting Δ into γ:

$$\gamma = \frac{b}{hL}\sum x_i$$

Introducing the average slip distance (x_{av}) given by:

$$x_{av} = \frac{\sum x_i}{N}$$

where N is the number of dislocations crossing the crystal and combining the previous equations, the total strain produced by the dislocations is:

$$\gamma = \boldsymbol{b} x_{av} \left(\frac{N}{hL} \right)$$

The term (N/hL) is called *dislocation density* and is represented by the symbol ρ_{dis}, thus the shear strain produced by the dislocation movement is:

$$\gamma = \boldsymbol{b} \rho_{dis} x_{av}$$

The following example gives an estimation of the dislocation density required to produce an appreciable plastic deformation.

A 1 cm per side cubic crystal, where $\boldsymbol{b} = 3 \times 10^{-8}$ *cm strains at* $\gamma = 0.1$. *Calculate the associated dislocation density.*

Solution If the spatial distribution of the dislocations within the crystal is random, it may be assumed that the average slip length is one half of the crystal width $x_{av} = 0.5$ cm. Taking $\boldsymbol{b} = 3 \times 10^{-8}$ cm, and solving for ρ_{dis}:

$$\rho_{dis} = \gamma / \boldsymbol{b}\, x = (0.1) / \left(3 \times 10^{-8}\,\text{cm}\right)(0.5\,\text{cm}) = 6.7 \times 10^{6}\,\text{Dislocations/cm}^2$$

This example shows that a great number of dislocations is required in order to deform a crystal. In fact, it has been experimentally determined that heavily cold worked materials have dislocation densities in the range of 10^{12} to 10^{14} Dislocations/cm^2. Obviously, such number of dislocation does not initially exist, because deformation-free crystals have dislocation densities of 10^3 to 10^4 Dislocations/cm^2 and this amount is not sufficient to produce an appreciable plastic deformation, therefore, there must be mechanisms to generate high numbers of dislocations in order to plastically deform a crystal.

The best known mechanism of dislocation multiplication is the Frank-Read source, which is schematically described in Fig. 3.10; its mechanism is as follows:

1. A dislocation segment is pinned between two obstacles.
2. Under the action of a shear stress on the slip plane, the dislocation starts to bow out up to a critical curvature, then, the ends of the loop attract each other, because they are of opposite signs, forming a dipole (a pair of opposite sign dislocations close to each other).

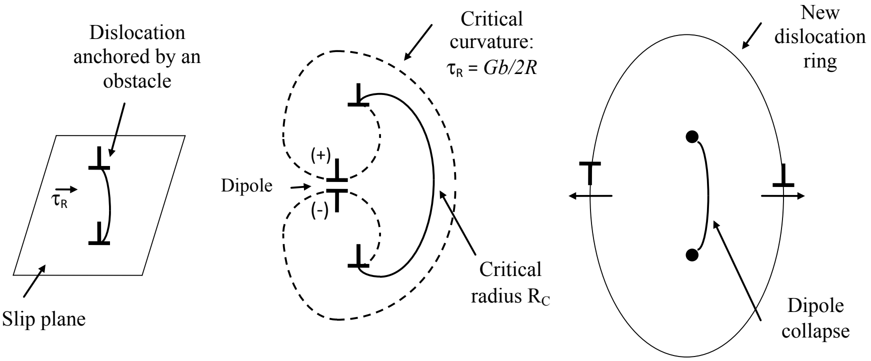


Fig. 3.10 Frank-Read source of dislocation multiplication

3. When the dislocation segments of the dipole are close enough, they collapse and the original dislocation is regenerated while the remaining segment of the loop forms a ring around the original dislocation.
4. The regenerated dislocation segment can continue generating new dislocation loops, while the dislocation loops can slip further into the crystal. This mechanism can repeat continuously until the crystal is saturated with dislocations. In fact, Frank and Read source is considered to be inexhaustible.

Slip strain rate: Although it has been said that cold work is independent of time, it actually occurs at certain rate, because the dislocations take some time to move through a crystal once a shear stress has been applied on the slip plane. The strain rate can be determined in the following way:

The shear strain rate ($d\gamma/dt$) can be determined by the total derivative:

$$\frac{d\gamma}{dt} = \boldsymbol{b}\rho_{dis}\left(\frac{dx}{dt}\right) + \boldsymbol{b}x\left(\frac{d\rho_{dis}}{dt}\right)$$

In the above equation, the term (dx/dt) is the dislocation slip rate, and the term ($d\rho_{dis}/dt$) is the dislocation multiplication rate, which is the amount of dislocations that have to be generated in order to obtain certain plastic deformation. At high strains, the crystal saturates with dislocations and (dp_{dis}/dt) comes close to zero, so the strain rate remains only in terms of slip rate:

$$\frac{d\gamma}{dt} = \boldsymbol{b}\rho_{dis}\left(\frac{dx}{dt}\right)$$

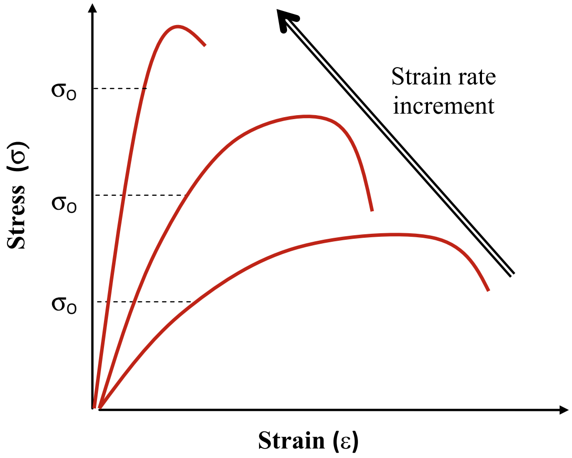


Fig. 3.11 Effect of the strain rate on tension behavior

Experimentally, it has been found that:

$$\frac{dx}{dt} = A\tau^{m'}$$

where A and m' are material's constants, thus:

$$\frac{d\gamma}{dt} = \boldsymbol{b}\rho_{dis}\left(A\tau^{m'}\right)$$

The previous equation indicates that the strain rate depends on the applied stress. Experimentally, it has been observed that when a crystalline material is deformed at high strain rates, the flow stress increases proportionally to the strain rate, while ductility is reduced, as shown in Fig. 3.11.

Resolved shear stress. Resolved shear stress (τ_R) is the shear component of a normal stress which actuates directly onto a slip plane. It is calculated from the applied tensile stress (P/A), as shown in the diagram of the Fig. 3.12.

$$\tau_R = \frac{P\cos\phi}{\frac{A}{\cos\lambda}} = \frac{P}{A}\cos\phi\cos\lambda$$

where the term $\cos\phi$ $\cos\lambda$ is called the *Schmid Factor*, therefore:

$$\tau_R = \sigma\cos\phi\cos\lambda$$

When τ_R is sufficiently high as to initiate slip, it is called *critical resolved shear stress* (τ_{RC}). Therefore, in a single crystal:

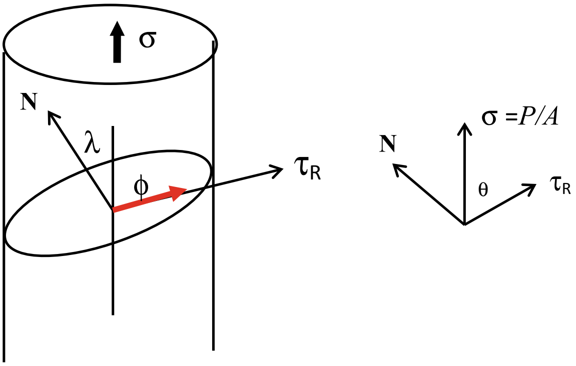


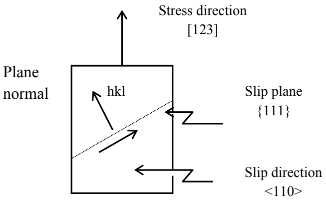
Fig. 3.12 Definition of resolved shear stress

$$\tau_{RC} = \sigma_0 \cos\phi \cos\lambda$$

where σ_o is the single crystal yield strength. When there are several active slip systems, slip will begin in the system with the highest Schmid Factor. If the Schmid Factor value is zero, the dislocation slip is impossible in such system.

Example An fcc single crystal is tension stressed along the [123] direction. Determine which slip system is first one to be active.

Solution A free-body diagram of the single crystal is:



The slip systems are:

$$(111)[110]$$
$$(111)[101]$$
$$(111)[110]$$

In a cubic crystal, the angle between planes or directions is calculated by:

$$\cos\ \phi = \frac{[hkl] * [mnp]}{\sqrt{h^2+k^2+l^2}\sqrt{m^2+n^2+p^2}}$$

For the (111) slip plane and the [123] stress direction:

$$\cos\lambda = (1\times 1 + 2\times 1 + 3\times 1)/(\surd 14\surd 3) = 6/\surd 42$$

For the slip direction [110] and the stress direction [123]:

$$\cos\phi = (1\times 1 - 1\times 2 + 0\times 3)/\surd 14\surd 2 = -1/\surd 28^2$$

The Schmid Factor is:

$$FS_1 = (6/\surd 42) * (-1/\surd 28) = -0.175$$

$$\begin{gathered} For\,system\,No.2 : FS_2 = -0.35 \\ For\,system\,No.3 : FS_3 = 0.52 \end{gathered}$$

Thus, system (111)[110] is the first to become active.

3.4 Direct Observation of Dislocations

The theory of dislocations succeeded in describing the plastic deformation mechanism of crystalline solids, even when dislocations had not been directly observed. The concept of dislocation was introduced by Volterra in the first years of the XX century, in those days the dislocations were observed in bubble rafts, stacks of pipes, block walls, and so forth. However, for over 40 years after the dislocation theory was published the direct observation of dislocations in engineering materials was not possible, until the work done Hirsch and collaborators in 1956, using the newly developed *transmission electron microscopy* technique.

Transmission electron microscopy (TEM) consists of an electron beam accelerated by hundreds of kilovolts, that is focused by magnetic fields that work as lenses to pass-through a thin foil of the material to be examined (typically of less than 1000 Angstroms thick). The hollow column that contains the electron beam must be at high vacuum to prevent air molecules from deflecting the beam. The electron beam, after passing through the specimen, is amplified and projected on a fluorescent screen placed on the lower part of the column, which can be observed through a transparent window. Figure 3.13 shows a scheme of a TEM.

The Bragg law states that the planes at an angle $\theta = \sin^{-1}(\lambda/2d)$, where λ is the electron beam wave length and $\boldsymbol{d}$ is the interplanar distance, will diffract the electron beam. At high voltages, λ has a very small value (3×10^{-12} m at 100 keV), so only those planes closely parallel to the beam will diffract. Therefore, when an electron beam passes through a crystal having a dislocation, the slight change of inclination of the planes produced by the dislocation will cause diffraction of the electron beam, then by eliminating the diffracted beams with an aperture plate

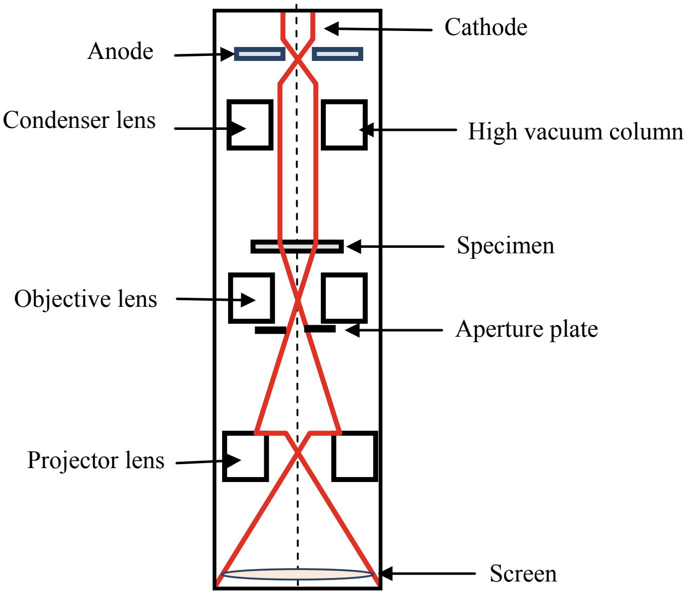
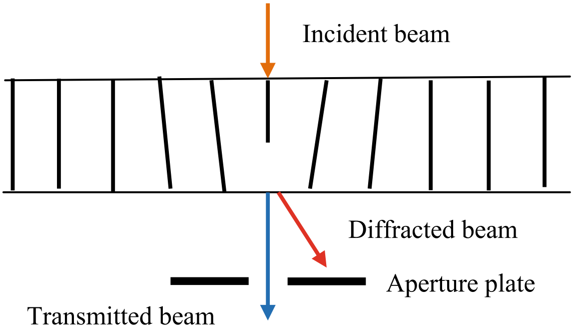


Fig. 3.13 Schematic construction of the transmission electron microscope

Fig. 3.14 Contrast mechanism that makes a dislocation visible in the TEM



placed just below the objective lens, a shadow will be observed in the exact same location where the dislocation line lies. This is schematically illustrated in Fig. 3.14.

Figure 3.15 shows an image of dislocations in an *fcc* metal, obtained with a TEM in bright field, meaning that the image is formed by the transmitted beam and eliminating the diffracted beam. The dislocation lines look wavy due to the ondulatory nature of electron diffraction, therefore if the crystal is tilted, the dislocation line shape may vary, or it may become invisible. The TEM images of dislocation lines are the segments that go from top to the bottom of the thin foil, and since the image is a bi-dimensional projection of a tri-dimensional space they look

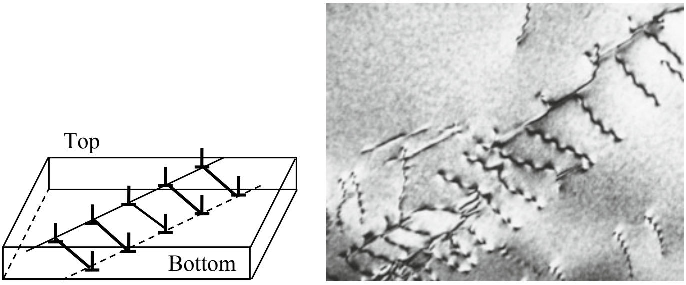


Fig. 3.15 Tri-dimensional lay out of dislocations segments on a thin foil and bi-dimensional projection image as seen in the TEM. Image courtesy of Dr. Héctor Dorantes R

short, as shown in Fig. 3.15. It is important to mention that, at a first glance, it cannot be determined which end is up or down.

3.5 Slip Modes

As it was seen before, plastic deformation is the result of the movement of a large number of dislocations over several crystallographic planes and directions, so there is a large number of interactions among single dislocations, as well as, among dislocation groups, which makes slip a complex phenomenon. The overall result of these interactions within a crystal is a particular configuration of dislocation groups, called *dislocation array*, which greatly influences the mechanical behavior of materials.

There are two basic modes in which a group of dislocations may move in the interior of a crystal, they are called *slip modes* and are described as follows:

Planar slip—In close-packed crystals, the dislocations tend to dissociate into partial dislocations that stay in the plane they were formed so they make up parallel aligned bands of dislocations. When the dislocations bands meet an obstacle, the distance between individual dislocations diminishes as they approach to the obstacle, thus producing arrays called *pile ups*, which look like the example shown in Fig. 3.16. Planar slip produces straight and well defined slip lines, as shown in Fig. 3.16.

Planar slip occurs is typical in *fcc* crystals, since a dislocation splits into two partial dislocations that have an associated stacking fault between them, that makes impossible for the dislocation to move out of the slip plane. The mechanism of partial dislocation formation in a *fcc* crystal is depicted in Fig. 3.17. To move on the (111) plane, a perfect dislocation must follow the [101] direction, thus, the dark atom must go over the gray atom. It is easy to see that the path around the gray atom, given by the directions a/c[112] and a/c[211] is more favorable, so the dislocation “splits” into two partials, forming a stacking fault, as shown in Fig. 3.18.

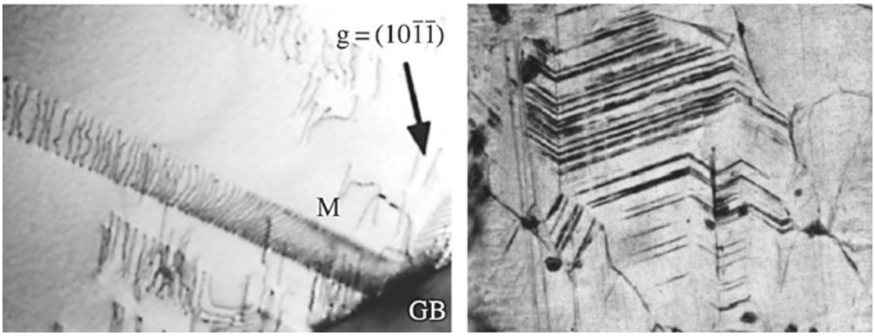


Fig. 3.16 Dislocation bands and straight slip lines produced by planar slip

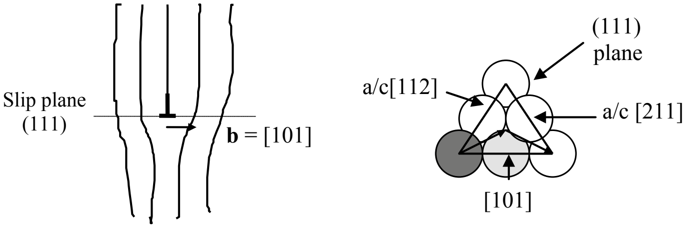


Fig. 3.17 Geometric model of an edge dislocation $b = [101]$ in a (111) plane in a *fcc* crystal, showing the split into two partial dislocations $b = [112]$ and $b = [211]$

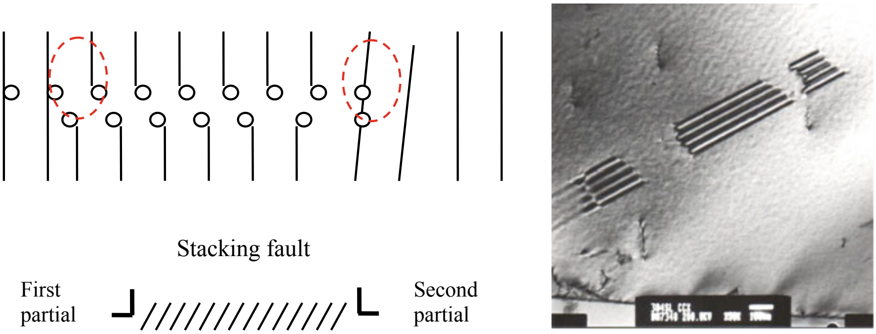


Fig. 3.18 Schematic representation of two partial dislocations with a stacking fault between them. The striped bands are stacking faults between partial dislocations as seen in the TEM

It is clear that stacking faults cannot change from one plane to another, so the partial dislocations are forced to stay on the planes where they were formed, favoring planar slip. The wider the stacking fault is, the less likely that a partial dislocation close. The separation of a stacking fault between two partial dislocations is result of the work balance between the rejection force of the partials (because they are of the same sign) and the stacking fault energy (*SFE*). The calculation of the stacking fault width is as follows:

The rejection force F_{re} between two partials is given by the following equation:

$$F_{re} = G\frac{\boldsymbol{b}_1\boldsymbol{b}_2}{2\pi l}$$

where l is the separation between partial dislocations and $\boldsymbol{b_1}$ and $\boldsymbol{b_2}$ are the corresponding Burgers vectors. The *SFE* is an energy by surface unit, therefore, the work necessary to form it is:

$$SFE = F_{re}(l)$$

Substituting F and solving for r, an equation to calculate l is obtained:

$$l = G\frac{b_1 b_2}{2\pi SFE}$$

This equation indicates that the lower the *SFE*, the wider the stacking fault between two partial dislocations and so it is more difficult to close them, which favors planar slip. Substitutional solid solutions usually feature lower *SFE*, thus metal alloys with *fcc* structure like stainless steels, bronzes, brasses and nickel alloys exhibit planar slip.

Wavy slip—Under normal circumstances, it is quite likely that a dislocation may change of slip plane, this is termed *cross-slip*. This is common in high *SFE* materials, such as pure metals and interstitial solution alloys, because they do not form partial dislocations and the single dislocations can cross-slip as long as their Burgers vector is compatible between the original and secondary slip planes. The slip lines produced when the dislocations cross-slip are diffuse and wavy, as shown in Fig. 3.19, so that is why this slip mode is called *wavy slip*.

Cross slip allows dislocations to move more freely than planar slip, which leads to the formation of complex dislocation arrays that evolve as the dislocation density

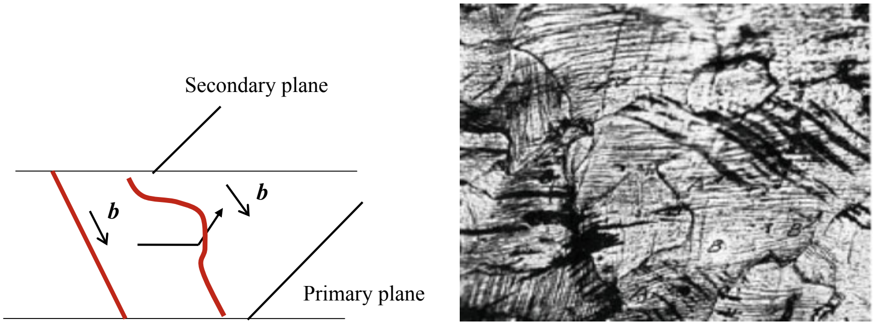


Fig. 3.19 Cross- slip of a dislocation and wavy slip lines as seen in a metallographic microscope

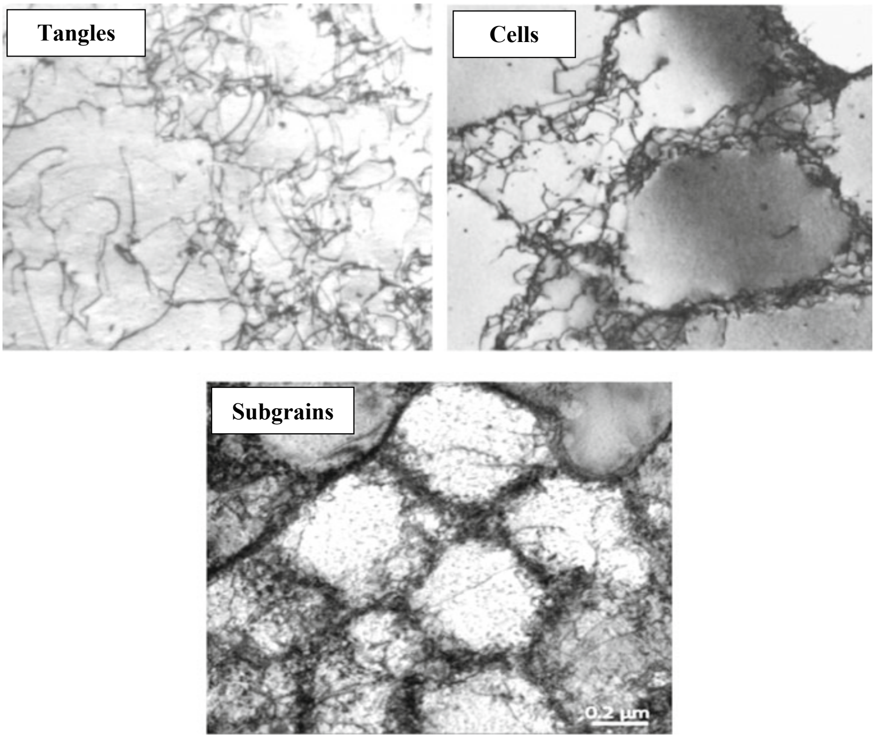


Fig. 3.20 Evolution of wavy-slip dislocation arrays. Initially, tangles are formed, which evolve into cells at higher deformation levels, up to a point where subgrains are formed

increases further. The evolution of dislocation arrays in the wavy slip begins with the formation of *tangles* which are dislocation bundles separated relatively long distances from each other. When the tangles get denser, the dislocation segments get closer to each other, up to a point where they rearrange to form *cells*, that are dense walls of tangled dislocations that surround or enclose low dislocation density zones. Finally, at high deformations, the dislocation cells produce a sufficiently large disorientation between neighboring zones, to lead to the formation of *subgrains*. Figure 3.20 shows the evolution of the wavy slip dislocation arrays as seen in the TEM.

The planar slip mode favors anisotropy of the mechanical behavior, because the plastic deformation is easier in the slip band directions, whereas wavy slip favors more homogeneous and isotropic deformation. Materials with planar slip also feature a more pronounced Bauschinger effect than materials with wavy slip because of the high back stress generated in the dislocation pile ups.

The slip mode also affects recrystallization, which is the formation of new deformation-free grains in cold worked materials after annealing, because the introduced heat causes the rearrangement and annihilation of dislocation arrays.

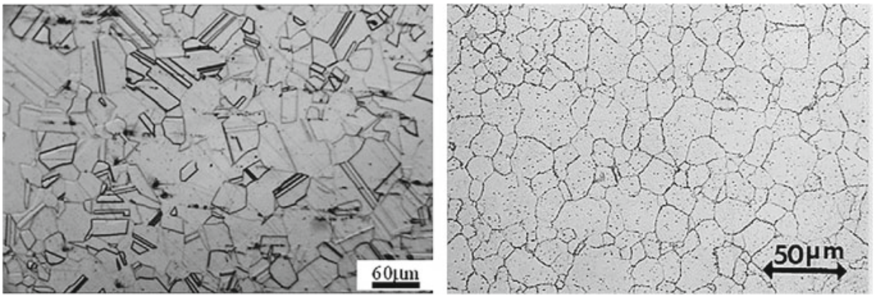


Fig. 3.21 Recrystallized grains of an alloy with planar slip (left) and an alloy with wavy slip (right). Left image taken from Metals 2017, 7(9), 348, https://doi.org/10.3390/met7090348

In planar slip materials, the dense dislocation bands favor the formation of annealing twins, and polygonal-shaped recrystallized grains, while in wavy slip materials, the subgrains favor the formation of refined-grain microstructures. Examples of the recrystallized microstructures of planar slip and wavy slip metals are given in Fig. 3.21.

3.6 Deformation by Twinning

Even though dislocation slip is the predominant plastic deformation mechanism at low homologous temperature, there are conditions where dislocation slip is not favored or it is completely inactive. Some of the conditions that inhibit dislocation slip are:

1. Few slip systems.
2. Unfavorable orientation of the slip plane (Schmid factor close to zero).
3. Strong obstacles.
4. High strain rates.
5. Very low temperatures (homologous temperature close to zero).

When any of these conditions is present, the *twinning* deformation mechanism appears, which consists of the shear strain of an entire section of a crystal, produced by the simultaneous movement of atoms in the sheared region. Figure 3.22 shows a schematic illustration of twining. Deformation by twinning is common in hexagonal metals such as magnesium and zinc, as well as on carbon steels, bronzes and other metals deformed at ballistic speeds, since deformation by twinning occurs at micro second rates, whereas slip rate is slower, in the order of milliseconds.

Twinning occurs on well-defined crystalline planes and crystalline directions, known as *twinning system.* The twinning systems in common crystalline structures are given in Table 3.4. Since twinning produces a region with different crystalline orientation, with plane boundaries, they are clearly visible in a metallographic

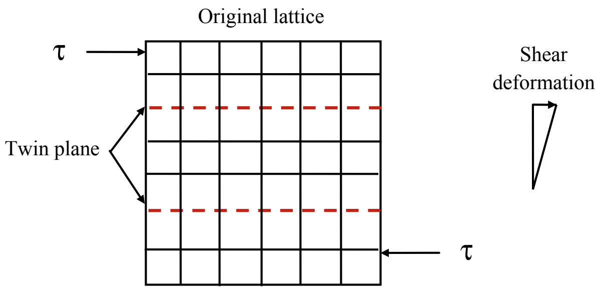
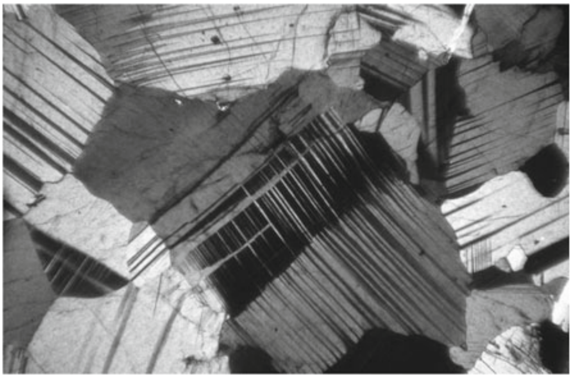


Fig. 3.22 Plastic deformation of a crystal by twinning

Table 3.4 Twinning systems in common crystalline structures

Structure	Twinning plane	Twinning direction	*S*
fcc	{111}	<112>	0.707
bcc	{112}	<111>	0.707
hc	{1012}, {1121}	<1011>, <1123>, <1126>	0.139

Fig. 3.23 Deformation twins as viewed in the metallographic microscope



preparation as straight bands across the grains. Deformation twins are fine and numerous, as shown in Fig. 3.23, while annealing twins in planar slip metals are formed by rearrangement of dislocation bands, forming coarser and less numerous bands, as shown in Fig. 3.24.

The magnitude of deformation by twinning can be calculated according to Hall (1954), by the following expression:

$$\Delta l = 1 - (1 + S \tan\chi)^{1/2}$$
$$\tan\chi = 1/2(S \pm \surd(S^2 + 4))$$

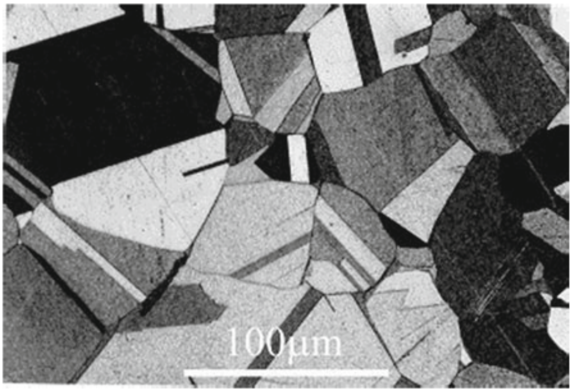


Fig. 3.24 Annealing twins as viewed in the metallographic microscope

where Δl is the elongation produced by twinning and S is a constant that depends on the structure and the twinning system (see Table 3.3). Based on this expression, the maximum elongation attainable by twinning is about 32%, but the average values obtained experimentally are rarely over 10%.

Twinning is rather unimportant as deformation mechanism because it occurs under quite special conditions and the obtained plastic deformation is not high. However, twinning is very important in the mechanical behavior, because by changing the crystal orientation, it may place some slip systems in favorable slip orientations and others on unfavorable orientations (i.e. low Schmid Factor). Additionally, twins are strong obstacles to dislocation movement, so they contribute to increase the mechanical strength.