Material Science

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Chapter 7. Dislocations and Strengthening Mechanisms

Learning objectives:

- To study the various strengthening mechanism

- To understand the role of dislocations in the strengthening

- To understand recovery, recrystallization and grain growth

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Highlights, Motivation and Critical Concepts:

Metallic materials in their natural form after extraction are suitable for limited engineering applications. For efficient use, these materials are usually either alloyed, heat treated, work hardened or strengthened by different mechanisms. One of the strengthening mechanisms is strain-hardening that involves plastic deformation. Almost all forming processes include plastic deformation. It is known that plastic deformation can occur in two ways – slip and twinning. However, dislocations will aid the plastic deformation process as they can move under applied stresses. If dislocation motion can be curtailed, metallic component can be used at higher loads i.e. component achieves more strength. In other words component got strengthened. Engineers, especially materials, structural and design engineers, need to understand the mechanisms involved in strengthening the metals for efficient use of metallic materials. It is also important to learn the after-process behavior of metals which may be subjected to heat treatment after plastic deformation. This is for further increasing the plasticity capacity of material. Phenomena that characterize the after-process are recovery, recrystallization, and grain growth. All the basic phenomena involved in the strengthening, plastic deformation and after-process are dealt in this chapter.

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6.1 Dislocations & Plastic deformation and Mechanisms of plastic deformation in metals

6.1.1 Dislocations & Plastic deformation

While some materials are elastic in nature up point of fracture, many engineering materials like metals and thermo-plastic polymers can undergo substantial permanent deformation. This characteristic property of materials makes it feasible to shape them. However, it imposes some limitations on the engineering usefulness of such materials. Permanent deformation is due to process of shear where particles change their neighbors. During this process inter-atomic or inter-molecular forces and structure plays important roles, although the former are much less significant than they are in elastic behavior. Permanent deformation is broadly two types – plastic deformation and viscous flow. Plastic deformation involves the relative sliding of atomic planes in organized manner in crystalline solids, while the viscous flow involves the switching of neighbors with much more freedom that does not exist in crystalline solids.

It is well known that dislocations can move under applied external stresses. Cumulative movement of dislocations leads to the gross plastic deformation. At microscopic level, dislocation motion involves rupture and reformation of inter-atomic bonds. The necessity of dislocation motion for ease of plastic deformation is well explained by the discrepancy between theoretical strength and real strength of solids, as explained in chapter-3. It has been concluded that one-dimensional crystal defects – dislocations – plays an important role in plastic deformation of crystalline solids. Their importance in plastic deformation is relevant to their characteristic nature of motion in specific directions (slip-directions) on specific planes (slip-planes), where edge dislocation move by slip and climb while screw dislocation can be moved by slip and cross-slip.

The onset of plastic deformation involves start of motion of existing dislocations in real crystal, while in perfect crystal it can be attributed to generation of dislocations and

subsequently their motion. During the motion, dislocations will tend to interact among themselves. Dislocation interaction is very complex as number of dislocations moving on number of slip planes in various directions. When they are in the same plane, they repel each other if they have the same sign, and annihilate if they have opposite signs (leaving behind a perfect crystal). In general, when dislocations are close and their strain fields add to a larger value, they repel, because being close increases the potential energy (it takes energy to strain a region of the material). When unlike dislocations are on closely spaced neighboring slip planes, complete annihilation cannot occur. In this situation, they combine to form a row of vacancies *or* an interstitial atom.

An important consequence interaction of dislocations that are not on parallel planes is that they intersect each other or inhibit each others motion. Intersection of two dislocations results in a sharp break in the dislocation line. These breaks can be of two kinds:

- (a) A jog is break in dislocation line moving it out of slip plane.
- (b) A kink is break in dislocation line that remains in slip plane.

Other hindrances to dislocation motion include interstitial and substitutional atoms, foreign particles, grain boundaries, external grain surface, and change in structure due to phase change. Important practical consequences of hindrance of dislocation motion are that dislocations are still movable but at higher stresses (or forces), and in most instances that leads to generation of more dislocations. Dislocations can spawn from existing dislocations, and from defects, grain boundaries and surface irregularities. Thus, the number of dislocations increases dramatically during plastic deformation. As further motion of dislocations requires increase of stress, material can be said to be strengthened i.e. materials can be strengthened by controlling the motion of dislocation.

6.1.2 Mechanisms of plastic deformation in metals

Plastic deformation, as explained in earlier section, involves motion of dislocations. There are two prominent mechanisms of plastic deformation, namely *slip* and *twinning*.

Slip is the prominent mechanism of plastic deformation in metals. It involves sliding of blocks of crystal over one other along definite crystallographic planes, called slip planes. In physical words it is analogous to a deck of cards when it is pushed from one end. Slip occurs when shear stress applied exceeds a critical value. During slip each atom usually moves same integral number of atomic distances along the slip plane producing a step, but the orientation of the crystal remains the same. Steps observable under microscope as straight lines are called slip lines.

Slip occurs most readily in specific directions (slip directions) on certain crystallographic planes. This is due to limitations imposed by the fact that single crystal remains homogeneous after deformation. Generally slip plane is the plane of greatest atomic density, and the slip direction is the close packed direction within the slip plane. It turns out that the planes of the highest atomic density are the most widely spaced planes, while

the close packed directions have the smallest translation distance. Feasible combination of a slip plane together with a slip direction is considered as a slip system. The common slip systems are given in *table-6.1*.

Table-6.1: *Slip systems for different crystal structures.*

Crystal	Occurrence	Slip planes	Slip directions
FCC		{111}	<110>
BCC	More common	{110}	<111>
	Less common	{112},{123}	
HCP	More common	Basal plane	Close packed
	Less common	Prismatic & Pyramidal planes	directions
NaCl		{110}	<110>

In a single crystal, plastic deformation is accomplished by the process called slip, and sometimes by twinning. The extent of slip depends on many factors including external load and the corresponding value of shear stress produced by it, the geometry of crystal structure, and the orientation of active slip planes with the direction of shearing stresses generated. Schmid first recognized that single crystals at different orientations but of same material require different stresses to produce slip. The dependence of various factors has been summarized using a parameter – critical resolved shear stress, τ_R, given as

$$\tau_R = \frac{P\cos\lambda}{A/\cos\phi} = \frac{P}{A}\cos\phi\cos\lambda = \sigma\cos\phi\cos\lambda$$
$$\Rightarrow m = \cos\phi\cos\lambda$$

where P – external load applied, A – cross-sectional area over which the load applied, λ – angle between slip direction and tensile axis, ϕ – angle between normal to the slip plane and the tensile axis and m – Schmid factor.

Shear stress is maximum for the condition where $\lambda = \phi = 45°$. If either of the angles are equal to 90°, resolved shear stress will be zero, and thus no slip occurs. If the conditions are such that either of the angles is close to 90°, crystal will tend to fracture rather than slip. Single crystal metals and alloys are used mainly for research purpose and only in a few cases of engineering applications.

Almost all engineering alloys are polycrystalline. Gross plastic deformation of a polycrystalline specimen corresponds to the comparable distortion of the individual grains by means of slip. Although some grains may be oriented favorably for slip, yielding cannot occur unless the unfavorably oriented neighboring grains can also slip. Thus in a polycrystalline aggregate, individual grains provide a mutual geometrical constraint on one other, and this precludes plastic deformation at low applied stresses. That is to initiate plastic deformation, polycrystalline metals require higher stresses than for equivalent single crystals, where stress depends on orientation of the crystal. Much of this increase is attributed to geometrical reasons.

Slip in polycrystalline material involves generation, movement and (re-)arrangement of dislocations. Because of dislocation motion on different planes in various directions, they may interact as well. This interaction can cause dislocation immobile or mobile at higher stresses. During deformation, mechanical integrity and coherency are maintained along the grain boundaries; that is, the grain boundaries are constrained, to some degree, in the shape it may assume by its neighboring grains. Once the yielding has occurred, continued plastic deformation is possible only if enough slip systems are simultaneously operative so as to accommodate grain shape changes while maintaining grain boundary integrity. According to von Mises criterion, a minimum of five independent slip systems must be operative for a polycrystalline solid to exhibit ductility and maintain grain boundary integrity. This arises from the fact that an arbitrary deformation is specified by the six components of strain tensor, but because of requirement of constant volume, there are only independent strain components. Crystals which do not possess five independent slip systems are never ductile in polycrystalline form, although small plastic elongation may be noticeable because of twinning or a favorable preferred orientation.

The second important mechanism of plastic deformation is <u>twinning</u>. It results when a portion of crystal takes up an orientation that is related to the orientation of the rest of the untwined lattice in a definite, symmetrical way. The twinned portion of the crystal is a mirror image of the parent crystal. The plane of symmetry is called twinning plane. Each atom in the twinned region moves by a homogeneous shear a distance proportional to its distance from the twin plane. The lattice strains involved in twinning are small, usually in order of fraction of inter-atomic distance, thus resulting in very small gross plastic deformation. The important role of twinning in plastic deformation is that it causes changes in plane orientation so that further slip can occur. If the surface is polished, the twin would be still visible after etching because it possesses a different orientation from the untwined region. This is in contrast with slip, where slip lines can be removed by polishing the specimen.

Twinning also occurs in a definite direction on a specific plane for each crystal structure. However, it is not known if there exists resolved shear stress for twinning. Twinning generally occurs when slip is restricted, because the stress necessary for twinning is usually higher than that for slip. Thus, some HCP metals with limited number of slip systems may preferably twin. Also, BCC metals twin at low temperatures because slip is difficult. Of course, twinning and slip may occur sequentially or even concurrently in some cases. Twinning systems for some metals are given in table-6.2.

Table-6.2: *Twin systems for different crystal structures.*

Crystal	Example	Twin plane	Twin direction
FCC	Ag, Au, Cu	(111)	[112]
BCC	α-Fe, Ta	(112)	[111]
HCP	Zn, Cd, Mg, Ti	(10¯12)	[¯1011]

Figure-6.1 presents schematic movement of atoms during plastic deformation in slip and during twinning.

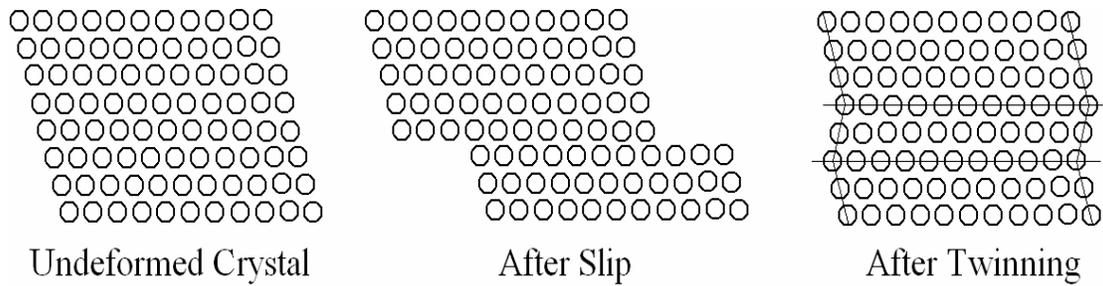Undeformed Crystal   After Slip   After Twinning

Figure-6.1: *Schematic presentation of different plastic deformation mechanism.*

In *table-6.3*, both the mechanisms of plastic deformations are compared with respect to their characteristics.

Table-6.3: *Comparison of mechanism of plastic deformation.*

	during/in slip	**during/in twinning**
Crystal orientation	Same above and below the slip plane	Differ across the twin plane
Size (in terms of inter-atomic distance)	Multiples	Fractions
Occurs on	Widely spread planes	Every plane of region involved
Time required	Milli seconds	Micro seconds
Occurrence	On many slip systems simultaneously	On a particular plane for each crystal

6.2 Strengthening mechanisms in Metals

Ability of a metal to deform plastically depends on ease of dislocation motion under applied external stresses. As mentioned in earlier section, strengthening of a metal consist hindering dislocation motion. Dislocation motion can be hindered in many ways, thus are strengthening mechanisms in metals. Strengthening by methods of grain-size reduction, solid-solution alloying and strain hardening applies for single-phase metals. Precipitation hardening, dispersion hardening, fiber strengthening and Martensite strengthening are applicable to multi-phase metallic materials.

6.2.1 Strengthening by Grain Size Reduction

This strengthening mechanism is based on the fact that crystallographic orientation changes abruptly in passing from one grain to the next across the grain boundary. Thus it is difficult for a dislocation moving on a common slip plane in one crystal to pass over to a similar slip plane in another grain, especially if the orientation is very misaligned. In addition, the crystals are separated by a thin non-crystalline region, which is the characteristic structure of a large angle grain boundary. Atomic disorder at the boundary causes discontinuity in slip planes. Hence dislocations are stopped by a grain boundary

and pile up against it. The smaller the grain size, the more frequent is the pile up of dislocations. A twin boundary can also act as an obstacle to dislocation motion.

A grain boundary can hinder the dislocation motion in two ways: (1) by forcing the dislocation to change its direction of motion and (2) discontinuity of slip plane because of disorder. Effectiveness of grain boundary depends on its characteristic misalignment, represented by an angle. The ordinary high-angle grain boundary (misalignment > 5˚) represents a region of random misfit between the grains on each side of the boundary. This structure contains grain-boundary dislocations which are immobile. However they group together within the boundary to form a step or grain boundary ledge. These ledges can act as effective sources of dislocations as the stress at end of slip plane may trigger new dislocations in adjacent grains. Small angle grain boundaries (misalignment < 1˚) are considered to be composed of a regular array of dislocations, and are not effective in blocking dislocations.

With decrease in grain size, the mean distance of a dislocation can travel decreases, and soon starts pile up of dislocations at grain boundaries. This leads to increase in yield strength of the material. E.O.Hall and N.J.Petch have derived the following relation, famously known as _Hall-Petch relation_, between yield strength (σ_y) and grain size (d):

$$\sigma_y = \sigma_i + kd^{-1/2}$$

where σ_i is the 'friction stress', representing the overall resistance of the crystal lattice to dislocation movement, k is the 'locking parameter' that measures the relative hardening contribution of the grain boundaries and d is the average grain diameter. Friction stress is interpreted as the stress needed to move unlocked dislocations along the slip plane. It depends strongly on temperature, strain, alloy and impurity content. Locking parameter is known to be independent of temperature. Thus friction stress and locking parameters are constants for particular material.

It is important to note that the above relation is not valid for both very large grain and extremely fine grain sizes. Grain size reduction improves not only strength, but also the toughness of many alloys. Grain size can be controlled by rate of cooling, and also by plastic deformation followed by appropriate heat treatment.

Grain size is usually measured using a light microscope to observe a polished specimen by counting the number of grains within a given area, by determining the number of grains that intersect a given length of random line, or by comparing with standard-grain-size charts. If d is average grain diameter, S_v is grain boundary area per unit volume, N_L is mean number of intercepts of grain boundaries per unit length of test line, N_A is number of grains per unit area on a polished surface; the all these are related as follows:

$$S_v = 2N_L, \quad d = \frac{3}{S_v} = \frac{3}{2N_L} \quad \text{and} \quad d = \sqrt{\frac{6}{\pi N_A}}$$

Another common method of measuring the grain size is by comparing the grains at a fixed magnification with standard grain size charts. Charts are coded with ASTM grain size number, G, and is related with n_a – number of grains per mm^2 at 1X magnification as

$$G = -2.9542 + 1.4427 \ln n_a$$

G represents number of grains per square inch (645 mm^2) at a magnification of 100X is equal to 2^{G-1}. Higher the ASTM grain number, smaller is the grain diameter. Grain diameter, D (in mm), and ASTM number, G, can be related as follows:

$$D = \frac{1}{100} \sqrt{\frac{645}{2^{G-1}}}$$

6.2.2 Solid Solution Strengthening

Adding atoms of another element that those occupy interstitial or substitutional positions in parent lattice increases the strength of parent material. This is because stress fields generated around the solute atoms interact with the stress fields of a moving dislocation, thereby increasing the stress required for plastic deformation i.e. the impurity atoms cause lattice strain which can "anchor" dislocations. This occurs when the strain caused by the alloying element compensates that of the dislocation, thus achieving a state of low potential energy. Since solid-solution alloy additions affect the entire stress-strain curve, it can be said that solute atoms have more influence on the frictional resistance to dislocation motion than on the static locking of dislocations. Pure metals are almost always softer than their alloys. Solute strengthening effectiveness depends on two factors – size difference between solute and parent atoms, and concentration of solute atoms.

Solute atoms are two categories with respect to their relative strengthening effect – (1) those produce non-spherical distortions, such as most interstitial atoms, have a relative strengthening effect per unit concentration of about three times their shear modulus, (2) those produce spherical distortion, such as substitutional atoms, have a relative strengthening of about G/10.

Solute atoms interact with dislocations in many ways, namely: elastic interaction; modulus interaction; stacking-fault interaction; electrical interaction; short-range order interaction; and long-range order interaction. Elastic, modulus, and long-range order interactions are of long-range i.e. they are relatively insensitive to temperature and continue to act about 0.6 T_m where T_m is the melting temperature in absolute Kelvin degrees.

Elastic interaction results from mutual interaction of elastic stress fields, while *modulus interaction* occurs if the presence of a solute atom locally alters the modulus of the crystal. *Stacking-fault interactions* arise because solute atoms may segregate to the

stacking-faults, thus lowering stacking-fault energy and widening partial dislocations. This interaction is also called as Suzuki or Chemical interaction. *Electrical interaction* arises if solute atoms of dissimilar valence interact with dislocations which have electrical dipoles. *Short-range order interaction* arises from the tendency for solute atoms to arrange themselves so that they have more then the equilibrium number of dissimilar neighbors. The opposite of short-range order is clustering. *Long-range order interaction* arises in alloys which form super-lattices, in which long-range periodic arrangement of dissimilar atoms gets disturbed, to form anti-phase boundaries, because of dislocation motion which leads to dissociation of dislocation into pairs of ordinary dislocations.

Some polycrystalline metals, such as mild steel, display a discrete yield point type of behavior where a higher stress is necessary to initiate plastic flow than to continue it. Thus, there exists a localized, heterogeneous type of transition from elastic to plastic deformation which produces a yield point in the stress-strain curve i.e. elastic-plastic transition is very well demarked and occurs abruptly in what is called *yield-point phenomenon*. During loading the load increases steadily with elastic strain, drops suddenly at the upper yield point where plastic deformation gets initiated. Continued deformation fluctuates slightly about some constant stress value – lower yield point. Subsequently stress increase with increasing strain. Elongation that occurs at constant load is called the yield-point elongation. Yield strength for metals with this phenomenon is taken as average of lower yield point, thus it is not necessary to employ strain offset method. A schematic picture in *figure-6.2* presents the exaggerated version of yield-point phenomenon.
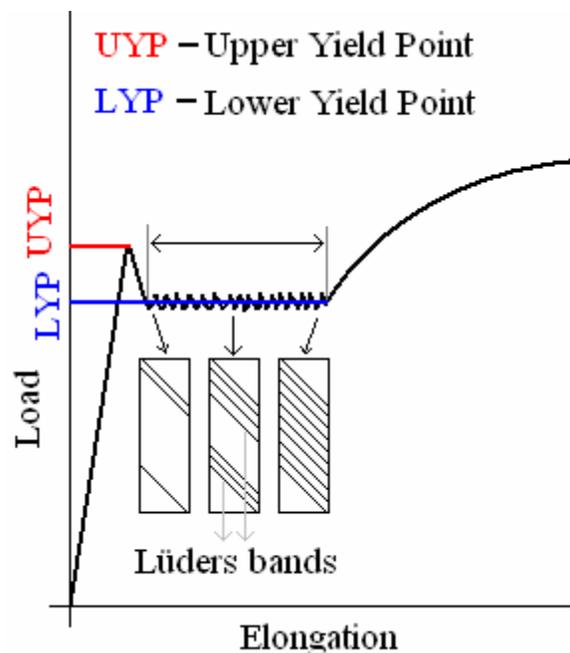

Figure-6.2: *Yield point phenomenon.*

At the upper yield point, deformed part of metal forms a discrete band usually visible to eye at a stress concentration. Many bands may form at the same time at different stress

concentration sites. Each band then propagates along the length of the specimen, causing yield-point elongation. These bands are called Lüders bands / Hartmann lines / stretcher stains, and generally are approximately 45 to the tensile axis. Occurrence of yield point is associated with presence of small amounts of interstitial or substitutional impurities. It's been found that either unlocking of dislocations by a high stress for the case of strong pinning or generation of new dislocations are the reasons for yield-point phenomenon. Thus when the dislocation line is pulled free from the influence of solute atoms, slip can occur at a lower stress. On the other hand, when dislocations are strongly pinned, new dislocations will generate followed by drop in flow stress. This is the origin of upper yield stress. Released dislocations tend to pile-up at grain boundaries, producing stress concentration which in addition to applied stress unlocks sources in next grain. This is the way band propagates. Magnitude of yield-point effect will depend on energy of interaction between solute atoms and dislocations and on the concentration of solute atoms at the dislocations.

6.2.3 Strain Hardening

Two most important industrial processes used to harden metals or alloys are: strain hardening and heat treatment. Strain hardening is used for hardening/strengthening materials that are not responsive to heat treatment. The phenomenon where ductile metals become stronger and harder when they are deformed plastically is called strain hardening *or* work hardening.

Intensity of strain hardening can be gaged from the slope of the flow curve, defined by the parameter strain hardening exponent, *n*. It is measure of the ability of a metal to strain harden. For a given amount of plastic strain, higher the value of n, greater is the strain hardening. Increasing temperature lowers the rate of strain hardening, and thus the treatment is given, usually, at temperatures well below the melting point of the material. Thus the treatment is also known as *cold working*. Most metals strain hardens at room temperature. The consequence of strain hardening a material is improved strength and hardness but material's ductility will be reduced.

As mentioned in earlier chapters, with plastic deformation dislocation density increases because under applied stress dislocation sources, like Frank-Reed source, becomes active. Thus lightly cold worked material may have a dislocation density in order of 10^{12} m^{-2}, while a very heavily cold worked material may have 10^{16} m^{-2} against 10^{10} m^{-2} for annealed material. High density of dislocations and thus increased interaction under applied loads during cold working is the cause for increase in strength of a material. In general rate of strain hardening is lower for HCP metals than for cubic metals where higher number of slip systems is active at an instant of time. Empirically shear stress (τ) to move a dislocation that increases with increase in dislocation density (ρ) is given as:

$$\tau = \tau_0 + A\sqrt{\rho}$$

It is convenient to express the degree of plastic deformation as percent cold work, defined as:

$$\%CW = \left[\frac{A_0 - A_d}{A_0}\right]x100$$

where A_0 is the original cross sectional area that experiences deformation, and A_d is the area after deformation.

Strain hardening is used commercially to enhance the mechanical properties of metals during fabrication procedures. In addition to mechanical properties, physical properties of a material also changes during cold working. There is usually a small decrease in density, an appreciable decrease in electrical conductivity, small increase in thermal coefficient of expansion and increased chemical reactivity (decrease in corrosion resistance).

The cold worked state is a condition of higher internal energy than the un-deformed metal. Although cold worked dislocation cell structure is mechanically stable, it is not thermodynamically stable. With increase in temperature state becomes more unstable, eventually reverts to strain-free condition. This process of heating to attain strain-free condition is called annealing heat treatment where effects of strain hardening may be removed. Annealing process can be divided into three distinct processes: recovery, recrystallization and grain growth. All these steps of the heat treatment process are explained in next section. It is usual industrial practice to use alternate cycles of strain hardening and annealing to deform most metals to a very great extent.

6.2.4-5 Precipitation Hardening and Dispersion strengthening

Small second-phase particles distributed in a ductile matrix can hinder the dislocation motion and thus increase the strength of a material. Second-phase particles either can be introduced by mixing and consolidation (dispersion strengthening) or precipitated in solid state (precipitation hardening).

Precipitation hardening or age hardening is produced by solution treating and quenching an alloy. Term 'Age hardening' is used to describe the process because strength develops with time. Requisite for precipitation hardening to take place is that second phase must be soluble at an elevated temperature but precipitates upon quenching and aging at a lower temperature. This limits the alloy systems which can be strengthened by precipitation hardening. For example: Al-alloys, Cu-Be alloys, Mg-Al alloys, Cu-Sn alloys. If the precipitation occurs at normal ambient temperatures, it is called natural aging. Some alloy systems needed to be aged at higher temperatures and the process is known as artificial aging. Most precipitation hardened alloys are limited in their maximum service temperatures, which may lose their strength at elevated temperatures due to over-aging.

In dispersion strengthening, hard particles are mixed with matrix powder and consolidated and processed by powder metallurgy techniques. Here second phase shall have very little solubility in the matrix, even at elevated temperatures. Because there is

very little solubility, the particles resist growth or over-aging to a much greater extent than the second phase particles in a precipitation hardening system. Theoretically, at least, it is possible to produce infinite number of dispersion-hardened systems by mixing finely divided metallic powders and second phase particles (oxides, carbides, nitrides, borides, etc).

Dislocations moving through the matrix, which is either precipitation hardened or dispersion strengthened, have two alternatives. They can either cut through the precipitate particles *or* bend around and bypass them. The first alternative is possible only when the slip plane is continuous from the matrix through the precipitate particle and when the stress to move a dislocation in precipitate is comparable to that in matrix. Cutting of particles is easier for small particles which can be considered as segregated solute atoms. There are many particle properties that can dictate the ease of shearing, for example: coherency strains, stacking-fault energy, ordered structure, modulus effect, interfacial energy, morphology and lattice friction stress.

Cutting of particles is not possible when there is an interface or an abrupt change in orientation. Under such instances, dislocations have to bend around them, and bypass. The mechanism involved in this instance is similar to the operation of a Frank-Reed source. Stress required to bend a dislocation is inversely proportional to the average interspacing (λ) of particles. Effective strengthening is achieved in the bending process, when the particles are submicroscopic in size.

$$\tau = Gb \Big/ \lambda$$

The degree of strengthening from second phase particles depends on particle distribution in the matrix. Particle dispersion along with the shape can be expressed by specifying the volume fraction, average particle diameter, and mean interspacing of particles which are interrelated. A simple expression for interspacing of particles is:

$$\lambda = \frac{4(1-f)r}{3f}$$

where f is the volume fraction of spherical particles of radius r.

The spacing between second phase particles should be typically a few hundred angstroms. Optimum strengthening occurs during aging once the right interspacing of particles is achieved. Smaller the particles, dislocations can cut through them at lower stresses. On the other hand, larger the particles they will be distributed at wider distances. Thus it is necessary to arrest the growth of particles in precipitation hardened alloys at right interspacing. Over-aging allows the fine particles to coalesce into larger and fewer particles. The interspacing is increased thereby and the yield stress decreases.

6.2.6 Fiber strengthening

Second phase material can also be introduced into matrix in form of fibers to strengthen it. However, mechanism of strengthening is different from either precipitation hardening or dispersion strengthening where second phase is introduced as fine particles.

Prerequisites are materials to be used as fibers include high strength and/or high strength-to-weight ratio. Fibers usually, thus, have high strength and high modulus while the matrix must be ductile and non-reactive with the fibers. Fibers may be long and continuous or they may be discontinuous. Examples for fiber material: Al_2O_3, boron, graphite, metal, glass, etc. Examples for matrix material: metals, polymers. Fiber reinforced materials are an important group of materials known as composite materials.

In fiber-reinforced materials, high modulus fibers carry essentially the entire load while the matrix serves to transmit the load to the fibers. Matrix also protects fibers from surface damage, serves to blunt cracks which arise from fiber breakage while it also serves to separate the fibers. It is understood that analysis of precipitation/dispersion strengthening is based on dislocation theory. On the other hand, the analysis of strength of fiber-reinforced material involves the direct application of continuum principles at the microscopic level as the material behavior is essentially elastic.

To achieve any benefit from presence of fibers, empirically it can be shown that there is a need for critical fiber volume which must be exceeded for fiber strengthening to occur.

$$f_{critical} = \frac{\sigma_{mu} - \sigma'_m}{\sigma_{fu} - \sigma'_m}$$

where σ_{mu} – strength of strain hardened matrix, σ'_m – flow stress of matrix at a strain equal to fiber breaking stress, σ_{fu} – ultimate tensile strength of the fiber.

However for small values of critical fiber volume, number of fibers present may not be sufficient to effectively restrain the elongation of the matrix. Thus a minimum volume fraction of fiber which must be exceeded to have real reinforcement is defined as follows:

$$f_{min} = \frac{\sigma_{mu} - \sigma'_m}{\sigma_{fu} + \sigma_{mu} - \sigma'_m}$$

When fiber-reinforced material is loaded, load transmits from matrix to fibers via interface. Thus interface between matrix and fiber play an important role in fiber strengthening.

Usually fibers are arranged in uni-directional manner, thus it is highly anisotropic in nature. One of the consequences of the anisotropy of fiber composites is that they exhibit shear coupling. This means that an axial load produces shear stresses, and shear load produces axial strains. In an isotropic material a uni-axially applied load produces only axial and transverse normal strains. Shear strains in fiber reinforced composites are compensated practically by using a cross-ply laminate.

6.2.7 Martensite Strengthening

This strengthening can be achieved in systems where a diffusion-controlled invariant transformation can be suppressed by rapid cooling. The Martensite strengthening process, thus, basically is a diffusion-less and displacive reaction. The martensitic phase is formed from the retained high temperature phase at temperatures lower then the equilibrium invariant transformation temperature. It occurs by a process of lattices shearing. Martensite under microscope appears as lenticular plates which divide and subdivide the grains of the parent phase. Always touching but never crossing one another. The characteristic lenticular shape minimizes the elastic distortion in the matrix. These platelets grow at about one-third the velocity of sound. This high speed means activation energy for growth is very low, and thus activation energy for nucleation determines the amount of Martensite that can form under given conditions.

Martensite platelets attain their shape by two successive shear displacements contained in boundaries coherent with the parent phase. The first displacement is a homogeneous shear throughout the plate which occurs parallel to a specific plane in the parent phase known as the habit plane. The second displacement, the lesser of the two, can take place by one of two mechanisms: slip as in Fe-C Martensite or twinning as in Fe-Ni Martensite. Martensite formation occurs in many systems like Fe-C, Fe-Ni, Fe-Ni-C, Cu-Zn, Au-Cd, and even in pure metals like Li, Zr and Co. However, only the alloys based on Fe and C show a pronounced strengthening effect.

High strength of Martensite is attributed to its typical crystal structure i.e. effective barriers to slip are provided by the fine twin structure or the high dislocation density. In Fe-C system, carbon atoms are also involved in strengthening. Super saturated carbon atoms strain the ferrite lattice, and this strain can be relieved by redistribution of carbon atoms by diffusion at room temperature. One result is that a strong binding is set up between dislocations and the carbon atoms. This hinders the motion of dislocations, thus increasing the strength.

This strengthening mechanism is one of the most common processes used in engineering materials. The name basically arises from the phase *Martensite* that forms in steels when they are quenched from the solutionizing temperature.

6.3 Recovery, Recrystallization and Grain Growth

As mentioned in earlier sections, annealing is an important industrial process to relieve the stresses from cold working. During cold working grain shape changes, while material strain hardens because of increase in dislocation density. Between 1-10% of the energy of plastic deformation is stored in material in the form of strain energy associated with point defects and dislocations. On annealing i.e. on heating the deformed material to higher temperatures and holding, material tends to lose the extra strain energy and revert to the original condition before deformation by the processes of recovery and recrystallization. Grain growth may follow these in some instances.

6.3.1 Recovery

This is the first stage of restoration after cold working where physical properties of the cold-worked material are restored without any observable change in microstructure. The properties that are mostly affected by recovery ate those sensitive to point defects, for example – thermal and electrical conductivities. During recovery, which takes place at low temperatures of annealing, some of the stored internal energy is relieved by virtue of dislocation motion as a result of enhanced atomic diffusion. There is some reduction, though not substantial, in dislocation density as well apart from formation of dislocation configurations with low strain energies. Excess point defects that are created during deformation are annihilated either by absorption at grain boundaries or dislocation climbing process. Stored energy of cold work is the driving force for recovery.

6.3.2 Recrystallization

This stage of annealing follows after recovery stage. Here also driving force is stored energy of cold work. Even after complete recovery, the grains are still in relatively high strain energy state. This stage, thus, involves replacement of cold-worked structure by a new set of strain-free, approximately equi-axed grains i.e. it is the process of nucleation and growth of new, strain-free crystals to replace all the deformed crystals. It starts on heating to temperatures in the range of 0.3-0.5 T_m, which is above the recovery stage. There is no crystal structure change during recrystallization. This process is characterized by recrystallization temperature which is defined as the temperature at which 50% of material recrystallizes in one hour time. The recrystallization temperature is strongly dependent on the purity of a material. Pure materials may recrystallizes around 0.3 T_m, while impure materials may recrystallizes around 0.5-0.7 T_m.

There are many variables that influence recrystallization behavior, namely amount of prior deformation, temperature, time, initial grain size, composition and amount of recovery prior to the start of the recrystallization. This dependence leads to following empirical laws:

- A minimum amount of deformation is needed to cause recrystallization.
- Smaller the degree of deformation, higher will be the recrystallization temperature.
- The finer is the initial grain size; lower will be the recrystallization temperature.
- The larger the initial grain size, the greater degree of deformation is required to produce an equivalent recrystallization temperature.
- Greater the degree of deformation and lower the annealing temperature, the smaller will be the recrystallized grain size.
- The higher is the temperature of cold working, the less is the strain energy stored and thus recrystallization temperature is correspondingly higher.
- The recrystallization rate increases exponentially with temperature.

During recrystallization, the mechanical properties that were changes during deformation are restored to their pre-cold-work values. Thus material becomes softer, weaker and ductile. During this stage of annealing impurity atoms tend to segregate at grain boundaries, and retard their motion and obstruct the processes of nucleation and growth. This *solute drag effect* can be used to retain cold worked strength at higher service temperatures. Presence of second phase particles causes slowing down of recrystallization – *pinning action of the particles*.

6.3.3 Grain growth

This stage follows complete crystallization if the material is left at elevated temperatures. However, grain growth does not need to be preceded by recovery and recrystallization; it may occur in all polycrystalline materials. During this stage newly formed strain-free grains tend to grow in size. This grain growth occurs by the migration of grain boundaries. Driving force for this process is reduction in grain boundary energy i.e. decreasing in free energy of the material. As the grains grow larger, the curvature of the boundaries becomes less. This results in a tendency for larger grains to grow at the expense of smaller grains. In practical applications, grain growth is not desirable. Incorporation of impurity atoms and insoluble second phase particles are effective in retarding grain growth.

Because the driving force for grain growth is lower than the driving force for recrystallization, grain growth occurs slowly at a temperature where recrystallization occurs at substantially high speeds. However, grain growth is strongly temperature dependent.

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Chapter 7. Dislocations and Strengthening Mechanisms

Mechanisms of plastic deformation in metals
What is the role of dislocation in strengthening?
Strengthening mechanisms in Metals
Strengthening by Grain Size Reduction
Solid Solution Strengthening
Strain Hardening
Precipitation Hardening and *Dispersion strengthening*
Fiber strengthening
Martensite Strengthening
Recovery, Recrystallization and Grain Growth
Want to know how good you are in this chapter?

Mechanisms of plastic deformation in metals

Plastic deformation, as explained in earlier section, involves motion of dislocations. There are two prominent mechanisms of plastic deformation, namely *slip* and *twinning*.

Slip is the prominent mechanism of plastic deformation in metals. It involves sliding of blocks of crystal over one other along definite crystallographic planes, called slip planes. In physical words it is analogous to a deck of cards when it is pushed from one end. Slip occurs when shear stress applied exceeds a critical value. During slip each atom usually moves same integral number of atomic distances along the slip plane producing a step, but the orientation of the crystal remains the same. Steps observable under microscope as straight lines are called slip lines.

Slip occurs most readily in specific directions (slip directions) on certain crystallographic planes. This is due to limitations imposed by the fact that single crystal remains homogeneous after deformation. Generally slip plane is the plane of greatest atomic

density, and the slip direction is the close packed direction within the slip plane. It turns out that the planes of the highest atomic density are the most widely spaced planes, while the close packed directions have the smallest translation distance. Feasible combination of a slip plane together with a slip direction is considered as a slip system. The common slip systems are given for different in *table-6.1*.

Table-6.1: *Slip systems for different crystal structures.*

Crystal	Occurrence	Slip planes	Slip directions
FCC		{111}	<110>
BCC	More common	{110}	<111>
	Less common	{112},{123}	
HCP	More common	Basal plane	Close packed
	Less common	Prismatic & Pyramidal planes	directions
NaCl		{110}	<110>

In a single crystal, plastic deformation is accomplished by the process called slip, and sometimes by twinning. The extent of slip depends on many factors including external load and the corresponding value of shear stress produced by it, the geometry of crystal structure, and the orientation of active slip planes with the direction of shearing stresses generated. Schmid first recognized that single crystals at different orientations but of same material require different stresses to produce slip. The dependence of various factors has been summarized using a parameter – critical resolved shear stress, τ_R, given as

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

$$\Rightarrow m = \cos\phi\cos\lambda$$

where P – external load applied, A – cross-sectional area over which the load applied, λ – angle between slip direction and tensile axis, ϕ – angle between normal to the slip plane and the tensile axis and m – Schmid factor.

Shear stress is maximum for the condition where $\lambda = \phi = 45°$. If either of the angles are equal to $90°$, resolved shear stress will be zero, and thus no slip occurs. If the conditions are such that either of the angles is close to $90°$, crystal will tend to fracture rather than slip. Single crystal metals and alloys are used mainly for research purpose and only in a few cases of engineering applications.

Slip in polycrystalline material involves generation, movement and (re-)arrangement of dislocations. Because of dislocation motion on different planes in various directions, they may interact as well. This interaction can cause dislocation immobile or mobile at higher stresses. During deformation, mechanical integrity and coherency are maintained along the grain boundaries; that is, the grain boundaries are constrained, to some degree, in the

shape it may assume by its neighboring grains. Once the yielding has occurred, continued plastic deformation is possible only if enough slip systems are simultaneously operative so as to accommodate grain shape changes while maintaining grain boundary integrity. According to Von Mises criterion, a minimum of five independent slip systems must be operative for a polycrystalline solid to exhibit ductility and maintain grain boundary integrity. This arises from the fact that an arbitrary deformation is specified by the six components of strain tensor, but because of requirement of constant volume, there are only independent strain components. Crystals which do not possess five independent slip systems are never ductile in polycrystalline form, although small plastic elongation may be noticeable because of twinning or a favorable preferred orientation.

The second important mechanism of plastic deformation is <u>twinning</u>. It results when a portion of crystal takes up an orientation that is related to the orientation of the rest of the untwined lattice in a definite, symmetrical way. The twinned portion of the crystal is a mirror image of the parent crystal. The plane of symmetry is called twinning plane. Each atom in the twinned region moves by a homogeneous shear a distance proportional to its distance from the twin plane. The lattice strains involved in twinning are small, usually in order of fraction of inter-atomic distance, thus resulting in very small gross plastic deformation. The important role of twinning in plastic deformation is that it causes changes in plane orientation so that further slip can occur. If the surface is polished, the twin would be still visible after etching because it possesses a different orientation from the untwined region. This is in contrast with slip, where slip lines can be removed by polishing the specimen.

Twinning also occurs in a definite direction on a specific plane for each crystal structure. However, it is not known if there exists resolved shear stress for twinning. Twinning generally occurs when slip is restricted, because the stress necessary for twinning is usually higher than that for slip. Thus, some HCP metals with limited number of slip systems may preferably twin. Also, BCC metals twin at low temperatures because slip is difficult. Of course, twinning and slip may occur sequentially or even concurrently in some cases. Twinning systems for some metals are given in table-6.2.

Table-6.2: *Twin systems for different crystal structures.*

Crystal	Example	Twin plane	Twin direction
FCC	Ag, Au, Cu	(111)	[112]
BCC	α-Fe, Ta	(112)	[111]
HCP	Zn, Cd, Mg, Ti	(10$\bar{1}$2)	[$\bar{1}$1011]

Figure-6.1 presents schematic movement of atoms during plastic deformation in slip and during twinning.
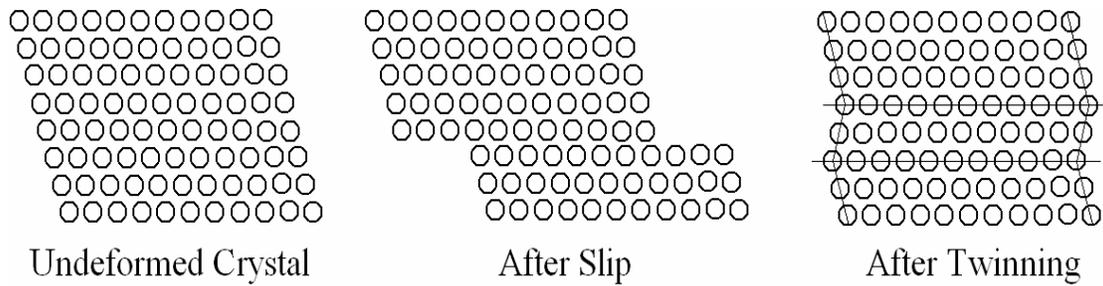
Figure-6.1: *Schematic presentation of different plastic deformation mechanism.*

In *table-6.3*, both the mechanisms of plastic deformations are compared with respect to their characteristics.

Table-6.3: *Comparison of mechanism of plastic deformation.*

	during/in slip	**during/in twinning**
Crystal orientation	Same above and below the slip plane	Differ across the twin plane
Size (in terms of inter-atomic distance)	Multiples	Fractions
Occurs on	Widely spread planes	Every plane of region involved
Time required	Milli seconds	Micro seconds
Occurrence	On many slip systems simultaneously	On a particular plane for each crystal

What is the role of dislocation in strengthening?

It is well known that dislocations can move under applied external stresses. Cumulative movement of dislocations leads to the gross plastic deformation. At microscopic level, dislocation motion involves rupture and reformation of inter-atomic bonds. The necessity of dislocation motion for ease of plastic deformation is well explained by the discrepancy between theoretical strength and real strength of solids. It has been known that one-dimensional crystal defects – dislocations – play an important role in plastic deformation of crystalline solids. Their importance in plastic deformation is relevant to their characteristic nature of motion in specific directions (slip-directions) on specific planes (slip-planes). Increasing the resistance to movement of dislocation by any means will lead to increase in load to cause the deformation.

The onset of plastic deformation involves start of motion of existing dislocations in real crystal, while in perfect crystal it can be attributed to generation of dislocations and subsequently their motion. During the motion, dislocations will tend to interact among themselves. Dislocation interaction is very complex as number of dislocations moving on

number of slip planes in various directions. When they are in the same plane, they repel each other if they have the same sign, and annihilate if they have opposite signs (leaving behind a perfect crystal). An important consequence interaction of dislocations that are not on parallel planes is that they intersect each other or inhibit each others motion. Intersection of two dislocations results in a sharp break in the dislocation line. These breaks can be of two kinds:

(a) A jog is break in dislocation line moving it out of slip plane.
(b) A kink is break in dislocation line that remains in slip plane.

Other hindrances to dislocation motion include interstitial and substitutional atoms, foreign particles, grain boundaries, external grain surface, and change in structure due to phase change. Important practical consequences of hindrance of dislocation motion are that dislocations are still movable but at higher stresses (or forces), and in most instances that leads to generation of more dislocations. Dislocations can spawn from existing dislocations, and from defects, grain boundaries and surface irregularities. Thus, the number of dislocations increases dramatically during plastic deformation. As further motion of dislocations requires increase of stress, material can be said to be strengthened i.e. materials can be strengthened by controlling the motion of dislocation.

Strengthening mechanisms in Metals

As mentioned in earlier section, strengthening of a metal consist hindering dislocation motion. Dislocation motion can be hindered in many ways, thus are strengthening mechanisms in metals. Strengthening by methods of grain-size reduction, solid-solution alloying and strain hardening applies for single-phase metals. Precipitation hardening, dispersion hardening, fiber strengthening and Martensite strengthening are applicable to multi-phase metallic materials.

Strengthening by Grain Size Reduction

This strengthening mechanism is based on the fact that crystallographic orientation changes abruptly in passing from one grain to the next across the grain boundary. Thus it is difficult for a dislocation moving on a common slip plane in one crystal to pass over to a similar slip plane in another grain, especially if the orientation is very misaligned. In addition, the crystals are separated by a thin non-crystalline region, which is the characteristic structure of a large angle grain boundary. Atomic disorder at the boundary causes discontinuity in slip planes. Hence dislocations are stopped by a grain boundary and pile up against it. The smaller the grain size, the more frequent is the pile up of dislocations. A twin boundary can also act as an obstacle to dislocation motion.

A grain boundary can hinder the dislocation motion in two ways: (1) by forcing the dislocation to change its direction of motion and (2) discontinuity of slip plane because of disorder. Effectiveness of grain boundary depends on its characteristic misalignment, represented by an angle. The ordinary high-angle grain boundary (misalignment > 5˚) represents a region of random misfit between the grains on each side of the boundary.

This structure contains grain-boundary dislocations which are immobile. However they group together within the boundary to form a step or grain boundary ledge. These ledges can act as effective sources of dislocations as the stress at end of slip plane may trigger new dislocations in adjacent grains. Small angle grain boundaries (misalignment < 1˚) are considered to be composed of a regular array of dislocations, and are not effective in blocking dislocations.

With decrease in grain size, the mean distance of a dislocation can travel decreases, and soon starts pile up of dislocations at grain boundaries. This leads to increase in yield strength of the material. E.O.Hall and N.J.Petch have derived the following relation, famously known as *Hall-Petch relation*, between yield strength (σ_y) and grain size (d):

$$\sigma_y = \sigma_i + kd^{-1/2}$$

where σ_i is the 'friction stress', representing the overall resistance of the crystal lattice to dislocation movement, k is the 'locking parameter' that measures the relative hardening contribution of the grain boundaries and d is the average grain diameter. Friction stress is interpreted as the stress needed to move unlocked dislocations along the slip plane. It depends strongly on temperature, strain, alloy and impurity content. Locking parameter is known to be independent of temperature. Thus friction stress and locking parameters are constants for particular material.

It is important to note that the above relation is not valid for both very large grain and extremely fine grain sizes. Grain size reduction improves not only strength, but also the toughness of many alloys. Grain size can be controlled by rate of cooling, and also by plastic deformation followed by appropriate heat treatment.

Grain size is usually measured using a light microscope to observe a polished specimen by counting the number of grains within a given area, by determining the number of grains that intersect a given length of random line, or by comparing with standard-grain-size charts. If d is average grain diameter, S_v is grain boundary area per unit volume, N_L is mean number of intercepts of grain boundaries per unit length of test line, N_A is number of grains per unit area on a polished surface; the all these are related as follows:

$$S_v = 2N_L, \quad d = \frac{3}{S_v} = \frac{3}{2N_L} \quad \text{and} \quad d = \sqrt{\frac{6}{\pi N_A}}$$

Another common method of measuring the grain size is by comparing the grains at a fixed magnification with standard grain size charts. Charts are coded with ASTM grain size number, G, and is related with n_a – number of grains per mm^2 at 1X magnification as

$$G = -2.9542 + 1.4427 \ln n_a$$

G represents number of grains per square inch (645 mm^2) at a magnification of 100X is equal to 2^{G-1}. Higher the ASTM grain number, smaller is the grain diameter. Grain diameter, D (in mm), and ASTM number, G, can be related as follows:

$$D = \frac{1}{100} \sqrt{\frac{645}{2^{G-1}}}$$

Solid Solution Strengthening

Adding atoms of another element that those occupy interstitial or substitutional positions in parent lattice increases the strength of parent material. This is because stress fields generated around the solute atoms interact with the stress fields of a moving dislocation, thereby increasing the stress required for plastic deformation i.e. the impurity atoms cause lattice strain which can "anchor" dislocations. This occurs when the strain caused by the alloying element compensates that of the dislocation, thus achieving a state of low potential energy. Since solid-solution alloy additions affect the entire stress-strain curve, it can be said that solute atoms have more influence on the frictional resistance to dislocation motion than on the static locking of dislocations. Pure metals are almost always softer than their alloys. Solute strengthening effectiveness depends on two factors – size difference between solute and parent atoms, and concentration of solute atoms.

Solute atoms are two categories with respect to their relative strengthening effect – (1) those produce non-spherical distortions, such as most interstitial atoms, have a relative strengthening effect per unit concentration of about three times their shear modulus, (2) those produce spherical distortion, such as substitutional atoms, have a relative strengthening of about G/10.

Solute atoms interact with dislocations in many ways, namely: elastic interaction; modulus interaction; stacking-fault interaction; electrical interaction; short-range order interaction; and long-range order interaction. Elastic, modulus, and long-range order interactions are of long-range i.e. they are relatively insensitive to temperature and continue to act about 0.6 T_m where T_m is the melting temperature in absolute Kelvin degrees.

Elastic interaction results from mutual interaction of elastic stress fields, while *modulus interaction* occurs if the presence of a solute atom locally alters the modulus of the crystal. *Stacking-fault interactions* arise because solute atoms may segregate to the stacking-faults, thus lowering stacking-fault energy and widening partial dislocations. This interaction is also called as Suzuki or Chemical interaction. *Electrical interaction* arises if solute atoms of dissimilar valence interact with dislocations which have electrical dipoles. *Short-range order interaction* arises from the tendency for solute atoms to arrange themselves so that they have more then the equilibrium number of dissimilar neighbors. The opposite of short-range order is clustering. *Long-range order interaction* arises in alloys which form super-lattices, in which long-range periodic arrangement of

dissimilar atoms gets disturbed, to form anti-phase boundaries, because of dislocation motion which leads to dissociation of dislocation into pairs of ordinary dislocations.

Some polycrystalline metals, such as mild steel, display a discrete yield point type of behavior where a higher stress is necessary to initiate plastic flow than to continue it. Thus, there exists a localized, heterogeneous type of transition from elastic to plastic deformation which produces a yield point in the stress-strain curve i.e. elastic-plastic transition is very well demarked and occurs abruptly in what is called *yield-point phenomenon*. During loading the load increases steadily with elastic strain, drops suddenly at the upper yield point where plastic deformation gets initiated. Continued deformation fluctuates slightly about some constant stress value – lower yield point. Subsequently stress increase with increasing strain. Elongation that occurs at constant load is called the yield-point elongation. Yield strength for metals with this phenomenon is taken as average of lower yield point, thus it is not necessary to employ strain offset method. A schematic picture in *figure-6.2* presents the exaggerated version of yield-point phenomenon.
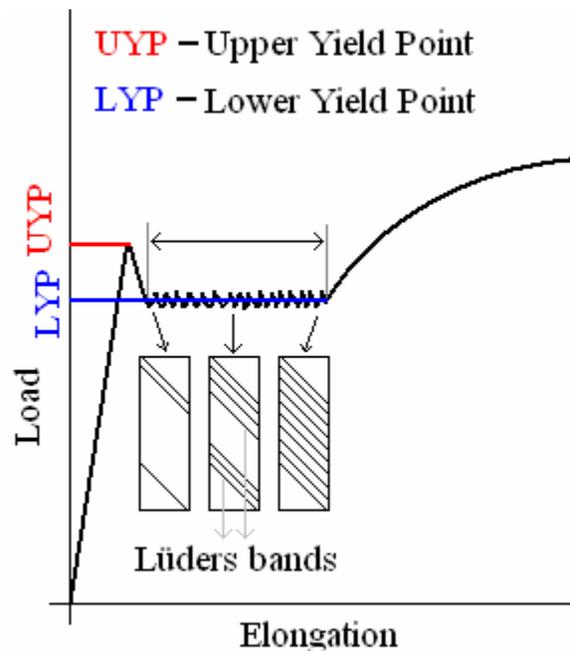

Figure-6.2: *Yield point phenomenon.*

At the upper yield point, deformed part of metal forms a discrete band usually visible to eye at a stress concentration. Many bands may form at the same time at different stress concentration sites. Each band then propagates along the length of the specimen, causing yield-point elongation. These bands are called Lüders bands / Hartmann lines / stretcher stains, and generally are approximately 45 to the tensile axis. Occurrence of yield point is associated with presence of small amounts of interstitial or substitutional impurities. It's been found that either unlocking of dislocations by a high stress for the case of strong pinning or generation of new dislocations are the reasons for yield-point phenomenon. Thus when the dislocation line is pulled free from the influence of solute atoms, slip can

occur at a lower stress. On the other hand, when dislocations are strongly pinned, new dislocations will generate followed by drop in flow stress. This is the origin of upper yield stress. Released dislocations tend to pile-up at grain boundaries, producing stress concentration which in addition to applied stress unlocks sources in next grain. This is the way band propagates. Magnitude of yield-point effect will depend on energy of interaction between solute atoms and dislocations and on the concentration of solute atoms at the dislocations.

Strain Hardening

Two most important industrial processes used to harden metals or alloys are: strain hardening and heat treatment. Strain hardening is used for hardening/strengthening materials that are not responsive to heat treatment. The phenomenon where ductile metals become stronger and harder when they are deformed plastically is called strain hardening *or* work hardening.

Intensity of strain hardening can be gaged from the slope of the flow curve, defined by the parameter strain hardening exponent, *n*. It is measure of the ability of a metal to strain harden. For a given amount of plastic strain, higher the value of n, greater is the strain hardening. Increasing temperature lowers the rate of strain hardening, and thus the treatment is given, usually, at temperatures well below the melting point of the material. Thus the treatment is also known as *cold working*. Most metals strain hardens at room temperature. The consequence of strain hardening a material is improved strength and hardness but material's ductility will be reduced.

As mentioned in earlier chapters, with plastic deformation dislocation density increases because under applied stress dislocation sources, like Frank-Reed source, becomes active. Thus lightly cold worked material may have a dislocation density in order of 10^{12} m^{-2}, while a very heavily cold worked material may have 10^{16} m^{-2} against 10^{10} m^{-2} for annealed material. High density of dislocations and thus increased interaction under applied loads during cold working is the cause for increase in strength of a material. In general rate of strain hardening is lower for HCP metals than for cubic metals where higher number of slip systems is active at an instant of time. Empirically shear stress (τ) to move a dislocation that increases with increase in dislocation density (ρ) is given as:

$$\tau = \tau_0 + A\sqrt{\rho}$$

It is convenient to express the degree of plastic deformation as percent cold work, defined as:

$$\%CW = \left[\frac{A_0 - A_d}{A_0}\right] x100$$

where A_0 is the original cross sectional area that experiences deformation, and A_d is the area after deformation.

Strain hardening is used commercially to enhance the mechanical properties of metals during fabrication procedures. In addition to mechanical properties, physical properties of a material also changes during cold working. There is usually a small decrease in density, an appreciable decrease in electrical conductivity, small increase in thermal coefficient of expansion and increased chemical reactivity (decrease in corrosion resistance).

The cold worked state is a condition of higher internal energy than the un-deformed metal. Although cold worked dislocation cell structure is mechanically stable, it is not thermodynamically stable. With increase in temperature state becomes more unstable, eventually reverts to strain-free condition. This process of heating to attain strain-free condition is called annealing heat treatment where effects of strain hardening may be removed. Annealing process can be divided into three distinct processes: recovery, recrystallization and grain growth. All these steps of the heat treatment process are explained in next section. It is usual industrial practice to use alternate cycles of strain hardening and annealing to deform most metals to a very great extent.

Precipitation Hardening and *Dispersion strengthening*

Small second-phase particles distributed in a ductile matrix can hinder the dislocation motion and thus increase the strength of a material. Second-phase particles either can be introduced by mixing and consolidation (dispersion strengthening) or precipitated in solid state (precipitation hardening).

Precipitation hardening or age hardening is produced by solution treating and quenching an alloy. Term 'Age hardening' is used to describe the process because strength develops with time. Requisite for precipitation hardening to take place is that second phase must be soluble at an elevated temperature but precipitates upon quenching and aging at a lower temperature. This limits the alloy systems which can be strengthened by precipitation hardening. For example: Al-alloys, Cu-Be alloys, Mg-Al alloys, Cu-Sn alloys. If the precipitation occurs at normal ambient temperatures, it is called natural aging. Some alloy systems needed to be aged at higher temperatures and the process is known as artificial aging. Most precipitation hardened alloys are limited in their maximum service temperatures, which may lose their strength at elevated temperatures due to over-aging.

In dispersion strengthening, hard particles are mixed with matrix powder and consolidated and processed by powder metallurgy techniques. Here second phase shall have very little solubility in the matrix, even at elevated temperatures. Because there is very little solubility, the particles resist growth or over-aging to a much greater extent than the second phase particles in a precipitation hardening system. Theoretically, at least, it is possible to produce infinite number of dispersion-hardened systems by mixing finely divided metallic powders and second phase particles (oxides, carbides, nitrides, borides, etc).

Dislocations moving through the matrix, which is either precipitation hardened or dispersion strengthened, have two alternatives. They can either cut through the precipitate particles *or* bend around and bypass them. The first alternative is possible only when the slip plane is continuous from the matrix through the precipitate particle and when the stress to move a dislocation in precipitate is comparable to that in matrix. Cutting of particles is easier for small particles which can be considered as segregated solute atoms. There are many particle properties that can dictate the ease of shearing, for example: coherency strains, stacking-fault energy, ordered structure, modulus effect, interfacial energy, morphology and lattice friction stress.

Cutting of particles is not possible when there is an interface or an abrupt change in orientation. Under such instances, dislocations have to bend around them, and bypass. The mechanism involved in this instance is similar to the operation of a Frank-Reed source. Stress required to bend a dislocation is inversely proportional to the average interspacing (λ) of particles. Effective strengthening is achieved in the bending process, when the particles are submicroscopic in size.

$$\tau = Gb\big/\lambda$$

The degree of strengthening from second phase particles depends on particle distribution in the matrix. Particle dispersion along with the shape can be expressed by specifying the volume fraction, average particle diameter, and mean interspacing of particles which are interrelated. A simple expression for interspacing of particles is:

$$\lambda = \frac{4(1-f)r}{3f}$$

where f is the volume fraction of spherical particles of radius r.

The spacing between second phase particles should be typically a few hundred angstroms. Optimum strengthening occurs during aging once the right interspacing of particles is achieved. Smaller the particles, dislocations can cut through them at lower stresses. On the other hand, larger the particles they will be distributed at wider distances. Thus it is necessary to arrest the growth of particles in precipitation hardened alloys at right interspacing. Over-aging allows the fine particles to coalesce into larger and fewer particles. The interspacing is increased thereby and the yield stress decreases.

Fiber strengthening

Second phase material can also be introduced into matrix in form of fibers to strengthen it. However, mechanism of strengthening is different from either precipitation hardening or dispersion strengthening where second phase is introduced as fine particles.

Prerequisites are materials to be used as fibers include high strength and/or high strength-to-weight ratio. Fibers usually, thus, have high strength and high modulus while the

matrix must be ductile and non-reactive with the fibers. Fibers may be long and continuous or they may be discontinuous. Examples for fiber material: Al_2O_3, boron, graphite, metal, glass, etc. Examples for matrix material: metals, polymers. Fiber reinforced materials are an important group of materials known as composite materials.

In fiber-reinforced materials, high modulus fibers carry essentially the entire load while the matrix serves to transmit the load to the fibers. Matrix also protects fibers from surface damage, serves to blunt cracks which arise from fiber breakage while it also serves to separate the fibers. It is understood that analysis of precipitation/dispersion strengthening is based on dislocation theory. On the other hand, the analysis of strength of fiber-reinforced material involves the direct application of continuum principles at the microscopic level as the material behavior is essentially elastic.

To achieve any benefit from presence of fibers, empirically it can be shown that there is a need for critical fiber volume which must be exceeded for fiber strengthening to occur.

$$f_{critical} = \frac{\sigma_{mu} - \sigma'_m}{\sigma_{fu} - \sigma'_m}$$

where σ_{mu} – strength of strain hardened matrix, σ'_m – flow stress of matrix at a strain equal to fiber breaking stress, σ_{fu} – ultimate tensile strength of the fiber.

However for small values of critical fiber volume, number of fibers present may not be sufficient to effectively restrain the elongation of the matrix. Thus a minimum volume fraction of fiber which must be exceeded to have real reinforcement is defined as follows:

$$f_{min} = \frac{\sigma_{mu} - \sigma'_m}{\sigma_{fu} + \sigma_{mu} - \sigma'_m}$$

When fiber-reinforced material is loaded, load transmits from matrix to fibers via interface. Thus interface between matrix and fiber play an important role in fiber strengthening.

Usually fibers are arranged in uni-directional manner, thus it is highly anisotropic in nature. One of the consequences of the anisotropy of fiber composites is that they exhibit shear coupling. This means that an axial load produces shear stresses, and shear load produces axial strains. In an isotropic material a uni-axially applied load produces only axial and transverse normal strains. Shear strains in fiber reinforced composites are compensated practically by using a cross-ply laminate.

Martensite Strengthening

This strengthening can be achieved in systems where a diffusion-controlled invariant transformation can be suppressed by rapid cooling. The Martensite strengthening process, thus, basically is a diffusion-less and displacive reaction. The martensitic phase is formed

from the retained high temperature phase at temperatures lower then the equilibrium invariant transformation temperature. It occurs by a process of lattices shearing. Martensite under microscope appears as lenticular plates which divide and subdivide the grains of the parent phase. Always touching but never crossing one another. The characteristic lenticular shape minimizes the elastic distortion in the matrix. These platelets grow at about one-third the velocity of sound. This high speed means activation energy for growth is very low, and thus activation energy for nucleation determines the amount of Martensite that can form under given conditions.

Martensite platelets attain their shape by two successive shear displacements contained in boundaries coherent with the parent phase. The first displacement is a homogeneous shear throughout the plate which occurs parallel to a specific plane in the parent phase known as the habit plane. The second displacement, the lesser of the two, can take place by one of two mechanisms: slip as in Fe-C Martensite or twinning as in Fe-Ni Martensite. Martensite formation occurs in many systems like Fe-C, Fe-Ni, Fe-Ni-C, Cu-Zn, Au-Cd, and even in pure metals like Li, Zr and Co. However, only the alloys based on Fe and C show a pronounced strengthening effect.

High strength of Martensite is attributed to its typical crystal structure i.e. effective barriers to slip are provided by the fine twin structure or the high dislocation density. In Fe-C system, carbon atoms are also involved in strengthening. Super saturated carbon atoms strain the ferrite lattice, and this strain can be relieved by redistribution of carbon atoms by diffusion at room temperature. One result is that a strong binding is set up between dislocations and the carbon atoms. This hinders the motion of dislocations, thus increasing the strength.

This strengthening mechanism is one of the most common processes used in engineering materials. The name basically arises from the phase *Martensite* that forms in steels when they are quenched from the solutionizing temperature.

Recovery, Recrystallization and Grain Growth

As mentioned in earlier sections, annealing is an important industrial process to relieve the stresses from cold working. During cold working grain shape changes, while material strain hardens because of increase in dislocation density. Between 1-10% of the energy of plastic deformation is stored in material in the form of strain energy associated with point defects and dislocations. On annealing i.e. on heating the deformed material to higher temperatures and holding, material tends to lose the extra strain energy and revert to the original condition before deformation by the processes of recovery and recrystallization. Grain growth may follow these in some instances.

Recovery: This is the first stage of restoration after cold working where physical properties of the cold-worked material are restored without any observable change in microstructure. The properties that are mostly affected by recovery ate those sensitive to point defects, for example – thermal and electrical conductivities. During recovery, which takes place at low temperatures of annealing, some of the stored internal energy is

relieved by virtue of dislocation motion as a result of enhanced atomic diffusion. There is some reduction, though not substantial, in dislocation density as well apart from formation of dislocation configurations with low strain energies. Excess point defects that are created during deformation are annihilated either by absorption at grain boundaries or dislocation climbing process. Stored energy of cold work is the driving force for recovery.

*Recrystallization:*This stage of annealing follows after recovery stage. Here also driving force is stored energy of cold work. Even after complete recovery, the grains are still in relatively high strain energy state. This stage, thus, involves replacement of cold-worked structure by a new set of strain-free, approximately equi-axed grains i.e. it is the process of nucleation and growth of new, strain-free crystals to replace all the deformed crystals. It starts on heating to temperatures in the range of 0.3-0.5 T_m, which is above the recovery stage. There is no crystal structure change during recrystallization. This process is characterized by recrystallization temperature which is defined as the temperature at which 50% of material recrystallizes in one hour time. The recrystallization temperature is strongly dependent on the purity of a material. Pure materials may recrystallizes around 0.3 T_m, while impure materials may recrystallizes around 0.5-0.7 T_m.

There are many variables that influence recrystallization behavior, namely amount of prior deformation, temperature, time, initial grain size, composition and amount of recovery prior to the start of the recrystallization. This dependence leads to following empirical laws:

- A minimum amount of deformation is needed to cause recrystallization.
- Smaller the degree of deformation, higher will be the recrystallization temperature.
- The finer is the initial grain size; lower will be the recrystallization temperature.
- The larger the initial grain size, the greater degree of deformation is required to produce an equivalent recrystallization temperature.
- Greater the degree of deformation and lower the annealing temperature, the smaller will be the recrystallized grain size.
- The higher is the temperature of cold working, the less is the strain energy stored and thus recrystallization temperature is correspondingly higher.
- The recrystallization rate increases exponentially with temperature.

During recrystallization, the mechanical properties that were changes during deformation are restored to their pre-cold-work values. Thus material becomes softer, weaker and ductile. During this stage of annealing impurity atoms tend to segregate at grain boundaries, and retard their motion and obstruct the processes of nucleation and growth. This *solute drag effect* can be used to retain cold worked strength at higher service temperatures. Presence of second phase particles causes slowing down of recrystallization – *pinning action of the particles*.

*Grain growth:*This stage follows complete crystallization if the material is left at elevated temperatures. However, grain growth does not need to be preceded by recovery and

recrystallization; it may occur in all polycrystalline materials. During this stage newly formed strain-free grains tend to grow in size. This grain growth occurs by the migration of grain boundaries. Driving force for this process is reduction in grain boundary energy i.e. decreasing in free energy of the material. As the grains grow larger, the curvature of the boundaries becomes less. This results in a tendency for larger grains to grow at the expense of smaller grains. In practical applications, grain growth is not desirable. Incorporation of impurity atoms and insoluble second phase particles are effective in retarding grain growth.

Because the driving force for grain growth is lower than the driving force for recrystallization, grain growth occurs slowly at a temperature where recrystallization occurs at substantially high speeds. However, grain growth is strongly temperature dependent.

Want to know how good you are in this chapter?

1. Sharp break in dislocation line that is in slip plane _____.

(a) Jog   (b) Kink   (c) Either jog or kink   (d) None

2. Minimum number of slip systems that must be operative during plastic deformation

(a) 3   (b) 4   (c) 5   (d) 6

3. Following strengthening mechanism applies to multi-phase material

(a) Grain size reduction   (b) Dispersion hardening
(c) Solid solution strengthening   (d) Strain hardening

4. If ASTM grain size number is 1, approximate grain diameter (in mm)

(a) 0.1   (b) 0.2   (c) 0.25   (d) 10

5. If volume fraction of spherical shaped second phase particles is 50% with radius of 3 μm, interspacing of particles is (in μm)

(a) 1   (b) 2   (c) 3   (d) 4

6. Characteristic shape of Martensite platelets

(a) Disc   (b) Lenticular   (c) Cylindrical   (d) Spheroids

7. Recrystallization temperature of pure materials _____ (in terms of homologous temperature)

(a) 0.1   (b) 0.2   (c) 0.3   (d) 0.4

8. Example for strengthening mechanism in single-phase material

(a) Strain hardening   (b) Precipitation hardening
(c) Fiber strengthening   (d) Dispersion strengthening

9. Higher the degree of deformation, recrystallization temperature is

(a) Higher   (b) Lower   (c) No effect   (d) Either higher or lower

10. Recrystallization rate varies in the following manner with temperature

(a) Linearly increasing   (b) Linearly decreasing
(c) Exponential   (d) Logarithmic

11. Methods to retard grain growth

(a) Solute drag   (b) Pinning action of particles   (c) Both   (d) None

12. Driving force for recrystallization process

(a) Stored energy of cold work   (b) Grain boundary energy
(c) Both   (d) Stacking-fault energy

13. Driving force for grain growth process

(a) Stored energy of cold work   (b) Grain boundary energy
(c) Both   (d) Stacking-fault energy

14. Fine grain size, usually, can not be obtained during the following process

(a) Slow cooling   (b) increasing nucleation rate (c) retarding grain growth   (d) fast cooling

15. Hardness during over-aging

(a) Decreases   (b) Increases   (c) Constant   (d) Decreases abruptly

16. Decrease in free energy during recrystallization is attributed to

(a) Excess point defects   (b) Excess dislocations
(c) Grain boundaries   (d) All

17. Decrease in free energy during recovery is attributed to

(a) Excess point defects   (b) Excess dislocations
(c) Grain boundaries   (d) All

Answers:

1. b
2. c
3. b
4. c
5. d
6. b
7. c
8. a
9. b
10. c
11. c
12. a
13. b
14. a
15. a
16. b
17. a