UNIT-III PHYSICAL ASPECTS OF FATIGUE AND FRACTURE

3 phases in fatigue life:

(i) Crack initiation or nucleation. i.e primary Stage

(ii) Secondary stage i.e crack propagation.

(iii) Final stage i.e failure by fracture or some other limiting factor.

Stage 1: Crack initiation:

It includes the early development of fatigue damage which can bee removed by suitable thermal annealing process.

Stage 2: Crack propagation

It involves the deepening of initial crack on planes of high shear stress. This is the stage I crack growth. Stage II crack growth involves the growth of well defid crack in dir'n normal to the maximum tensile stress.

Stage 3: Final fracture

It occurs when crack reaches sufficient length so that the remaining cross section cannot support the applied load.

Crack initiation is based on the microcrack formation. Crack growth depends on macrocrack formation

∴ Total fatigue life $(N)_T$ = No. of cycle for crack initiation + No. of cycles for crack growth to produce ultimate failure.

$$N_T = N_i + N_g$$

where N_i depends upon micromechanism, so it is very difficult to find out the initiation cycle.

N_g – determined by fatigue-crack propagation.

Normally fatigue crack initiated at the three surfaces.

Fatigue crack initiation process:

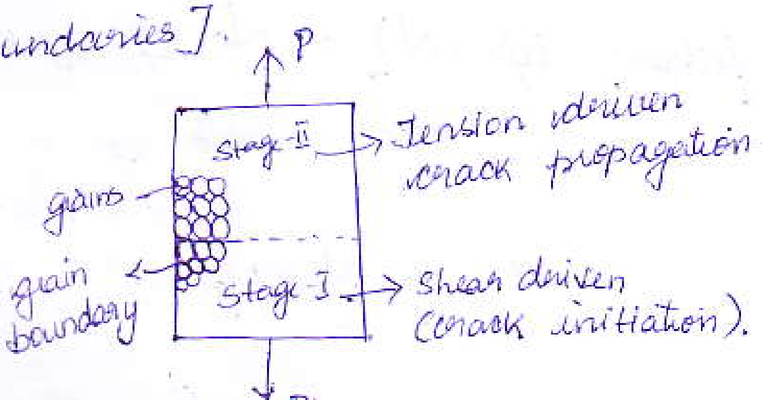
Slip lines are produced in the material during first few ~~lines~~ thousand cycle of stress. Successive cycle produces additional slip-bands. [No., of slip bands is not proportional to the no. of cycle of stress].

Cracks are usually found in the region of heavy deformation parallel to a original slip band. Initially fatigue crack propagates along the direction normal to max. applied stress (tensile).

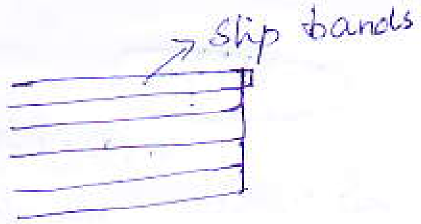
Normally fatigue crack propagation is transgrannular, not intergranular.

[Transgrannular – crack propagates along the grains, not on the grain boundaries.

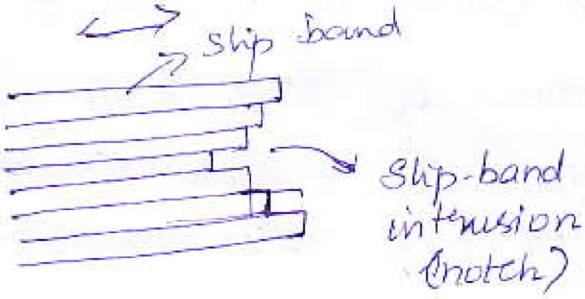
Intergrannular – crack propagates along grain boundaries].

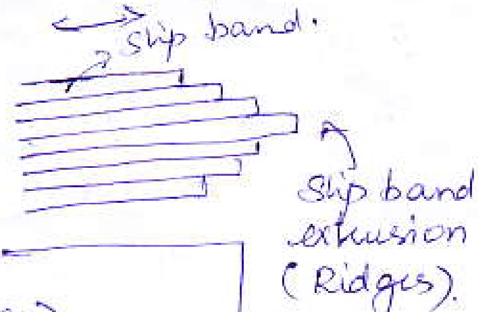


Formation of fatigue deformation in the surface of ridges & grooves are called slip-band extrusion and slip band intrusion. This concept was explained by Woods micromechanism.



static deformation





By using electron microscope (or) Scanning electron microscope, the microstructure of atoms are scanned for knowing the crack initiation process.

Based on this observation, slip produced by fatigue has indicating that slip-bands are the result of systematic built up of fine slip movements in the order of less than 1nm, rather than 100-1000nm which are observed for static loading.

Stage-I crack growth

In the first stage crack propagation rate is very slow in the order of nanometer per cycle. In this stage it is very difficult to get fracture surface.

Stage-II crack growth:

In this stage, crack propagation rate is high, in the order of micrometer per cycle. Fracture surfaces are in the form of ripples or fatigue fracture strilations

Each ripples represent successive portion of advancing the crack. ~~f~~ (That is normal to the tensile stress).
Each ripples was produced by single cycle stress. For observing ripples through the cycle is very difficult which can be observed only after complete fracture.

The crack propagation in the stage II process is given below.

This is a plastic blunting process at the starting of the loading cycle, the crack ripple is sharp as shown in fig.(1)

As the tensile load is applied, the small double notch at the crack will concentrate the slip along the plane at 45° to the plane of crack in fig(2)

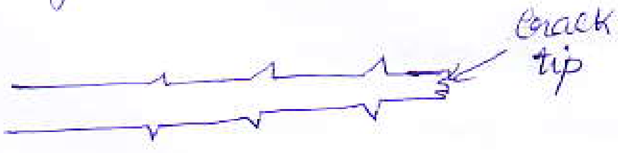


Fig (1)

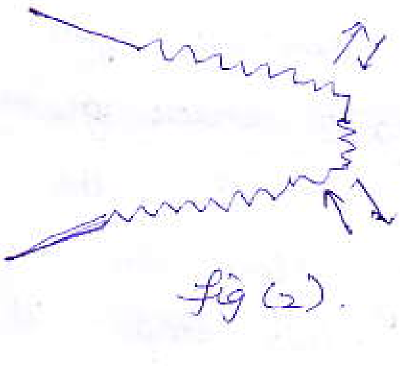


fig (2).

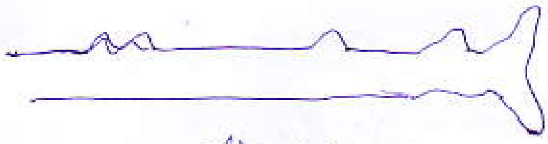


fig (5)

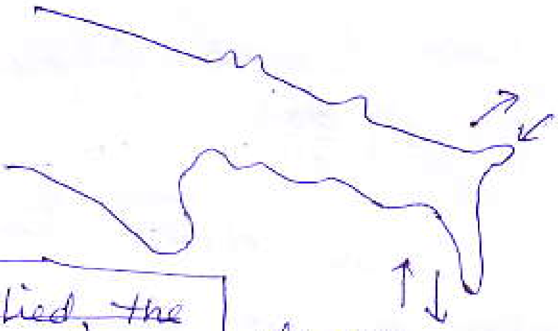


fig (6)

~~As the tensile load is applied, the~~

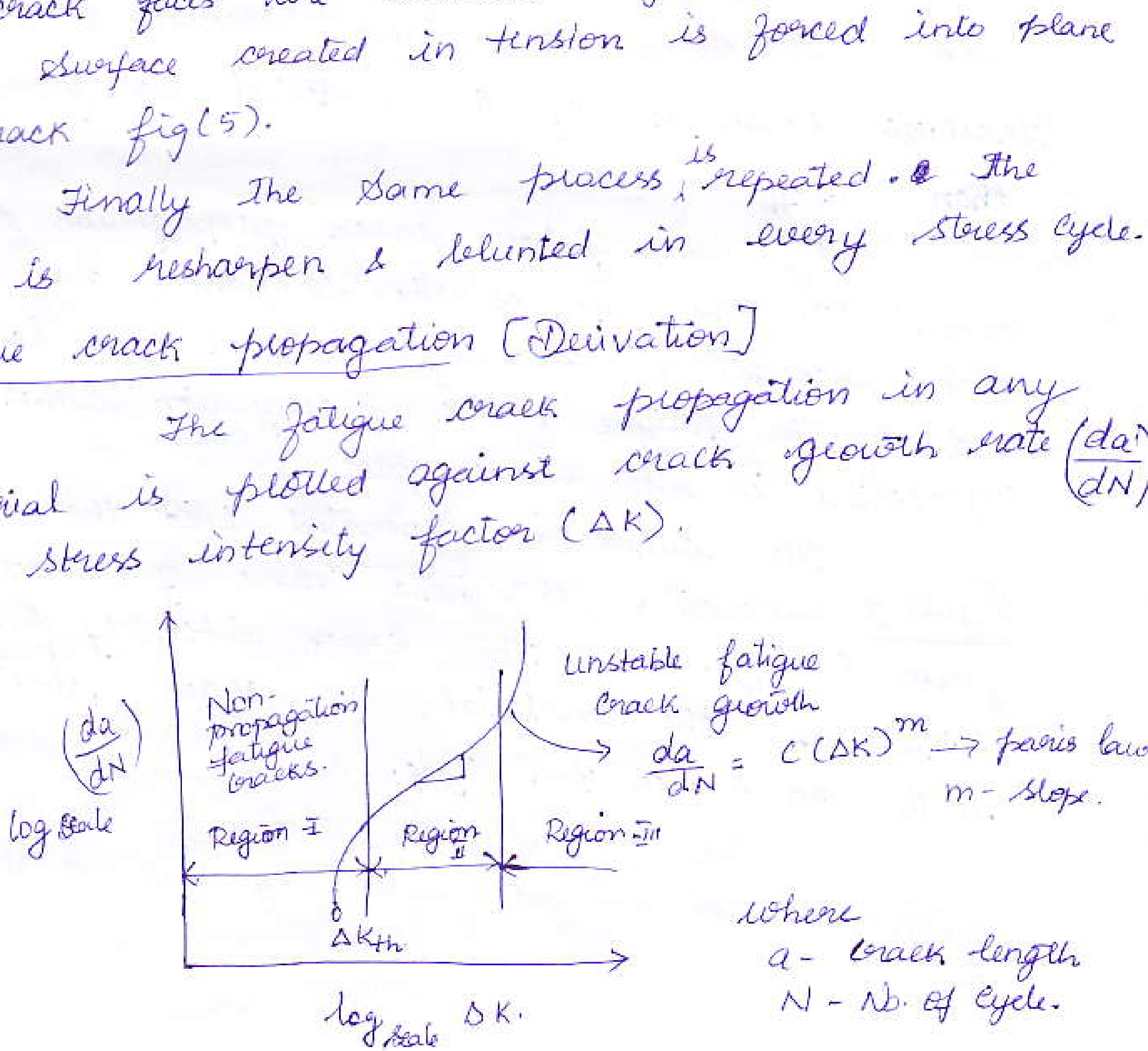
As the crack widens to its max. extension it is longer by plastic shearing and at the same time, its tip becomes blunter fig (3).

When the load is changed to compression, the slip dir'n in the end zones is reversed fig (4) The crack faces are crushed together and the new crack surface created in tension is forced into plane of crack fig (5).

Finally the same process is repeated. The crack is resharpen & blunted in every stress cycle.

Fatigue crack propagation [Derivation]

The fatigue crack propagation in any material is plotted against crack growth rate $\left(\frac{da}{dN}\right)$ and stress intensity factor (ΔK).



$$\frac{da}{dN} = C(\Delta K)^m \rightarrow \text{paris law}$$

m - slope.

where

a - crack length

N - No. of cycle.

Due to fatigue load, the crack length continously grows $\left(\frac{da}{dN}\right)$. ΔK — stress concentration

When objects having any crack, the stress concentration factor is not suitable. so we represent the stress in the tip of the crack by stress intensity factor

$$\Delta K = \Delta\sigma \sqrt{\pi a}$$

$$K = \sigma\sqrt{\pi a}$$ → Measuring the crack at the tip.

σ - applied stress ,

a - crack length ;

ΔK - Range of stress intensity factor.

$$\Delta K = K_{max} - K_{min}.$$

ΔK_{th} - Threshold of stress intensity factor.

Fracture toughness of the material is greater than ΔK_{th}

By using fatigue crack propagation theory, we can initiate the life of the component and also crack propagation. For any material and loading condition, the fatigue crack propagation curve is represented in the above plot.

The curve is seperated into three regions. Region I indicates threshold value of stress intensity factor (ΔK_{th}). When the crack intensity factor reaches ΔK_{th} of the material, the crack starts propagate with nanometer per cycle.

If the value of K is below the thresold value, the crack will not propagate under any type of loadings.

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Region - II :

In this region, due to tensile and compressive loading, the crack propagates slow & steadily with respect to $\frac{da}{dN}$ & ΔK. In this region, the crack propagation is easily measured because it is in micrometer / cycle.

By using paris law, the crack propagation is measured

$$\frac{da}{dN} = C(\Delta K)^m$$

where m - slope of the curve

For steel, m is in the range of 0-3

For Al alloys, m is 3-4.

For Composites, m is higher.

where $\Delta K = K_{max} - K_{min}$

$$K_{max} = \sigma_{max}\sqrt{\pi a}, \quad K_{min} = \sigma_{min}\sqrt{\pi a}$$

For compressive and zero loads, the K_{min} is taken as zero. Because it won't give any effect in the crack propagation. But when material subjected to Variable loading, compressive load is also included for calculating the crack retardation.

By using fatigue mean stress cycle, stress ratio (R), which affect the crack propagation in region I and region II. But in region II, it gives very less effect.

Including R in the paris eqn, the relationship is given by $\frac{da}{dN} = \frac{C(\Delta K)^m}{(1-R)K_c - \Delta K}$

K_c is critical stress intensity factor ⓞⓡ fracture Toughness of the material.

→ The crack propagation is also calculated by total strain in the tip of the crack, by using single powerlaw expression for elastic and plastic strain region.

$$\frac{da}{dN} = c_1 \epsilon^{m_1}$$

By using paris law, the crack propagation life of the component is determined. paris law is a link b/w fracture mechanics and fatigue.

By using elastic stress intensity factor, the fatigue crack propagation is calculated. This approach is called linear elastic fracture mechanics [LEFM].

This is used for high ductile materials

The initial & final crack size are represented by a_i & a_f, ~~we~~

WKT, $\Delta K = \Delta\sigma \sqrt{\pi a}$

where $\Delta\sigma = \sigma_{max} - \sigma_{min}$

By using shape factor (f) in the above eqn $\Delta K = f \Delta\sigma \sqrt{\pi a} \rightarrow ①$ $\quad f = a/w$.

We know the fatigue crack propagation (paris law) $\frac{da}{dN} = c(\Delta K)^m \rightarrow ②$

sub ΔK from ① in ②, we get

$$\frac{da}{dN} = c\left[f \Delta\sigma \sqrt{\pi a}\right]^m$$

$$= c \cdot f^{m} \Delta\sigma^{m} (\pi a)^{m/2} \longrightarrow ③$$

WKT, the critical crack or final crack

$$K = \sigma_{max} \sqrt{\pi a}$$

$$K = f \sigma_{max} \sqrt{\pi a}$$

$$K_c = f \sigma_{max} \sqrt{\pi a_f}$$ [a_f - final crack size]

$$\therefore a_f = \frac{1}{\pi} \left[\frac{K_c}{f \sigma_{max}} \right]^2$$

Fatigue crack propagation cycle or failure cycle (N

is $$N_p = \int_{0}^{N_p} dN \longrightarrow ④$$

Substitute dN value from ③

$$N_p = \int_{0}^{N_p} \frac{da}{c \cdot f^{m} \Delta\sigma^{m} (\pi a)^{m/2}}$$

$$N_p = \int_{a_i}^{a_f} \frac{a^{-m/2} \, da}{c f^{m} \Delta\sigma^{m} (\pi)^{m/2}}$$

put $m=2$ in the above eqn

$$= \int_{a_i}^{a_f} \frac{a^{-1} da}{c f^{m} \Delta\sigma^{m} \pi^{2/2}}$$

$$= \frac{1}{c f^{2} \Delta\sigma^{2} \pi} \int_{a_i}^{a_f} \frac{da}{a}$$

$$= \frac{1}{c f^{2} \Delta\sigma^{2} \pi} \left[\ln a_f - \ln a_i \right]$$

$$N_p = \frac{\ln \left[a_f / a_i \right]}{c f^{2} \Delta\sigma^{2} \pi}$$

and if $m \neq 2$

$$N_f = \frac{2}{(m-2)\, C\, f^m (\Delta\sigma)^m \pi^{m/2}} \left\{ \frac{1}{(a_0)^{(m-2)/2}} - \frac{1}{(a_f)^{(m-2)/2}} \right\}$$

GRIFFITH CRACK THEORY:

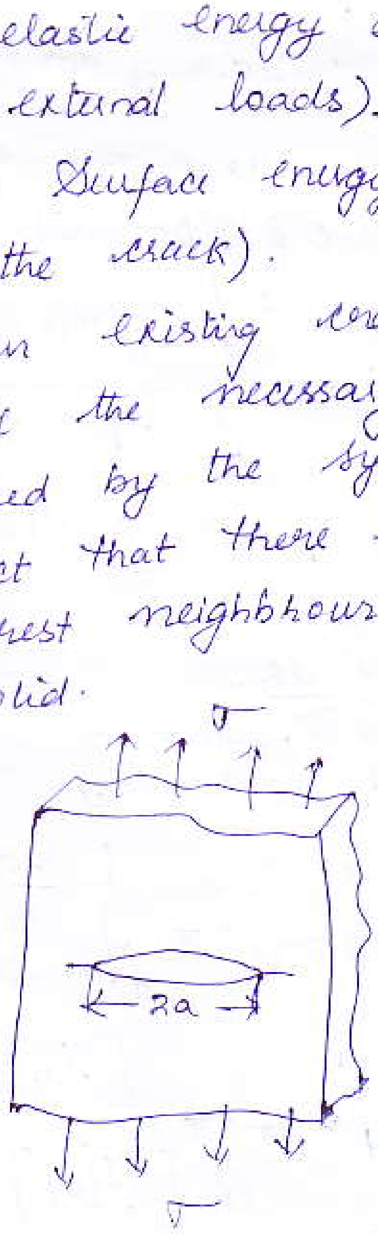
→ A.A. Griffith, some 75 yrs ago, initially stated the quantitative relations that Engineers & Scientists use today in determining the fracture of cracked solids.

→ Griffith noted that when a crack is introduced to a stressed plate of elastic material, a balance must be struck between

* The ↓se in potential energy (related to the release of stored elastic energy and work done by movement of the external loads).

* The ↑se in surface energy (resulting from the presence of the crack).

→ Likewise, an existing crack would grow by some increment, if the necessary additional surface energy were supplied by the system. This surface energy arises from the fact that there is a non-equilibrium configuration of nearest neighbour atoms at any surface in a solid.



→ For the configuration in fig, Griffith estimated the surface energy term to be the product of the total crack surface area (2a. 2t) & the specific surface energy (γ_s) which has units of energy per unit area.

→ He then used the stress analysis of Inglis for the case of an infinitely large plate containing an elliptical crack and computed the decrease in potential energy of the cracked plate to be $\frac{\pi\sigma^2a^2t}{E}$

→ Hence the change in potential energy of the plate associated with the introduction of a crack may be given by

$$U = U_0 - U_a + U_\gamma$$

$$U_a = \frac{\pi\sigma^2a^2t}{E} \; ; \quad U_\gamma = 4at\gamma_s$$

$$U = U_0 - \frac{\pi\sigma^2a^2t}{E} + 4at\gamma_s$$

where U = P.E of a body with crack.

U_0 = PE of the body without crack.

U_a = ↓se in P.E. of the cracked plate

σ = applied stress.

a = Semi-crack length

t = plate thickness.

E = Modulus of elasticity

γ_s = Specific surface energy.

→ To determine the condition of equilibrium, differentiate the PE (U) w.r.t the crack length and set it equal to zero

$$\frac{\partial U}{\partial a} = 4t\gamma_s - \frac{2\pi\sigma^2 at}{E} = 0$$

$\frac{\partial U_0}{\partial a} = 0$, since U_0 accounts for the PE of the body without a crack and does not vary with crack length, therefore $\boxed{2\gamma_s = \frac{\pi\sigma^2 a}{E}}$ → It represents the equilib'm condition.

↳①

→ The nature of the equilibrium is determined by the second derivative $\frac{\partial^2 U}{\partial a^2}$. Since $\frac{\partial^2 U}{\partial a^2} = \frac{-2\pi\sigma^2 t}{E}$ is negative, the equilibrium condition is unstable and the crack will always grow.

→ Griffith rewrote the above eqn ① in the form as $\sigma = \sqrt{\frac{2E\gamma_s}{\pi a}} \rightarrow ②$ for the case of plane stress and

$\sigma = \sqrt{\frac{2E\gamma_s}{\pi a(1-\nu^2)}}$ for the case of plane strain.

→ Since $\nu \simeq 0.25$ to 0.33 for many materials, the difference in allowable stress level in a given material subjected to plane strain or plane-stress conditions does not appear to be large.

→ It is important that Griffith relation was derived for an elastic material containing a very sharp crack.

→ This is not true for metals and polymers, where the fracture energy is found to be several orders of magnitude greater than the surface energy of a given material.

→ Orowan recognized this fact and suggested that eqn ② to be modified to include the energy of plastic deformation in the fracture processes, so that

$$\sigma = \sqrt{\frac{2E(\gamma_s+\gamma_p)}{\pi a}} = \sqrt{\frac{2E\gamma_s}{\pi a}\left(1+\frac{\gamma_p}{\gamma_s}\right)}$$

Where γ_p = plastic deformation energy & $\gamma_p >> \gamma_s$

Under these conditions $\sigma = \sqrt{\frac{2E\gamma_s}{\pi a}\left(\frac{\gamma_p}{\gamma_s}\right)}$

→ Fracture of real materials will depend on the sharpness of the crack & the relative amount of plastic deformation. It was shown that

$$\sigma = \sqrt{\frac{2E\gamma_s}{\pi a}\left(\frac{\pi\rho}{8a_0}\right)}$$

→ Later Irwin showed that $\sigma = \sqrt{\frac{EG}{\pi a}}$ where G is the strain energy release rate.

UNIT – III

CRYSTAL IMPERFECTIONS

By "crystalline defect" is meant a lattice irregularity having one or more of its dimensions on the order of an atomic diameter. Classification of crystalline imperfections is frequently made according to geometry or dimensionality of the defect and is as follows:

1. Point imperfections
2. Line imperfections or dislocations
3. Surface imperfections, and
4. Bulk or Volume imperfections.

POINT IMPERFECTIONS

Point imperfections are also referred to as *zero dimensional imperfections*. As the name implies, they are imperfect point-like regions in the crystal. One or two atomic diameters is the typical size of a point imperfections. Different kinds of point imperfections are as follows:

a) Vacancies and self-interstitials
b) Interstitialcies
c) Impurities
d) Electronic defects

Vacancies And Self-Interstitials

A *vacancy* refers to an atomic site from where the atom is missing (as shown in fig.1), i.e., they are vacant sites unoccupied by any atom. This is the simplest of the point defects. Vacancies are formed during solidification, and also as a result of atomic vibrations, which can cause the displacement of atoms from their normal lattice sites. The number of vacancies increases exponentially with temperature.

A *self-interstitial* is an atom from the crystal that is crowded into an interstitial site, a small void space that under ordinary circumstances is not occupied (as shown in fig.2). In metals, a self-interstitial introduces relatively large distortions in the surrounding lattice because the atom is substantially larger than the interstitial position in which it is situated. Consequently, the formation of this defect is not highly probable, and it exists in very small concentrations, which are significantly lower than for vacancies.

Impurities in solids

No metal is pure; impurities or foreign atoms will always be present in any metal. The addition of impurity atoms to a metal will result in the formation of a *solid solution* and/or a new *second phase*, depending on the kinds of impurity, their concentrations, and the temperature of the alloy. "*Solvent*" represents the element or compound that is present in the greatest amount;

solvent atoms are also called as host atoms. "*Solute*" is used to denote an element or compound present in a minor concentration.

A *substitutional* impurity (or solute impurity) is a point imperfection. It refers to a solute or impurity atoms that substitute for or replaces a parent (or host) atom in the crystal, as shown in fig.2. Aluminium and phosphorus doped in silicon are substitutional impurities in the crystal.

An *interstitial* impurity is also a point imperfection. It is a small sized solute or impurity atoms filling the voids or interstices among the parent or host atoms in the crystal, without dislodging any of the parent atoms from their sites, as shown in fig.3. A solute or impurity atom can enter the interstitial void space only when it is substantially smaller than the parent atom.

In ionic crystals, because the atoms exist as charged ions, the formation of point imperfections is subject to the requirement that the *overall electro-neutrality is maintained*. Electro-neutrality is the state that exists when there are equal numbers of positive and negative charges from the ions. As a consequence, defects in ceramics do not occur alone. One such type of defect involves a cation-vacancy and a cation-interstitial pair. This is called a *Frenkel* defect. (Refer fig. 5). It might be thought of as being formed by a cation leaving its normal position and moving into an interstitial site. There is no change in charge because the cation maintains the same positive charge as an interstitial. As cations are generally the smaller ions, it is possible for them to get displaced into the void space. Anions do not get displaced like this, as the void space is just too small for their size. The point imperfections in silver halides and CaF_2 are of the Frenkel type.

Another type of defect found is a cation vacancy – anion vacancy pair knows as a *Schottky* defect (refer fig. 5). This defect might be thought of as being created by removing one cation and one anion from the interior of the crystal and then placing them both at an external surface. Since both cations and anions have the same charge, and since for every anion vacancy there exists a cation vacancy, the electro-neutrality of the crystal is maintained. This type is dominant in alkali halides.

LINE IMPERFECTIONS OR DISLOCATIONS

They are linear or one-dimensional defects around which some of the atoms are misaligned. A dislocation may be defined as a disturbed region between two substantially perfect parts of a crystal. It is a line defect in a crystal structure whereby a part-plane of atoms is displaced from its symmetrically stable positions in the array. The dislocation is responsible for the phenomenon of *slip* by which most metals deform plastically. Further it is also connected with other mechanical phenomena such as strain hardening, the yield point, creep, fatigue, and brittle fracture and are helpful in describing crystal growth and electrical conductivity. One may conclude that dislocation is the region of localized lattice disturbance separating the slipped and un-slipped regions of a crystal. Dislocations arise in crystals as a result of

(i) growth accidents
(ii) thermal stresses
(iii) external stresses causing plastic flow
(iv) phase transformations
(v) segregation of solute atoms causing mismatches, etc.

The voids and vacant sites in the crystals favor the generation of dislocations.

The two basic types of dislocations are *edge* dislocation and *screw* dislocation.

Edge Dislocation

An edge dislocation is created in the crystal by introducing an extra portion of a plane of atoms, or half plane, the edge of which terminates *within* the crystal (i.e., any extra plane that does not extend up to the base of the crystal). It is a linear defect that centers around the line that is defined along the end of the extra half-plane of atoms. This is sometimes termed the dislocation line. For edge dislocation, the dislocation line is perpendicular to the plane of the page. Within the region around the dislocation line there is some localized lattice distortion. The atoms above the dislocation line are squeezed together, and those below are pulled apart.

Screw Dislocation

A screw dislocation results from a displacement of the atoms in one plane of a crystal relative to the rest of the crystal, forming a spiral ramp around the dislocation line. It may be thought of as being formed by a shear stress that is applied to produce the distortion shown in fig.6. the upper front region of the crystal is shifted one atomic distance to the right relative to the bottom portion. The atomic distortion associated with a screw dislocation is also linear and along a dislocation line, line AB in fig. The screw dislocation derives its name from the spiral or helical path or ramp that is traced around the dislocation line by the atomic planes of atoms.

Mixed Dislocation

Most dislocations found in crystalline materials are probably neither pure edge nor pure screw, but exhibit components of both types; these are termed mixed dislocations.

DIFFERENCES BETWEEN EDGE DISLOCATION AND SCREW DISLOCATION

Edge Dislocation	Screw Dislocation
1. An edge of an atomic plane is formed internal to the crystal.	Only a distortion of the lattice cells in the immediate vicinity is produced.
2. It involves an extra row of atoms either above or below the slip plane.	The distortion follows a helical or screw path and both right hand and left hand senses are possible.
3. The forces required to form and to move an edge dislocation are smaller.	The forces required to form and to move a screw dislocation are greater.
4. Speed of movement of an edge dislocation is greater.	Speed of movement of a screw dislocation is lesser.
5. It is particularly useful in explaining slip in plastic flow during mechanical working.	It is especially useful in explaining crystal growth as well as slip in plastic deformation.

SURFACE DEFECTS OR INTERFACIAL DEFECTS

Surface defects or interfacial defects are boundaries that have two dimensions and normally separate regions of the materials that have different crystal structures and/or crystallographic orientations. They are also called as plane defects. We can classify these defects into two main types, namely (a) external surface imperfections and (b) internal surface imperfections.

(a) External Surface Imperfections

The external surface of a crystal is an imperfection in itself, as the atomic bonds do not extend beyond the surface. Although we may visualize an external surface as simply a terminus of the crystal structure, the atoms on that surface cannot be compared with the atoms within the crystal because the external surface atoms have neighbors on one side only, while atoms inside the crystal have neighbors on either side of them. Since the external surface atoms are not entirely surrounded by others they possess higher energy than that of internal atoms.

(b) Internal Surface Imperfections

Internal surface imperfections arise from a change in the stacking of atomic planes across a boundary. The change may be one of the orientation or of the stacking sequence of the planes. Some important internal surface imperfections are given below.

(i). grain boundaries

(ii) twin boundaries

(iii) stacking faults

(iv) phase boundaries, and

(v) ferromagnetic domain walls.

(i) Grain Boundaries

For a crystalline solid, when the periodic and repeated arrangement of atoms is perfect or extends throughout the entirety of the specimen without interruption, the result is *single crystal*. All unit cells interlock in the same way and have the same orientation. Most crystalline solids are composed of a collection of many small crystals or grains; such materials are termed *polycrystalline*. Various stages in the solidification of a polycrystalline specimen are represented schematically in fig. 8.

Initially, small crystals or nuclei form at various positions. These have random crystallographic orientations, as indicated by the square grids. The small grains grow by the successive addition from the surrounding liquid of atoms to the structure of each. The extremities of adjacent grains impinge on one another as the solidification process approaches completion. As indicated in the figure, the crystallographic orientation varies from grain to grain. Also there exists some atomic mismatch within the region where two grains meet; this area is called the *grain boundary*. So a grain boundary is defined as the boundary separating two small grains or crystals having different crystallographic orientations in polycrystalline materials.

There is some atomic mismatch in a transition from the crystalline orientation of one grain to that of an adjacent one. Various degrees of crystallographic misalignment between adjacent grains are possible. When this orientation mismatch is slight, on the order of a few degrees, then the term *small-angle grain boundary* is used. These boundaries can be described in terms of dislocation arrays. One simple small-angle grain boundary is formed when the edge dislocations are aligned in the manner of fig. 9. This type is called a *tilt boundary*; the angle of misorientation, θ, is also indicated in the figure. When the angle of misorientation is parallel to the boundary, a *twist boundary* results, which can be described by an array of screw dislocations.

(ii) Twin Boundaries

A twin boundary is a special type of grain boundary across which there is a specific mirror lattice symmetry; that is, atoms on one side of the boundary are located in mirror image positions of the atoms on the other side (refer fig.10). The region of material between these boundaries is appropriately termed a *twin*. Twins result from atomic displacements that are produced from applied mechanical shear forces (mechanical twins), and also during annealing heat treatments following deformations (annealing twins). Twinning occurs on a definite crystallographic plane and in a specific direction, both of which depend on the crystal structure. Annealing twins are typically found in metals that have the FCC crystal structure, while mechanical twins are observed in BCC and HCP metals.

Stacking faults are found in FCC metals when there is an interruption in the stacking sequence of close-packed planes. Phase boundaries exist in multiphase materials across which there is a sudden change in physical and/or chemical characteristics. For *ferro*magnetic and *ferri*magnetic materials, the boundary that separates regions having different directions of magnetization is termed a domain wall. Associated with each of the defects discussed above is an interfacial energy, the magnitude of which depends on boundary type, and which will vary from material to material. Normally, the interfacial energy will be greatest for external surfaces and least for domain walls.

BULK OR VOLUME DEFECTS

Bulk or volume defects exist in all solid materials that are much larger than those heretofore discussed. These include pores, cracks, foreign inclusions, and other phases. They are normally introduced during processing and fabrication steps.