HEAT TREATMENT TECHNOLOGY

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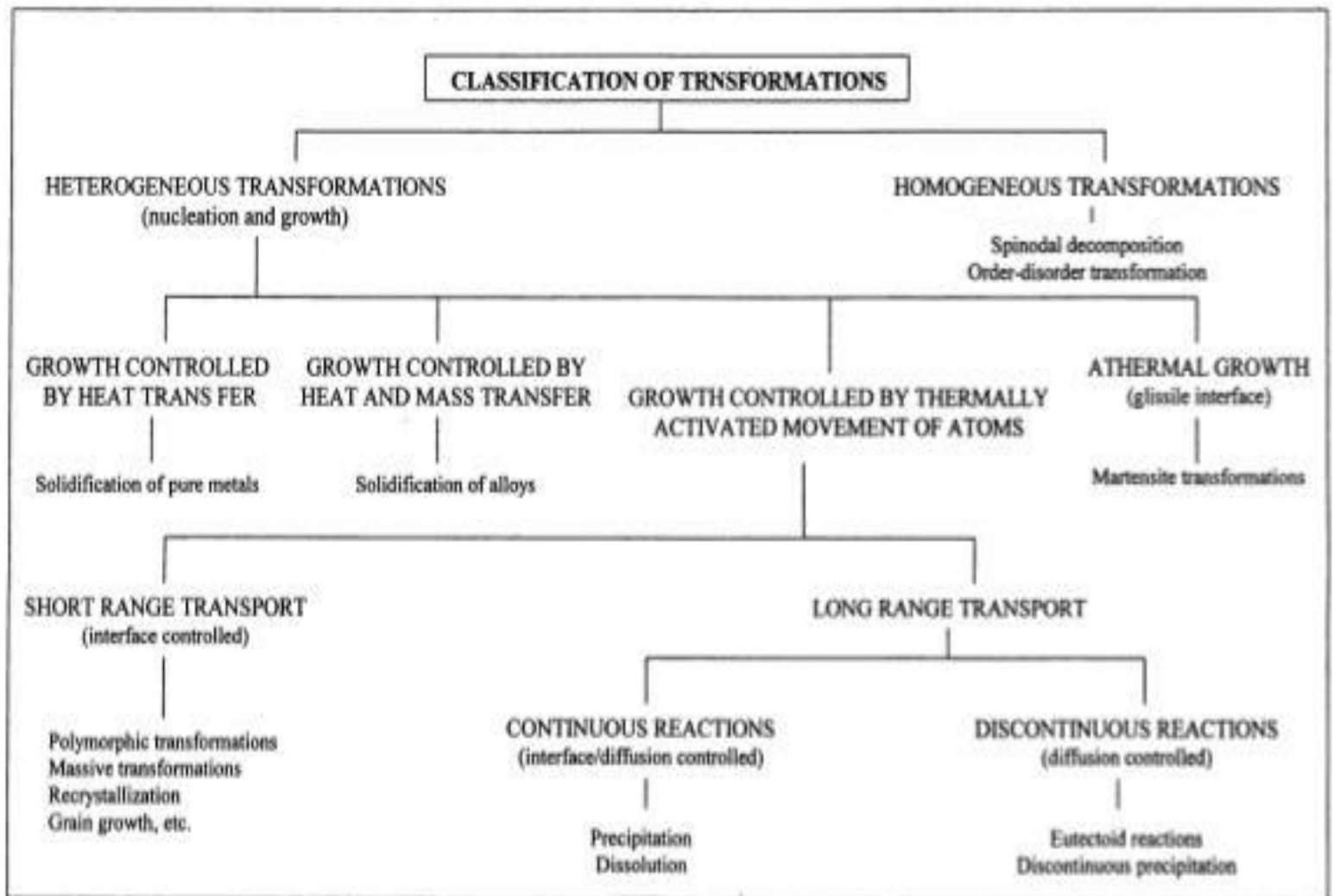

GOVERNMENT POLYTECHNIC MAYURBHANJ

TIKARPADA

# HEAT TREATMENT TECHNOLOGY

MODULE-1

1.0   SOLID STATE PHASE TRANSFORMATION:

CLASSIFICATION OF TRNSFORMATIONS

HETEROGENEOUS TRANSFORMATIONS
(nucleation and growth)

HOMOGENEOUS TRANSFORMATIONS
|
Spinodal decomposition
Order-disorder transformation

GROWTH CONTROLLED BY HEAT TRANS FER

GROWTH CONTROLLED BY HEAT AND MASS TRANSFER

GROWTH CONTROLLED BY THERMALLY ACTIVATED MOVEMENT OF ATOMS

ATHERMAL GROWTH
(glissile interface)

Solidification of pure metals

Solidification of alloys

Martensite transformations

SHORT RANGE TRANSPORT
(interface controlled)

LONG RANGE TRANSPORT

Polymorphic transformations
Massive transformations
Recrystallization
Grain growth, etc.

CONTINUOUS REACTIONS
(interface/diffusion controlled)

DISCONTINUOUS REACTIONS
(diffusion controlled)

Precipitation
Dissolution

Eutectoid reactions
Discontinuous precipitation

1.1 DIFFUSION

At any temperature different from absolute zero all atoms, irrespective of their state of aggregation (gaseous, liquid or solid), are constantly in motion. Since the movement of particles is associated with collisions, the path of a single particle is a zigzag one. However, an aggregation of "diffusing" particles has an observable drift from places of higher to places of lower concentration . For this reason diffusion is known as a transport phenomenon
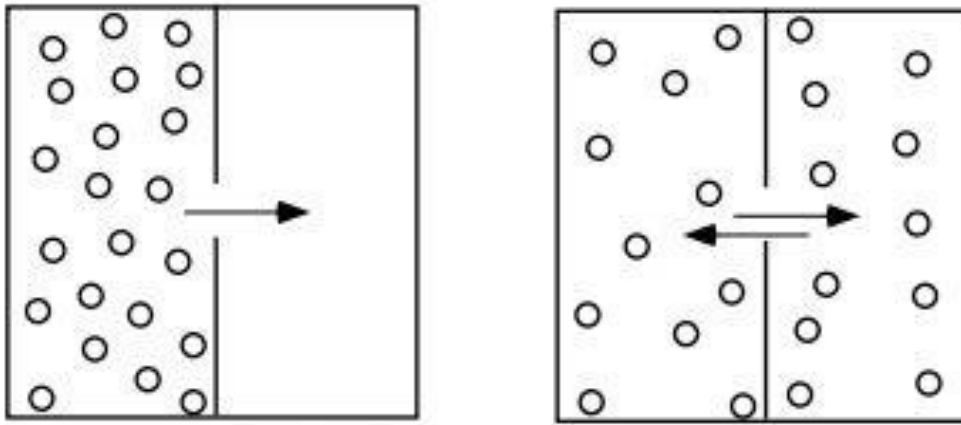
Mass transport, diffusion as a consequence
of existing spacial differences in concentration.

In each diffusion reaction (heat flow, for example, is also a diffusion process), the flux (of matter, heat, electricity, etc.) follows the general relation

Flux = (conductivity) x (driving force)

 In the case of atomic or molecular diffusion, the "conductivity" is referred to as the diffusivity or the diffusion constant, and is represented by the symbol D. We realize from the above considerations that this diffusion constant (D) reflects the mobility of the diffusing species in the given environment and accordingly assumes larger values in gases, smaller ones in liquids, and extremely small ones in solids.

The "driving force" for many types of diffusion is the existence of a concentration gradient. The term $\Delta C/\Delta x$ $\rightarrow dC/dx$ "gradient" describes the variation of a given property as a function of distance in the x-direction. If a material exhibits a linear variation of concentration with distance in the distance x-direction, we speak of

a constant Concentration gradient ΔC /Δx concentration x-direction. The gradient itself is the rate of change of the concentration with distance (dc/dx), which is the same as the slope of a graph of concentration vs. position (Δc/Δx)

concentration

ΔC

$\Delta C/\Delta x \rightarrow dC/dx$

Δx

distance
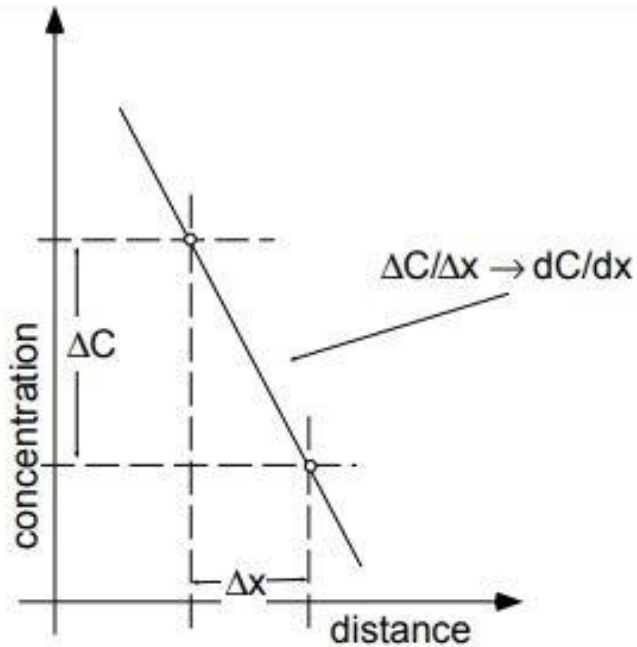
Concentration gradient (constant) in the x direction

Steady State and Nonsteady Diffusion Diffusion processes may be divided into two types:

(a) steady state and

(b) nonsteady state.

Steady state diffusion takes place at a constant rate - that is, once the process starts the number of atoms (or moles) crossing a given interface (the flux) is constant with time. This means that throughout the system dc/dx = constant and dc/dt = 0.

Nonsteady state diffusion is a time dependent process in which the rate of diffusion is a function of time. Thus dc/dx varies with time and dc/dt ≠ 0. Both types of diffusion are described quantitatively by Fick's laws of diffusion. The first law concerns both steady state and nonsteady state diffusion, while the second law deals only with nonsteady state diffusion.

2. STEADY STATE DIFFUSION (FICK'S FIRST LAW)

On the basis of the above considerations, *Fick's First Law* may be formulated as:

$$J = -D\left(\frac{dc}{dx}\right)$$

In words: The diffusive flux is proportional to the existing concentration gradient.

The negative sign in this relationship indicates that particle flow occurs in a "down" gradient direction, i.e. from regions of higher to regions of lower concentration. The flux J can be given in units of atoms/cm^{2}s, moles/cm^{2}s, or equivalents. Correspondingly, the diffusivity (D) will assume the dimensions cm^2/s, as can be seen from a dimensional analysis:

$$J\left(\frac{moles}{cm^2 s}\right) = -D\left(\frac{dc}{dx}\right)\left(\frac{moles \cdot cm^{-3}}{cm}\right)$$

Thus: $D = cm^2/s$

Like chemical reactions, diffusion is a thermally activated process and the temperature dependence of diffusion appears in the diffusivity as an "Arrhenius-type" equation:

$$D = D_o e^{-E_a/RT}$$

where D_o (the equivalent of A in the previously discussed temperature dependence of the rate constant) includes such factors as the jump distance, the vibrational frequency of the diffusing species and so on. Selected values of D, D_o and E_a are given in Table 1 (a) and (b).

TABLE 1

(a) Selected Values of Diffusion Constants (D)

Diffusing Substance	Solvent	T (°C)	D (cm²·s⁻¹)
Au	Cu	400	5×10^{-13}
Cu (Self-Diffusing)	(Cu)	650	3.2×10^{-12}
C	Fe (FCC)	950	10^{-7}
Methanol	H_2O	18	1.4×10^{-5}
O_2	Air	0	0.178
H_2	Air	0	0.611

(b) Selected Values of D_o and E_a for Diffusion Systems

Solute	Solvent (host structure)	D_o, cm²s	E_a, kJoules/mole
1. Carbon	fcc iron	0.2100	142
2. Carbon	bcc iron	0.0079	76
3. Iron	fcc iron	0.5800	285
4. Iron	bcc iron	5.8000	251
5. Nickel	fcc iron	0.5000	276
6. Manganese	fcc iron	0.3500	283
7. Zinc	copper	0.0330	159
8. Copper	aluminum	2.0000	142
9. Copper	copper	11.0000	240
10. Silver	silver	0.7200	188

A typical application of Fick's first law: Determine the rate at which helium (He), held at 5 atm and 200°C in a Pyrex glass bulb of 50 cm diameter and a wall thickness (x) of 0.1 cm, diffuses through the Pyrex to the outside. Assume that the pressure outside the tube at all times remains negligible (see fig. 3). (For the diffusion of gases it is

Figure 3   Conditions for outdiffusion of He from a glass bulb.

customary, although not necessary, to replace the diffusion constant D with the Permeation constant K, normally given in units of $cm^2/s \cdot atm$. Using the gas laws, K is readily converted to D if so desired.)

In the present system

$$K = 1 \times 10^{-9}\ cm^2/s \cdot atm$$

We can now set up the diffusion equation:

$$J = -K\left(\frac{dP}{dx}\right)$$

(and operate with pressures instead of concentrations)

We may now formally separate the variables and integrate:

$$Jdx = -KdP$$

$$\int_{x=0}^{x=0.1} Jdx = -\int_{p_2=5}^{p_1=0} KdP$$

$$J = K\frac{5.0}{0.1}$$

We can forego the integration since (dP/dx) = (ΔP/Δx) and we may immediately write:

$$J \text{ (total)} = -K\frac{\Delta P}{\Delta x} \times A = 5 \times 10^{-8} \times 7.9 \times 10^{3}$$

$$\boxed{J = 3.9 \times 10^{-4} \text{ ??}}$$

The units of the flux may be obtained from a dimensional analysis:

$$J = -K \times \frac{\Delta P}{\Delta x} \times A = \frac{cm^2}{s \cdot atm} \frac{atm}{cm} \frac{cm^2}{1} = \boxed{\frac{cm^3}{s}}$$

The total flux is 3.9×10^{-4} cm^3/s (with the gas volume given for 0°C and 1 atm). If the total gas flow by diffusion were to be determined for a specified time interval, the volume would be multiplied by the indicated time.

3. NONSTEADY STATE DIFFUSION (FICK'S SECOND LAW)

The quantitative treatment of nonsteady state diffusion processes is formulated as a partial differential equation. It is beyond the scope of 3.091 to treat the equations in detail but we can consider the second law qualitatively and examine some relevant solutions quantitatively.

The difference between steady state and nonsteady state diffusion conditions can readily be visualized (fig. 4). In the first case we have, for example, the diffusion of gas
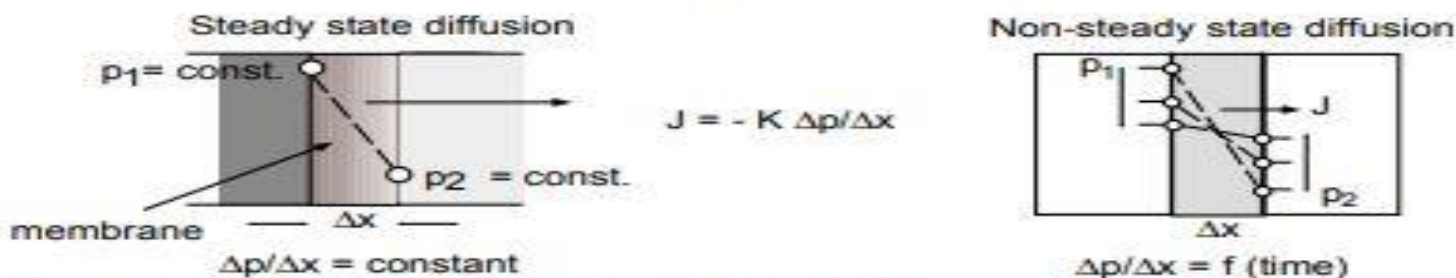

Figure 4   Steady state and Non-steady state diffusion

from an infinite volume (P$_1$ const) through a membrane into an infinite volume (P$_2$ const). The pressure gradient across the membrane remains constant as does the

diffusive flux. In the second case we deal with diffusion from a finite volume through a membrane into a finite volume. The pressures in the reservoirs involved change with time as does, consequently, the pressure gradient across the membrane.

(You are not required to be familiar with the following derivation of the Second Fick's Law, but you must know its final form.)

Figure 5
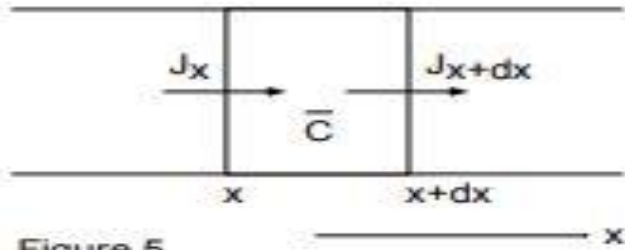
Consider a volume element (between x and x+dx of unit cross sectional area) of a membrane separating two finite volumes involved in a diffusion system (fig. 5). The flux of a given material into the volume element minus the flux out of the volume element equals the rate of accumulation of the material into this volume element:

$$J_x - J_{x+dx} = \frac{\partial \bar{c}}{\partial t} dx$$

[$\bar{c}$ is the average concentration in the volume element and $\bar{c}dx$ is the total amount of the diffusing material in the element at time (t).]

Using a Taylor series we can expand J_{x+dx} about x and obtain:

$$J_{x+dx} = J_x + \frac{\partial J_x}{\partial x} dx + \frac{\partial^2 J_x}{\partial x^2} \frac{dx^2}{2} + \ldots$$

Accordingly, as dx → 0:

$$\frac{\partial}{\partial x} \left(D \frac{\partial c}{\partial x} \right) = \frac{\partial c}{\partial t}$$

and if D does not vary with x (which is normally the case) we have the formulation of Fick's Second Law:

$$\frac{\partial c}{\partial t} = D \frac{\partial^2 c}{\partial x^2} \qquad \text{(Fick's Second Law)}$$

PRACTICE PROBLEM

For a steel alloy it has been determined that a carburizing heat treatment of 10-h duration will raise the carbon concentration to 0.45 wt% at a point 2.5 mm from the surface. Estimate the time necessary to

achieve the same concentration at a 5.0-mm position for an identical steel and at the same carburizing temperature

Ans: 40 hr

1.2 FORMATION OF AUSTENITE

Formation of austenite is a preliminary step for any heat treatment process. Austenite is formed on heating an aggregate of ferrite and pearlite, ferrite and cementite or cementite and pearlite, depending on whether the steel is of hypoeutectoid, eutectoid or hypereutectoid type, respectively. Formation of austenite in eutectoid steel differs from that of hypoeutectoid and hypereutectoid steels in the sense that in the former case it occurs at a particular temperature (Ac_1) whereas for the latter it takes place over a range of temperature (see Figure 4.1).

Let us consider the formation of austenite from a mixture of ferrite-cementite in a eutectoid steel. Normally, this mixture in eutectoid steel occurs as pearlite. Figure 4.2 shows the microstructure of lamellar pearlite which consists of alternate lamellae of ferrite and cementite. The orientation of these layers varies for different grains. These layers are in direct contact with each other. The intermetallic compound cementite has 6.67 weight percent carbon, whereas ferrite is almost pure iron, free of carbon. Though a carbon gradient exists, the structure is thermodynamically stable at room temperature and above. This is so because carbon diffusion rates are practically insignificant up to 200°C. As the temperature increases to higher values, but below eutectoid temperature, carbon atoms have a tendency to diffuse into ferrite. But the

Figure 4.2   Microstructure showing lamellar nature of pearlite (×1200).

These processes, i.e. the formation of primary austenitic grains and their growth by dissolving ferrite, continue till all ferrite and cementite dissolve to form austenite. The growth
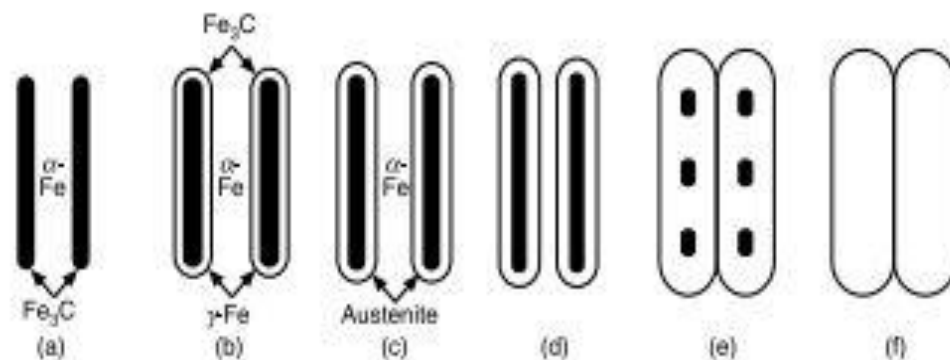
Figure 4.3 Steps associated with transformation of pearlite to austenite.

rate of austenite is higher than the rate of dissolution of the cementite into austenite because austenitic grain growth takes place because of the transformation of alpha iron to gamma iron and diffusion of carbon atoms from austenite to ferrite. This explains the experimentally observed fact that dissolving of ferrite is completed before that of cementite. The austenite thus formed at eutectoid temperature is not homogeneous. The carbon concentration is higher in these regions which are adjacent to the original cementite lamellae than those which are adjacent to the ferrite mass. Chemically homogeneous austenitic grains are obtained by holding steel above the eutectoid temperature. The holding time should be sufficient so that carbon atoms may diffuse and result in uniform distribution of carbon atoms. Figure 4.4 is a pictorial representation of the statement.
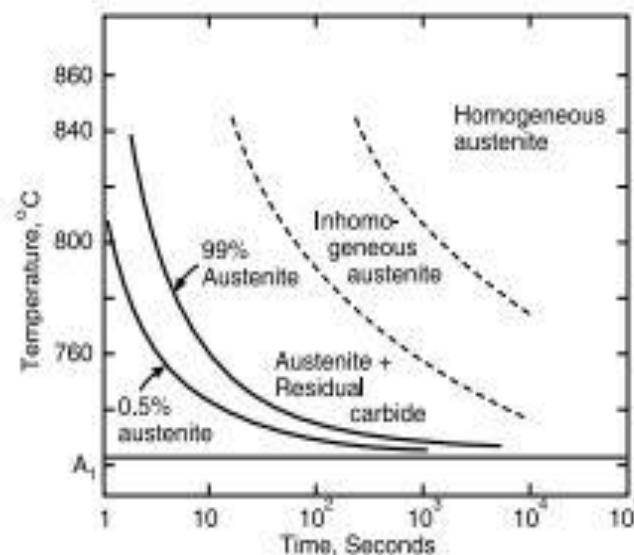

Figure 4.4 Effect of temperature and time on the austenite formed from pearlite in 0.3% carbon steel.

Transformation to austenite on heating hypoeutectoid and hypereutectoid steels is somewhat different from the above condition because of the presence of proeutectoid ferrite and proeutectoid cementite, respectively, along with pearlite. Transformation of proeutectoid ferrite and pearlite, in the case of hypoeutectoid steels, can be understood with the help of the

Fe-Cementite phase diagram. On very slow heating, austenite nuclei are formed just above the eutectoid temperature. More nuclei will form with increase in temperature. Therefore, at first the austenitic grains will grow by the growth of initially formed austenitic grains and then by the growth of newly formed austenite nuclei. The process will continue till the upper critical temperature (A_3) is reached. The austenite present at this temperature will be non-homogeneous due to the presence of embedded cementite particles within the austenitic grains. For hypoeutectoid steels, growth of primary austenitic grains take place at the expense of proeutectoid ferrite. Further, austenite nuclei are also possibly formed at grain boundaries of ferritic grains. In hypoeutectoid steels, cementite dissolves into the ferrite which in turn transforms into austenite. In the case of hypereutectoid steels, the transformation proceeds in a similar way with the difference that austenitic grains grow by dissolving proeutectoid cementite.

Theoretically, though pearlite must transform to austenite completely at eutectoid temperature, it does not happen so in practice. Complete dissolution of cementite of pearlite into austenite takes place over a range of temperatures. On parallel lines, it has been experimentally observed that dissolution of proeutectoid ferrite or proeutectoid cementite is not completed at A_3 or A_{cm} respectively. It, therefore, becomes essential to heat eutectoid, hypoeutectoid and hypereutectoid steels above A_1, A_3 and A_{cm}, respectively, in order to get homogeneous austenite.

1.9

Pearlitic transformation:

Pearlite is a product of eutectoid transformation in steel. It forms within austenite having a fixed amount of carbon (0.8%C) during isothermal hold at a temperature below 723°C (A1) but above 550°C (the knee of the TTT diagram). It has a lamellar structure consisting of alternate plates of cementite and ferrite. The distance between two consecutive cementite plates is). The cementite plates are more widely spaced in coarseλknown as its inter lamellar spacing (pearlite. The finer details of coarse pearlitic structure are visible under optical microscopes. Nital (2% nitric acid in alcohol) the most common etching reagent for steel attacks only the boundary between ferrite and cementite. However the width of a cementite plate is so thin that it appears as a dark line. Its width is approximately 1/7th the width of a ferrite plate. Let us now look at the mechanism of decomposition of austenite into pearlite in a little more detail. This is shown in slide 3 by a set of sketches representing the redistribution of carbon within austenite resulting in the nucleation of a pearlite nodule.
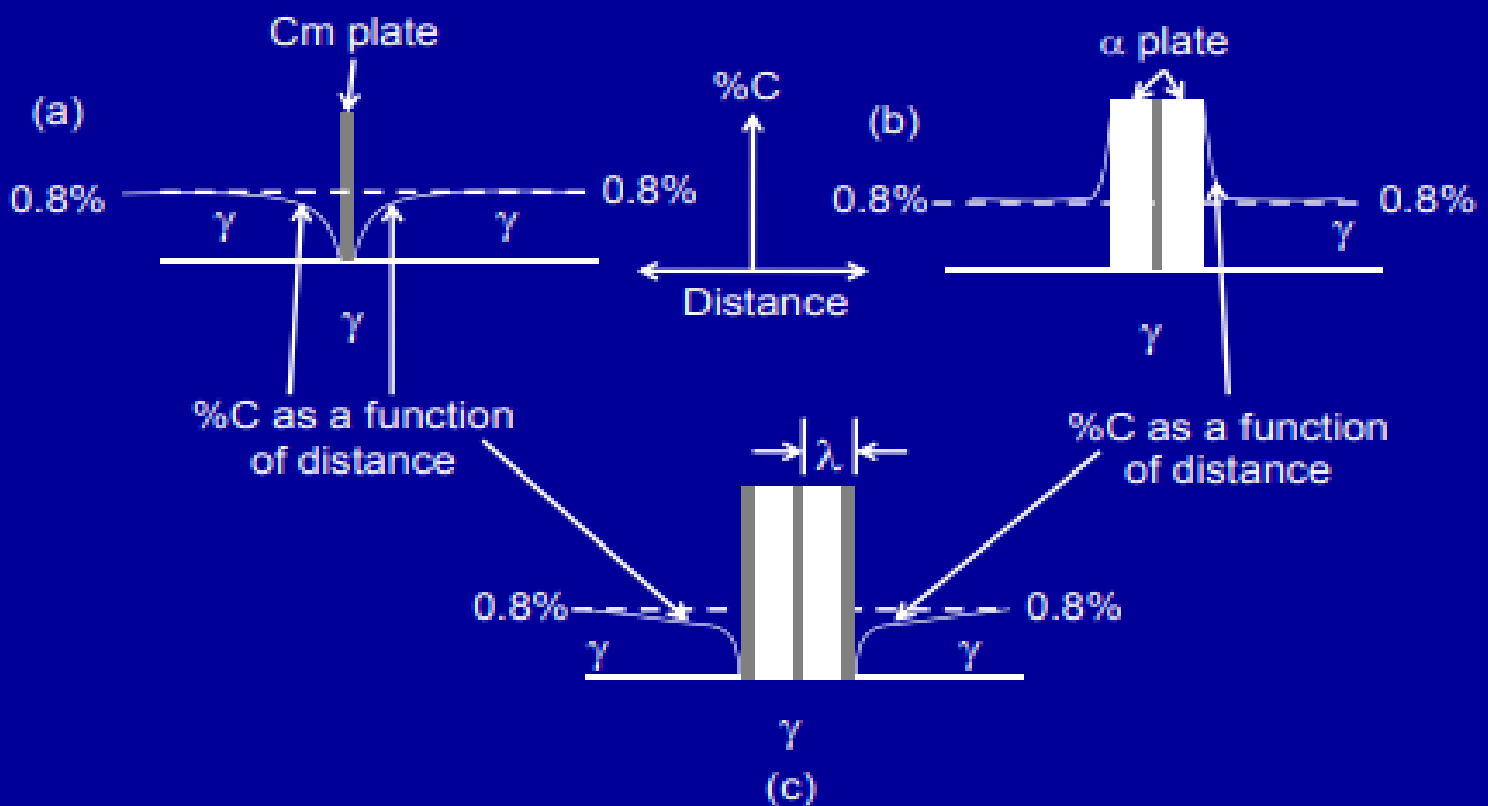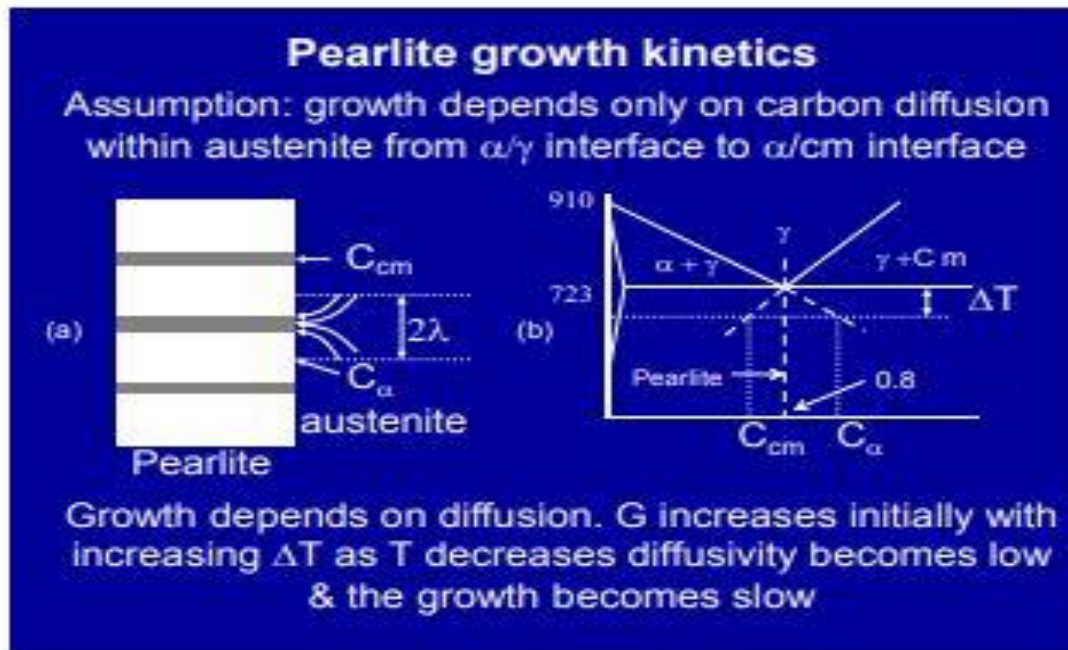
The sketch (a) in slide 3 shows how % carbon in austenite would change when a tiny plate of cementite (Cm) forms. Recall that % C in eutectoid austenite and Cm are 0.8 and 6.67 respectively. When a tiny plate of Cm forms the carbon content at the interface drops significantly. However %C in austenite away from the interface still remains at 0.8%. This is represented schematically as a plot of %C as a function of distance on the two sides of a Cm plate. The carbon profile keeps changing with time as the Cm plate becomes thicker. When %C at the interface drops below a critical limit ferrite plates nucleate on the two sides of the Cm plate. %C in ferrite is as low as 0.02%. Therefore nucleation of ferrite is accompanied by the rejection of excess carbon into austenite. The %C in austenite increases beyond 0.8%. The sketch (b) in slide 3 shows the variation of %C as a function of distance from the ferrite austenite interface. This too keeps changing with time. When it exceeds beyond a critical point Cm plates nucleate at the two ferrite austenite interfaces as shown in sketch (c) of slide 3. This too is accompanied by a change in carbon concentration at the interfaces. This is how the pearlitic transformation in steel may be assumed to be taking place.   On the basis of this) of pearlite and theλassumption it is possible to find a relation between the lamellar spacing (temperature of isothermal hold. This is illustrated in slide 4

The sketch (a) in slide 4 shows a pearlite nodule consisting of alternate layers of ferrite and cementite. The sketch (b) shows the relevant part of the Fe-Fe3C phase diagram. The dotted horizontal line denotes the

temperature at which the pearlite nodule has formed.   The line representing the composition of austenite that can coexist with ferrite as a function of temperature has been extended to intersect the dotted horizontal line denoting temperature at

a point C_α^γ. This represents the concentration of carbon at the α/γ interface. The line representing the composition of austenite that can coexist with cementite as a function of temperature has been extended to intersect the dotted horizontal line at a point C_{Cm}^γ. This represents the concentration of carbon at the Cm/γ interface. Note that $C_\alpha^\gamma > C_{Cm}^\gamma$. Therefore the growth of the nodule should be determined by the flow of carbon atoms from the α/γ interface to the Cm/γ interface. The real interfaces are curved not straight (see Fig 4).
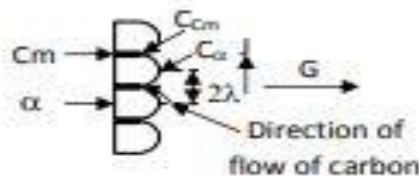

Fig 4: The interface between austenite and ferrite is not straight as shown in slide 4 but curved as in this figure. The cementite plate is too thin. Therefore the curvature is not visible.

Figure 4 indicates the direction of growth of pearlite. The growth may be assumed to be determined by the flow of carbon atom. This is governed by Fick's law of diffusion. J denotes the flux of carbon atoms. Its direction is denoted by the arrow labeled J in figure 4.

$$J = -D_C^\gamma \frac{dC}{dx} = D_C^\gamma \frac{C_\alpha^\gamma - C_{Cm}^\gamma}{\lambda} \qquad (1)$$

The growth rate (G) is is determined by J. It is directly proportional to the diffusivity of carbon in austenite and the concentration gradient that depends on the difference in carbon concentration within a specified zone of austenite. Note that $(C_\alpha^\gamma - C_{Cm}^\gamma) \propto \Delta T$ (see sketch (b) in slide 4). It is therefore obvious from equation 1 that as ΔT approaches zero (the isothermal transformation temperature is close to the eutectoid temperature) the growth rate is expected to be very slow. This is why the magnitudes of both t_s & t_f are large. As ΔT increases the growth rate increases; the magnitude of t_s & t_f decreases. When the temperature becomes too low the transformation rate is dominated by the diffusivity of carbon in austenite. Therefore the growth rate again becomes slow and the magnitudes of t_s & t_f start increasing. This explains the 'C' shaped nature of the transformation diagram.

1.11

ISOTHERMAL TRANSFORMATION OF AUSTENITE TO PEARLITE / TTT :

We have just seen how a nucleus of pearlite once formed could grow within a grain of super cooled austenite. The effect of nucleation rate was ignored. Yet it is possible to explain why the time temperature transformation diagram of a diffusion controlled transformation has a characteristic C shape. The nucleation rate too like the growth rate depends on the degree of super cooling. Let us now look at the process of transformation at a constant temperature. The kinetics of transformation at a constant temperature is best followed by monitoring fraction transformed as a function of time. Module 24 explains various ways of monitoring the kinetics of such transformation. Slide 5 shows the isothermal transformation (or TTT) diagram of eutectoid steel and beside it there is a plot of fraction transformed (f) as a function of time (t) at a constant temperature (T). The temperature at which the transformation has been monitored is shown as a horizontal dotted line on the TTT diagram (sketch (a) in slide (5)). The sketch (b) in slide 5 gives a plot of f as a function of time at a constant temperature T. Note that it has a characteristic 'S' shape. The transformation starts when the hold time t exceeds t_s. Initially the rate of transformation is slow but it increases until it reaches a maximum. Thereafter the rate of transformation keeps decreasing until the transformation is complete at time $t = t_f$

Pearlitic transformation
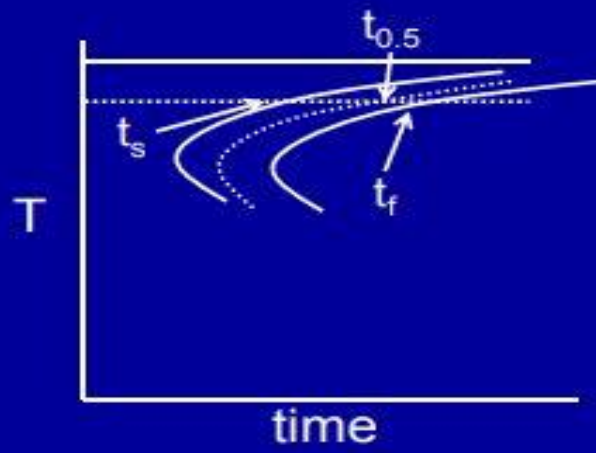
(a)
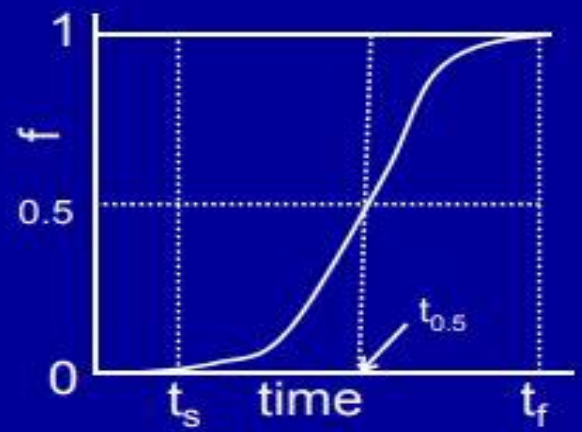
(b)

1.12 BAINITE: LOW TEMPERATURE EUTECTOID:

Bainite forms in steel if it is quenched from the state of austenite to a temperature below the nose of the TTT diagram but above the Ms (temperature) and held there long enough for the transformation to go to completion (see fig 1).

It consists of an intimate mixture of ferrite and extremely fine carbides. Unlike pearlitic transformation it starts with the nucleation of ferrite laths or plates within which the tiny particles of carbides are dispersed.

The precipitation occurs simultaneously as in the case of pearlitic transformation. It can therefore be viewed as a product of low temperature eutectoid transformation.

Like pearlite the kinetics of bainitic transformation too is controlled by diffusion.

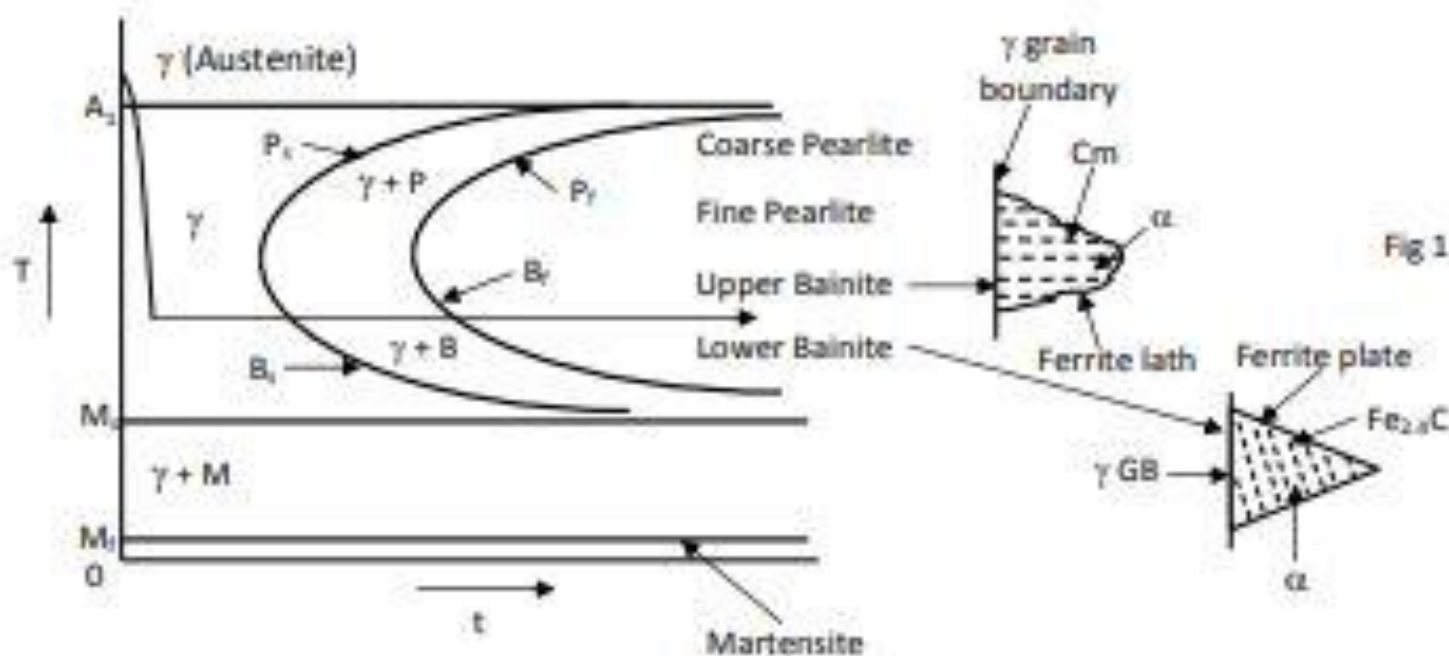However the nucleation of bainite laths or plates require higher degree of super-cooling.

Fig 1

Figure 1 shows the TTT diagram of eutectoid steel. It is reproduced here to show the main features of bainitic transformation. Bainite that forms near the nose of the TTT diagram is known **as upper or feathery bainite**.

It forms with the nucleation of ferrite laths on specific crystallographic planes of austenite and carbide platelets form on the boundary of the lath as it keeps extending. This leads to a microstructure as shown in fig 1.

The gap between the platelets is too short to be seen under optical microscope. There is a close similarity between the structure of fine pearlite and upper bainite.

The main difference is in the size of the cementite plate. In pearlite, cementite plates are long whereas in bainite it is made of a large number of broken platelets aligned along the axis of the ferrite lath
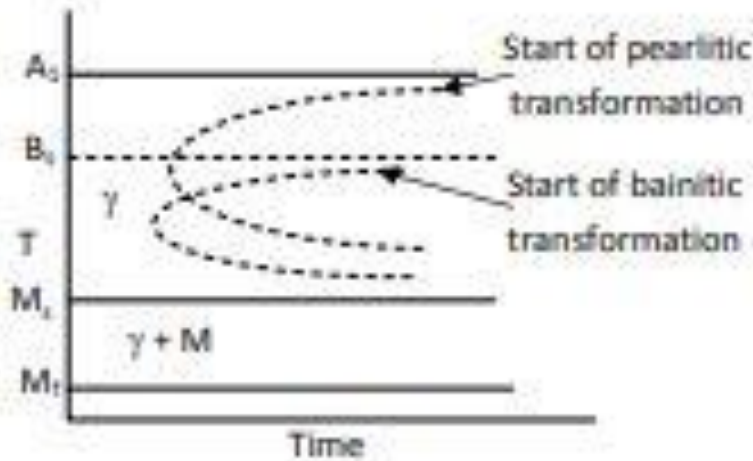
Fig 2

The time it takes for pearlitic or bainitic transformations to start can be represented by two different 'C' shaped plots. It is shown in fig 2. Pearlitic transformation occurs just below the eutectoid temperature (A1). Bs is the temperature below which austenite can transform into bainite. There should be a pair of similar 'C' shaped plots representing the time it takes for the transformations to be completed. These have not been shown to avoid overcrowding of too many curves within the sketch in fig 2. The lower bainite has a distinctly different structure. Its main features are shown with the help of a sketch in fig 1.

 Lower bainite, also known as acicular bainite, forms with the nucleation of ferrite plates with a fine dispersion of thin rod shaped carbides. The composition of the carbide corresponds to Fe2.4C. It is known as epsilon carbide).

 These are aligned at an angle with respect to the axis of the ferrite plate. We would later ε(come across similar structure having a dispersion of epsilon carbide in a ferrite matrix during tempering of steel.

This is why some time lower bainite might get identified as tempered martensite. This can be avoided by a careful examination under TEM or SEM at higher magnification.

 Bainite is also characterized by the shorter distance between neighboring carbides. This suggests that diffusion of carbon during bainitic transformation is limited over a very short distance. We know that the diffusion distance is given by $\sqrt{Dt}$ where D denotes diffusivity of carbon in austenite.

 Bainite forms at a relatively low temperature where the diffusivity of carbon is low. This is why carbides in bainite are in the forms of short platelets or thin & short rods.

1.13 MARTENSITIC TRANSFORMATION:

 Martensite forms in steel if it is quenched from the austenitic state to a temperature below Ms.   Unlike pearlite and bainite it is a homogeneous (single) phase. Austenite is FCC. It is stable at high temperature. The solubility of carbon in austenite is much more than that in ferrite the low temperature form of iron which is BCC. If the steel is cooled slowly carbon gets enough time to diffuse within austenite. As a result it transforms into a mixture of ferrite and carbide. However on quenching atoms of carbon are retained within the lattice even if the crystal structure transforms from FCC to BCC. The presence of excess carbon in BCC lattice is responsible for tetragonal distortion. This is why the crystal structure of martensite is BCT. It may be '). The most importantαconsidered as a super saturated solid solution of carbon in alpha iron (

Isothermal transformation   Athermal transformation
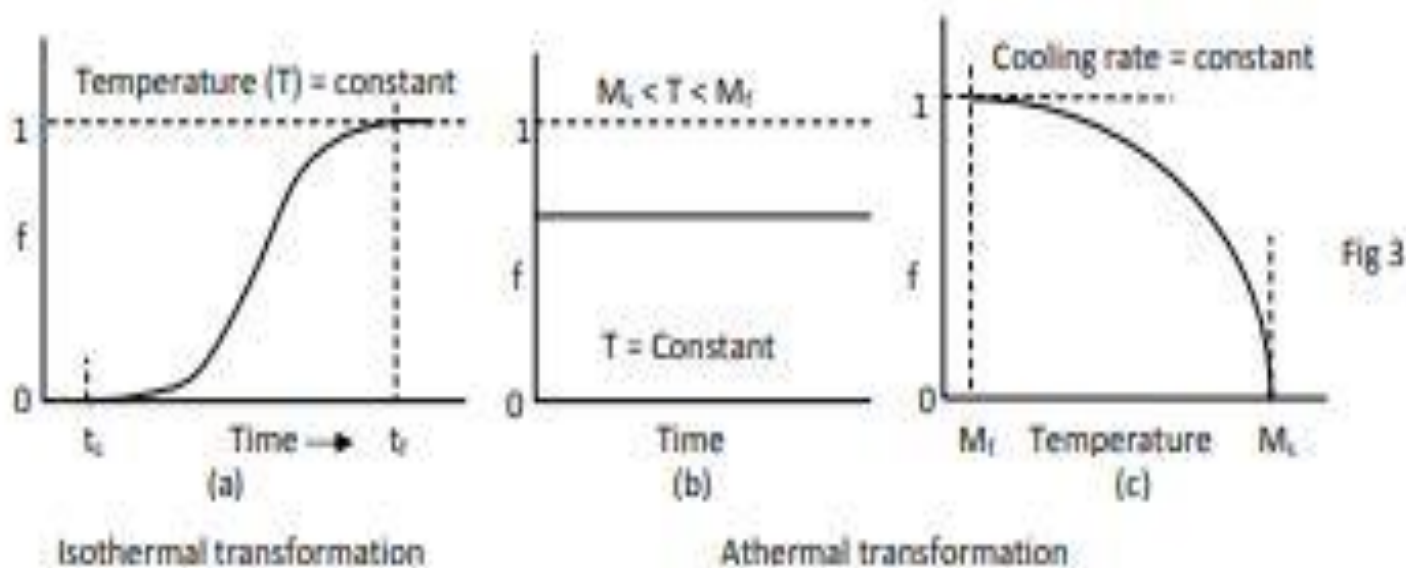
characteristics of martensitic transformation are as follows:

- **Athermal:** Transformation occurs as long as the temperature keeps dropping.

- **Diffusion-less:** The composition of martensite is exactly same as that of austenite.

- **Crystal structure**: Body centered tetragonal. Its c/a ratio is a function of carbon content.

- **Hardness**: It is hard& brittle. Hardness increases with %C. It reaches RC 64 at %C = 0.6

- **Morphology**: Depends on %C. It is acicular or needle like if %C > 0.3. Lath if %C < 0.3.

- **Speed**: Extremely fast (~ velocity of sound / elastic wave). Takes place by shear.
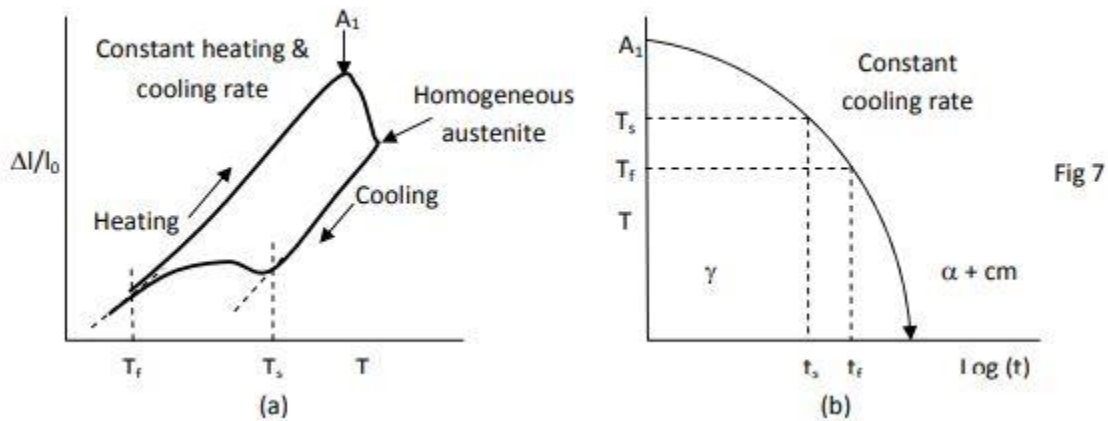
- **CCT DIAGRAM**

Figure 7(a) shows how from the dilation versus T plot the transformation temperatures can be obtained. During heating the length of the sample increases almost linearly due to thermal expansion.

As T goes a little beyond the eutectoid temperature ferrite carbide structure start transforming into austenite. Since austenite has higher packing density than ferrite it is accompanied by contraction.

By the time T reaches 20°-30°C more than A1 it completely transforms into homogeneous austenite (in case of eutectoid steel). Thereafter the cooling stahe begins.

 Initially the sample contracts with decreasing temperature.

The dilation versus T plot is nearly linear till it drops to Ts. Here onwards the length of the sample starts increasing. This is because of **volume expansion accompanying the transformation of austenite into a mixture of ferrite and carbide**.

 The expansion continues until T reaches Tf where the transformation is complete. Thus from the dilation plot one can get temperatures at which the transformation begins (Ts) and ends (Tf).

Once these are known the time at which the transformation begins (ts) and ends (tf) at this cooling rate can be directly read from the cooling curve as shown in fig 7(b).

CCT diagram
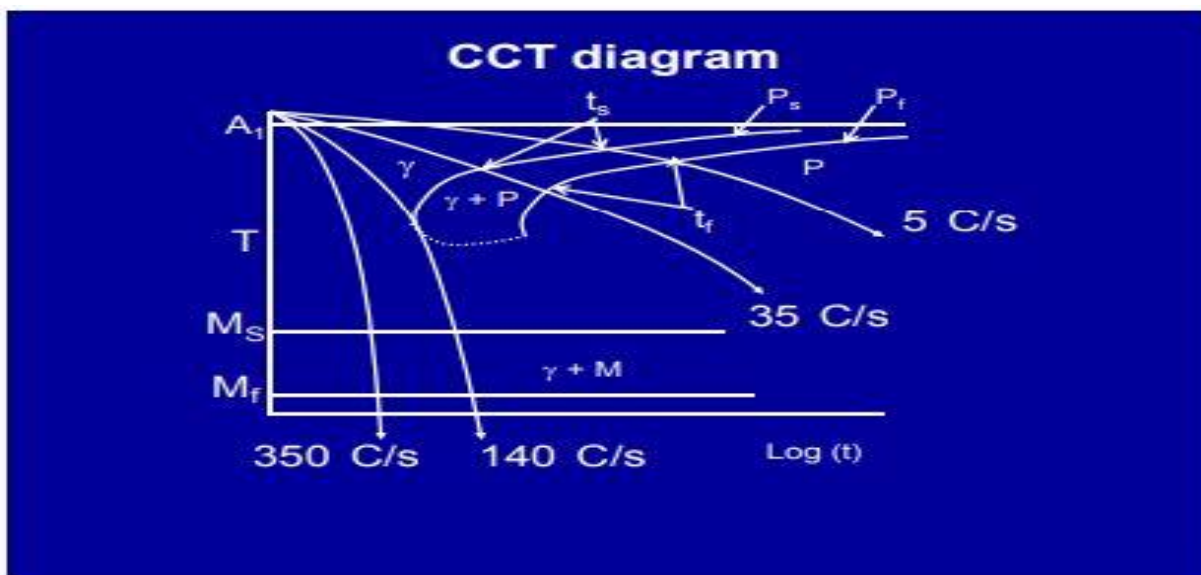
Slide 6

Repeat the experiment just described for a set of different cooling rates. Prepare a table giving the Ts, Tf, ts and tf as a function of the cooling rates. The temperature at which the transformation begins and ends depends on cooling rate if it is slow.

However beyond a critical cooling rate it becomes independent of cooling rate. This defines the cooling rate above which the nature of the transformation is athermal. Plot these data to generate a transformation diagram as shown in slide 6.

Note how ts and tf for the two cooling rates (5°C/s and 35°C/s) are marked on the corresponding T versus t plots. Beyond a cooling rate Ts and Tf are independent of cooling rate. They represent Ms and Mf temperatures.

The line joining the points denoting ts at various cooling rates represent the time at which the transformation of austenite to pearlite begins. The curve is labeled as Ps. Similarly the line joining the points denoting tf at various cooling rates represent the time at which the transformation of austenite to pearlite is complete. The curve is labeled as Pf. The sketch shown in slide 6 is known as CCT or the continuous cooling transformation diagram

RELATIONSHIP BETWEEN TTT AND CCT

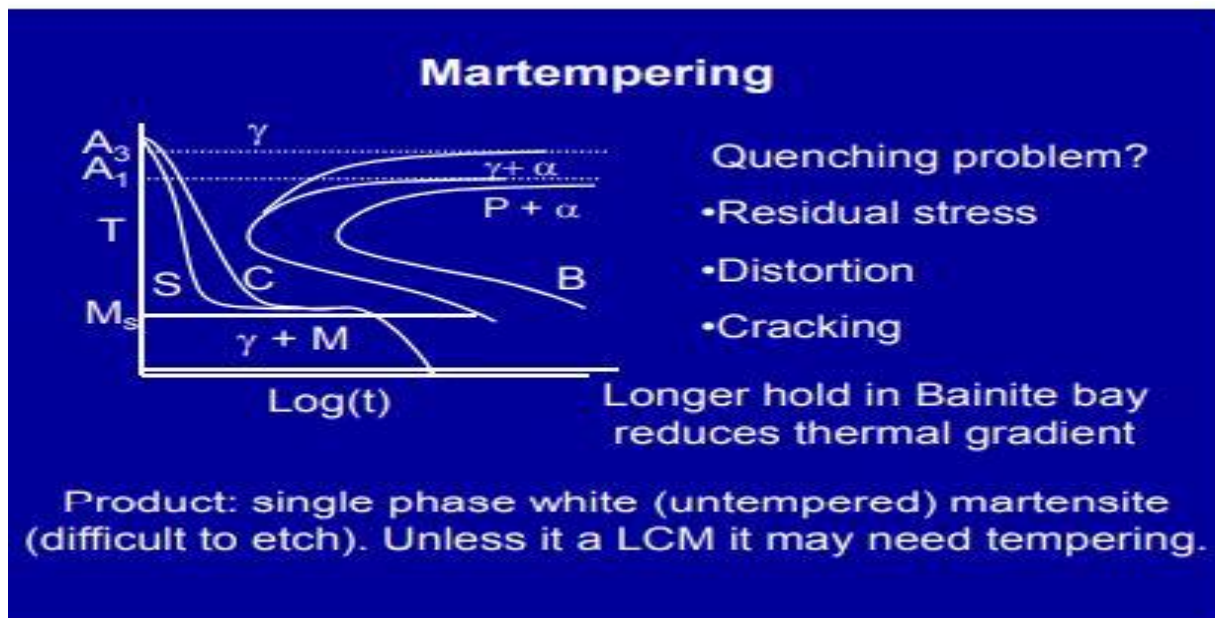Fig 8: Shows the relation between TTT and CCT diagram. A pair of dotted lines denotes the starting and finishing times for isothermal transformations. The curves labeled as P_s & P_f denote the starting and finishing points for pearlitic transformation during continuous cooling. These are at located at a little lower temperature than the corresponding plots for isothermal transformation. The M_s and M_f temperatures are still the same. Bainite does not form during continuous cooling in carbon steel.

Figure 8 shows the relation between the locations of the starting and the finishing lines in the CCT and TTT diagrams of eutectoid steel.

The curves labeled Ps & Pf correspond to the starting and finishing points of pearlitic transformation in the CCT diagram. The dotted 'C' curves are the corresponding plots in its TTT diagram.

There is continuous cooling curve superimposed on the diagram. It intersects the TTT curves at two points: a & b; whereas a' and b' are the points of intersection with CCT plots.

When the cooling curve intersects the dotted line denoting the start of pearlitic transformation the effective hold time at this temperature is not long enough for the transformation to begin.

The temperature should drop a little more for the pearlitic transformation to begin. This is why the **point a' is located at a lower temperature** and longer time in comparison to those corresponding to the point a.

Similarly when the cooling curve intersects the dotted line denoting the finishing points for isothermal transformation the effective hold time at this temperature is not long enough for the completion of the transformation. **The point b' (see fig 8) on the cooling curve therefore, is located at a little lower temperature.**

2.11

MARTEMPERING

This is a special heat treatment technique to decrease the severity of quenching to minimize residual stress. Instead of direct quenching to room temperature the part is first quenched in a bath maintained at a temperature a little above Ms. It is kept there for a while before final quenching. The isothermal hold helps reduce the temperature difference between the surface T is not expected to be low.Δ(S) and the core (C) of the part. During the final quenching when The surface and the core may undergo martensitic transformation more or less at the same time. This is illustrated with the help of a diagram given in slide 3
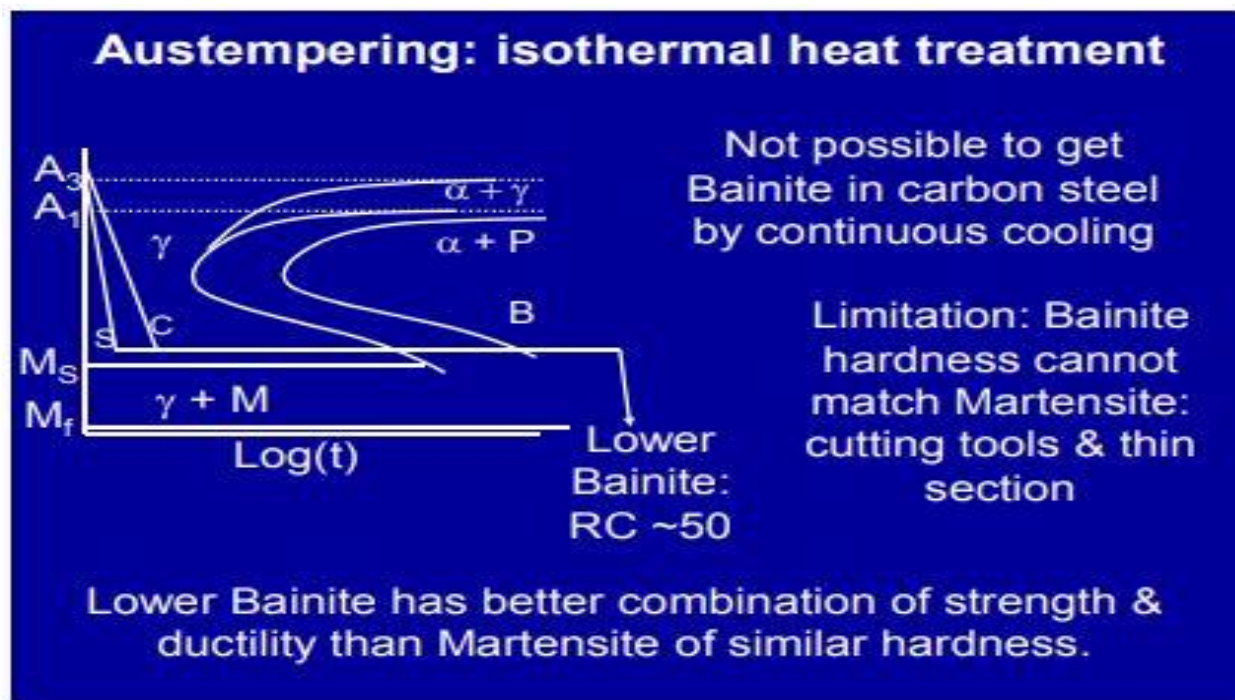

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The sketch in slide 3 has cooling curves of the surface and the centre superimposed on the TTT diagram of hypo-eutectoid steel. Isothermal hold in the bainitic bay helps reduce the temperature difference. The treatment is known as martempering.   It helps reduce residual stress, susceptibility to distortion and cracking. One of the objectives of tempering is to reduce quenching stress. It can be avoided if %C in the steel is low. The final structure would consist of 100% un-tempered martensite. It is difficult to etch. If %C in martensite is high it should be tempered.

AUSTEMPERING

It is an isothermal heat treatment process where you get 100% bainite all through the section of a part. Bainitic structure has certain advantages. It is hard yet tough. Hardness of lower bainite can be as high as Rc50. It has better combination of strength and toughness than that of martensite of similar hardness. Maternsite needs to be tempered. On tempering its structure becomes a mixture of ferrite and carbide. Bainite being is a mixture of ferrite and carbide. This is similar to that of tempered martensite. Therefore austempered structure need not be tempered again. Besides this the isothermal hold allows enough time for temperature within the part to become uniform. As a result transformation occurs simultaneously at the surface and the centre. There is no chance of developing any residual stress. Besides this it is the only way you could get bainite in plain carbon steel.   Slide 4 illustrates with the help of a TTT diagram important steps to be followed during austempering of hypo-eutectoid steel.


The sketch in slide 4 has cooling curves of the surface and the centre superimposed on the TTT diagram of hypo-eutectoid steel. Isothermal hold in the bainitic bay helps reduce the temperature difference between the surface(S) and the centre(C).

3.0 HARDENABILITY

Qantitative parameter that defines the ability of steel to harden on quenching. It should be independent of the size and shape of the component and the quenching medium.

  Hardenability may be defined as the ability of steel to become hard all through the section on quenching from its austenitic state

(a)   (b)   (c)
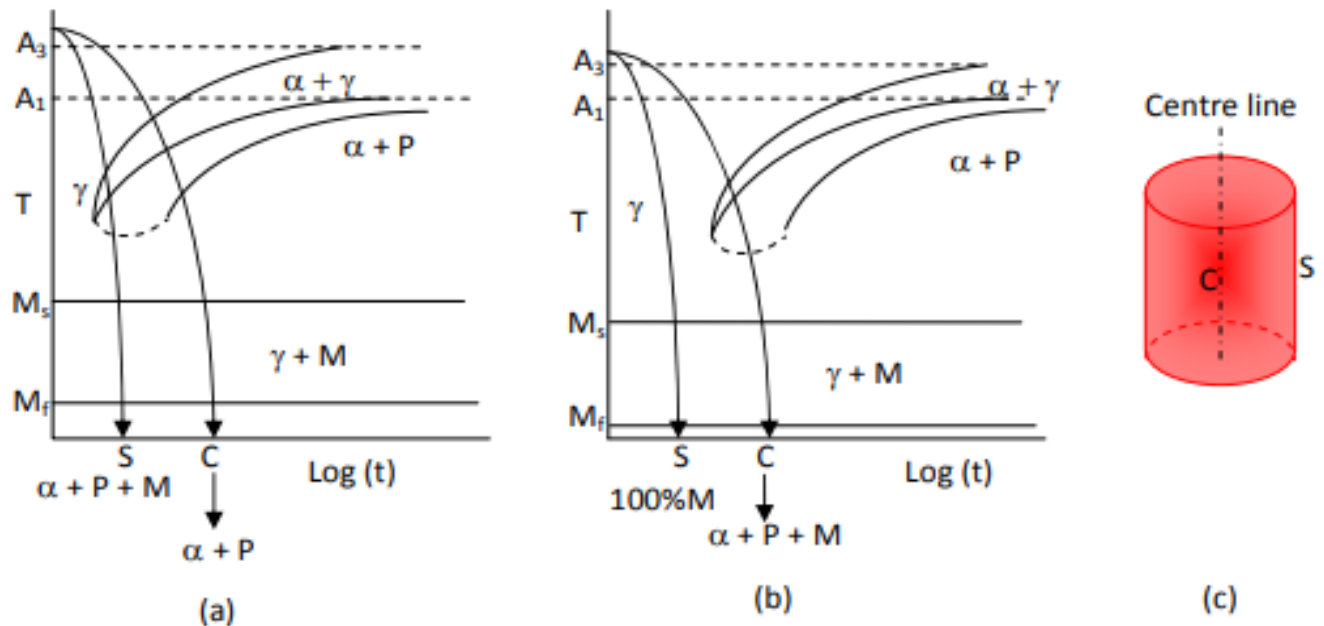
Hardenability may be defined as the ability of steel to become hard all through the section on quenching from its austenitic state. When you quench steel in cold water a blanket of steam would immediately surround it. This has relatively poor conductivity. Therefore the initial heat extraction rate may not be very high. However as the steam blanket disintegrates into small bubbles that float up due to the difference in density the cooling rate increases significantly.   This stage is known as nucleate boiling. It is promoted by agitation. It may soon reach a steady state when the heat flow from the core to the surface of the part by conduction becomes equal to the heat being extracted by convection current set up within the water due to boiling and agitation. Clearly under such a situation the cooling rates at the centre and the surface cannot be the same. The difference between the two is a function of the size of the part made of the same steel.   This is illustrated with help of a set of diagrams given in fig 3. Note that on quenching the surfaces of the two parts come in contact with water at the same temperature. We may therefore assume that the cooling rates at the surface to be the same. However the cooling rates at the centers are widely different

Ideal critical diameter (DI): It is the diameter of the cylindrical sample which on quenching from the appropriate austenitizing temperature in an ideal quenching medium gives 50% martensite and 50% fine pearlite at its centre. The magnitude of H for such a medium is infinity. Clearly DI should be greater than Dcrit for a particular quenching medium. There are charts and tables that are obtained empirically or by numerical solution of heat transfer equations under appropriate convective boundary conditions at the surface to convert Dcrit to DI. Slide 12 gives an example of the same. The chart Dc versus DI has a set of lines representing the correlation between the two. It illustrates how to convert Dc to DI with the help of a pair of dotted lines if it denotes the critical diameter for water quenching.

Ideal critical diameter adequately describes hardenabilty of steel. It is a material property. It does not depend on the size and geometry of steel. It gives an estimate of the depth of hardness of steel on quenching in an ideal quenching medium having infinite severity of quench (H). However it depends on the composition of steel, austenitizing temperature (or austenite grain size) and the homogeneity of austenite