

KINETICS

KINETIC OF PHASE TRANSFORMATIONS

Equilibrium \Rightarrow PHASE DIAGRAMS

NON equilibrium \Rightarrow KINETIC OF PHASE TRANSFORMATIONS

Kinetic = things change (evolve) \Rightarrow KINETIC = evolution
 = rate of speed at which the phenomenon occurs.

Example: liquid \rightarrow solid \Rightarrow transformation has time to occur

\Rightarrow liquid copper to solid copper \Rightarrow NO ENERGY STRAIN
 \Rightarrow volume adapts inside liquid

but solid \Rightarrow solid \Rightarrow transformation has troubles

\Rightarrow plenty of ENERGY STRAIN \rightarrow volume of α must squeeze β to make space for α (expand)

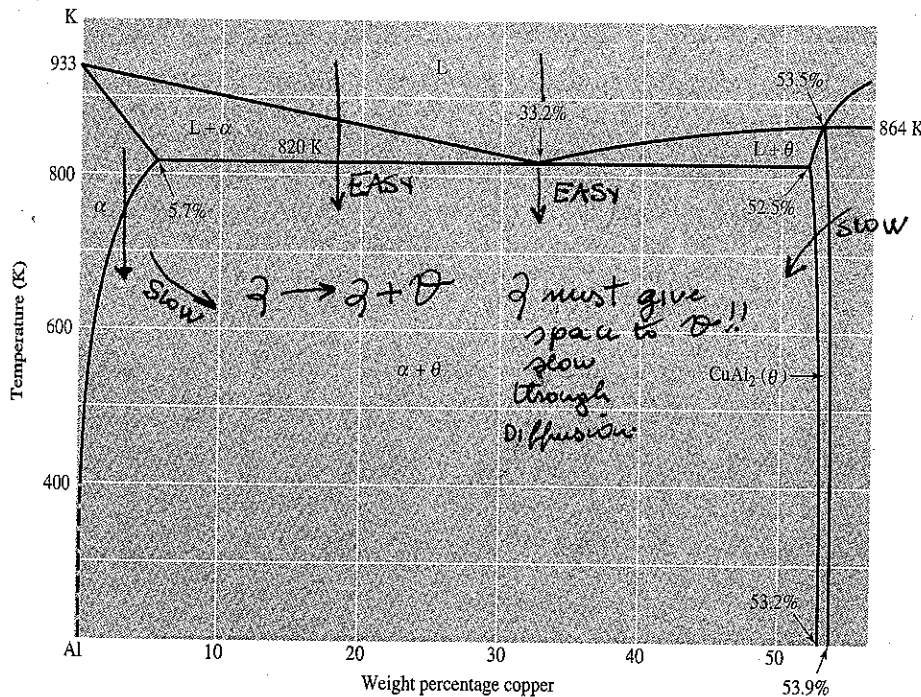


FIGURE 8.2-1
 The aluminum-rich end of the Al-Cu binary phase diagram.

SLOW
 Slow KINETIC
 EASY
 FAST KINETIC

K1

IF I remove that too fast (T \downarrow FAST)
 then I create some microstructures that I might not want

for practical alloys, many heat treatments are necessary! (to get rid of microstructures)
form

It's all a matter of temperature \Rightarrow NOT E but G !!

DRIVING FORCE OF A TRANSFORM.

multiply
 gibbs free
 energy

$$E \rightarrow dE = -pdV + Tds$$

$$H = E + PV \Rightarrow dH = Vdp + Tds$$

$$G = E + PV - TS \quad dG = Vdp - SdT$$

$\underbrace{H - TS}$

\rightarrow min(G) \rightarrow stable equilibrium

\rightarrow for const V & S workers \Rightarrow min(E)
 (or min(E) instead of $-pdV$)

for constant pressure & entropy problems \Rightarrow min(H)

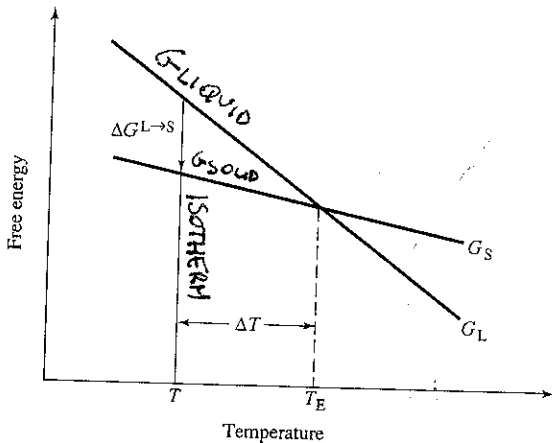
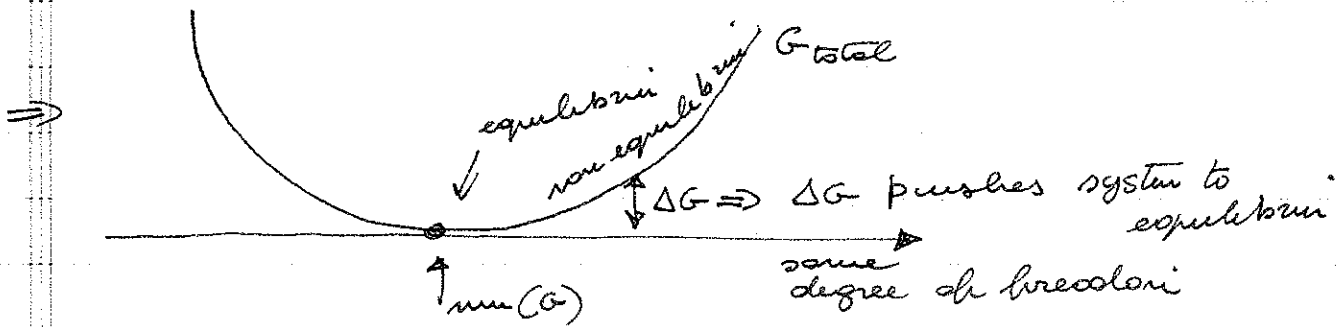
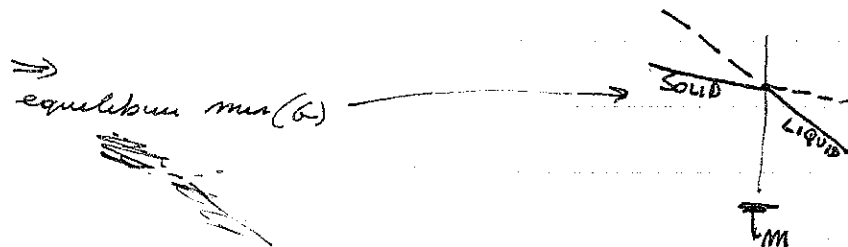


FIGURE 8.2-2
 An approximation of the variation in free energy for the liquid and solid phases close to the equilibrium melting temperature.



K2

LIQUID & SOLIDS have different S & E
 \Rightarrow different slopes!!

Transformation LIQ \rightarrow SOLID

$\Delta G^{L \rightarrow S}$
is driving force

at const T

$$G = H - TS$$

$$\Delta G^{L \rightarrow S} = \Delta H^{L \rightarrow S} - T \Delta S^{L \rightarrow S}$$

FINAL-INITIAL

$$\Delta G^{L \rightarrow S} = G^S - G^{Liq}$$

$$\Delta S^{L \rightarrow S} = S^S - S^L$$

$$\Delta H^{L \rightarrow S} = H^S - H^L$$

Remember

$$dH = Vdp + Tds$$

so if ya keep always $P=1$
 $= \text{const}$

$$\Rightarrow dH = Tds = \delta Q = \text{internal heat}$$

$$\Rightarrow \Delta H_{P=1}^{L \rightarrow S} = \delta Q^{L \rightarrow S}$$

Release or absorption
of heat during $L \rightarrow S$
reaction !!

LATENT HEAT
EXOTHERMIC
ENDOTHERMIC

$$\Delta G^{L \rightarrow S} = \Delta H^{L \rightarrow S} - T \Delta S^{L \rightarrow S}$$

@ melting T_M

$$\Delta G^{L \rightarrow S} = 0 \Rightarrow \text{system switches status}$$

$$\Delta H^{L \rightarrow S} - T_M \Delta S^{L \rightarrow S} = 0$$

$$\Delta S^{L \rightarrow S} = \frac{\Delta H^{L \rightarrow S}}{T_M} = \frac{\delta Q^{L \rightarrow S}}{T_M}$$

$$\Rightarrow \Delta G^{L \rightarrow S} = \Delta H^{L \rightarrow S} - T \frac{\Delta H^{L \rightarrow S}}{T_M}$$

$$\Delta T \equiv T_M - T$$

$$\Delta G^{L \rightarrow S} = \Delta H^{L \rightarrow S} \left(1 - \frac{T}{T_M} \right) = \Delta H^{L \rightarrow S} \frac{\Delta T}{T_M}$$

$$\Delta G^{L \rightarrow S} = \frac{\Delta H^{L \rightarrow S}}{T_M} \Delta T$$

$$\Delta T = T_M - T$$

$$\Delta H^{L \rightarrow S} = \delta Q$$

K3

Which is Favorite?

depends on sign of $\Delta H^{L \rightarrow S}$

$$\Delta H^{L \rightarrow S} = H^S - H^L = \delta Q^{L \rightarrow S}$$

$$\Delta G = \frac{\Delta H^{L \rightarrow S}}{T_M} \Delta T$$

higher (lower) is the temperature
(& bigger is ΔT) \Rightarrow the
stronger is $\Delta G \Rightarrow$ stronger
is driving force

Temperature is EVERYTHING

- 1) T gives defects (necessary)
- 2) T gives entropy (diversity) ch configurations
- 3) ΔT gives two of them driving force

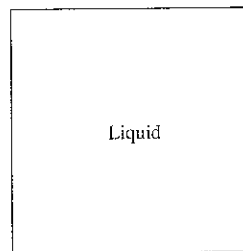
NUCLEATION

HOMOGENEOUS
the particles
appear randomly

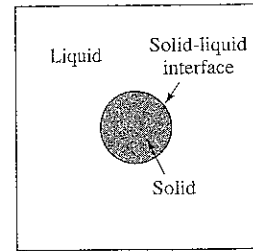
HETEROGENEOUS
the particles
appear in specific
places

FIGURE 8.2-3

Schematic showing (a) a region containing only liquid, and (b) a region containing liquid with a solid and a solid-liquid (S/L) interface.



(a)



(b)

EXAMPLE = PURE SUBSTANCE

pure substance

liquid has more energy in L
than in S $\Rightarrow H^S < H^L$

$\Rightarrow \delta Q^{L \rightarrow S} < 0$ HEAT $\downarrow \Rightarrow$
INTERNAL
 \Rightarrow HEAT IS EXPELED

EXOTHERMIC

$$\Delta H^{L \rightarrow S} < 0$$

at high temp $T > T_M$

$$\Delta T = T_M - T < 0$$

$$\Rightarrow \Delta G^{L \rightarrow S} = \frac{\Delta H^{L \rightarrow S}}{T_M} \Delta T < 0$$

$\Rightarrow L \rightarrow S$ IS DISFAVORED

$\Rightarrow S \rightarrow L$ TRANSFORMATION

at low temperature $T < T_M$

$$\Delta T > 0$$

$$\Delta G^{L \rightarrow S} = \frac{\Delta H^{L \rightarrow S}}{T_M} \Delta T > 0$$

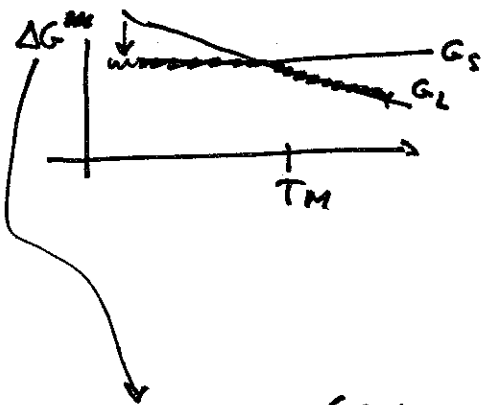
G \downarrow gets below T_M
minimum with $L \rightarrow S$

$L \rightarrow S$ IS FAVORED !!

K4

HOMOGENEOUS NUCLEATION

- HAVE LIQUID
- COOL IT UNDER T_M
wants to SOLIDIFY



$$\Delta G^{L \rightarrow S} = \Delta H^{L-S} \frac{\Delta T}{T_M}$$

$$\Delta T = T_M - T$$

$$\Delta G^{L \rightarrow S} = G^S - G^L$$

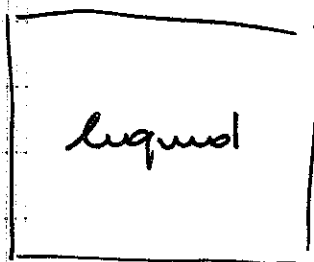
EXOTHERMIC ~ release heat
 $\Rightarrow \Delta H^{L \rightarrow S} < 0$
 (wants to solidify)

(like E, S, ...)

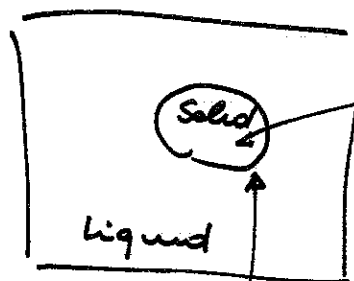
this ΔG is extensive ($2V \rightarrow 2\Delta G$)
 \Rightarrow let's introduce $\Delta G_v = \Delta$ Free energy per unit volume.

$$\Rightarrow \Delta G = V \Delta G_v \quad \text{// simpl.} \quad < 0 \quad \text{(system wants to solidify)}$$

(J/m^3)



Homogeneous melting



approx sphere of radius r

Formation of a surface

CREATION OF SURFACE \Rightarrow REQUIRES ENERGY

(like making a chewing-gum bubble)

$$\text{Energy} \propto \text{surface} = \gamma_s \cdot \text{surface}$$

(J/m^2)
 $\gamma_s = \text{surface energy}$

⇒ to create (nucleate) sphere of radius r

$$\text{surface} = 4\pi r^2$$

$$\text{Volume} = \frac{4}{3}\pi r^3$$

$$\frac{4}{3}\pi r^3 \Delta H \xrightarrow{L \rightarrow S} \frac{\Delta T}{T_m}$$

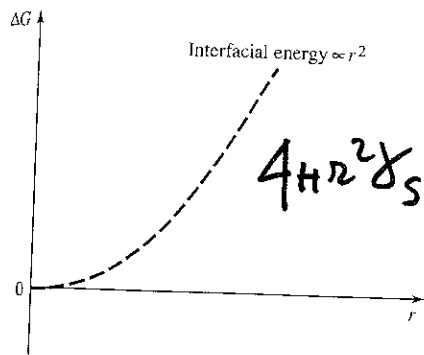
$$\Rightarrow \Delta G = 4\pi r^2 \gamma_s + \frac{4}{3}\pi r^3 \Delta G_v$$

positive > 0
the system ~~loves~~
HATES SURFACE

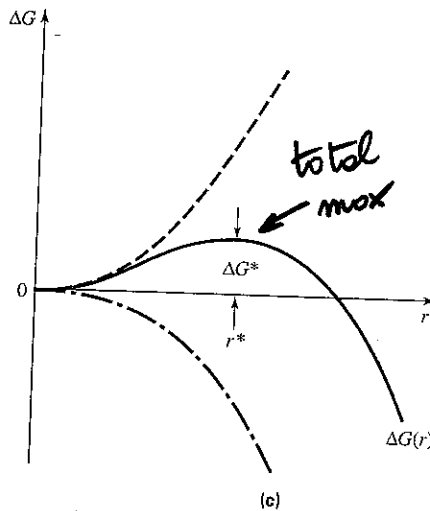
NEGATIVE < 0
the system
LOVES TO SOLIDIFY

r^2

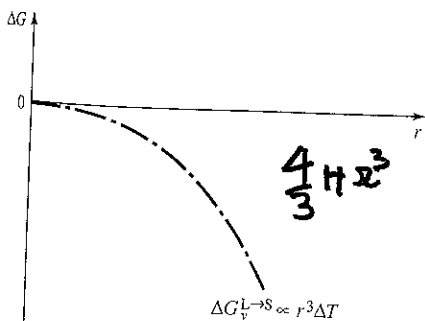
r^3



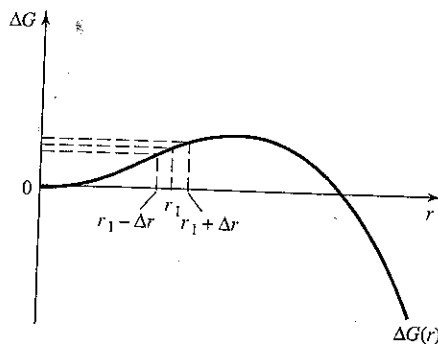
(a)



(c)



(b)



(d)

FIGURE 8.2-4

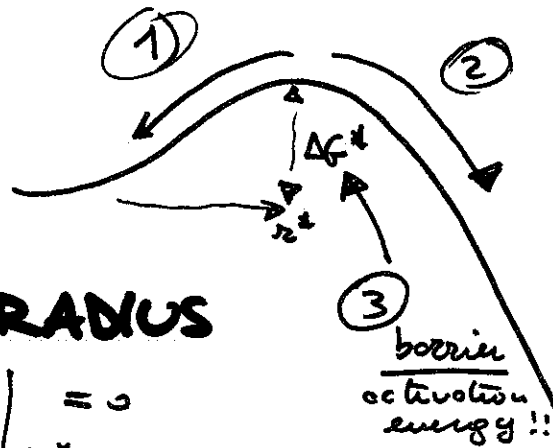
The dependence of the various energy terms associated with nucleation as a function of the radius of the growing phase: (a) the relationship between cluster radius and surface energy of growing spherical solid phase in a liquid, (b) the relationship between the cluster radius and $\Delta G_v^{L \rightarrow S}$, (c) the sum of the previous two curves, and (d) an annotated version of ΔG plotted versus r (see footnote 1 on page 294).

K6

Remember ~~that~~ that

$\min(G) = \text{thermodynamical equilibrium}$

$\Rightarrow G$ WANTS TO DECREASE



1) all particles $r < r^*$ will disappear, become liquid & DIE ...

2) all particles $r > r^*$ will grow, eating liquid and becoming bigger and bigger solids!!

3) we need to feed the system with more than ΔG^* to

start nucleation
 \downarrow
 sort of activation energy

CRITICAL RADIUS

$$r^* \Rightarrow \left. \frac{\partial \Delta G}{\partial r} \right|_{r^*} = 0$$

$$0 = \frac{\partial \Delta G}{\partial r} = \frac{\partial}{\partial r} \left[4\pi r^2 \gamma_s + \frac{4}{3} \pi r^3 \Delta G_{\nu}^{L \rightarrow S} \right]$$

$$= 8\pi r \gamma_s + 4\pi r^2 \Delta G_{\nu}^{L \rightarrow S} = 0 \Rightarrow$$

$$\Downarrow$$

$$2\gamma_s + r \Delta G_{\nu}^{L \rightarrow S} = 0$$

$$r^* = - \frac{2\gamma_s}{\Delta G_{\nu}^{L \rightarrow S}}$$

$$= - \frac{2\gamma_s T_M}{\Delta H_{\nu}^{L \rightarrow S} \Delta T}$$

$$r^* \propto \frac{1}{\Delta T}$$

$\Delta T \uparrow \Rightarrow r^* \downarrow \Rightarrow \text{easy to nucleate}$

K7

So

$$r^* = \frac{-2\gamma_s}{\Delta H_v} \frac{T_m}{\Delta T}$$

$\Delta G^* = \Delta G(r^*) \Rightarrow$ plug r^* into $\Delta G(r)$ and obtain

$$\Delta G^* = \frac{16\pi\gamma_s^3 T_m^2}{3(\Delta H_v)^2} * \frac{1}{\Delta T^2}$$

$\Delta T \uparrow \Delta G^* \downarrow$
barrier.

UNDERCOOLING INCREASES ($\Delta T \uparrow$), THE ENERGY BARRIER DECREASES.

another problem: ~~mobility~~ MOBILITY OF LIQUID DUE BY difference of concentration



Particles in the liquid must touch the surface to become solid (particles move thru DIFFUSION)
 \Rightarrow at low temperature
 Diffusion $\propto \exp(-Q/T)$
 $T \downarrow D \downarrow \Rightarrow$ small # of particles! to feed the solid.

at very low T solid solidification decreases.

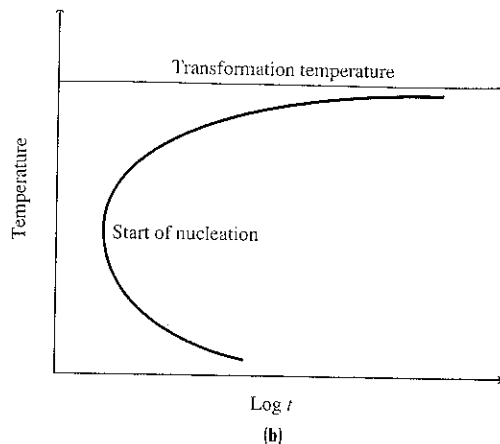
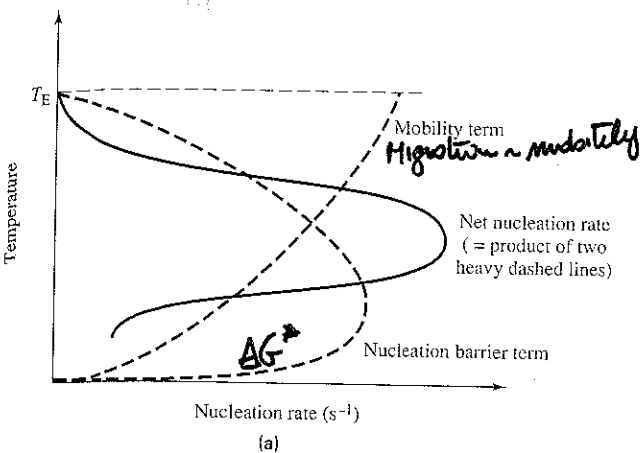


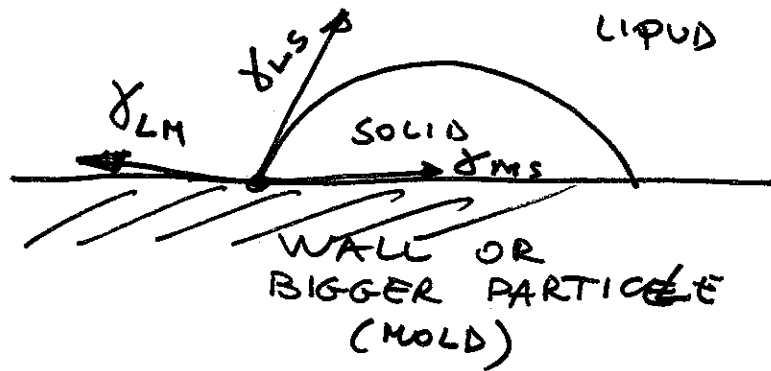
FIGURE 8.2-5 (a) The influence of temperature on the mobility term and the nucleation barrier term. The opposing processes result in a maximum in the nucleation rate at an intermediate temperature. (b) Since the time for nucleation is inversely related to the nucleation rate, the time curve exhibits a minimum at an intermediate temperature. Because of its shape, this curve is often referred to as a C curve.

HOMOGENEOUS
 ~ NEVER.
 always start
 in some place
 HOMOGENEOUS + NEED SPACE
K8

HETEROGENEOUS NUCLEATION

- IF WE ADD SOME SUBSTANCE (ANOTHER BIG PARTICLE) TO HELP NUCLEATION
 - WALLS, CORNERS
- $\delta \downarrow$
 $\delta \downarrow$ } HOW?

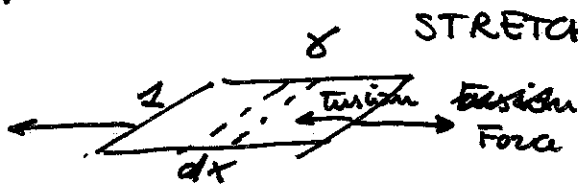
WETTING PROBLEM



3 surface energies:
 Liquid \leftrightarrow Solid
 liquid \rightarrow mold (wall)
 mold \rightarrow solid

Surface energy = surface tension

ρ



$$\underbrace{-\text{tension} \cdot dx}_{\text{WORK MADE}} = \underbrace{\delta \cdot dA}_{\text{energy gained}} = \delta dx$$

Surface tension \leftarrow

= (surface energy) with direction of ~~growth~~ growth to direction when surface grows

SOLID-LIQUID surface growth



LIQ.-MOLD surface growth

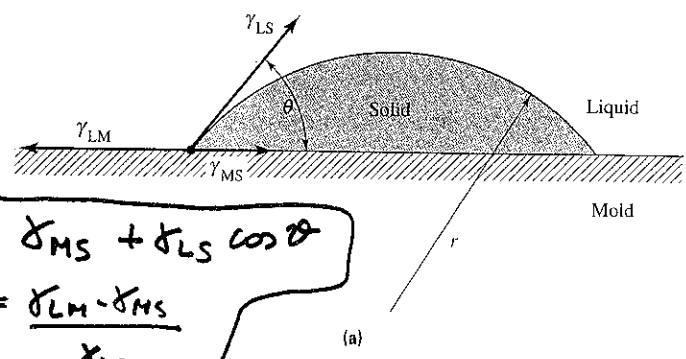


SOLID-MOLD surface growth

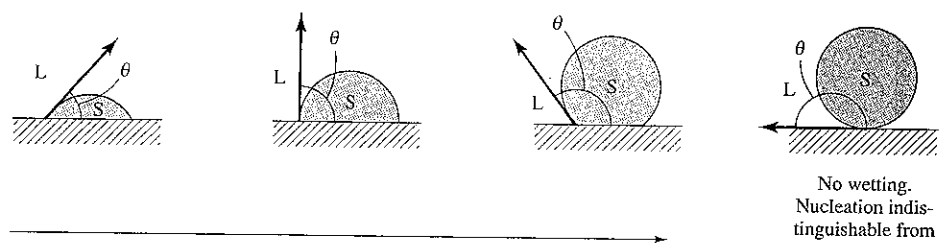
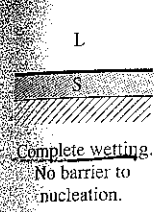


K9

⇒ and $\gamma_{LS}, \gamma_{LM}, \gamma_{SM} \Rightarrow$ determine slope to bubble.



Mech Equilibrium
YOUNG'S LAW 1805
 $\gamma_{LM} = \gamma_{MS} + \gamma_{LS} \cos \theta$
 $\Rightarrow \cos \theta = \frac{\gamma_{LM} - \gamma_{MS}}{\gamma_{LS}}$



No wetting.
Nucleation indistinguishable from homogeneous nucleation.

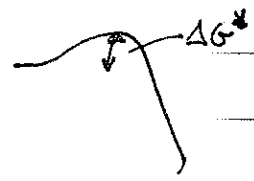
ADDING
Volume does NOT
ADD SURFACE LS

$\theta = 0$ $f(\theta) = 0$	$\theta = \pi/4$ $f = 0.058$	$\theta = \pi/2$ $f = 0.500$	$\theta = 3\pi/4$ $f = 0.943$	$\theta = \pi$ $f = 1.000$
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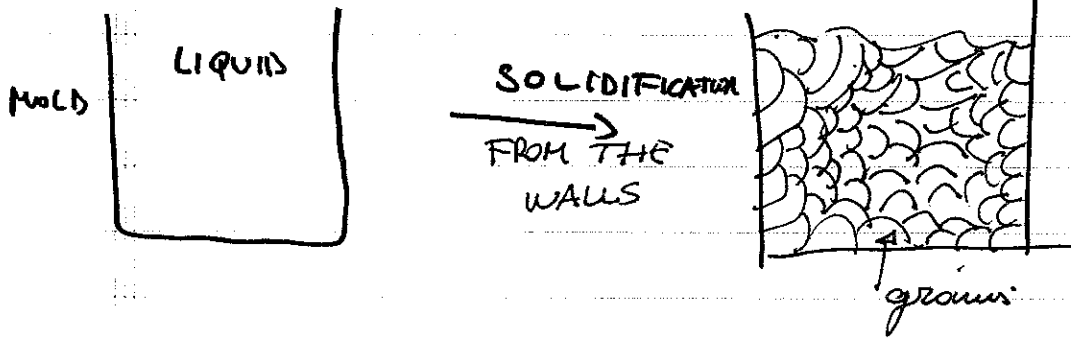
FIGURE 8.2-6 The wetting characteristics of a number of solid-liquid systems: (a) The angle θ , as defined in the illustration, is an indication of the wettability of the system; (b) as θ increase from $\theta = 0$ to $\theta = \pi$, the wettability decreases; (c) the function $f(\theta)$ varies from 0 to 1 as θ varies from 0 to π .

$$\Rightarrow \Delta G_{HET}^* = \Delta G_{HOM}^* \cdot f(\theta)$$

↳ $\theta = 0$ (complete wetting)
NO BARRIER TO SOLIDIFICATION



K16



IF WE PREVENT

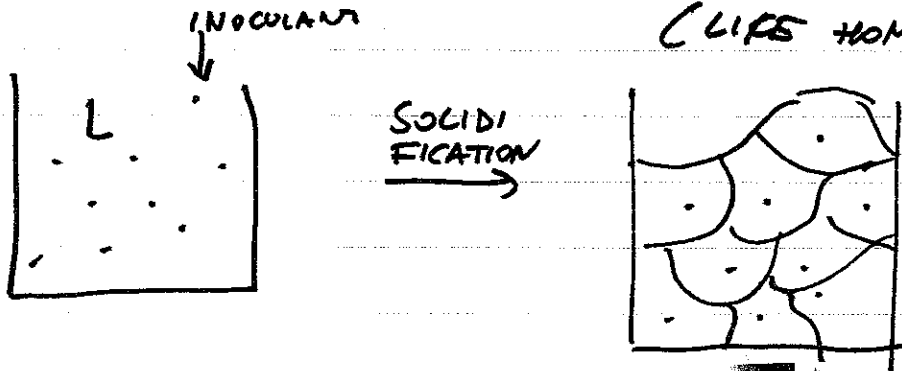
WETTING ON SURFACE OF THE MOLD

(TiB₂ particles in Al)
INOCULANT



NO WETTING ⇒ ~~ABSENCE~~
ON SURFACE ⇒

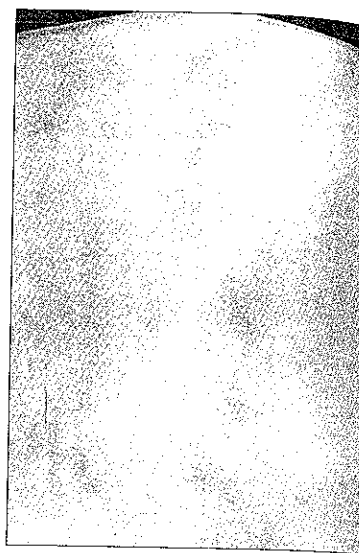
NUCLEATION HETEROGENEOUS
A IN RANDOM POSITIONS
(LIFE HOMOGENEOUS)



bigger
grains ⇒ better
CAST



(a)



(b)

FIGURE 8.2-7 Photomicrographs of a transverse section from a 20-in x 54-in cast aluminum alloy: (a) with no ingot grain refiner, and (b) with an ingot grain refiner. Illustration is half the cross section.

K11

SOLID-SOLID INTERFACES (WHEN IN SOLID 2 or + PHASE REGIONS)

Liquid \leftrightarrow Solid interfaces & γ 's generate different nucleation.

$\frac{3}{2+P}$

Solid \leftrightarrow Solid interfaces, what happens in I love a big particle inside a sea (matrix) of atoms ($A_x B_y$) A (precip)

MATRIX \leftrightarrow PRECIPITATE interfaces

COHERENT (α, β same lattice type)

COHERENT INTERFACES: one to one atom correspondence

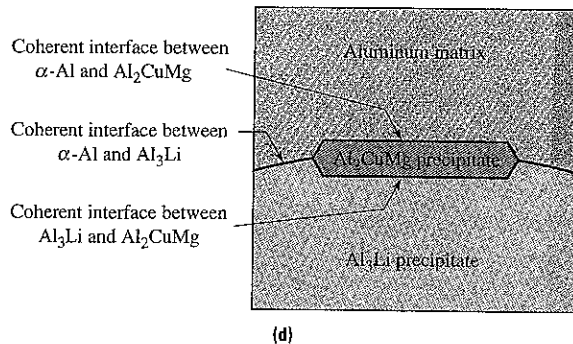
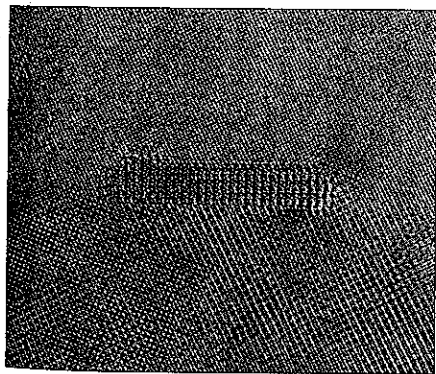
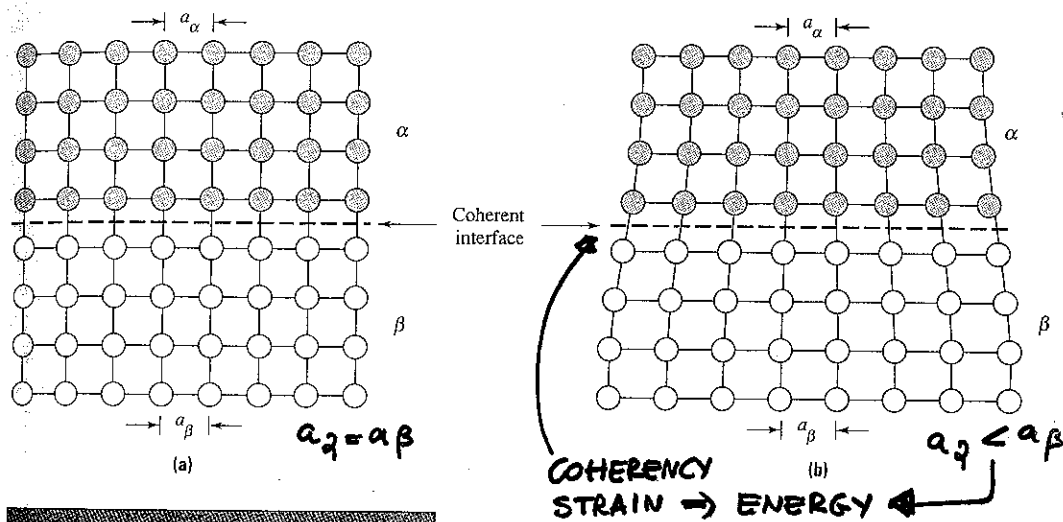


FIGURE 8.2-8 Schematic illustration of coherent interface between two phase α and β . (a) The crystal structures and lattice parameters are identical. (b) The crystal structures are identical but the lattice parameters are different. The difference in lattice parameters leads to coherency strains. (c) High-resolution image showing the coherent interface that exists between the aluminum matrix and two precipitates that form during artificial aging of an Al-Cu-Mg-Li alloy. The white dots are columns of atoms that form the phase. (d) A schematic of the photomicrograph in part c identifying the phases and the coherent interfaces.

K12

SEMI COHERENT (α, β same lattice type)
 $a_\alpha < a_\beta$, $\text{push } a_\beta \uparrow \Rightarrow$ every once in awhile one atom slips out the COHERENCY STRAIN
 \Rightarrow INTRODUCE PERIODIC DISLOCATION (depend on DIRECTION)

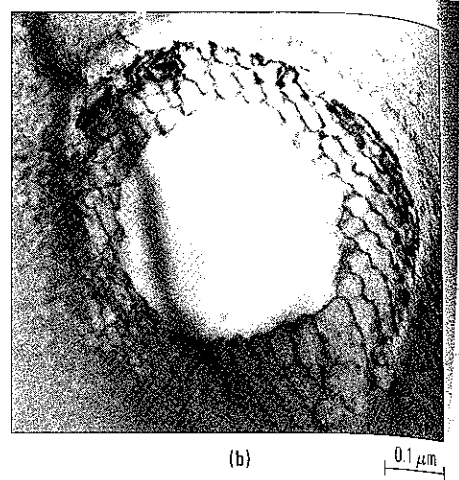
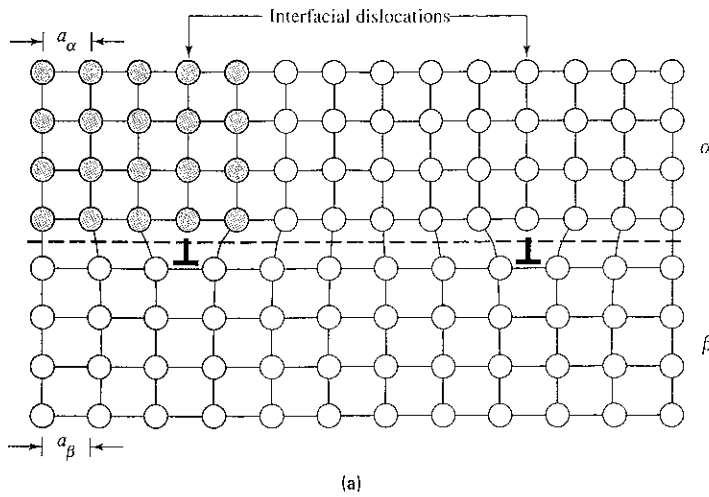


FIGURE 8.2-9 (a) Schematic of a semicoherent interface between two phases. The insertion of the periodically spaced dislocations eliminates coherency strains but increases the interfacial energy. (b) Regular array of dislocations at the matrix/precipitate interface of a semicoherent precipitate in Astroloy, a high-temperature Ni-base superalloy. (Source: Photomicrograph courtesy of Mario Luis Maciá).

INCOHERENT (α, β different lattice types and $a_\alpha \neq a_\beta$)

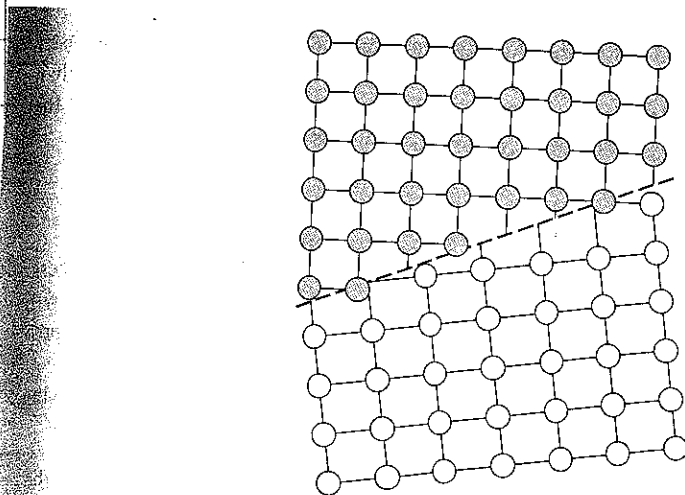


FIGURE 8.2-10 An incoherent interface occurs when the crystal structures and lattice parameters of two phases are different.

largest interfacial energy $\gamma_{\alpha\beta}$

Fe₃C in Fe
 COMPOSITE \leftrightarrow BCC
 FERRITE
 PEARLITE

K13

REMEMBER

$$\Delta G^* = \frac{16\pi (\gamma_{SL})^3 T_M^2}{3 (\Delta H_{f \rightarrow s})^2} + \frac{1}{(\Delta T)^2} \quad \text{for solidification}$$

TABLE 8.2.-1 Relationship among interfacial energy, the nature of the nucleation process, and the precipitate.

Type of interface	$\gamma_{\alpha\beta}$ Interfacial energy	Nucleation process	Location of precipitates	Number of precipitates per unit volume
Coherent	Increasing	Homogeneous	Throughout the matrix	$\sim 10^{18}/\text{cm}^3$ many
Semicoherent		Heterogeneous	Mostly at dislocations	Orders of magnitude less than homogeneous nucleation
Noncoherent		Heterogeneous	At grain boundaries	Orders of magnitude less than homogeneous nucleation

$\gamma_{\alpha\beta}$
 $\sim 200 \text{ mJ/m}^2$
 $\sim 500 \text{ mJ/m}^2$
 $\sim 1000 \text{ mJ/m}^2$

for solid \rightarrow solid solutical $\Delta G_{\alpha \rightarrow \beta}^*$, $\gamma_{\alpha\beta}$, $\Delta H_{f \rightarrow s}$

$$\Delta G_{\alpha \rightarrow \beta}^* \propto (\gamma_{\alpha\beta})^3 \rightarrow \text{NON COHERENT REALLY DOES NOT WANT TO SPONTANEOUSLY APPEAR}$$

ΔG^*
 COHERENT \rightarrow precipitates precipitate everywhere (EASY)

SEMICOHERENT \rightarrow precipitates precipitate around dislocations (which relieve $\gamma \downarrow$) (MEDIUM)

NON-COHERENT HETEROGENEOUS \rightarrow precipitates precipitate ONLY IN regions of grain boundaries where plenty of γ relieve (HARD)

K14

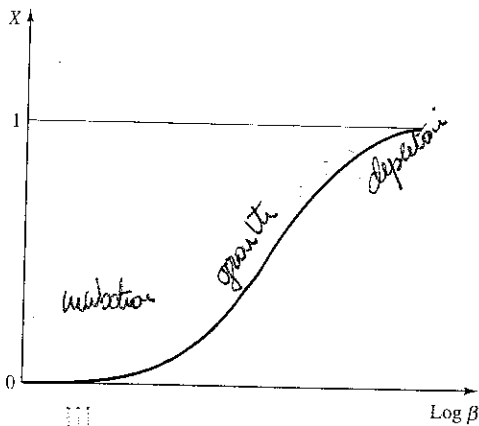
HETEROGENEOUS FORMING PARTICLES GROWTH FAST BECAUSE DEFECTS ARE DIFFUSION CHANNELS

GROWTH OF A PHASE

Nucleation \Rightarrow growth of phase:
how it works in time

INCUBATION
 \downarrow
 GROWTH
 \downarrow
 SATURATION
 FRACTION TRANSFORMED

1. There is a time period, often called the incubation time, required to nucleate the phase; the incubation time is a function of undercooling. No measurable phase transformation occurs during the incubation time.
2. Once nucleated, the phase begins to grow and there is a rapid increase in the amount of new phase present.
3. Eventually, the growth rate of the new phase decreases because of either depletion of solute or physical impingement of the growing phase.



AVRAMI EQUATION

$$X = 1 - \exp\left[-\underbrace{(kt)^m}_\beta\right]$$

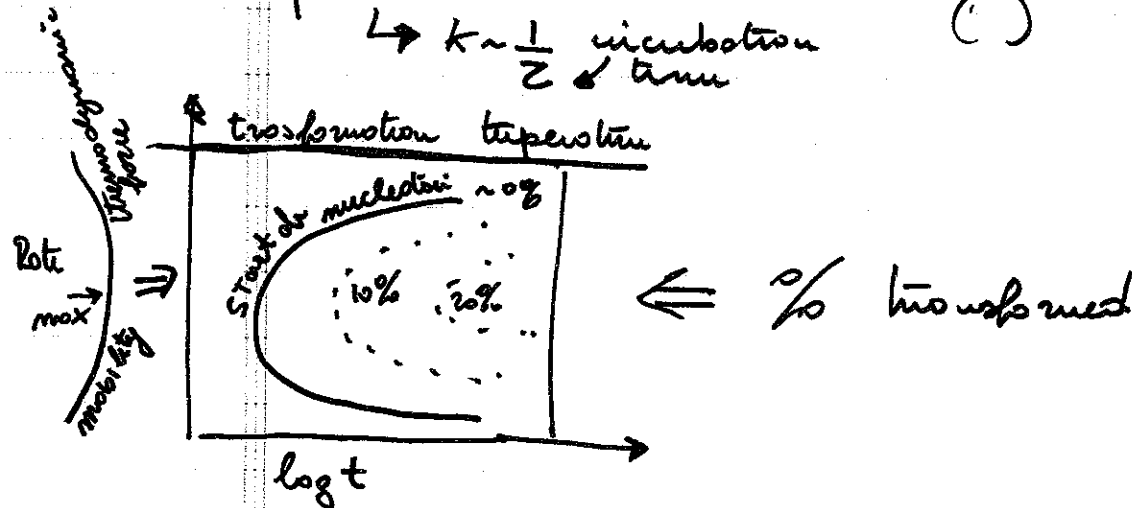
$\beta \equiv kt$

INCUBATION = time required for a stable nucleus to appear with $r > r^*$

$\beta = kT$

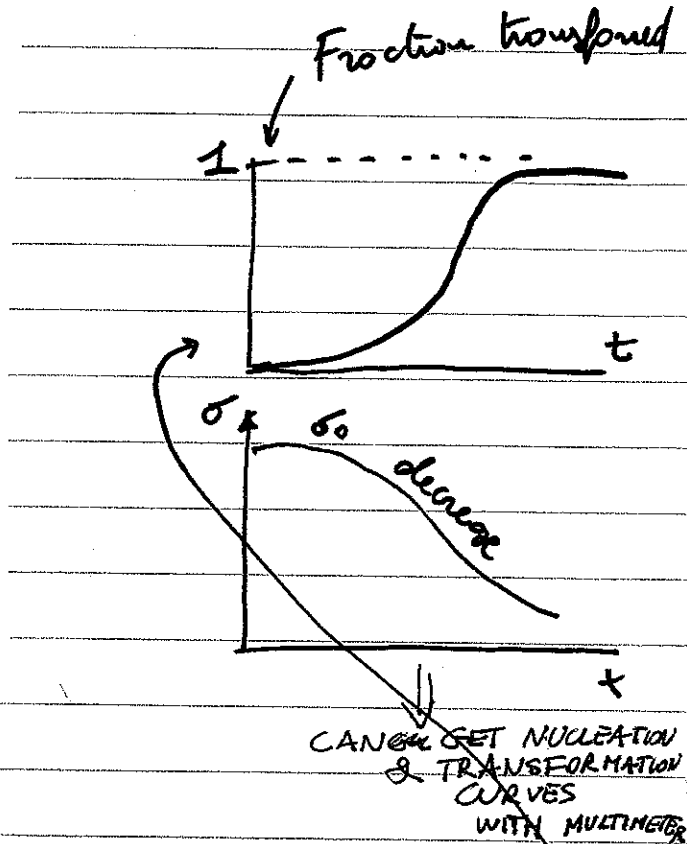
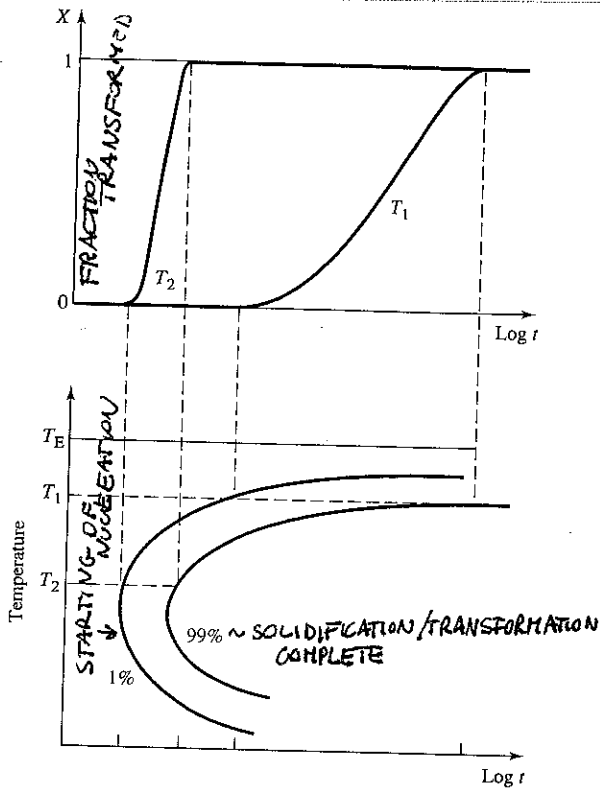
$\hookrightarrow k \sim \frac{1}{\tau}$ incubation time

$(m) \rightsquigarrow 0.5 \text{ to } 5$
depend on system & structure

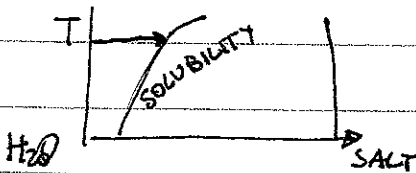


K15

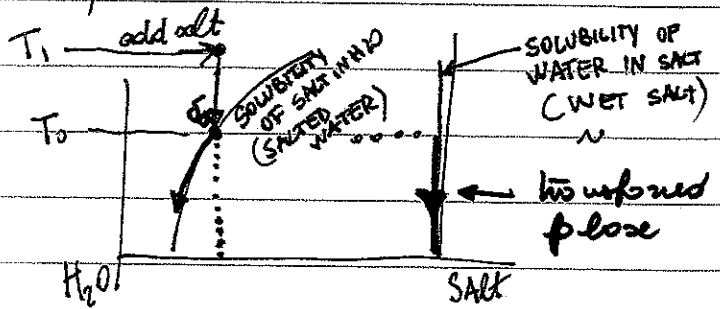
⇒ ISOTHERMAL Growth (T=const)



EXAMPLE : WATER + SALT



- 1) • HEAT WATER
- 2) • PUT SALT (A LOT)
- 3) • GO DOWN IN TEMP
- 4) • SALT WILL PRECIPITATE



- 6) SALT PARTICLES DO NOT CONDUCE
- 7) SALTED WATER CONDUCE

$$\sigma = \frac{\mu n q}{\approx \text{const}}$$

↑ mobility ↑ const charge

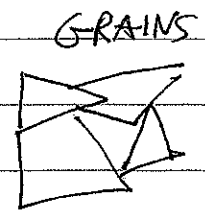
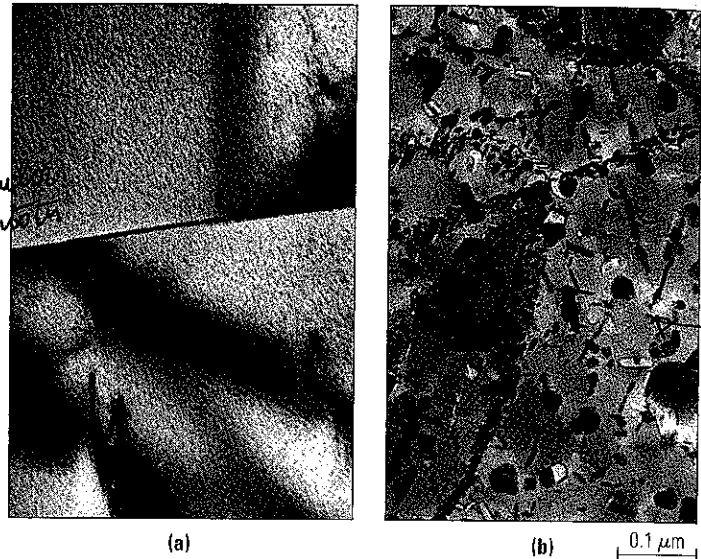
$$n(T) = \text{conc of salt in water}$$

5) @ T1, salted water loses salt & SALT NUCLEATE

$$\Rightarrow = 1 - \text{fraction of transformed phase} \Rightarrow$$

K16

Homogeneous Nucleation
 → intergranular precipitates



Heterogeneous nucleation
 ⇒ starts precipitates start from defect:

ULTRA GRANULAR (Bethen grains)

FIGURE 8.2-13 The distribution of precipitates in an aluminum alloy: (a) precipitates that homogeneously nucleated, and (b) those that heterogeneously nucleated. In part a, the precipitates are the fine, pepperlike features that are scattered throughout.

This example illustrates two key points: (1) altering the heat treatment has a significant effect on the size and spatial distribution of the precipitates through its influence on the nucleation and growth rate, and (2) since macroscopic material properties depends strongly on the size and spatial distribution of the phases, heat treatment is a powerful tool for modifying the properties of engineering materials.

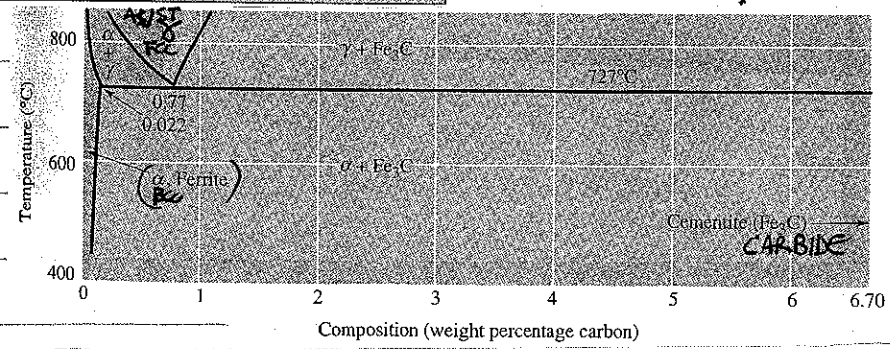
⇒ Heat treatment twist results!!

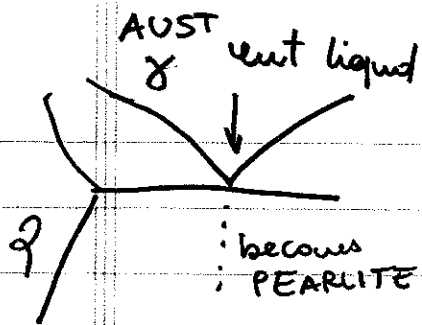
PHASES IN STEEL

Microstructure name	Description
δ -ferrite	An interstitial solid solution of carbon in δ iron (BCC).
Austenite	An interstitial solid solution of carbon in γ iron (FCC).
α -ferrite	An interstitial solid solution of carbon in α iron (BCC).
Pearlite	Eutectoid of α -ferrite and cementite with a lamellar microstructure of alternate α -Fe and cementite plates.
Bainite	Eutectoid of α -ferrite and cementite. The α -ferrite either has a feathery appearance or occurs as plates. Carbide particles lie between the α -ferrite regions.
Spheroidite	Spherical particles of cementite in a matrix of α -ferrite.
Martensite	An interstitial solid solution of carbon in a body centered tetragonal (BCT) Fe crystal structure.
Hypoeutectoid steels	Alloys with compositions to the left of the eutectoid reaction.
Hypereutectoid steels	Alloys with compositions to the right of the eutectoid reaction.
Proeutectoid ferrite	Ferrite that forms prior to the eutectoid ferrite.
Proeutectoid cementite	Cementite that forms prior to the eutectoid cementite.

EUTECTOID
 $\gamma \rightarrow \alpha + Fe_3C$
 (NO SOLUBILITY OF Fe in Fe_3C compound)

K17



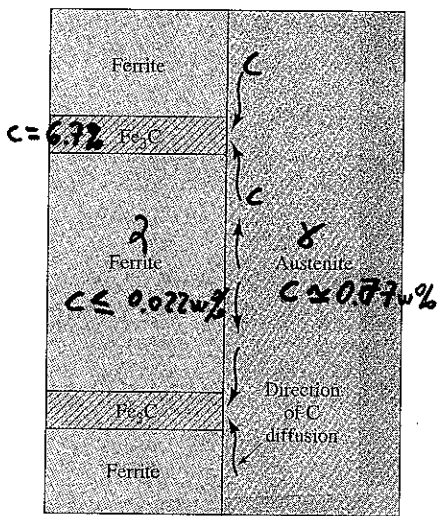


becomes PEARLITE = lamellar alternation of α -ferrite & cementite

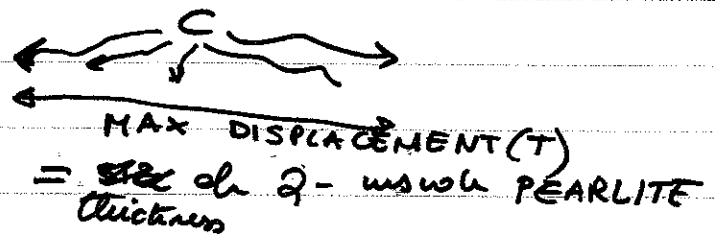
INTERFACE

γ -austenite & pearlite MOVES during phase transformation

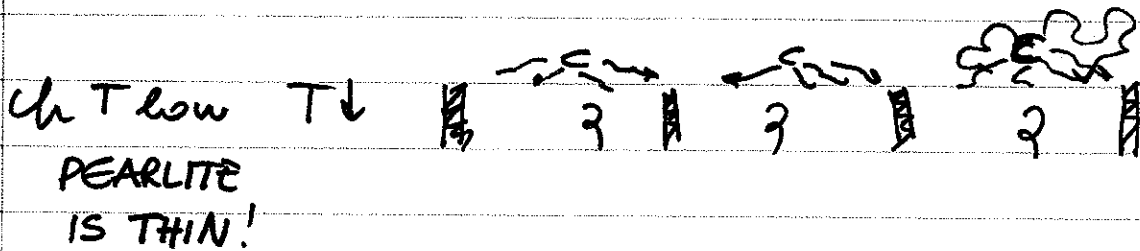
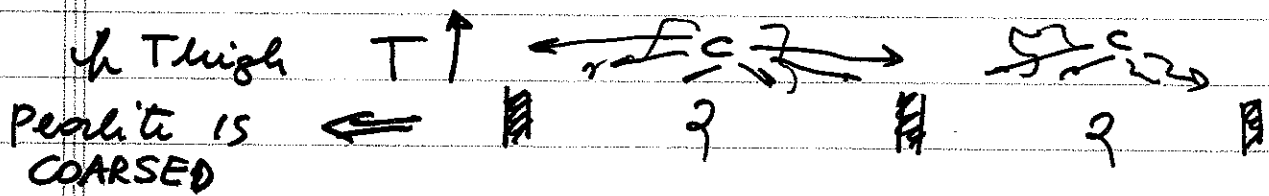
Direction of $\alpha + Fe_3C$ growth



MIGRATION OF C inside γ -austenite toward $Fe_3C \Rightarrow$
C cannot move more than a distance



MIGRATION = DIFFUSION (T)



different EXTERNAL Temperature \Rightarrow DIFFERENT PEARLITE density!!!

K18

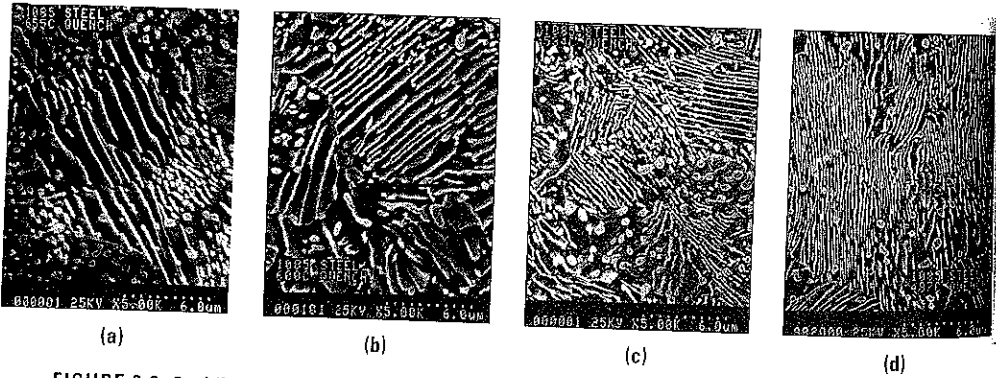
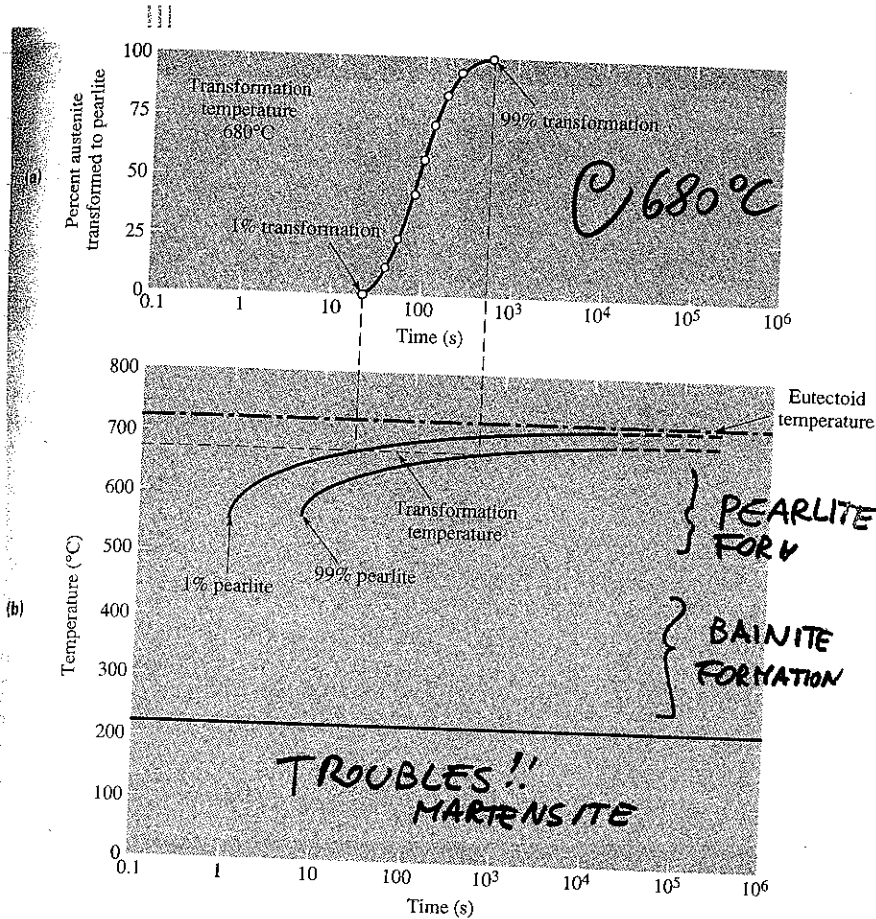


FIGURE 8.3-2 Microstructure of pearlite formed at different isothermal hold temperatures: (a) 655°C, (b) 600°C, (c) 534°C, and (d) 487°C. Notice that the morphologies of the two-phase structure are similar but their spacings decrease with decreasing isothermal hold temperature.

lower T (and bigger ΔT)
and smaller
are the size of
the carbides (Fe_3C)
 \Rightarrow smaller, finer
is PEARLITE

PEARLITE
FORMS FROM
 δ -AUSTENITE
@ EUTECTOID
COMPOSITION
ONLY IF
UNDERCOOLING
IS MODERATE
 $\approx 550^\circ C$



IF UNDERCOOLING IS
BELOW PARATE, EUTECTOID δ -AUSTENITE
550°C BECOMES BAINITE ($\alpha + Fe_3C$)

α -Ferrite LATHS OR PLATES

with CEMENTITE BETWEEN α -Ferrite
PLANES!!!

K19

⇒ THE ISOTHERMAL TRANSFORMATION DIAGRAM LOOKS LIKE

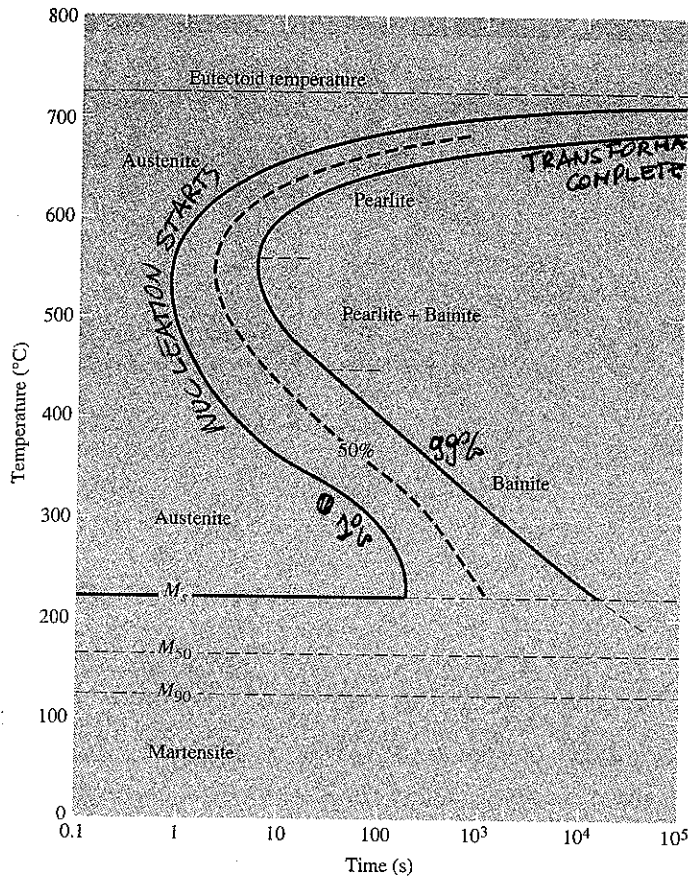


FIGURE 8.3-6 Schematic of an isothermal transformation (IT) diagram for a eutectoid steel.

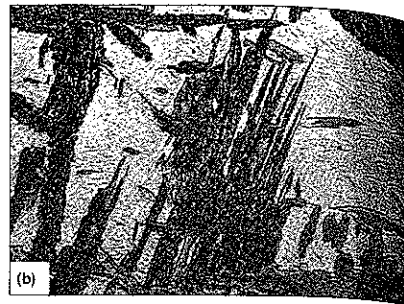
TEMPERING
HEAT UP STEEL
AND COOL IT DOWN
DRAMATICALLY
WITH COLD WATER
OR OIL !!

ASK TO STUDENTS

& BAINITE (PLANES, LATHS) + PLENTY OF DISLOCATION



(a)



(b)

FIGURE 8.3-5 The microstructure of bainite in an alloy steel isothermally transformed at: (a) 495°C, and (b) 410°C. (Source: R. F. Hehman, "Ferrous and Nonferrous Bainite Structures," Metals Handbook, 8th ed., pp. 194-96. ASM International, Materials Park, Ohio. Reprinted with permission of the publisher.)

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K20

WITHOUT

- Pearlite & Bainite Form IF UNDERCOOLING
has T. between 220°C to 727°C (EUTECTOID
TEMP.)

C inside δ -AUSTENITE
the δ structure FCC to DIFFUSES through
to reach ITS FRIENDLY
CARBIDE (Fe_3C)

- IF UNDERCOOLING has T below $\sim 200^{\circ}\text{C}$
 \Rightarrow C has NO TIME TO DIFFUSE (COOLING
& solidification over fast, T \downarrow migration ~ 0)
 \Rightarrow There is a change of phase without diffusion

DIFFUSIONLESS TRANSFORMATION

δ -AUSTENITE (FCC)



MARTENSITE (BCC)

ATOMS FREEZE
in THEIR POSITIONS
BUT LATTICE DEFORMS!!

This is what you make if you put hot-red IRON
in COLD WATER \Rightarrow MARTENSITE

K21

MARTENSITIC TRANSFORMATION

- From γ -AUSTENITE TO MARTENSITE (FCC \rightarrow BCC)
- DIFFUSIONLESS (C does not move)
- DISPLACIVE (change slope of LATTICE, deform without diffusing atoms)
- happens if UNDERCOOLING IS BIG
 $T \leq 220^\circ\text{C} \Rightarrow \Delta T > 500^\circ\text{C}$ (HUGE)

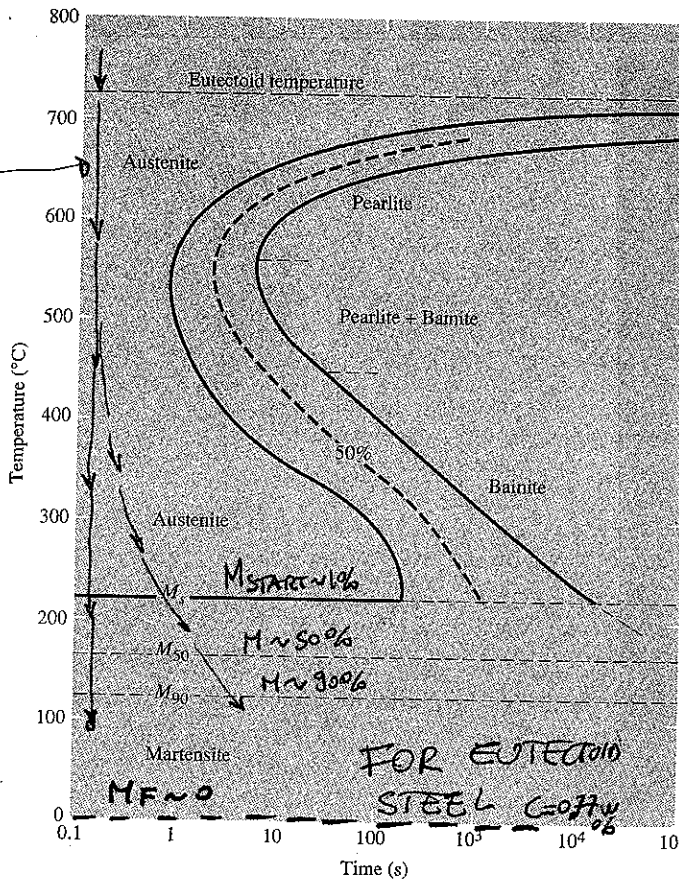


FIGURE 8.3-6 Schematic of an isothermal transformation (IT) diagram for a eutectoid steel.

ARROW OF THE "IRON AGE" WARRIOR

FOR EUTECTOID STEEL

$$M_s \sim 220^\circ\text{C}$$

$$M_{50} \sim 170^\circ\text{C}$$

$$M_{90} \sim 120^\circ\text{C}$$

$$M_f = M_{100} \sim 0^\circ\text{C}$$

DURING "IRON AGE" ONLY HUMAN IN THE NORTH HAD ICE TO COOL STEEL

VIOLENT COOLING TO 100°C MADE BY THE "IRON AGE" METALLURGIST WARRIOR

K22

IF WE CHANGE C% we can get MARTENSITE ALL OVER THE PLACE

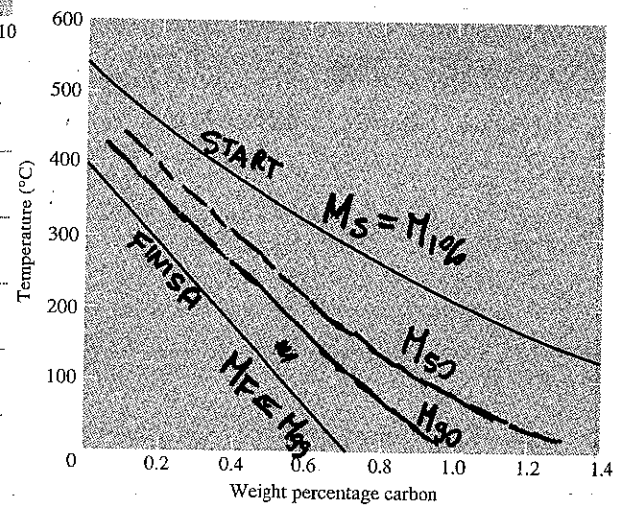


FIGURE 8.3-7 The effect of carbon on the M_s and M_f temperatures for a plain carbon steel.

MARTENSITE IS DIFFUSIONLESS

⇒ SAME CONCENTRATION OF STARTING δ -AUSTENITE (HOT IRON)

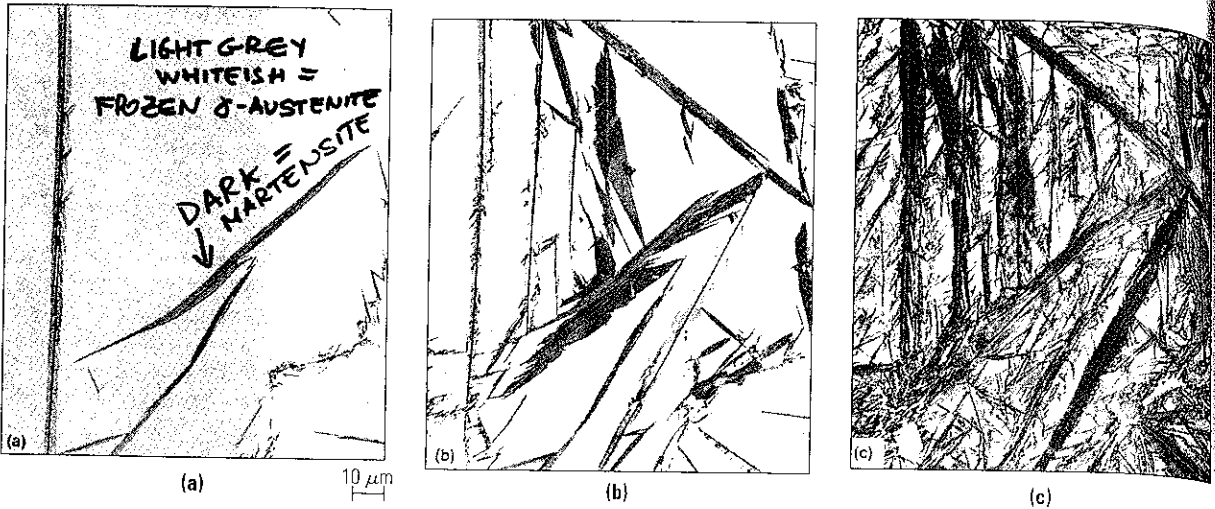


FIGURE 8.3-8 Progress of athermal martensitic transformation in an Fe-1.8 wt. % C alloy after cooling to (a) 24°C, (b) -60°C, and (c) -100°C. (Source: G. Kraus and A. R. Marder, The Morphology of Martensite in Iron Alloys. Metallurgical Transactions, Vol. 2, 1971, pp. 2343-57. Reprinted by permission of the publisher.)

NEEDLE LIKE → DIRECTION OF M are same from δ

HOW TO GO FROM δ -AUST TO MARTENSITE?

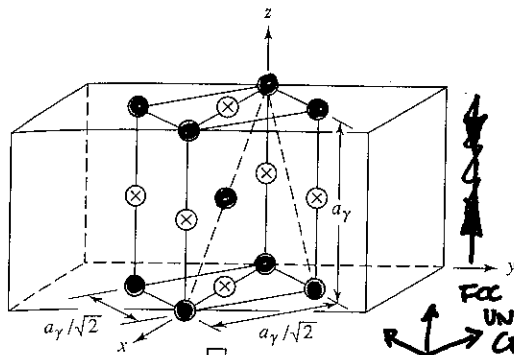
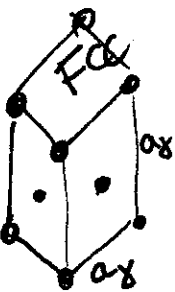
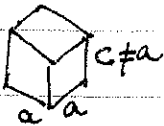
γ -AUSTENITE \Rightarrow MARTENSITE

Only LATTICE ^{FCC} DISTORTIONS!

Body Centered

1. An FCC to BCT transformation.
2. Distortion of the BCT lattice parameters to match those of the martensite. This is called the Bain distortion after the person who first proposed it.
3. Rotation to produce the appropriate habit planes and directions between the martensite and the parent austenite.

BCT = Body Center Tetragonal

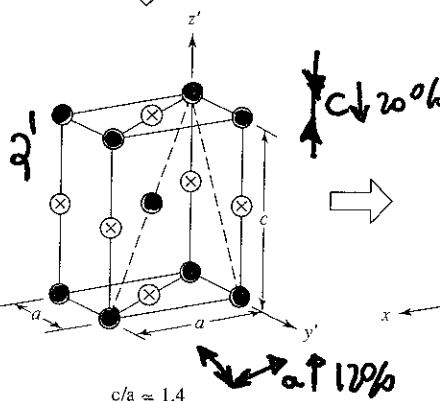


γ -AUSTENITE

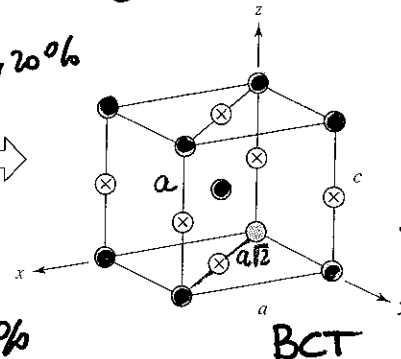
● = Fe in FCC

⊗ = Possible C OR nothing to FILL 'UP CONCENTRATION

FCC UNIT CELL OF γ



BAIN



BCT

$a_y \Rightarrow c \downarrow$
 \Rightarrow CONTRACT 20%

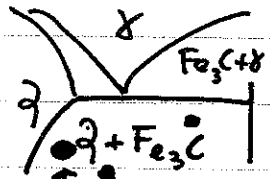
$a_x = a \uparrow 12\%$

$\gamma \rightarrow \alpha$
Volume increase!!

BCT

FIGURE 8.3-9 The relationship between the FCC γ phase and the BCT martensite phase. The possible sites for carbon atoms are located at the x's, and the iron atoms are located at the open circles. To obtain the α' phase from the γ unit cell, the c axis must contract about 20% and the a axis must expand about 12%.

BUT @ low Temp

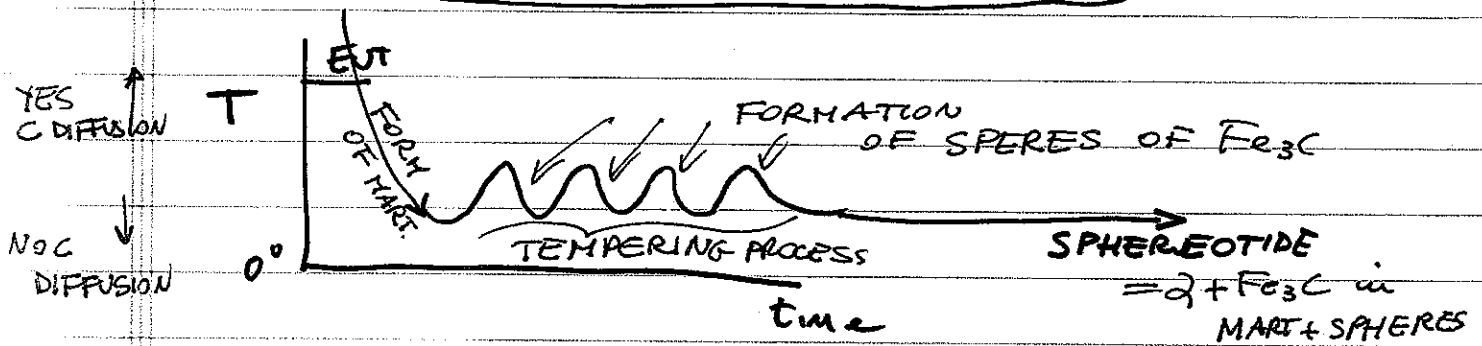


ALL THESE MARTENSITES ARE UNSTABLE TOWARD DECOMPOSITION OF $\alpha + Fe_3C$

K24

TEMPERING

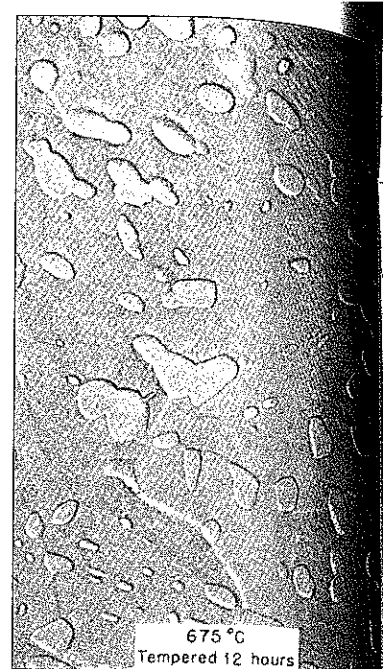
IF I KEEP MARTENSITE COLD WILL REMAIN MART.
 BUT IF I HEAT IT UP A LITTLE IT WILL
 ALLOW C TO DIFFUSE & PRECIPITATE ON
 ~ SPHERES OF Fe_3C (CEMENTITE) (AND BCC becomes BCC)
 SUCH SPHERES ARE CALLED SPHERO
 SPHEROIDITE = MARTENSITE + SPHERES OF CEMENTITE



(a)



(b)



(c)

FIGURE 8.3-10 Microstructure of a tempered martensite (spheroidite) in a steel with 0.7 wt. % C. (Source: Transactions of the Metallurgical Society (212), 1958, a publication of the Minerals, Metals & Materials Society, Warrendale, PA. Reprinted by permission of the publisher.)

K25

EXAMPLE

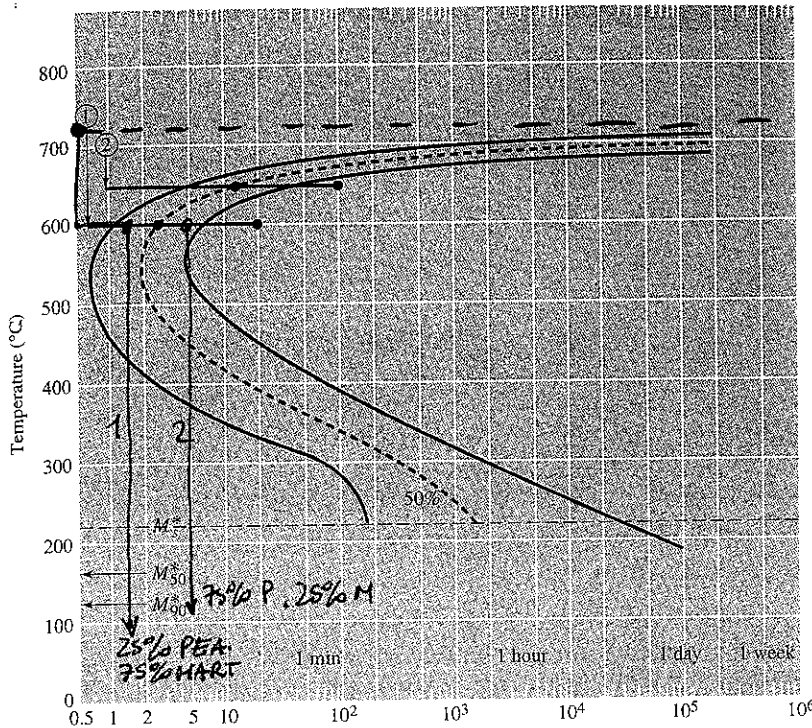


FIGURE 8.3-11 Examples of time-temperature paths on the isothermal transformation diagram of a eutectoid steel.

- 1) @ $t=0$ $T=727^{\circ}\text{C}$
 - COOL FAST (QUENCH) to 600°C
 - WAIT 1.5 secs
 - COOL FAST TO 100°C \Rightarrow $\approx 25\%$ PEARLITE
 $\approx 75\%$ MARTENSITE BCT.
- 2) • COOL FAST to 600°C , WAIT 5 secs, COOL FAST TO 100°C
 - \Rightarrow 75% PEARL
25% MART
- 3) • COOL FAST TO 650°C , WAIT 2 MINS, COOL TO 100°C
 - \Rightarrow 100% PEARL
- 4) COOL SUPER SLOWLY (to ALLOW TH. EQ) to 500°C , WAIT 2 YEARS, COOL TO 100°C
 - \Rightarrow ALL α + Fe_3C
- 5) DO LIKE 1) BUT THEN HEAT UP TO 200°C , WAIT 10 SECS & QUENCH, REPEAT MANY TIMES
 - \Rightarrow 25% P, MUCH Fe_3C , SOME α , SOME BCC M,