

**2017 Fall**

# **“Phase Transformation *in* Materials”**

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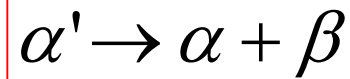
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**Office hours: by an appointment**

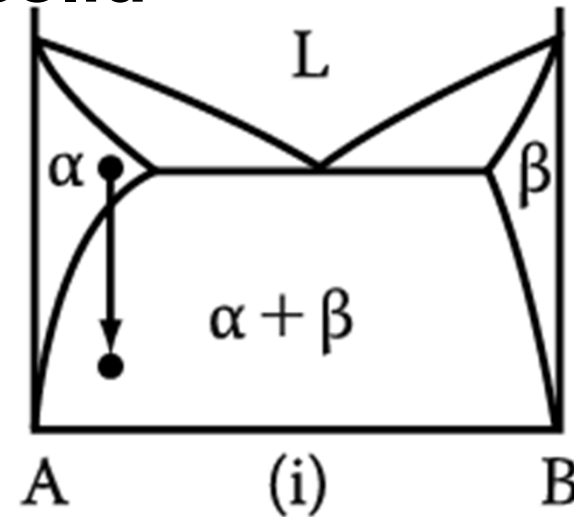
# 5. Diffusion Transformations in solid

: diffusional nucleation & growth

## (a) Precipitation



Metastable supersaturated  
Solid solution



### Homogeneous Nucleation

➡ Effect of misfit strain energy

$$\Delta G = -V\Delta G_V + A\gamma + V\Delta G_S$$

$$r^* = \frac{2\gamma}{(\Delta G_V - \Delta G_S)} \quad \Delta G^* = \frac{16\pi\gamma^3}{3(\Delta G_V - \Delta G_S)^2}$$

$$N_{\text{hom}} = \omega C_0 \exp\left(-\frac{\Delta G_m}{kT}\right) \exp\left(-\frac{\Delta G^*}{kT}\right)$$

### Heterogeneous Nucleation

➡ suitable nucleation sites ~ nonequilibrium defects  
(creation of nucleus ~ destruction of a defect (-ΔG<sub>d</sub>))

$$\Delta G_{\text{het}} = -V(\Delta G_V - \Delta G_S) + A\gamma - \Delta G_d$$

$$\frac{\Delta G^*_{\text{het}}}{\Delta G^*_{\text{hom}}} = \frac{V^*_{\text{het}}}{V^*_{\text{hom}}} = S(\theta)$$

$$\frac{N_{\text{het}}}{N_{\text{hom}}} = \frac{C_1}{C_0} \exp\left(\frac{\Delta G^*_{\text{hom}} - \Delta G^*_{\text{het}}}{kT}\right)$$

## 5.5.3. Quenched-in Vacancies

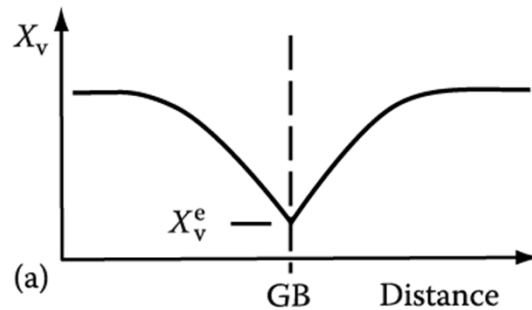
If  $X_v < X_v^c$  critical vacancy supersaturation,  
Precipitate nucleation  $X \rightarrow$  formation of PFZ

In the vicinity of grain boundaries on subsequent aging,

### a) Precipitate-Free Zone (PFZ) due to Vacancy Diffusion during quenching

Solute concentration within the zone  $\sim$  largely unchanged, but no precipitate at GB  
 $\therefore$  a critical vacancy supersaturation must be exceeded for nucleation to occur.

- a) Excess  $\text{V} \rightarrow \text{D}$  nucleation and moving  $\uparrow$ :  
Heterogeneous nucleation sites  $\uparrow$
- b) Excess  $\text{V} \rightarrow$  atomic mobility  $\uparrow$  at ageing temp:  
speeds up the process of nucleation and growth
- ex) rapid formation of GP zones at the relatively low  
aging temperature. (possible to RT aging in Al-Cu alloy)



Similar PFZs can also form at inclusions and dislocations.

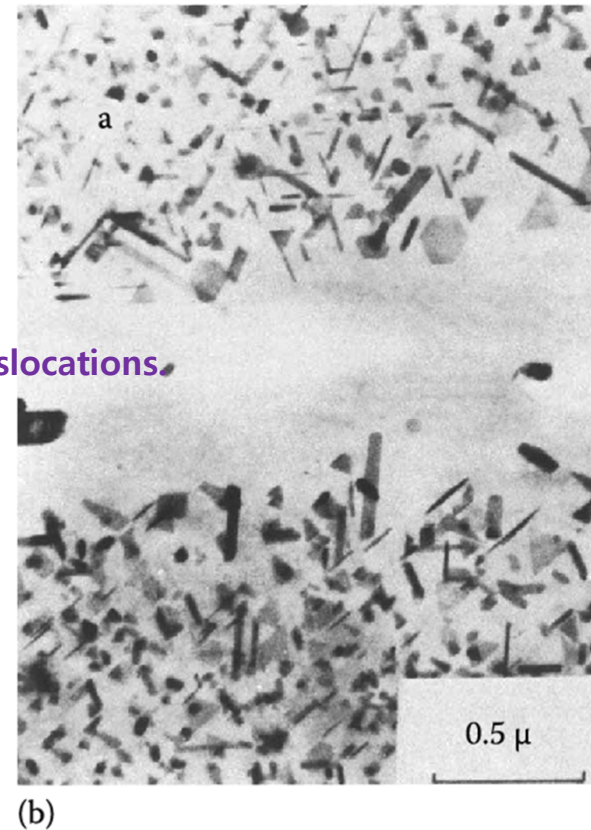
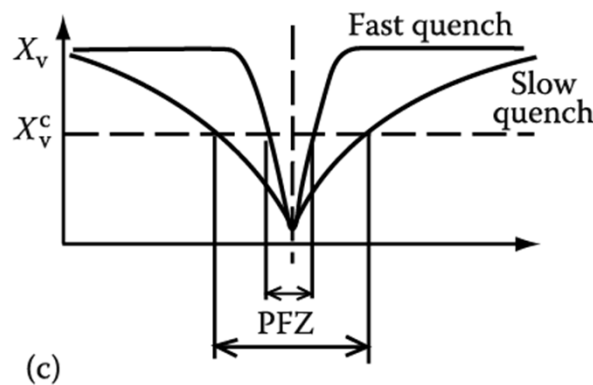


Fig. 5.35 A PFZ due to vacancy diffusion to a grain boundary during quenching.

(a) Vacancy concentration profile. (b) A PFZ in an Al-Ge alloy (x 20,000)

(c) Dependence of PFZ width on critical vacancy concentration  $X_v^c$  and rate of quenching.

b) Another cause of PFZs can be **the nucleation and growth of GB precipitates** during cooling from the solution treatment temperature.

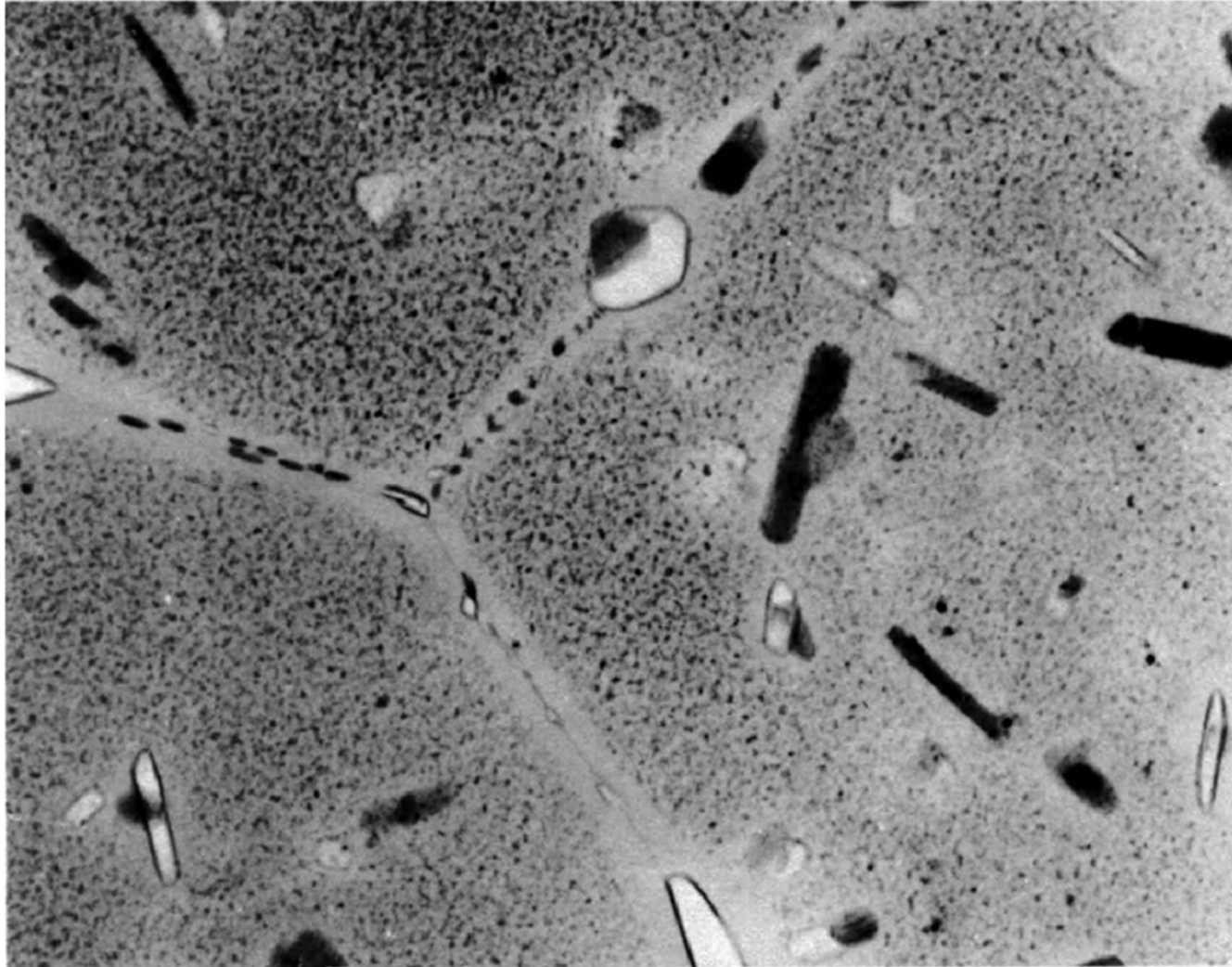


Fig. 5.36 PFZs around grain boundaries in a high-strength commercial Al-Zn-Mg-Cu alloy. Precipitates on grain boundaries have extracted solute from surrounding matrix. (x 59,200)

## 5.5.4. Age Hardening

### Transition phases

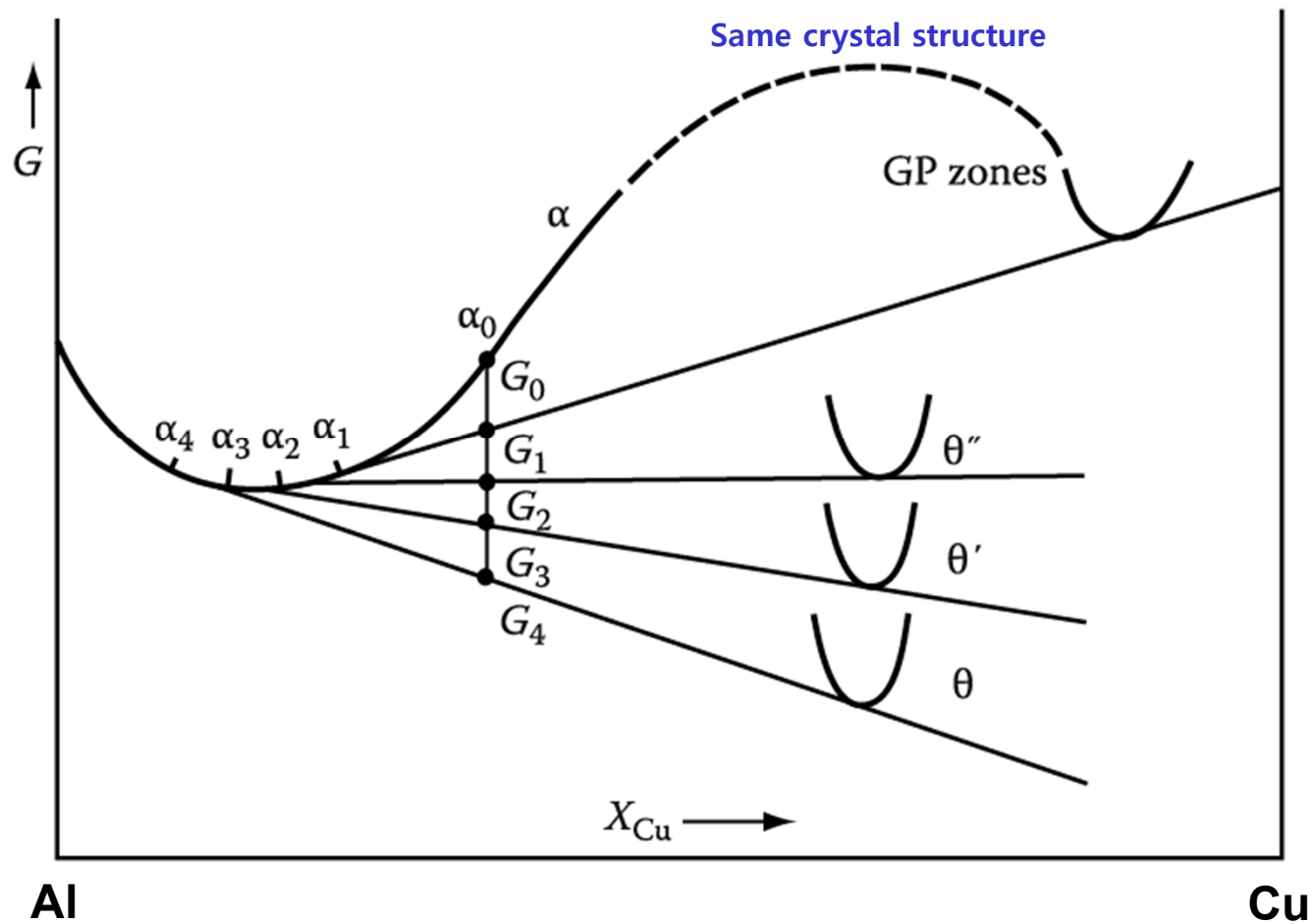
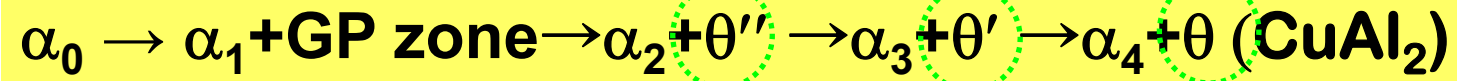


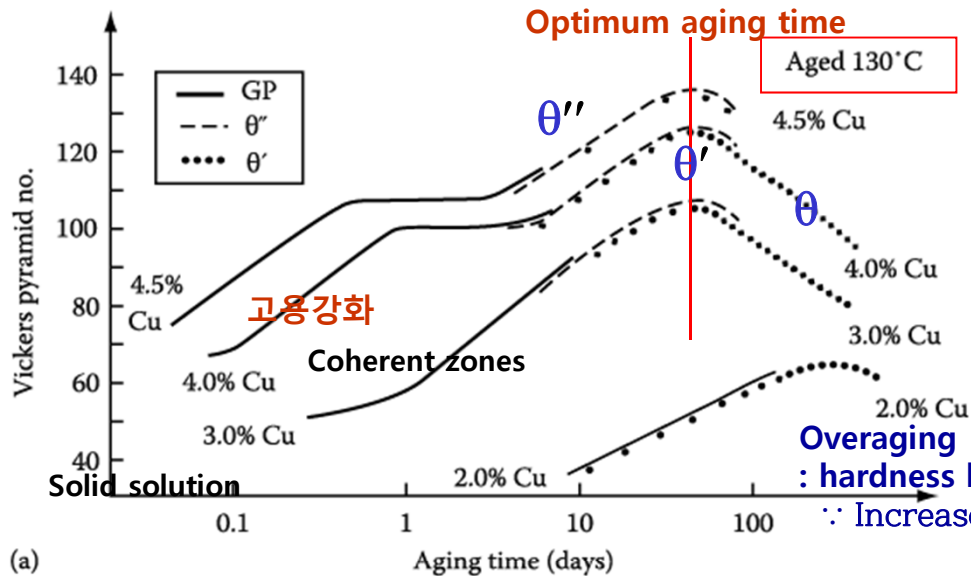
Fig. 5.27 A schematic molar free energy diagram for the Al-Cu system.

## 5.5.4. Age Hardening

Transition phase precipitation → great improvement in the mechanical properties

Coherent precipitates → highly strained matrix → the main resistance to the  $\odot$  movement: solid solution hardening

### Hardness vs. Time by Ageing



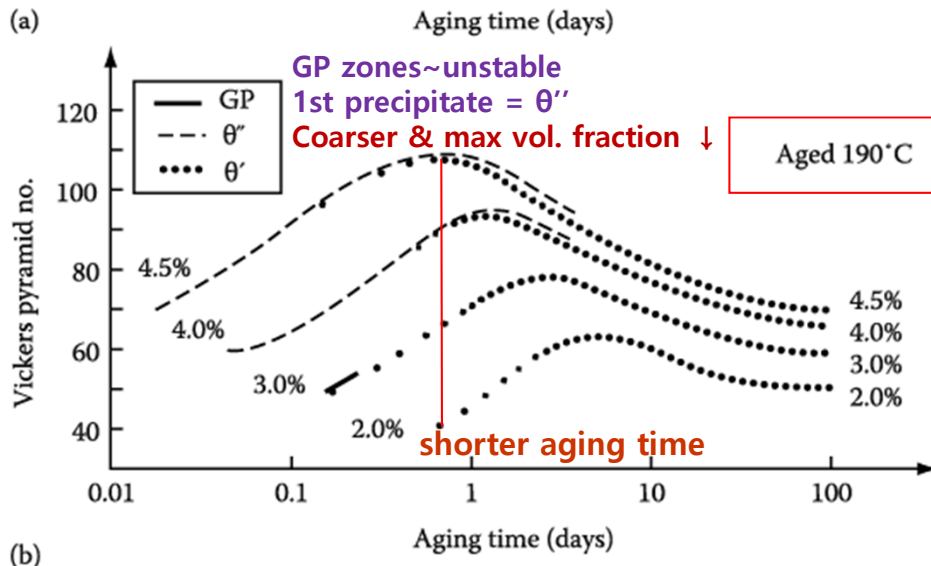
Maximum hardness ~ largest fraction of  $\theta''$  (coherent precipitates)

Ageing at 130°C produces higher maximum hardness than ageing at 190°C.

At 130°C, however, it takes **too a long time** (several tens of days).

Overaging : hardness begins to decrease

∴ Increases the distance btw precipitates making  $\odot$  bowing easier



How can you get the high hardness for the relatively short ageing time (up to 24h)?

**Double ageing treatment**  
first below the GP zone solvus → fine dispersion of GP zones then ageing at higher T.

: Engineering alloys are not heat treated for max. strength alone. → to optimize other properties **best heat treatment in practice**

Fig. 5. 37 Hardness vs. time for various Al-Cu alloys at (a) 130 °C (b) 190 °C

## 5.5.5 Spinodal Decomposition

### Spinodal mode of transformation has no barrier to nucleation

: describing the transformation of a system of two or more components in a metastable phase into two stable phases

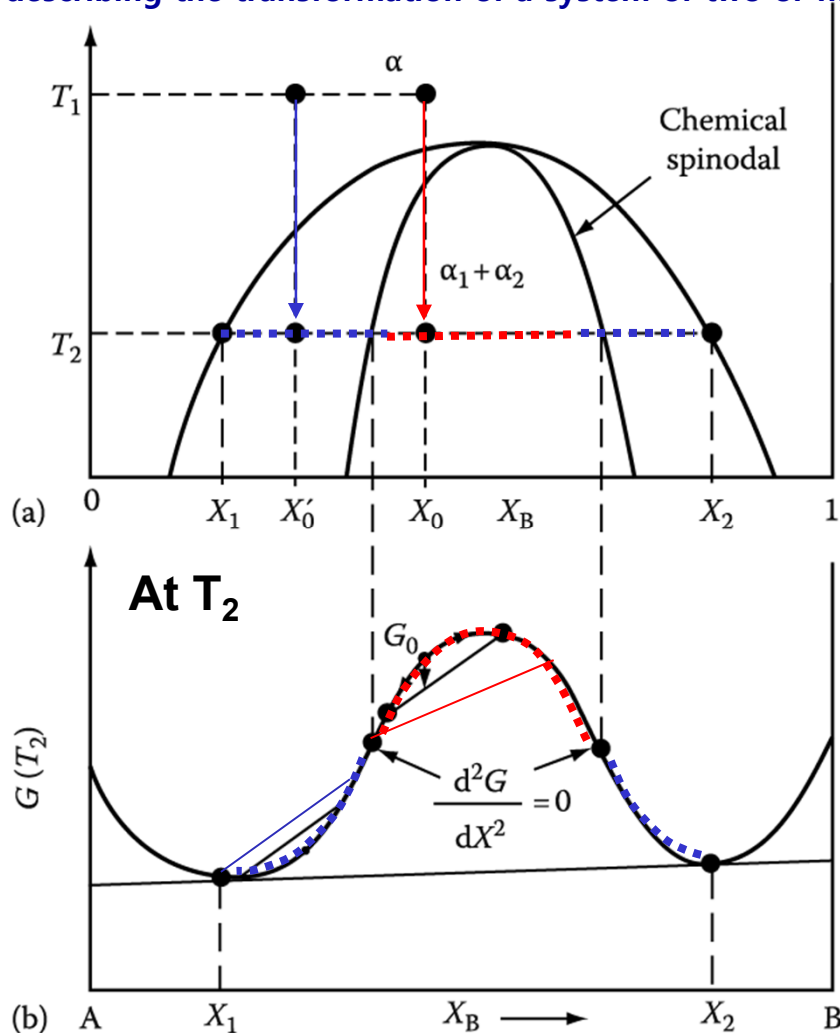


Fig. 5.38 Alloys between the spinodal points are unstable and can decompose into two coherent phases  $\alpha_1$  and  $\alpha_2$  without overcoming an activation energy barrier. Alloys between the coherent miscibility gaps and the spinodal are metastable and can decompose only after nucleation of the other phase.

How does it differ between **inside** and **outside the inflection point** of Gibbs free energy curve?

1) **Within the spinodal**  $\frac{d^2G}{dX^2} < 0$

: phase separation by small fluctuations in composition/  
"up-hill diffusion"

2) If the alloy lies **outside the spinodal**, small variation in composition leads to an increase in free energy and the alloy is therefore **metastable**.

The free energy can only be decreased if nuclei are formed with a composition very different from the matrix.

→ **nucleation and growth**  
: "down-hill diffusion"

**a) Composition fluctuations within the spinodal**

**b) Normal down-hill diffusion outside the spinodal**

**up-hill diffusion**

**down-hill diffusion**

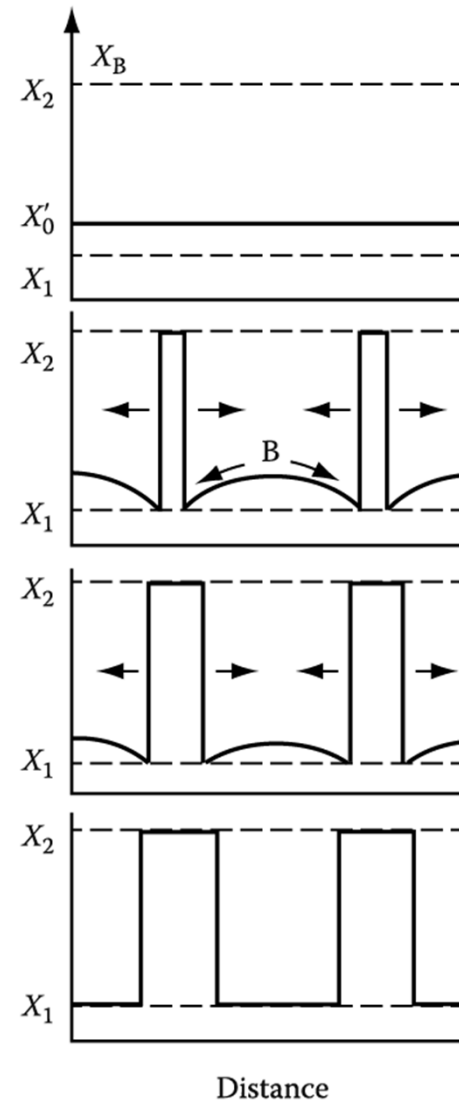
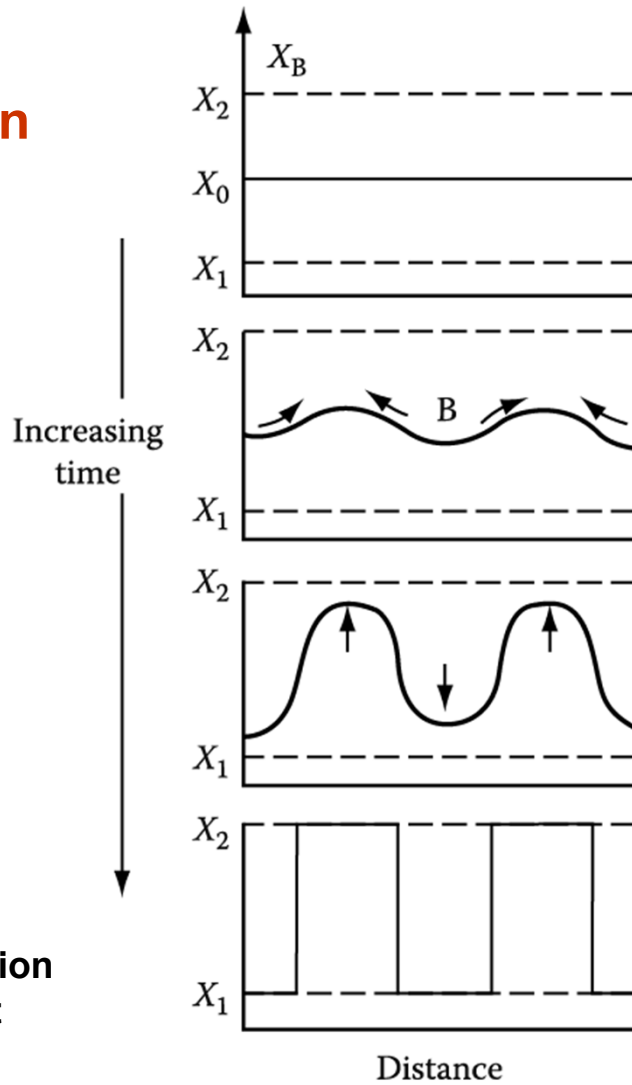


Fig. 5.39 & 5.40 schematic composition profiles at increasing times in (a) an alloy quenched into the spinodal region ( $X_0$  in Figure 5.38) and (b) an alloy outside the spinodal points ( $X'_0$  in Figure 5.38)



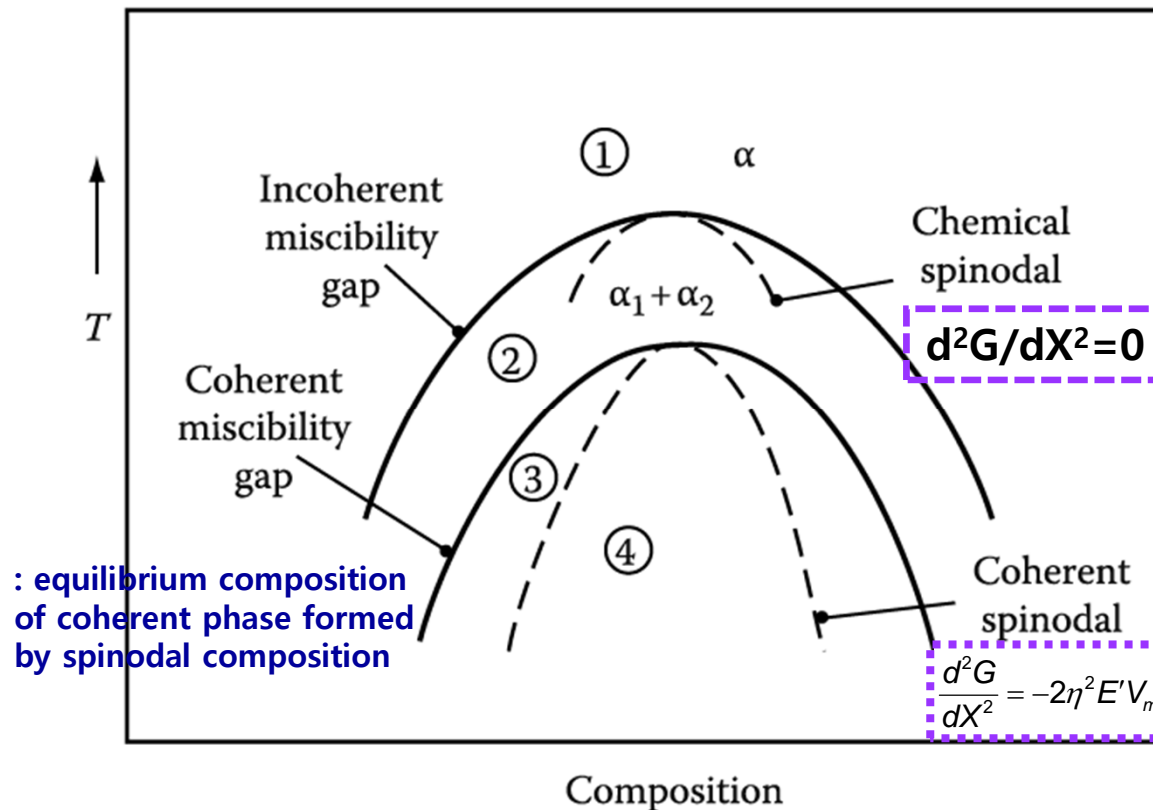
This figure include the lines defining the equilibrium compositions of the coherent/ incoherent phases that result from spinodal decomposition.

**\* Incoherent(or equilibrium) miscibility gap:  $\Delta H > 0$**

The miscibility gap the normally appears on an equilibrium phase is the incoherent (or equilibrium) miscibility gap. → equilibrium compositions of incoherent phases without strain fields.

a) chemical spinodal:  $d^2G/dX^2=0$ \_no practical importance X

b) Area ② ,  $\Delta G_V - \Delta G_S < 0$  → only incoherent strain-free nuclei can form.



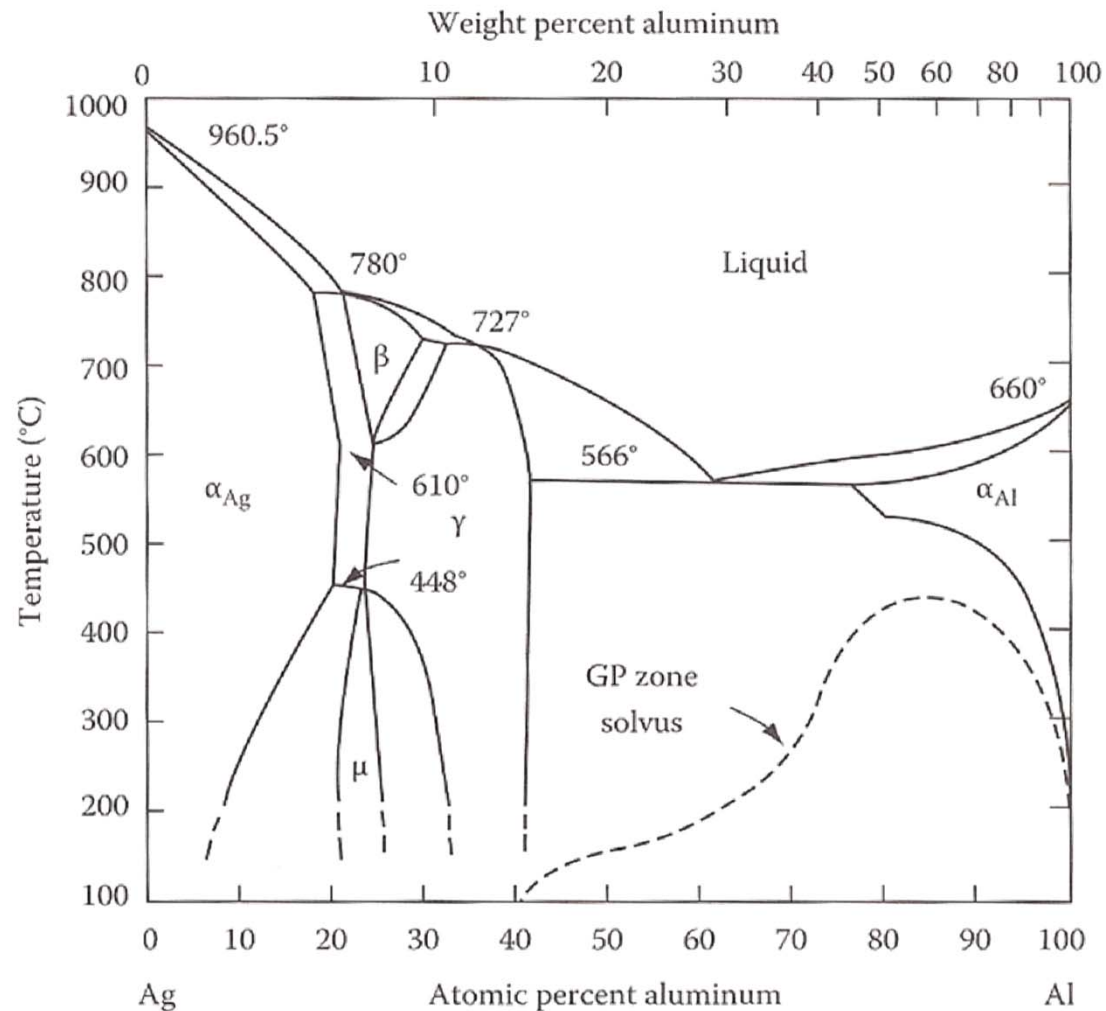
**Figure 5.41 Schematic phase diagram for a clustering system.**

Region 1: homogeneous  $\alpha$  stable. Region 2: homogeneous  $\alpha$  metastable, only incoherent phases can nucleate. Region 3: homogeneous  $\alpha$  metastable, coherent phase can nucleate. Region 4: homogeneous  $\alpha$  unstable, no nucleation barrier, spinodal decomposition occurs.

Spinodal decomposition is not only limited to systems containing a stable miscibility gap

All systems in which GP zones form, for example, containing a metastable coherent miscibility gap, i.e., the GP zone solvus.

→ at high supersaturation, GP zone can form by the spinodal mechanism.



**Figure 5.34**

Al-Ag phase diagram showing metastable two-phase field corresponding to GP zones.

Microstructure of a two phase alloy is always unstable if the total interfacial free E is not a minimum. →

## 5.5.6. Particle Coarsening (smaller total interfacial area → loss of strength or disappearance of GB pinning effect → particular concern in the design of materials for high temp. applications)

### Two Adjacent Spherical Precipitates with Different Diameters

(Gibbs-Thomson effect: radius of curvature ↓ →  $X_B$  ↑)

Assumption: volume diffusion is the rate controlling factor

$$(\bar{r})^3 - r_0^3 = kt$$

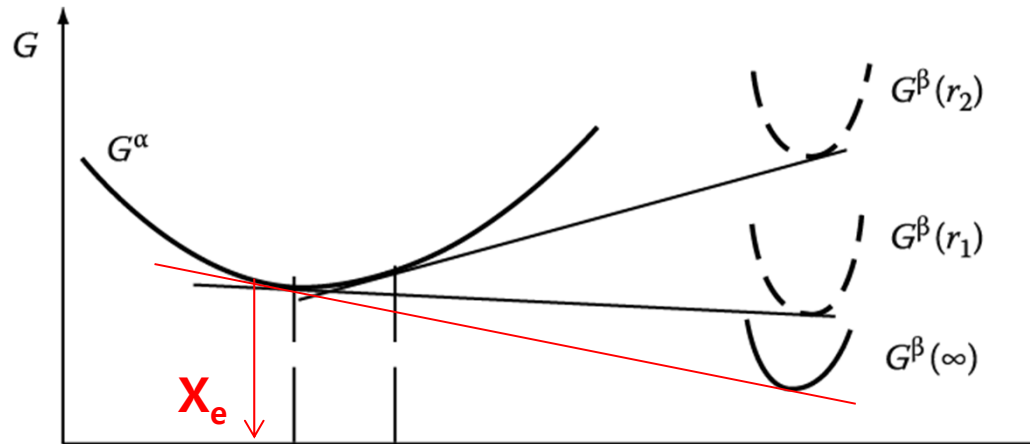
Average radius

where  $k \propto D_\gamma X_e$

( $X_e$ : Equil. solubility of very large particles)

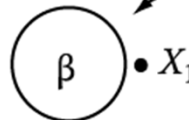
Coarsening rate

$$\frac{d\bar{r}}{dt} \propto \frac{k}{\bar{r}^2}$$

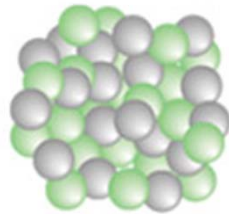


(b)  $X_1 < X_2$   $X_B$  →  
A-rich B-rich

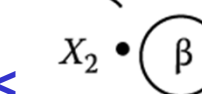
Lower density of larger particles



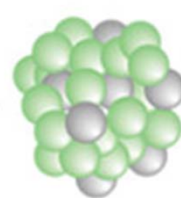
A-rich  $r_1$



B



B-rich  $r_2$

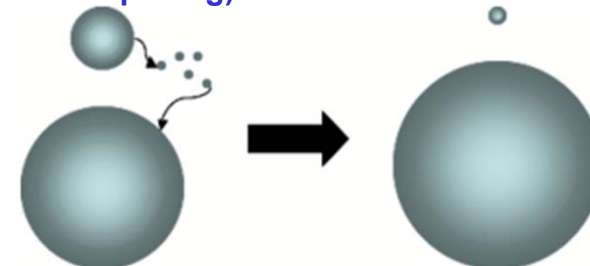


High density of small precipitates

( $D$  and  $X_e \sim \exp(-Q/RT)$ )  
 $\bar{r}$  Rapidly increase with Increasing temp.

⇒ CR ↑

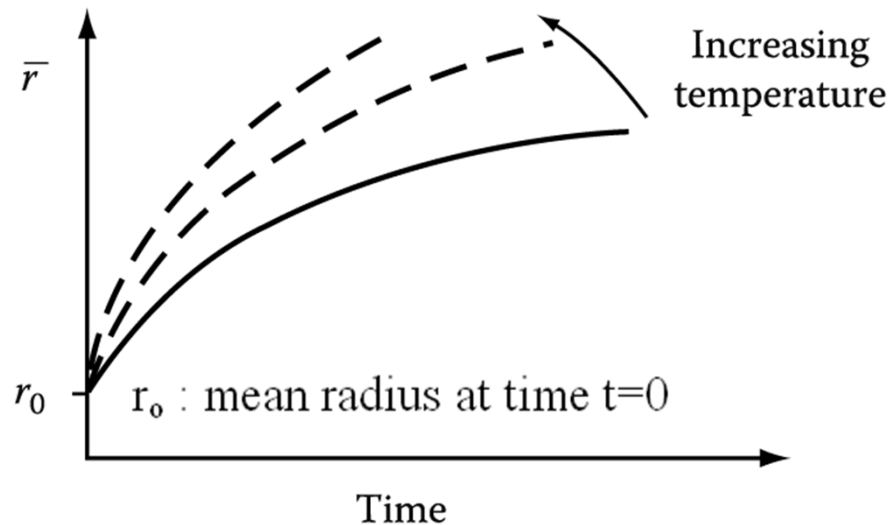
(Ostwald Ripening)



: Concentration gradient in matrix → diffusion → small particle\_shrink/ large particle\_grow

## 5.5.6. Particle Coarsening

### The Rate of Coarsening with Increasing Time and Temp.



$\bar{r}$  ~ Particular concern in the design of materials for high temperature applications

Undesirable degradation of properties:  
less strength/ disappearance of GB pinning effects

How can you design an alloy with high strength at high T?

→ fine precipitate dispersion

hint)  $\frac{d\bar{r}}{dt} \propto \frac{k}{\bar{r}^2}$      $k \propto D\gamma X_e$

### 1) low $\gamma$

heat-resistant Nimonic alloys

based on Ni-rich Ni-Cr → ordered fcc

$\text{Ni}_3(\text{Ti,Al})$  in Ni-rich matrix → high strength

Ni/ $\gamma'$  interface ~ “fully coherent” (10 ~ 30 mJ m<sup>-2</sup>)

Maintain a fine structure at high temperature

→ improve creep-rupture life

### 2) low $X_e$ (Oxide ~ very insoluble in metals)

: fine oxide dispersion in a metal matrix

Ex) dispersed fine  $\text{ThO}_2$  (thoria) in W and Ni

→ strengthened for high temperature

### 3) low $D$

Cementite dispersions in tempered steel

→ high  $D$  of carbon → very quickly coarsening

a. substitutional alloying element

→ segregates to carbide → slow coarsening

b. strong carbide-forming elements

→ more stable carbides → lower  $X_e$

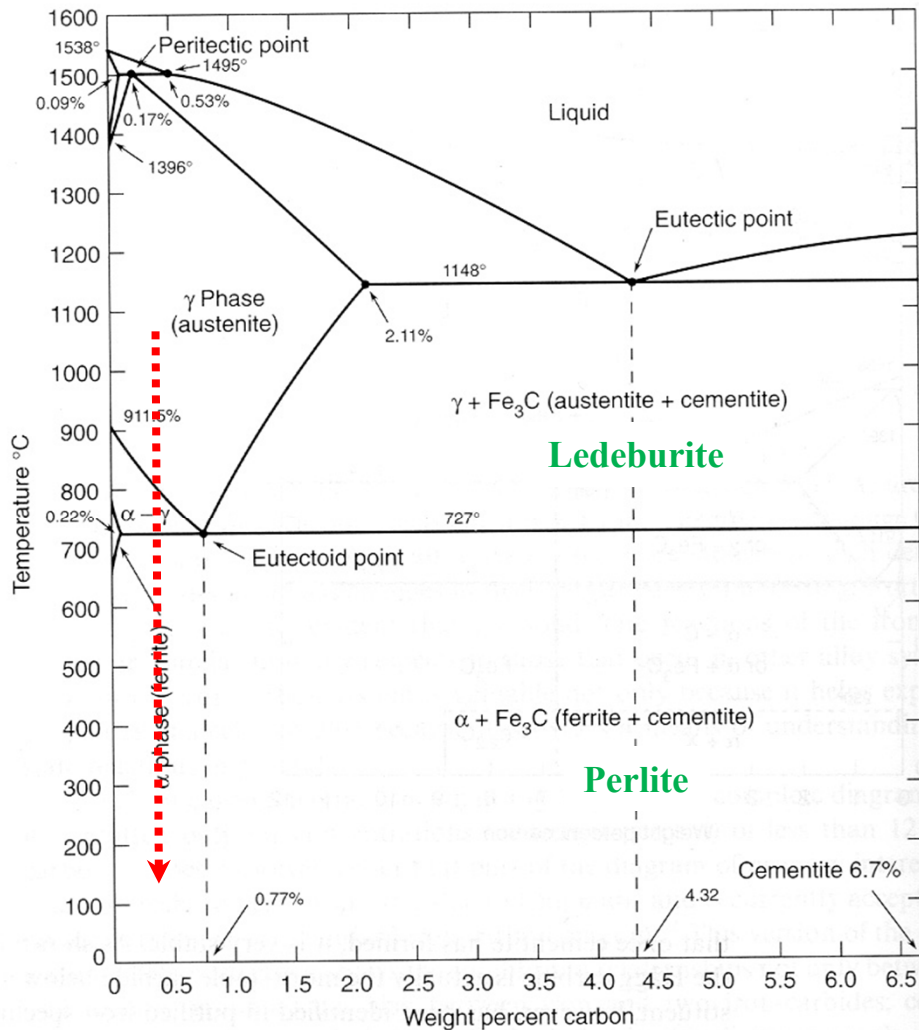
## Q7: Precipitation of Ferrite from Austenite ( $\gamma \rightarrow \alpha$ )

### 3) Precipitation of equilibrium phase by diffusional transformation

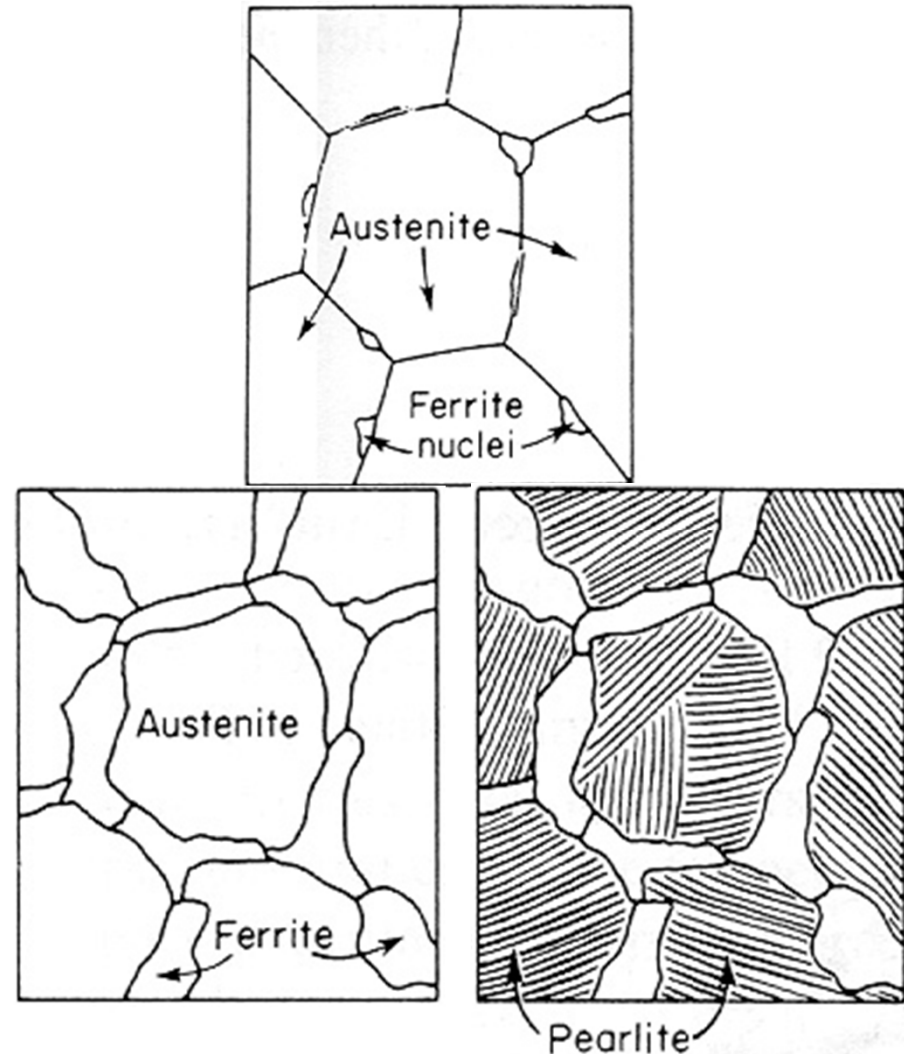
## 5.6. The Precipitation of Ferrite from Austenite ( $\gamma \rightarrow \alpha$ )

(Most important nucleation site: Grain boundary and the surface of inclusions)

### The Iron-Carbon Phase Diagram

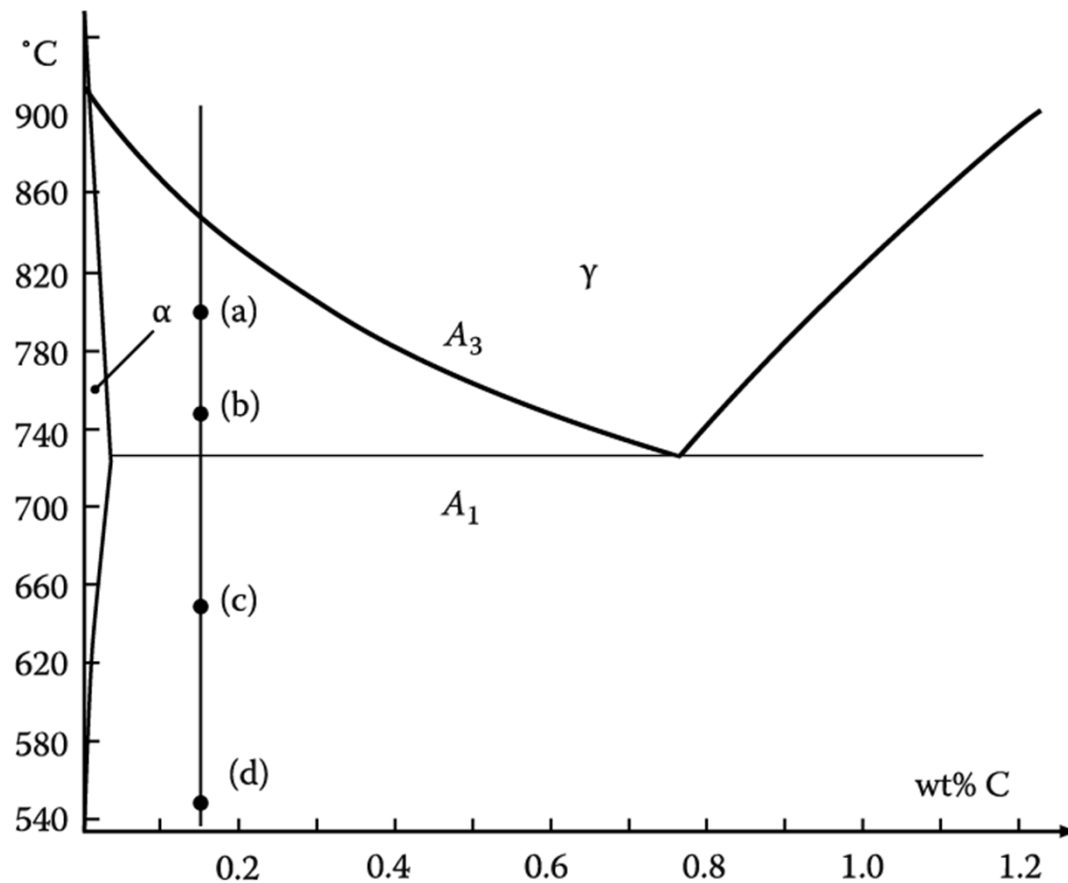


Microstructure (0.4 wt%C) evolved by slow cooling (air, furnace) ?



## 5.6. The Precipitation of Ferrite from Austenite

### Diffusional Transformation of Austenite into Ferrite



**Fe-0.15 wt%C**

After being austenitized, held at

(a) 800°C for 150 s

(b) 750°C for 40 s

(c) 650°C for 9 s

(d) 550°C for 2 s and  
then quenched to room T.

**What would be the  
microstructures?**

Figure 5.45 Holding temperature for steel in Figure. 5.46

# Microstructures of an austenitized Fe-0.15%C alloy (x 100 except (d, x300))

White:  $\alpha$  ferrite/ Gray: M formed from untransformed  $\gamma$ / fine constituent: a mixture of ferrite and carbide

Primary ferrite allotriomorphs with a few plates  $\Rightarrow$  Many more plates, mostly growing from GBs/ inside  $\alpha$  grain

Smaller  $\Delta T$

(a)  
800°C  
for 150 s

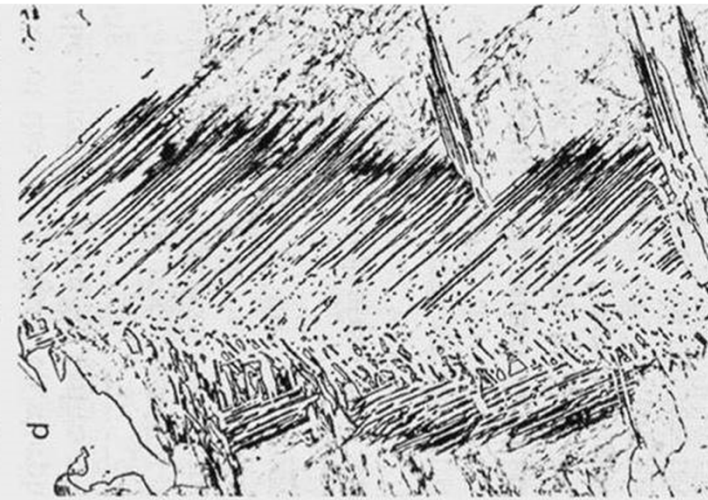
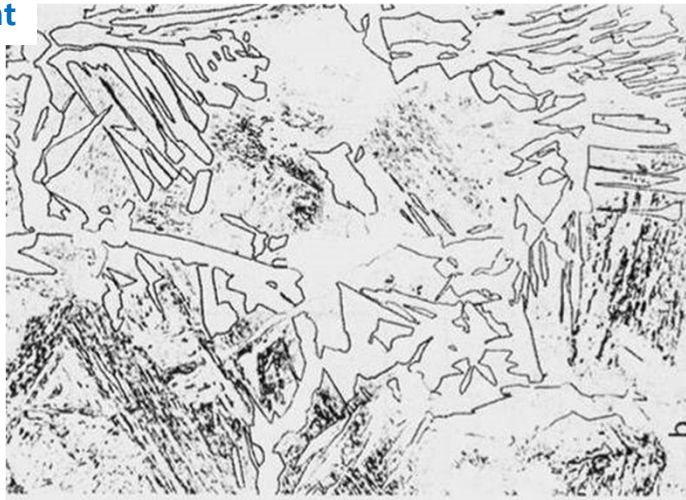


(입계타형) G.B. allotriomorphs  
"blocky" manner/  
Smoothly curved  
& faceted  $\alpha/\gamma$   
Interface are present



(c)  
650°C  
for 9 s

(b)  
750°C  
for 40 s



(d)  
550°C  
for 2 s

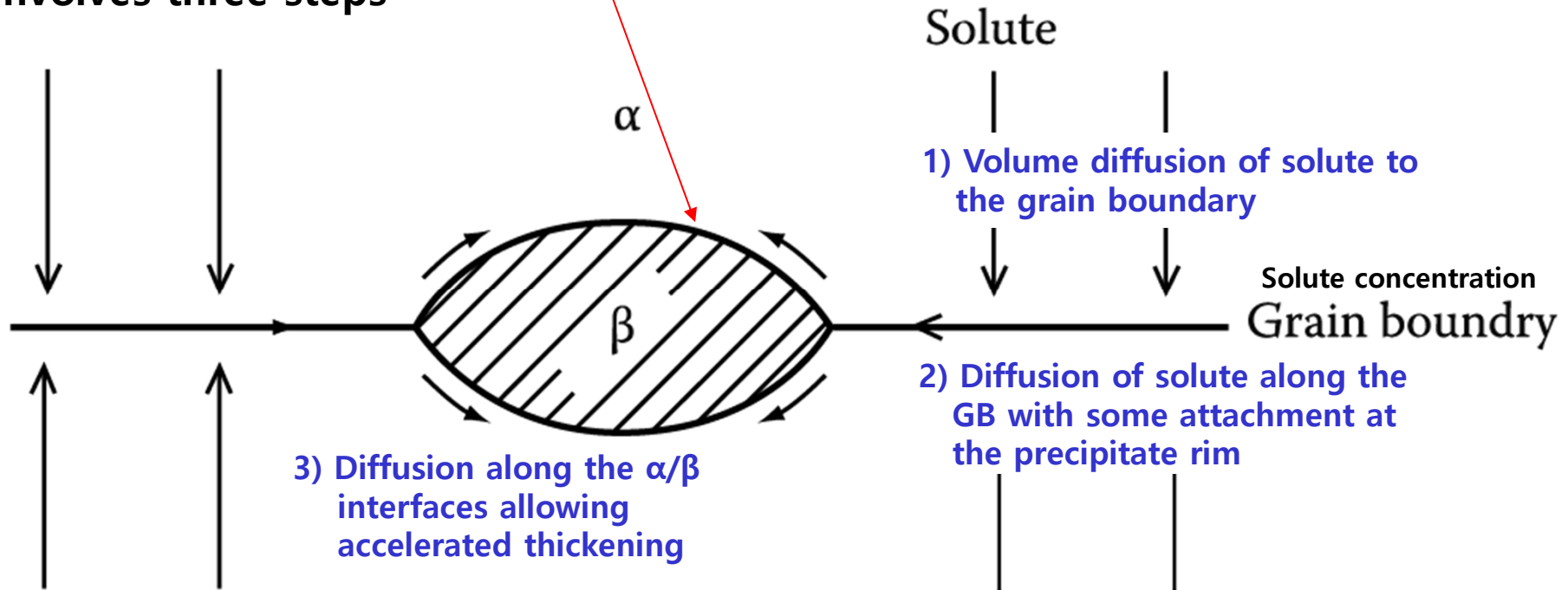
larger  $\Delta T$

Widmanstätten ferrite side-plates (b), (c), (d) – Finer & faceted coherent interface with increasing "undercooling"



## \* Grain boundary allotriomorph 입계타형

Grain boundary precipitation involves three steps **→ Faster than allowed by volume diffusion**



치환형 확산이 일어나는 경우 매우 중요/ 침입형 고용체에서는 체적 확산 속도가 크기 때문에 입계나 전위를 통한 단거리 확산은 상대적으로 중요하지 않음.

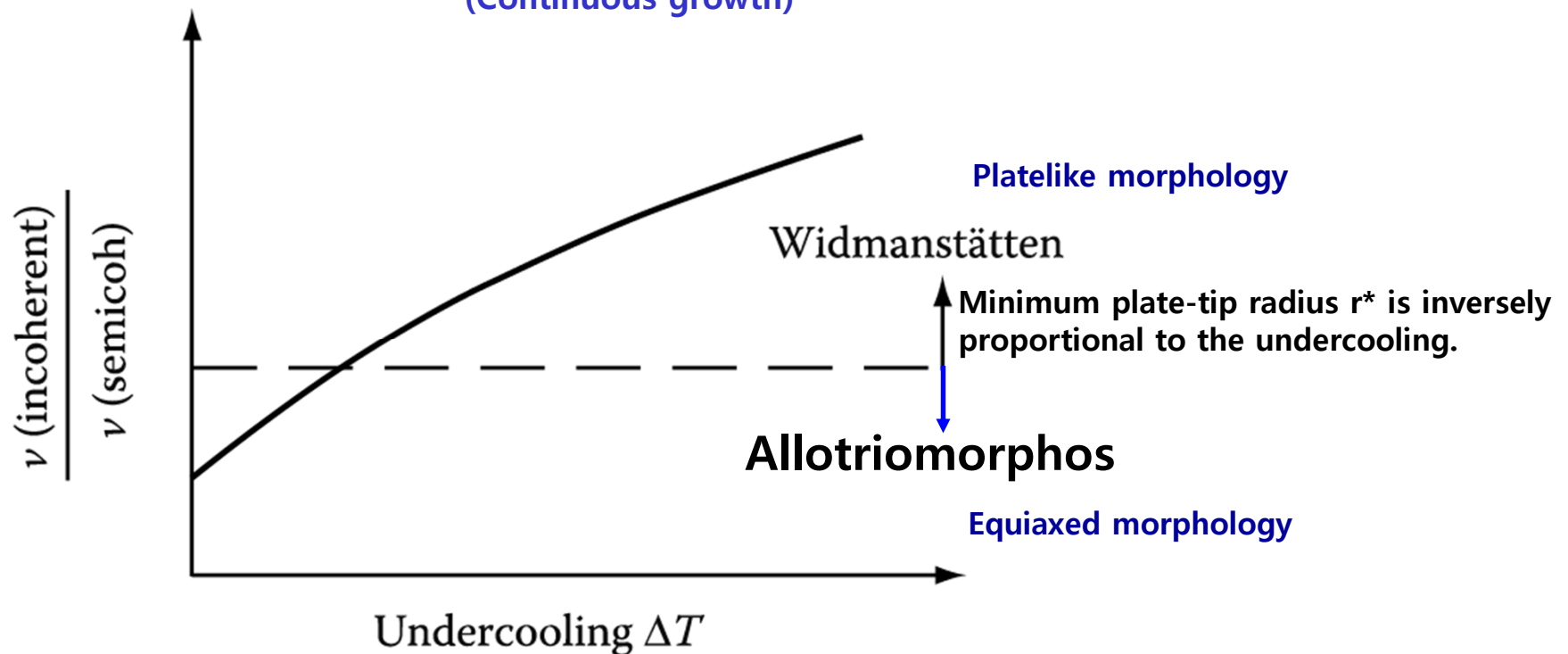
**Fig. 5.18 Grain-boundary diffusion can lead to rapid lengthening and thickening of grain boundary precipitates, especially by substitutional diffusion.**

**The reason** for the transition from grain boundary allotriomorphs to Widmanstätten side-plates with increasing undercooling is not fully understood.

→ possible answer: **Relative Velocity of Incoherent & Semicoherent Interfaces vary with undercooling**

a) At small undercoolings, both semi-coherent and incoherent interfaces ~similar rates

b) At large undercoolings, only incoherent interfaces ~full use of increased driving force  
(Continuous growth)



\* **Intragranular ferrite in large-grained specimen**

: ferrite can also precipitate within the austenite grains (Fig. in page 17)

suitable heterogeneous nucleation site ~ inclusions and dislocations

generally **equiaxed** at low undercooling ↔ **more platelike** at higher undercoolings

# 5.6. The Precipitation of Ferrite from Austenite

Typical TTT curve for  $\gamma \rightarrow \alpha$  transformation  $\rightarrow f(t, T)$

J-M-A Eq.

$$f = 1 - \exp(-kt^n)$$

$k$ : sensitive with  $T$   $f(I, v) \sim \frac{\pi}{3} I v^3$   
 $n$ : 1 ~ 4 (depend on nucleation mechanism)

- a) Time for a given percentage transformation will decrease as the constant  $k$  increase
  - b)  $k$  increases with increases in  $\Delta T$  or total # of nucleation sites
- $\rightarrow$  Thus, decreasing the austenite grain size has the effect of shifting the C curve to shorter transformation times.

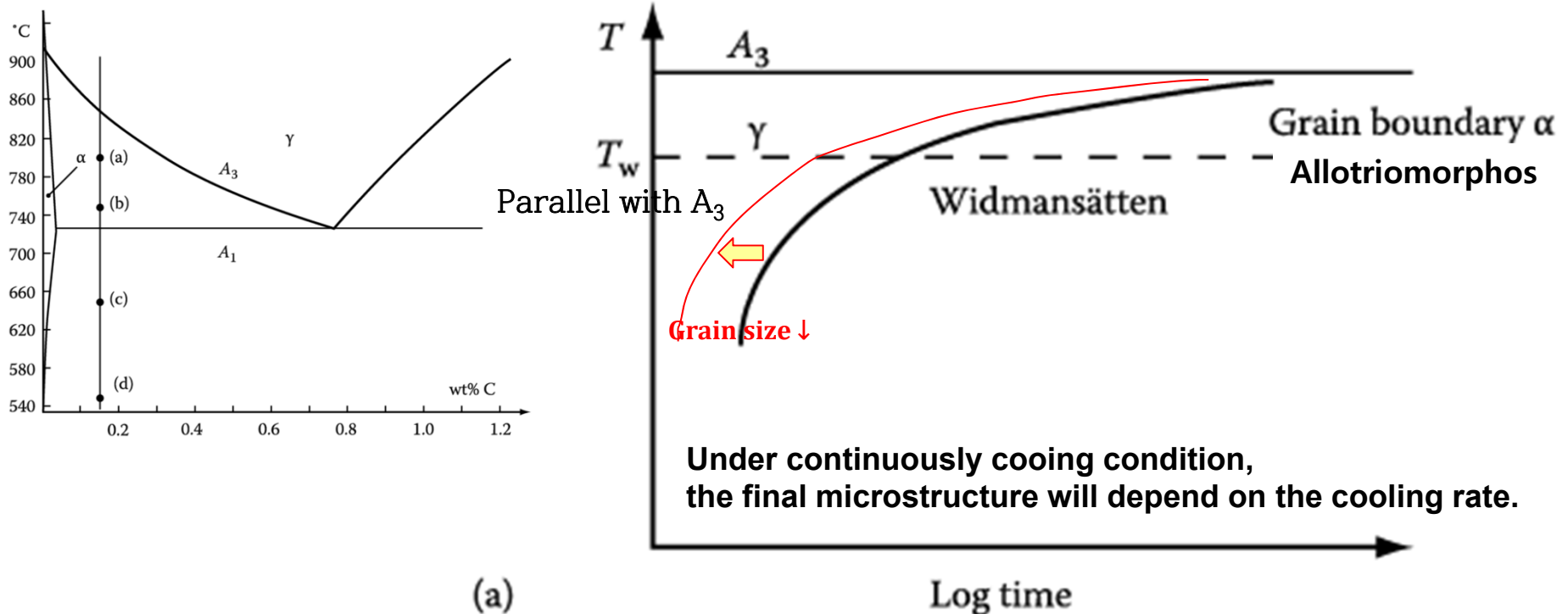
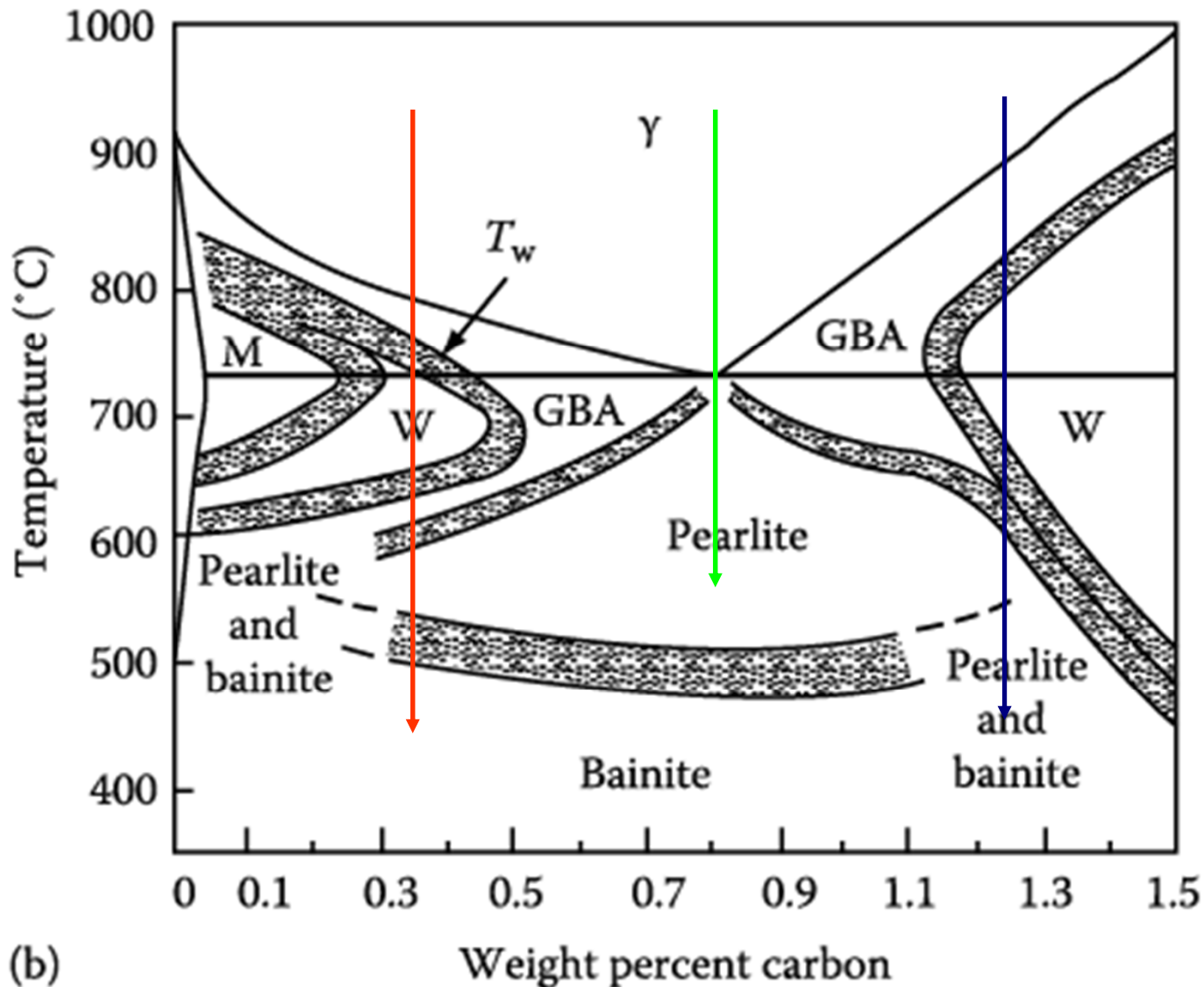


Figure 5.48 (a) typical TTT curve for  $\gamma \rightarrow \alpha$  transformation in a hypoeutectoid steel: a typical C shape.

For alloys of different carbon content,  $A_3$  and  $T_w$  vary and show parallel manner each other.



(GBA: GB allotriomorphs, W: Widmanstätten sideplates/intermolecular plates, M: Massive ferrite)

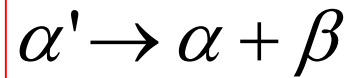
Figure 5.48 (b) Temperature-composition regions in which the various morphologies are dominant at late reaction times in specimens with ASTM grain size Nos. 0-1. 20

## **5.6.1 & 5.7 skip**

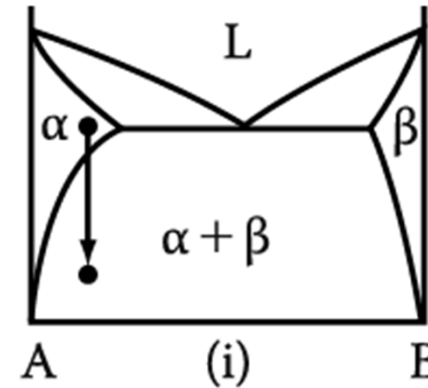
# 5. Diffusion Transformations in solid

: diffusional nucleation & growth

## (a) Precipitation



Metastable supersaturated  
Solid solution



### Homogeneous Nucleation

$$\Delta G = -V\Delta G_V + A\gamma + V\Delta G_S$$

### Heterogeneous Nucleation

$$\Delta G_{het} = -V(\Delta G_V - \Delta G_S) + A\gamma - \Delta G_d$$

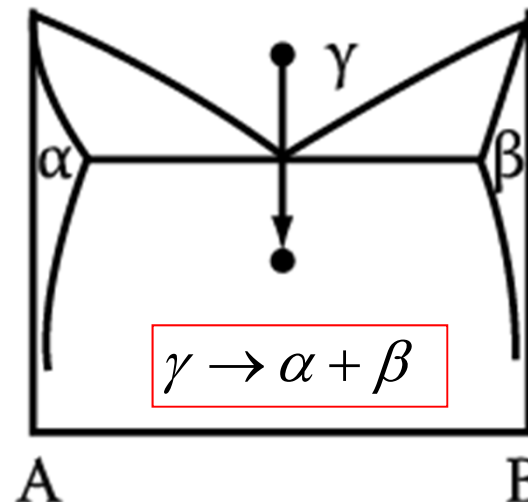
$$N_{hom} = \omega C_0 \exp\left(-\frac{\Delta G_m}{kT}\right) \exp\left(-\frac{\Delta G^*}{kT}\right)$$

→ suitable nucleation sites ~ nonequilibrium defects  
(creation of nucleus ~ destruction of a defect (-ΔG<sub>d</sub>))

## (b) Eutectoid Transformation

Composition of product phases  
differs from that of a parent phase.

→ long-range diffusion



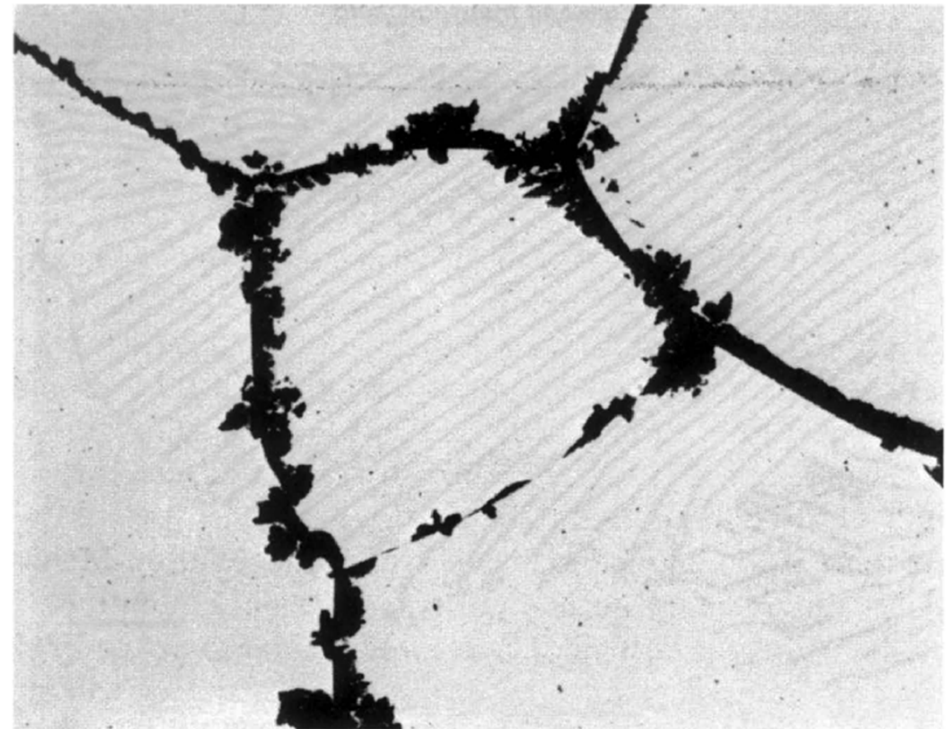
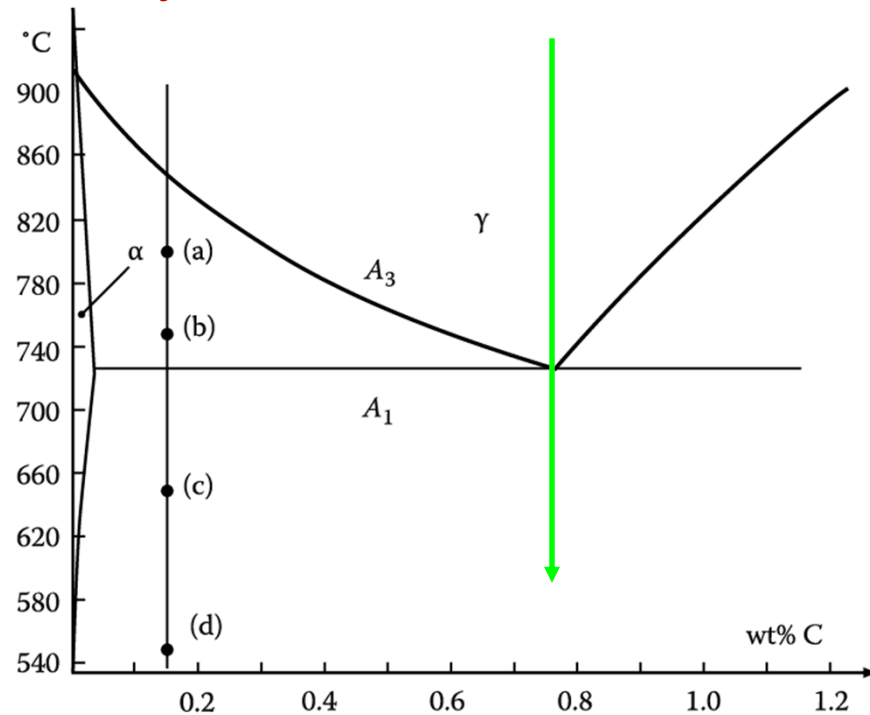
# 5.8. Eutectoid Transformation

## 5.8.1 Pearlite Reaction in Fe-C Alloys



Pearlite nodule nucleate on GBs and grow with a roughly constant radial velocity into the surrounding austenite grains.

Very similar to a eutectic transformation



\* **At large undercooling,**

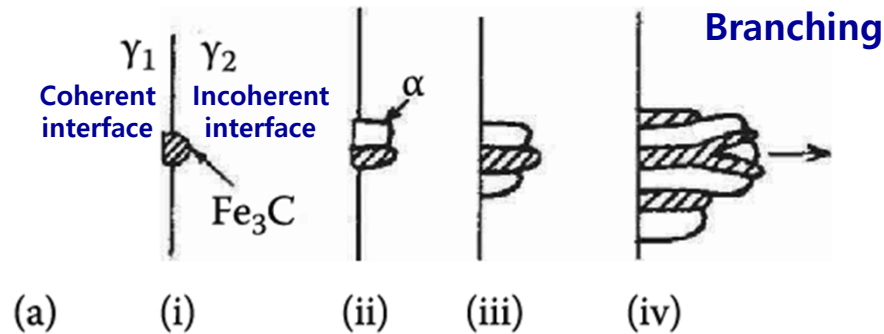
: the nucleation rate is much higher and site saturation occurs, that is all GBs become quickly covered with nodules which grow together forming layers of pearlite, Figure 5.61.

\* **At small undercooling below A<sub>1</sub>,**

: the number of pearlite nodules that nucleate is relatively small, and the nodules can grow as hemispheres or spheres without interfering with each other.

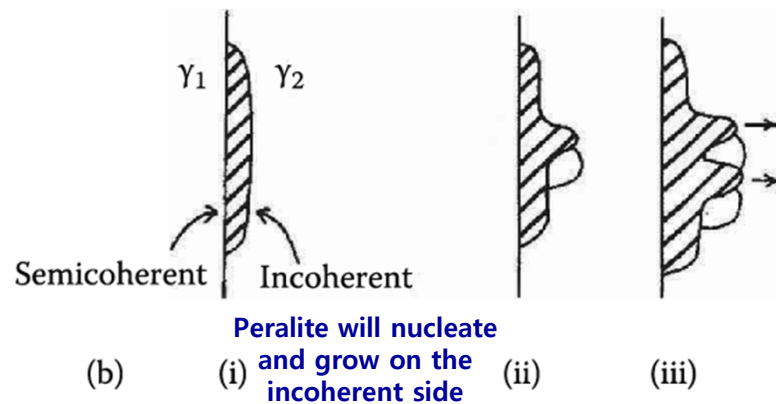
# Pearlite Reaction in Fe-C Alloys: nucleation and growth

**Nucleation:** depend on **GB structures and composition**

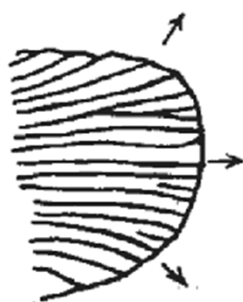


(a) On a “clean” GB.

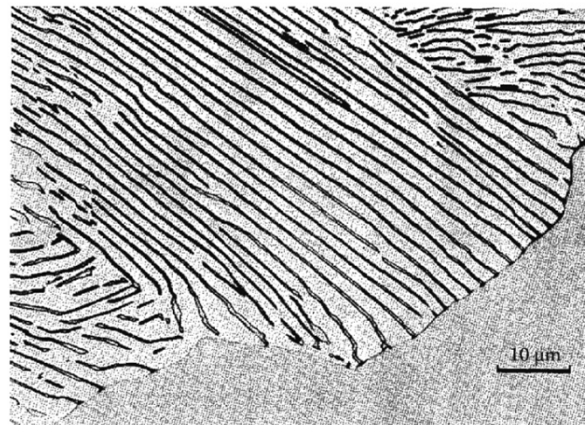
- (i) **Cementite nucleates on GB** with coherent interface and orientation relationship with  $\gamma_1$  and incoherent interface with  $\gamma_2$ .
- (ii)  **$\alpha$  nucleates adjacent to cementite** also with a coherent interface and orientation relationship with  $\gamma_1$ . (This also produces an orientation relationship between the cementite and the ferrite).
- (iii) The **nucleation process repeats side ways**, while incoherent interfaces grow into  $\gamma_2$ .
- (iv) New plates can also form by a **branching mechanism**.



(b) When a proeutectoid phase (cementite or ferrite) already exists on that boundary, pearlite will **nucleate and grow on the incoherent side**. A different orientation relationship between the cementite and the ferrite results in this case.



(c) A pearlite colony at a later stage of growth



(c) **Pearlite colony** at a latest stage of growth. Pearlite grows into the austenite grain with which it does not have an orientation relationship.



# Growth of Pearlite: analogous to the growth of a lamellar eutectic

Min. possible:  $(S^*) \propto 1/\Delta T$  / Growth rate : mainly lattice diffusion  $v = kD_c \gamma (\Delta T)^2$

Interlamellar spacing of pearlite colonies : mainly boundary diffusion  $v = kD_b (\Delta T)^3$

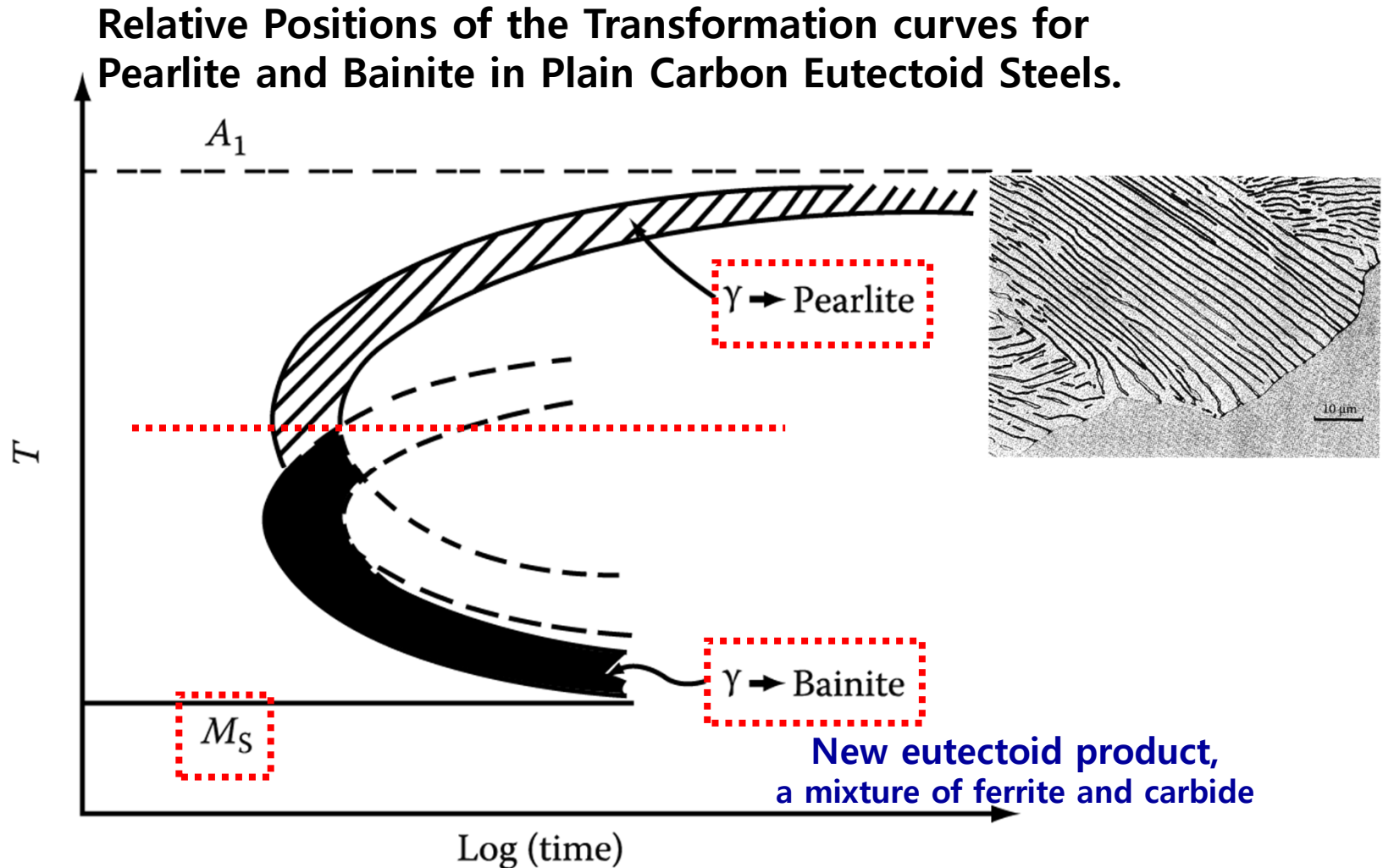


Figure 5.64 Schematic diagram showing relative positions of the transformation curves for pearlite and bainite in plain carbon eutectoid steel.

# 5.8.2 Bainite Transformation

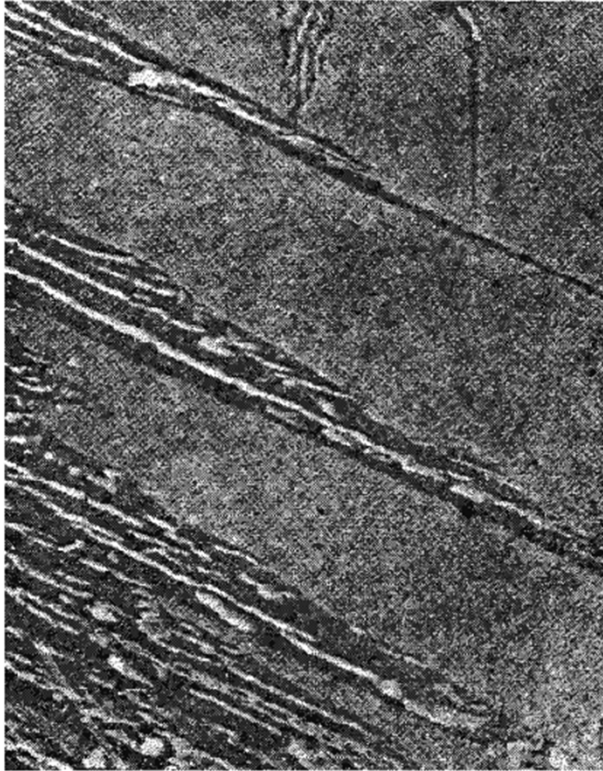
The microstructure of bainite depends mainly on the temperature at which it forms.

## Upper Banite in medium-carbon steel

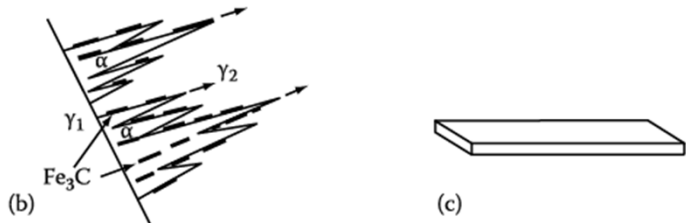
## Lower Bainite in 0.69wt% C low-alloy steel

At high temp. 350 ~ 550°C, ferrite laths, K-S relationship, similar to Widmanstätten plates

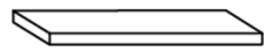
At sufficiently low temp. laths → plates  
Carbide dispersion becomes much finer, rather like in tempered M.



(a)

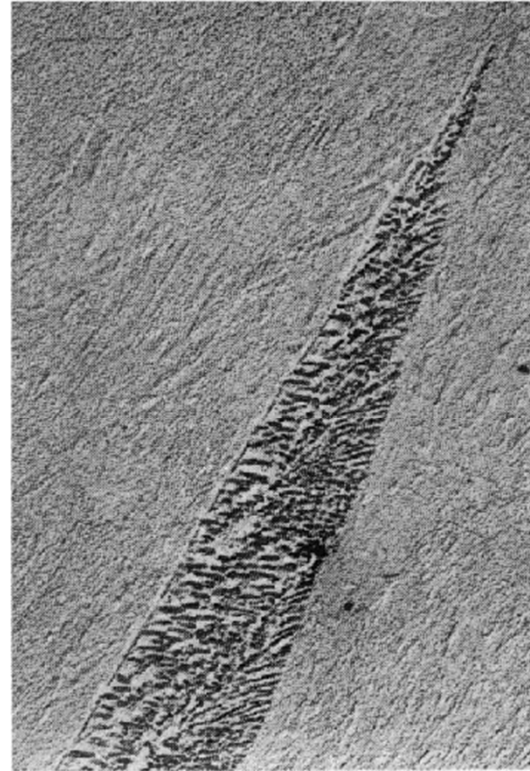


(b)

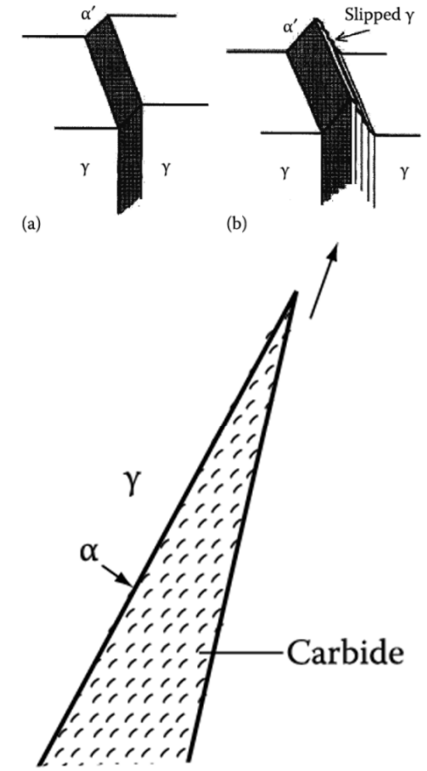


(c)

(b) Schematic of growth mechanism. Widmanstätten ferrite laths grow into  $\gamma_2$ . Cementite plates nucleate in carbon-enriched austenite.



(a)



(b)

Surface tilts by bainite trans. like M trans.  
Due to Shear mechanism/ordered military manner

(b) A possible growth mechanism.  $\alpha/\gamma$  interface advances as fast as carbides precipitate at interface thereby removing the excess carbon in front of the  $\alpha$ .

At the highest temp. where pearlite and bainite grow competitively.

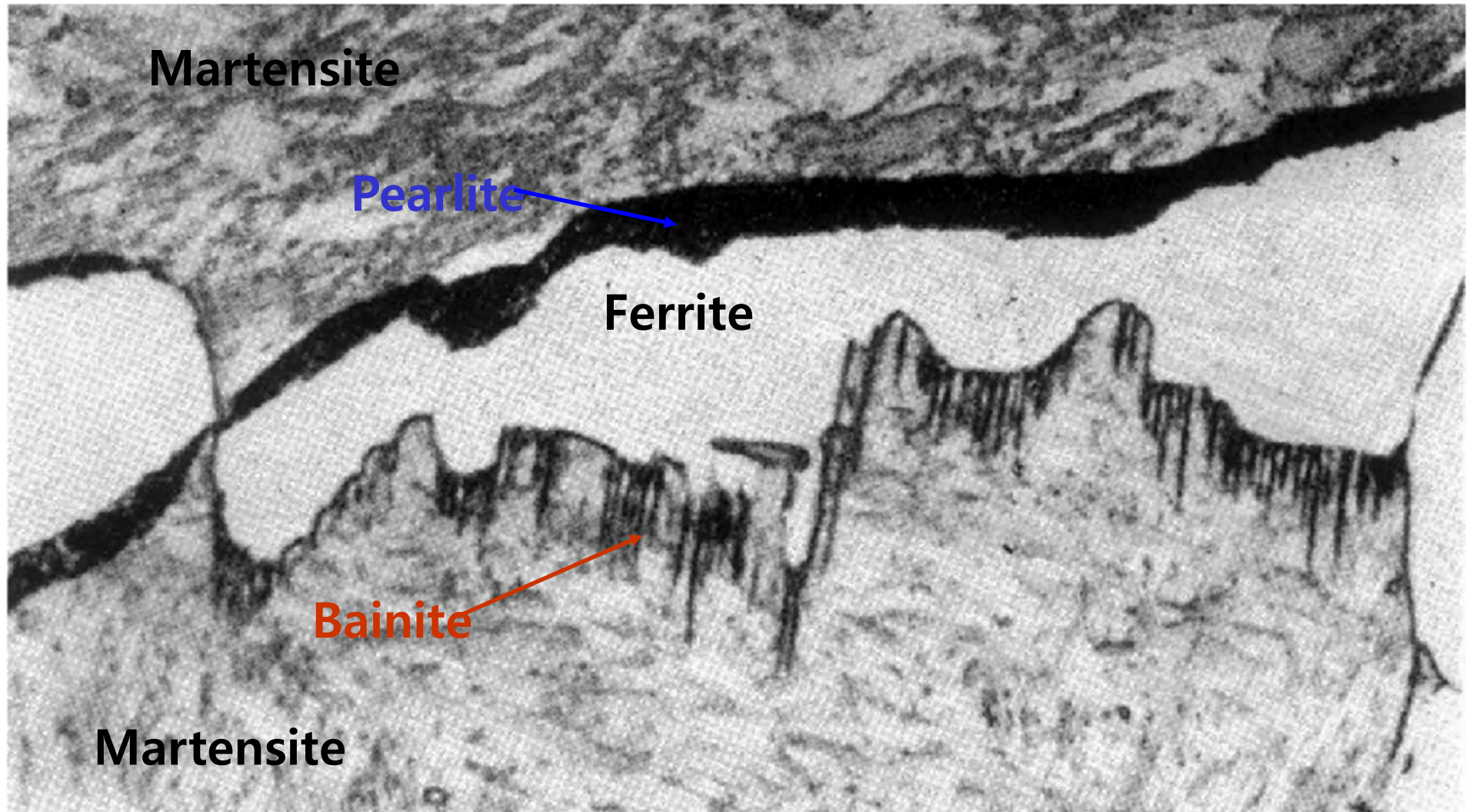


Fig. 5.67 Hypoeutectoid steel (0.6% C) partially transformed for 30 min at 710 °C. Inefficiently quenched. Bainitic growth into lower grain of austenite and pearlitic growth into upper grain during quench (x1800).

Pearlite : no specific orientation relationship

Bainite : orientation relationship

## 5.8.3 The effect of alloying elements on hardenability

: adding alloying elements to steels → delay to time required for the decomposition into ferrite and pearlite → M trans under slower cooling rate → increase hardenability

- \* **Main factor limiting hardenability is the rate of formation of pearlite at the nose of the C curve in the TTT diagram.**
- Austenite stabilizer (Mn, Cu, Ni) – depress A3 temperature
- Ferrite stabilizer (Cr, Mo, Si) – increase A3 temperature

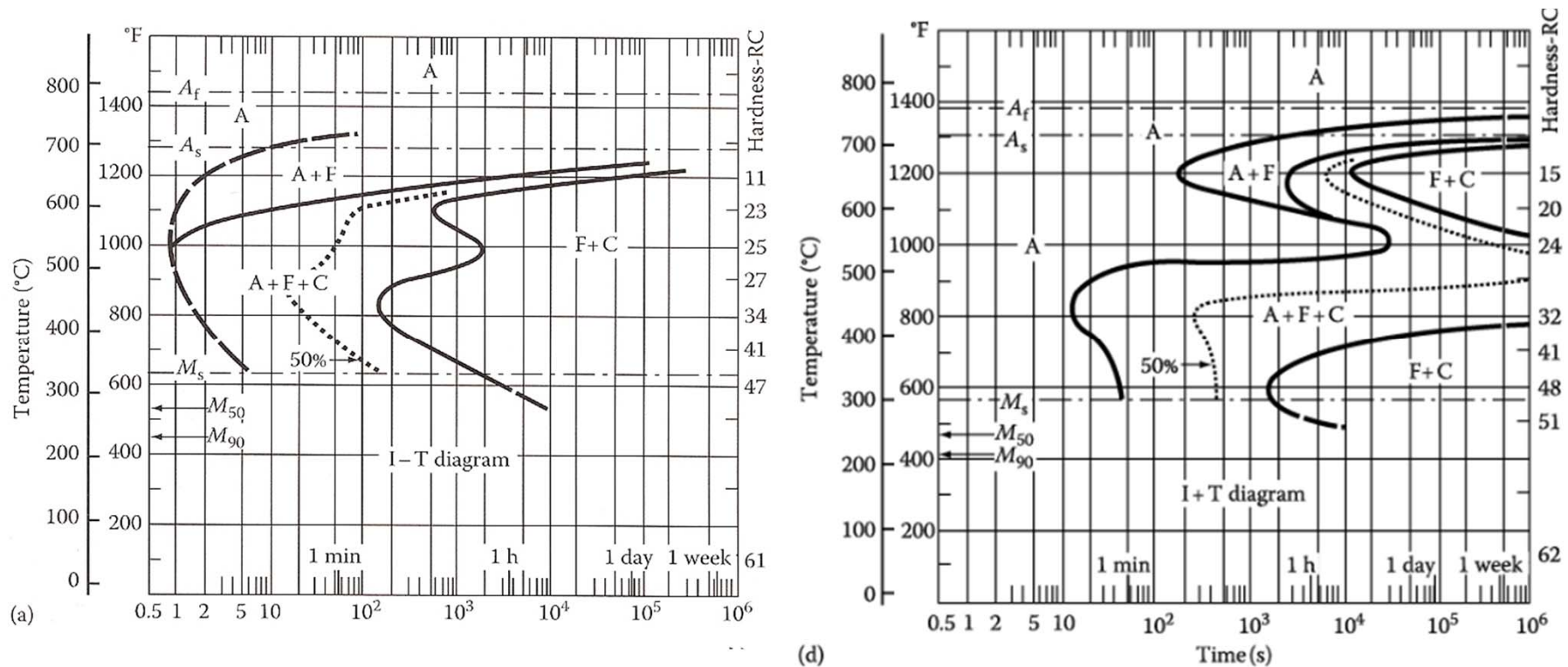
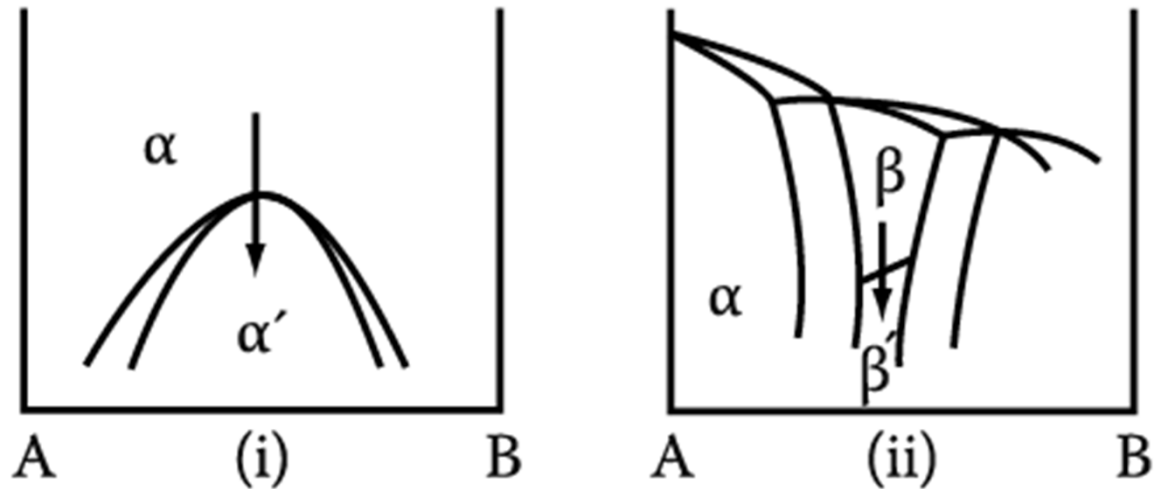
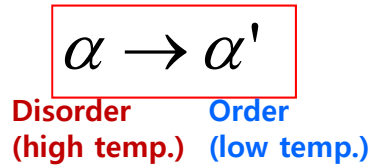


Figure 5.73 TTT diagrams for two commercial low-alloy steels all of which (a) contain roughly 0.4% C and 1% Mn and (b) contains 0.8% Cr, 0.3% Mo, and 1.8% Ni

**5.8.4 - 5.8.6 skip**

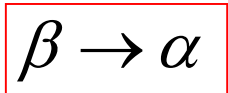
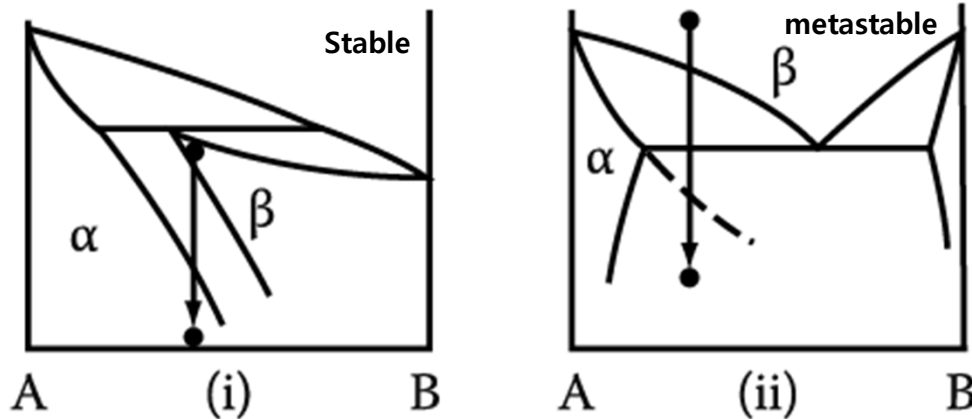
# 5. Diffusion Transformations in solid

## (c) Order-Disorder Transformation

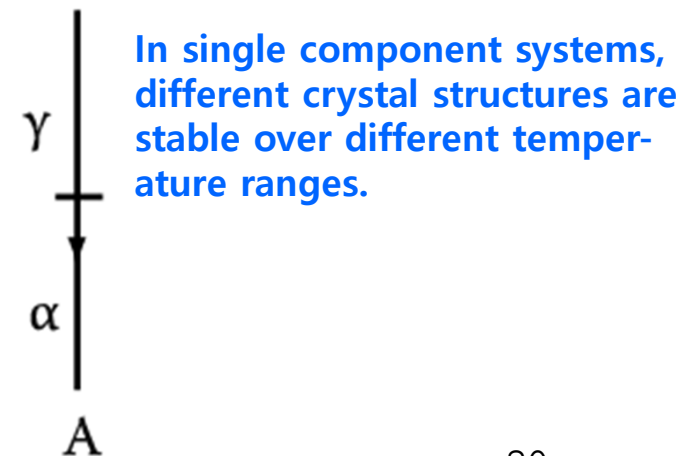


## (d) Massive Transformation

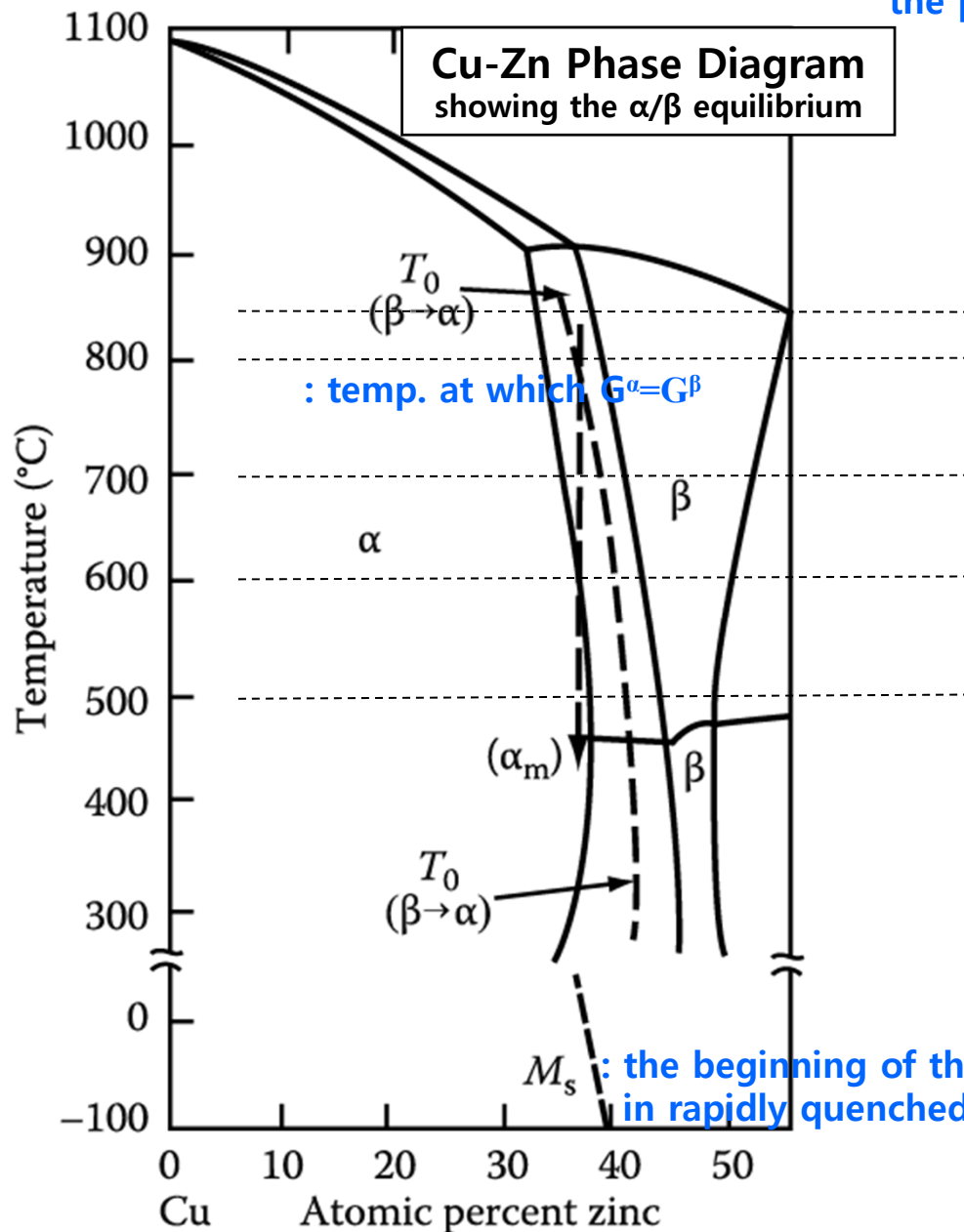
: The original phase decomposes into one or more new phases which have the same composition as the parent phase, but different crystal structures.



## (e) Polymorphic Transformation

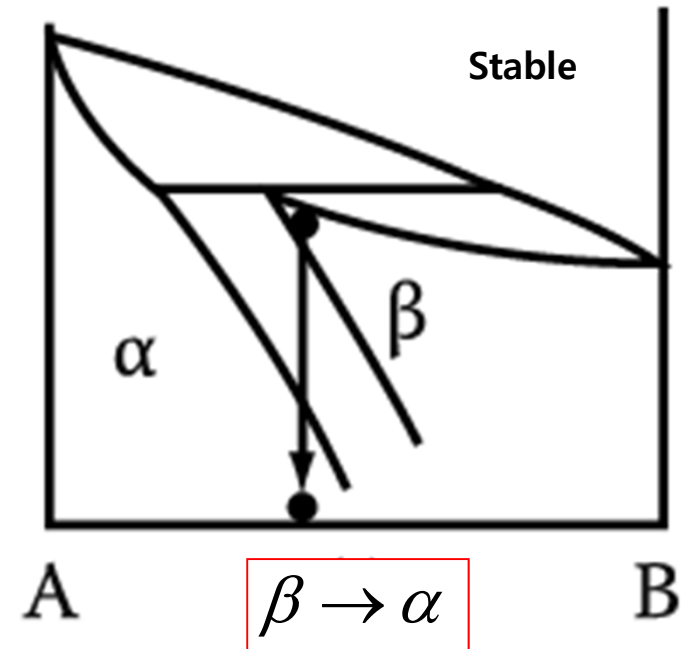


# 5.9 Massive Transformation : The original phase decomposes into one or more new phases which have the same composition as the parent phase, but different crystal structures.



: temp. at which  $G^{\alpha} = G^{\beta}$

: the beginning of the M transformation in rapidly quenched specimens.



Free energy-composition curves for  $\alpha$  and  $\beta$  at 850 $^{\circ}\text{C}$ , 800 $^{\circ}\text{C}$ , 700 $^{\circ}\text{C}$  and 600 $^{\circ}\text{C}$ ?

# 5.9 Massive Transformation

Free energy-composition curves for  $\alpha$  and  $\beta$

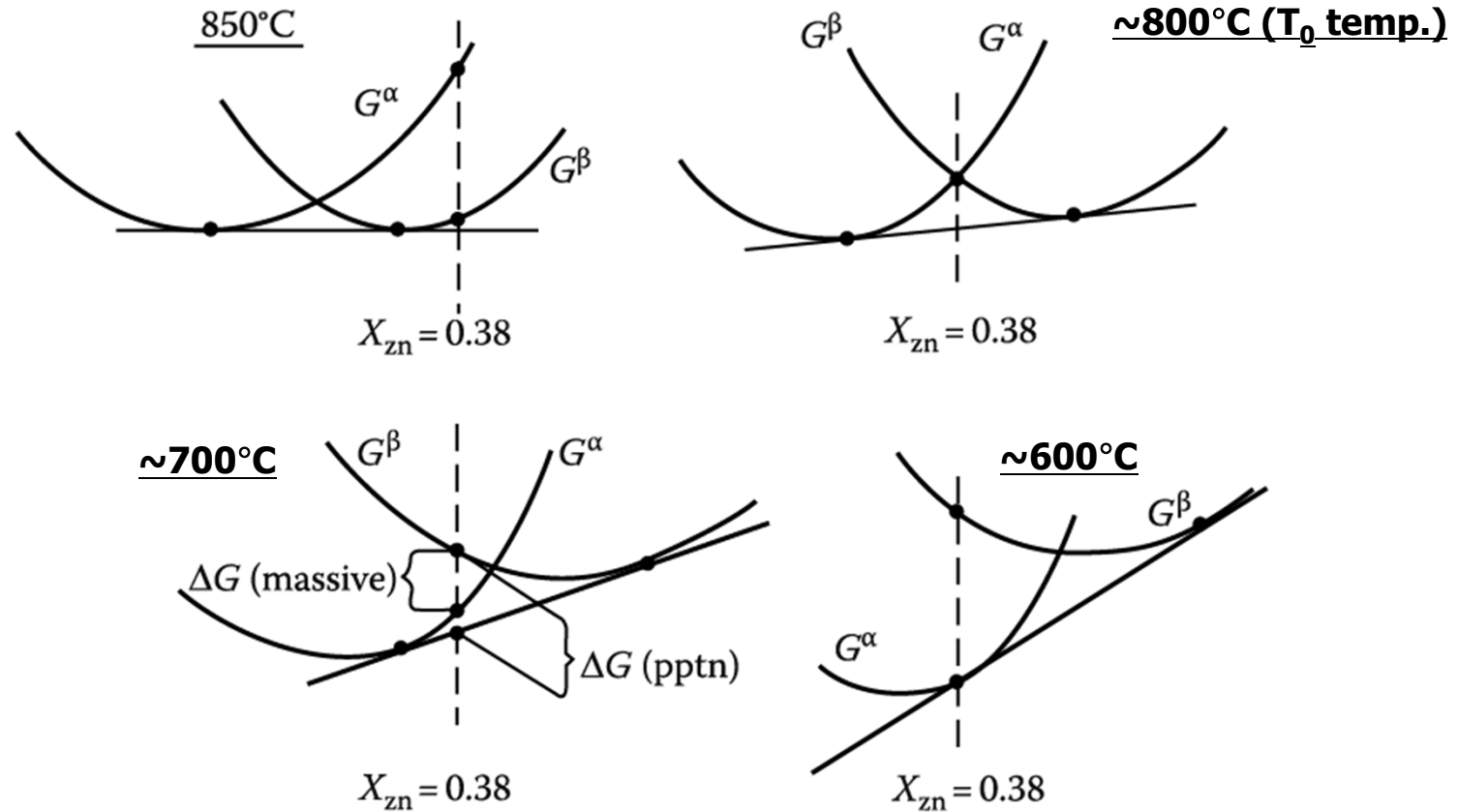


Fig. 5.86 A schematic representation of the free energy-composition curves for  $\alpha$  and  $\beta$  in the Cu-Zn system at various temperatures.

At the thermodynamic point of view, it may possible for a massive trans. to occur within the two-phase region of the phase dia. anywhere below the  $T_0$  temp.. But, in practice, there is evidence that massive trans. usually occur only within the single-phase region of the phase diagram



## 5.9 Massive Transformation

Massive  $\alpha$  formed at the GBs of  $\beta$  and grow rapidly into the surrounding  $\beta$

: a diffusionless civilian transformation (change of crystal structure without a change of composition)

Migration of the  $\alpha/\beta$  interfaces~ very similar to the migration of GBs during recrystallization of single-phase material but, driving force ~ orders of magnitude greater than for recrystallization→ rapid growth: a characteristic irregular appearance.

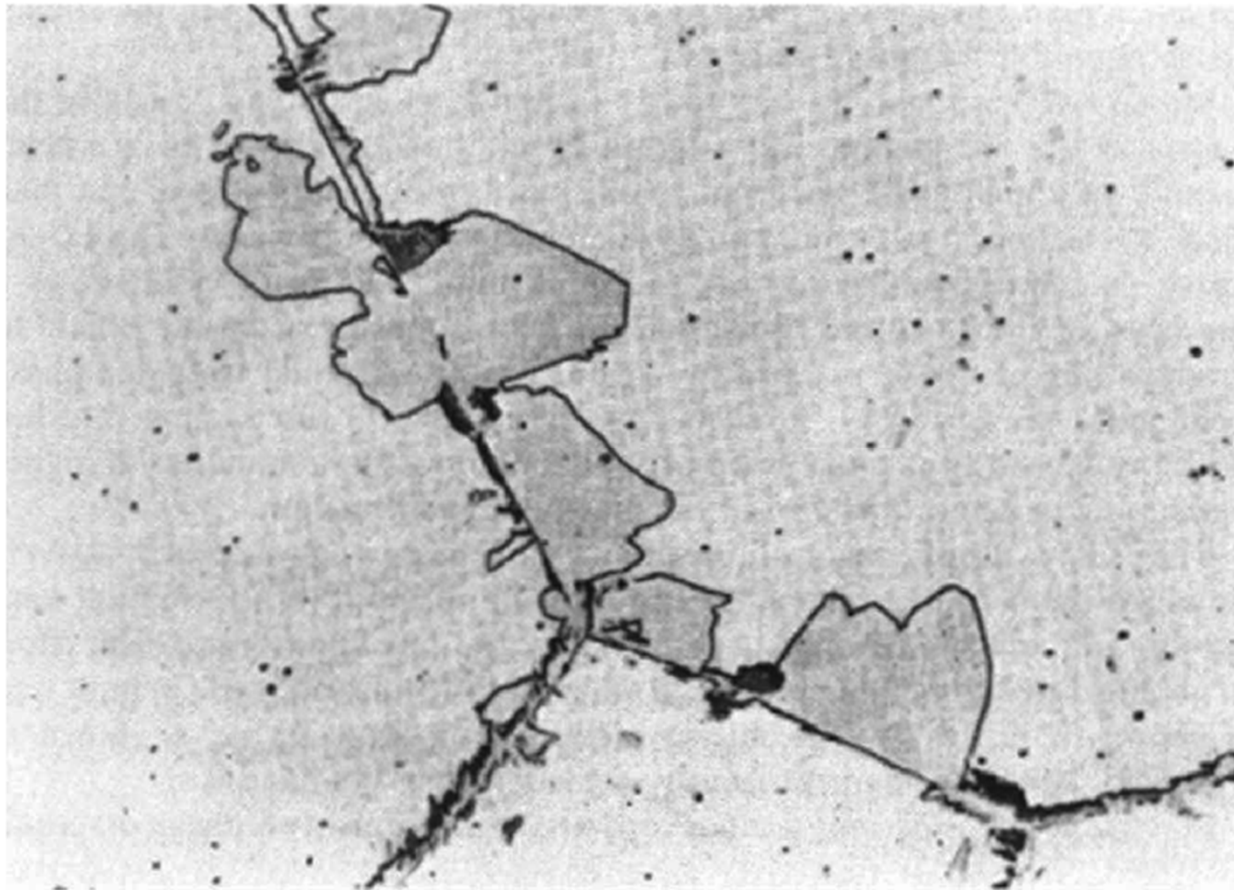


Figure 5.85 Massive  $\alpha$  formed at the GBs of  $\beta$  in Cu-38.7wt% Zn quenched from 850 °C in brine at 0 °C. Some high temperature precipitation has also occurred on the boundaries.

# \* Massive, Martensite Transformation

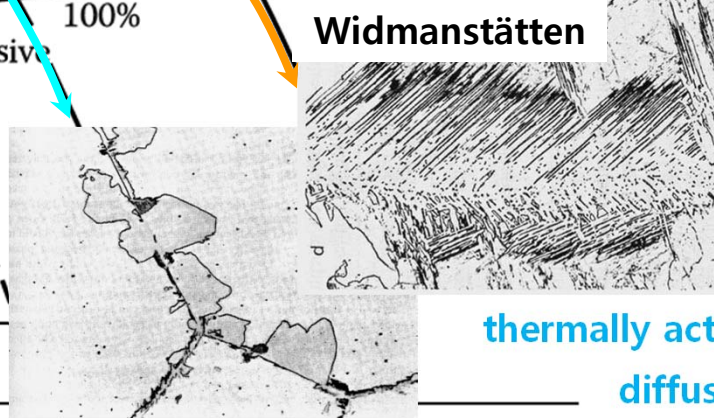
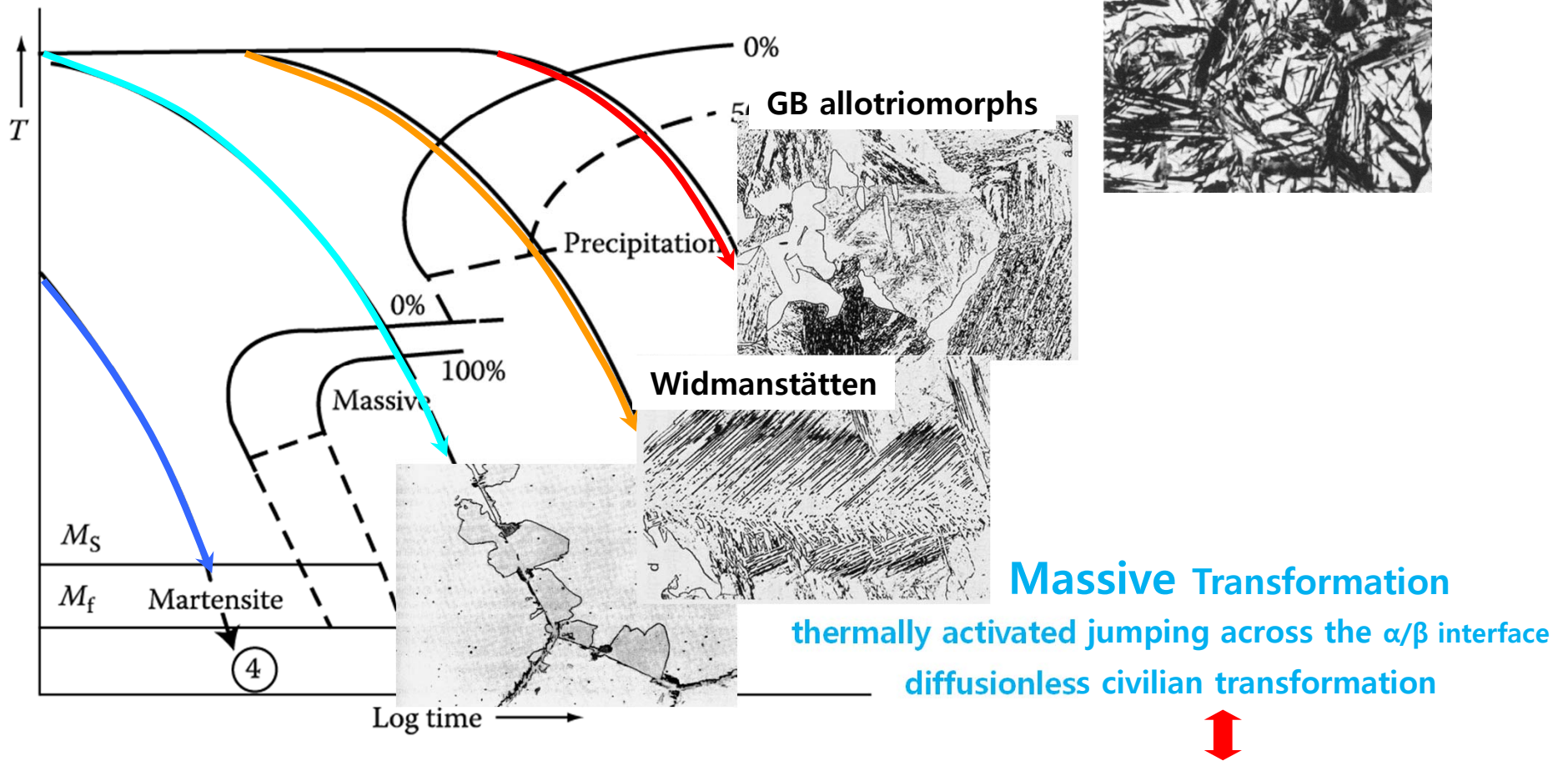
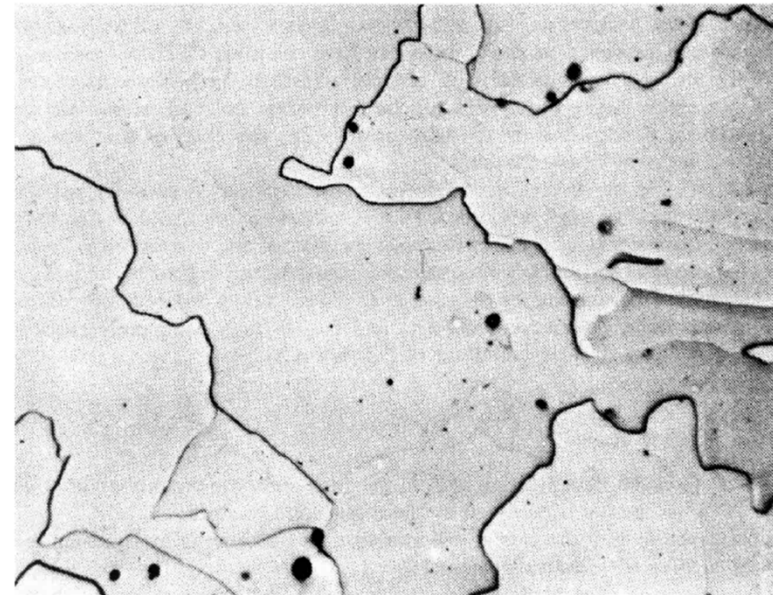
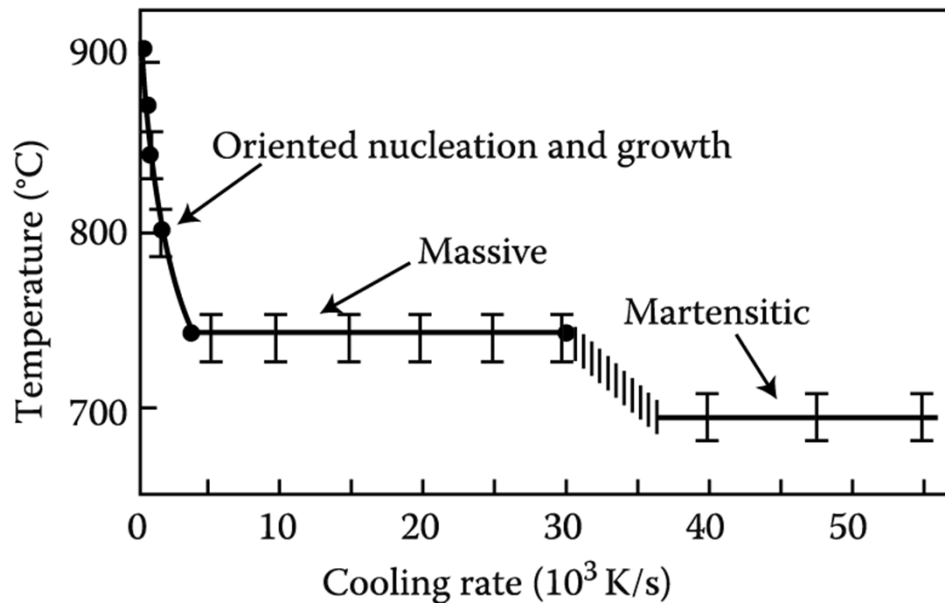


Fig. 5.75 A possible CCT diagram for systems showing a massive transformation. Slow cooling (1) produces equiaxed  $\alpha$ . Widmanstätten morphologies result from faster cooling (2). Moderately rapid quenching (3) produces the massive transformation, while the highest quench rate (4) leads to a martensitic transformation.

$\beta$  is sheared into  $\alpha$  by the cooperative movement of atoms across a glissile interface  
 diffusionless military transformation  
**Martensite Transformation**

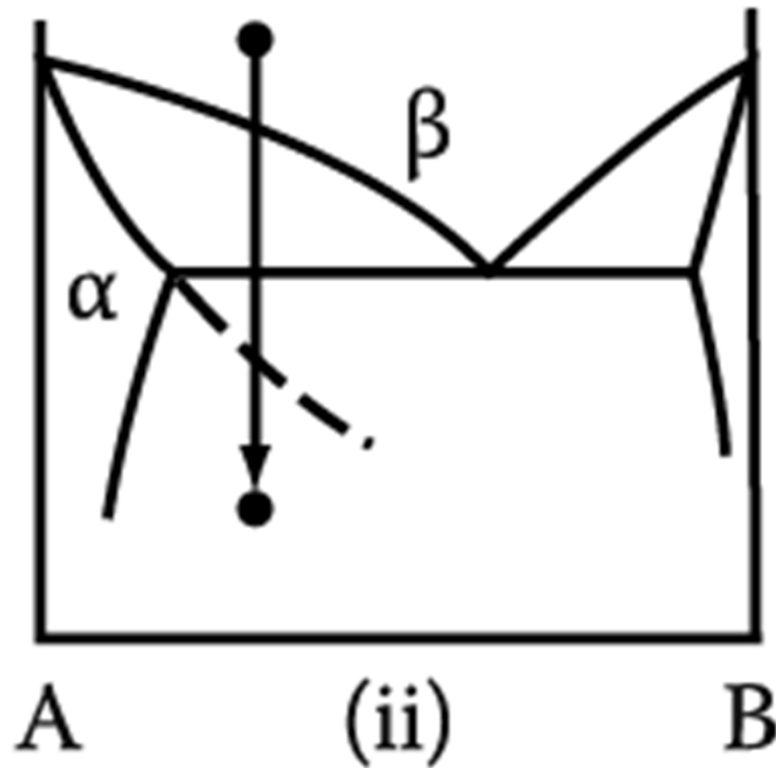
## 5.9 Massive Transformation: $\gamma \rightarrow \alpha$ transformation in iron and its alloy

Effect of Cooling Rate on the Transformation Temperature at which transformation starts in pure iron



**Massive  $\alpha$  in an Fe-0.002wt%C**  
Quenched into iced brine from 1000 °C  
: characteristically irregular  $\alpha/\alpha$  GBs.

## 5.9 Massive Transformation



**Metastable phases can also form massively.**

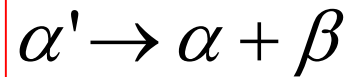
It is not even necessary for the transformation product to be a single phase: **two phases, at least one of which must be metastable**, can form simultaneously provided they have the same composition as the parent phase.

## **5.10 & 5.11 skip**

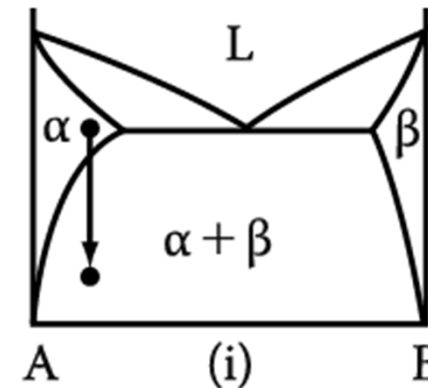
# 5. Diffusion Transformations in solid

: diffusional nucleation & growth

## (a) Precipitation



Metastable supersaturated  
Solid solution



### Homogeneous Nucleation

$$\Delta G = -V\Delta G_V + A\gamma + V\Delta G_S$$

### Heterogeneous Nucleation

$$\Delta G_{het} = -V(\Delta G_V - \Delta G_S) + A\gamma - \Delta G_d$$

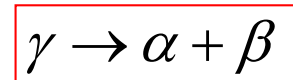
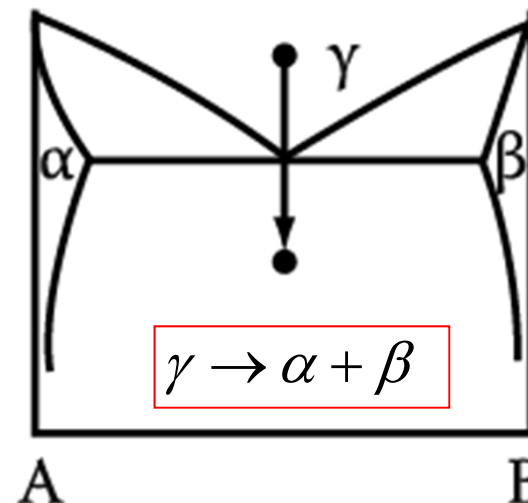
$$N_{hom} = \omega C_0 \exp\left(-\frac{\Delta G_m}{kT}\right) \exp\left(-\frac{\Delta G^*}{kT}\right)$$

→ suitable nucleation sites ~ nonequilibrium defects  
(creation of nucleus ~ destruction of a defect (- $\Delta G_d$ ))

## (b) Eutectoid Transformation

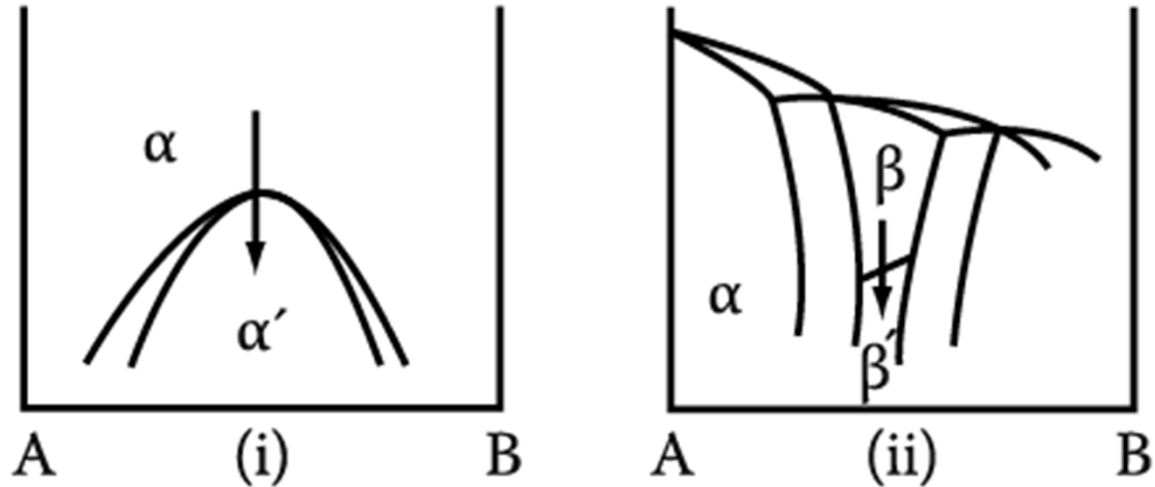
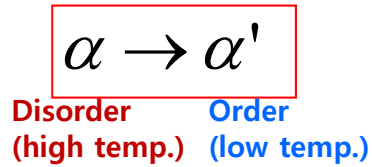
Composition of product phases  
differs from that of a parent phase.  
→ **long-range diffusion**

Which transformation proceeds  
by short-range diffusion?



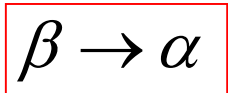
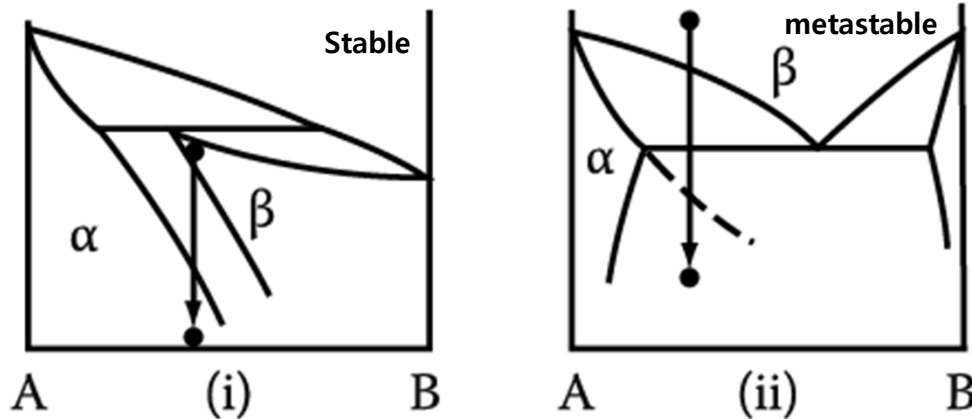
# 5. Diffusion Transformations in solid

## (c) Order-Disorder Transformation

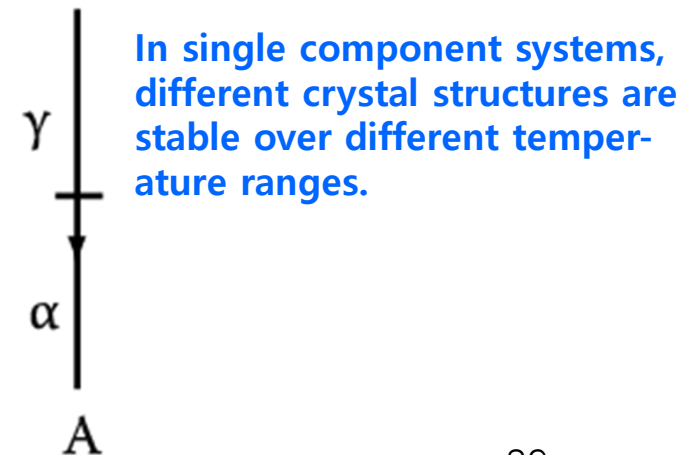


## (d) Massive Transformation

: The original phase decomposes into one or more new phases which have the same composition as the parent phase, but different crystal structures.



## (e) Polymorphic Transformation



**\* Homework 5 : Exercises 5 (pages 379-381)  
until 18th December (before exam)**

**Good Luck!!**