2015 Fall

“Phase Transformation *in* Materials”

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Contents for today's class

3.4 Interphase Interfaces in Solids

Interphase boundary - different two phases : different crystal structure
different composition

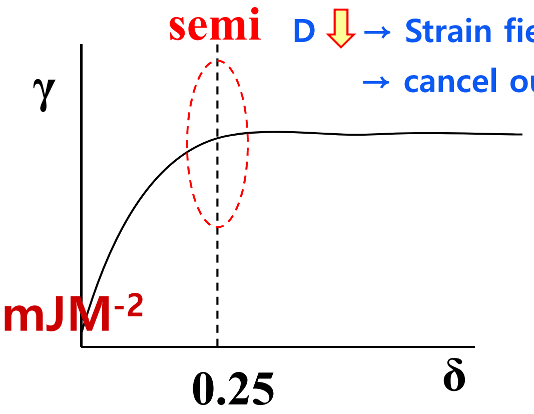
coherent, **Perfect atomic matching at interface**

γ (coherent) = γ_{ch} **γ (coherent) ~ 200 mJM^{-2}**

semicoherent $$\gamma(semicoherent) = \gamma_{ch} + \gamma_{st}$$

γ_{st} → due to structural distortions caused by the misfit dislocations

γ(semicoherent) ~ 200~500 mJM^{-2}



D ⇩ → Strain field overlap
→ cancel out

δ=4: 1 dislocation per 4 lattices

incoherent

1) δ > 0.25 No possibility of good matching across the interface

2) different crystal structure (in general)

γ (incoherent) ~ 500~1000 mJM^{-2}

Complex Semicoherent Interfaces

Nishiyama-Wasserman (N-W) Relationship **Kurdjumov-Sachs (K-S) Relationships**

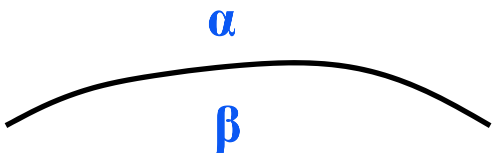
(The only difference between these two is a rotation in the closest-packed planes of 5.26°.)

The degree of coherency can, however, be greatly increased if a macroscopically irrational interface is formed.

Contents for today's class

3.4 Interphase Interfaces in Solids (α/β)

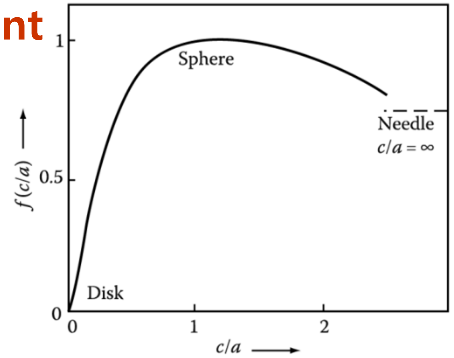
1) Interphase boundary - **different two phases : different crystal structure**
different composition

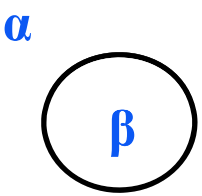


- **Coherent/ Semicoherent/ Incoherent**
- **Complex Semicoherent**

2) Second-Phase Shape

$$\boxed{\sum A_i \gamma_i + \Delta G_S = minimum}$$





Interface energy + Misfit strain energy

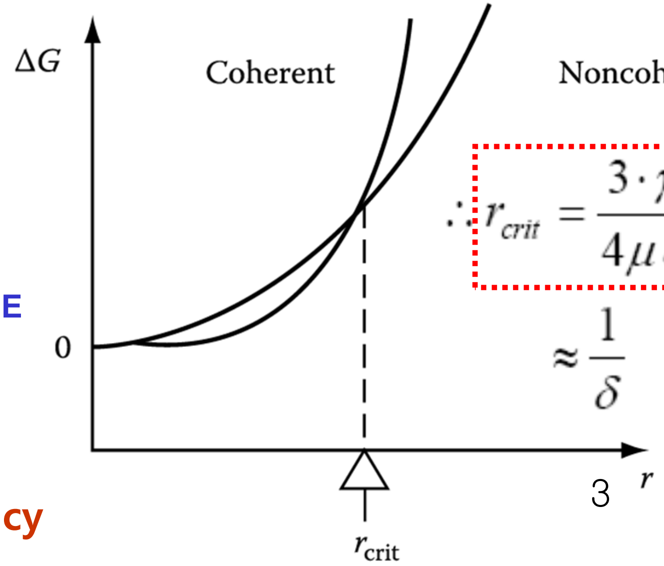
$$\Delta G_S = 4\mu\delta^2 \cdot V$$ **(If v=1/3)** ⬄ $$\Delta G_S = \frac{2}{3}\mu\Delta^2 \cdot V \cdot \boxed{f(c/a)}$$

Fully coherent precipitates **Incoherent inclusions**

3) Coherency Loss

$$\Delta G(coherent) = \underbrace{4\mu\delta^2 \cdot \frac{4}{3}\pi r^3}_{\text{Coherency strain energy}} + \underbrace{4\pi r^2 \cdot \gamma_{ch}}_{\text{Chemical interfacial E}}$$

$$\Leftrightarrow \Delta G(non-coherent) = \underbrace{4\pi r^2 \cdot (\gamma_{ch} + \gamma_{st})}_{\text{Chemical and structural interfacial E}}$$



$$\boxed{\therefore r_{crit} = \frac{3 \cdot \gamma_{st}}{4\mu\delta^2}}$$

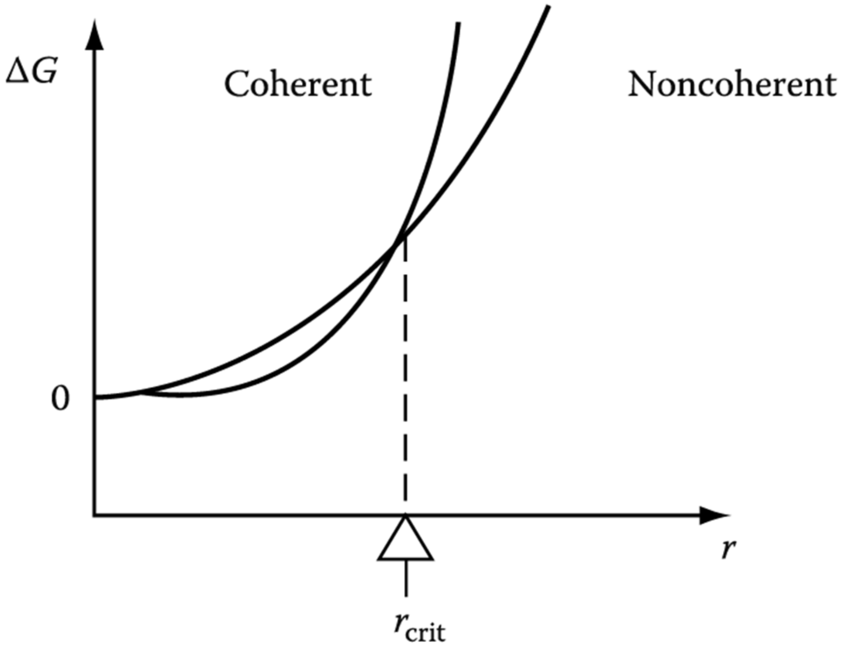
$$\approx \frac{1}{\delta}$$

If a coherent precipitate grows, it should lose coherency to maintain minimum interfacial free E.

Q: Which state produces the lowest total E for a spherical precipitate?

"Coherency loss"

**If a coherent precipitate grows, during aging for example,
It should lose coherency when it exceeds r_{crit}.**

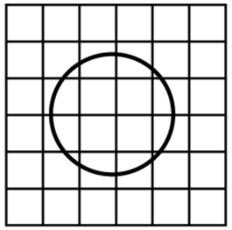


Coherency Loss

- Precipitates with coherent interfaces=low interfacial E + coherency strain E
- Precipitates with non-coherent interfaces=higher interfacial E

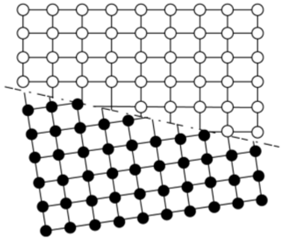
If a coherent precipitate grows, it should lose coherency to maintain minimum interfacial free E.

$$\Delta G(coherent) = 4\mu\delta^2 \cdot \frac{4}{3}\pi \boxed{r^3} + 4\pi \boxed{r^2} \cdot \gamma_{ch} \Leftrightarrow \Delta G(non-coherent) = 4\pi \boxed{r^2} \cdot (\gamma_{ch} + \gamma_{st})$$



Coherency strain energy
Eq. 3.39

Chemical interfacial E



Chemical and structural interfacial E

$$\frac{4}{3}\pi r^3 (4\pi\mu\delta^2) + 4\pi r^2 (\gamma_{ch}) = 4\pi r^2 (\gamma_{st} + \gamma_{ch})$$

coherent ΔG_s-relaxed

$$\therefore \boxed{r_{crit} = \frac{3 \cdot \gamma_{st}}{4\mu\delta^2}}$$

$$for\ small\ \delta, \quad \gamma_{st} \propto \delta$$

(semi-coherent interface)

$$\approx \frac{1}{\delta} \quad \mathbf{(\delta = (d_\beta - d_\alpha)/\ d_\alpha : misfit)}$$

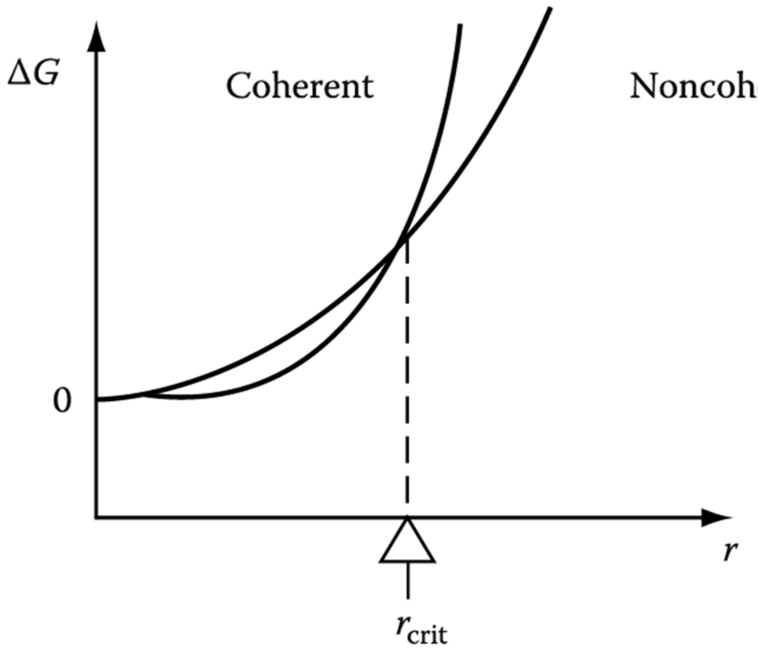
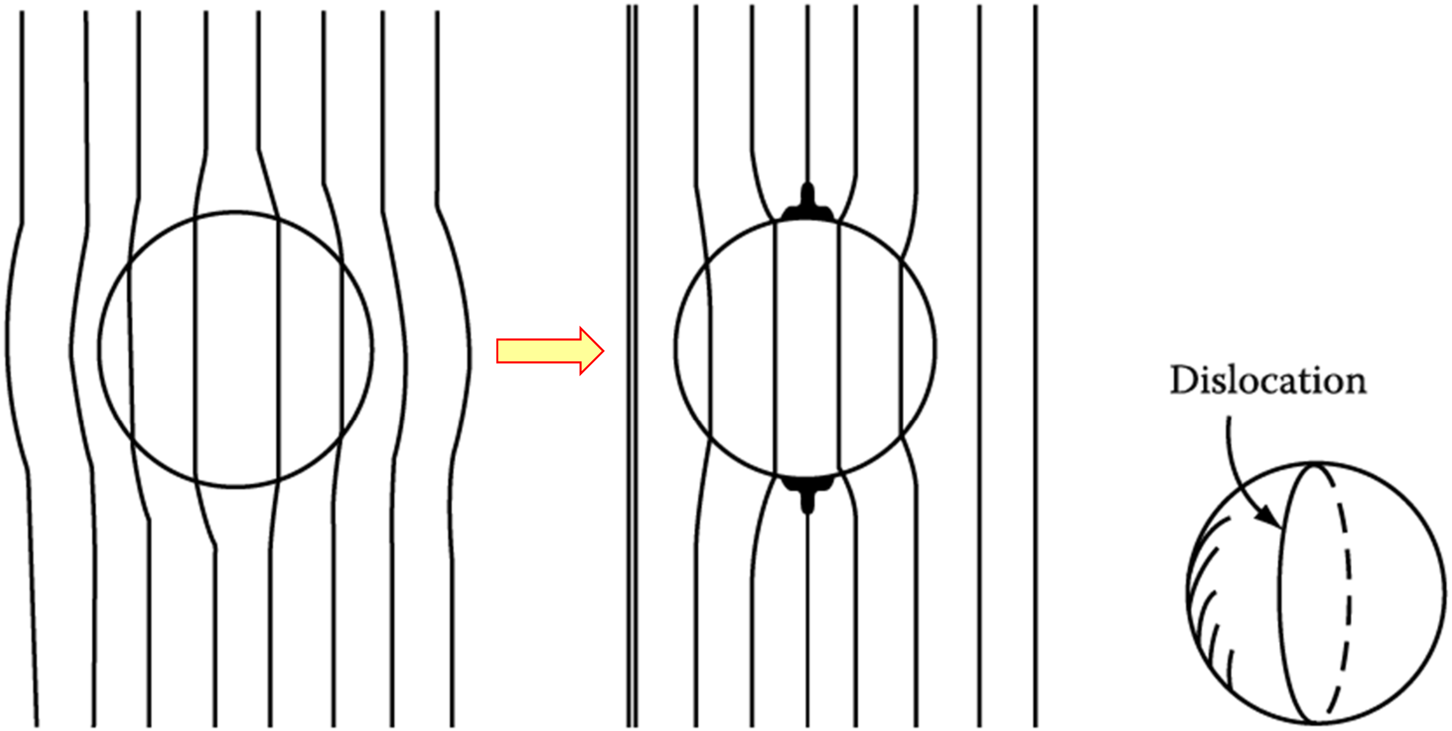


Fig. 3. 52 The total energy of matrix + precipitate vs. precipitate radius for spherical coherent and non-coherent (semicoherent of incoherent) precipitates.

**If a coherent precipitate grows, during aging for example,
It should lose coherency when it exceeds r_{crit}.**



(a) **Coherent with strain E** (b) **Coherency strain replace by dislocation loop.** (c) **Precipitate with dislocation**

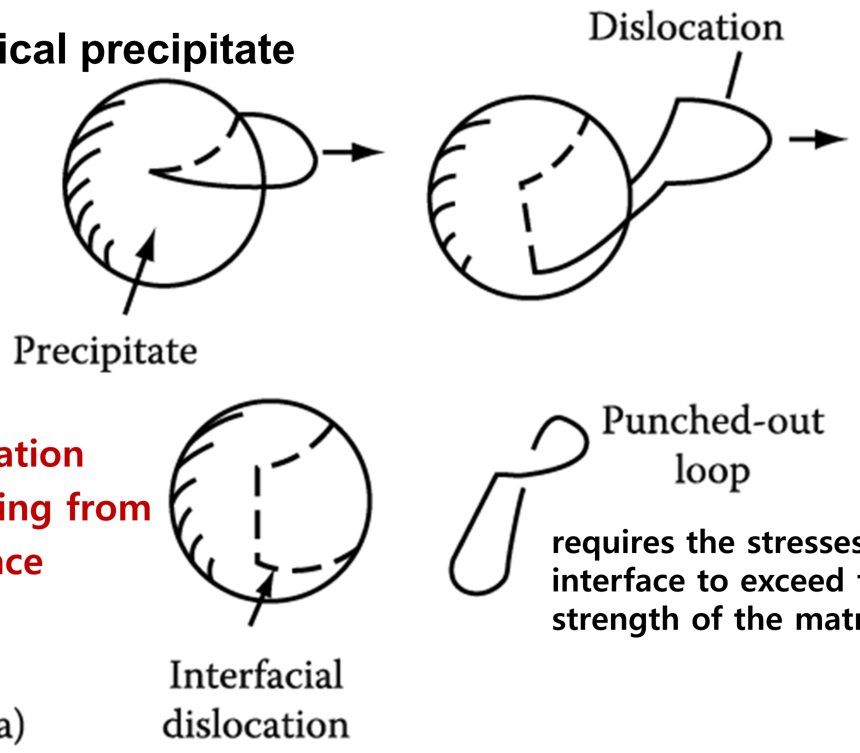
Fig. 3.53. Coherency loss for a spherical precipitate

In practice, this phenomena can be rather difficult to achieve.

→ **Coherent precipitates are often found with sizes much larger than r_{crit}.**

"Mechanisms for coherency loss": all require the precipitate to reach a larger size than r_{crit}

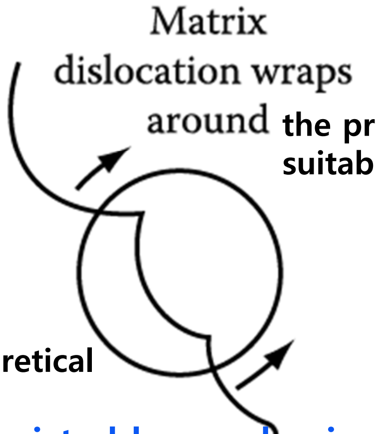
1) spherical precipitate



Dislocation punching from interface

requires the stresses at the interface to exceed the theoretical strength of the matrix

* Punching stress (P_s) ~ independent of size, but $P_s \propto$ constrained misfit, ε ($>\varepsilon_{crit}\sim 0.05$), → "precipitates with a smaller ε cannot lose coherency by (a), no matter how large."

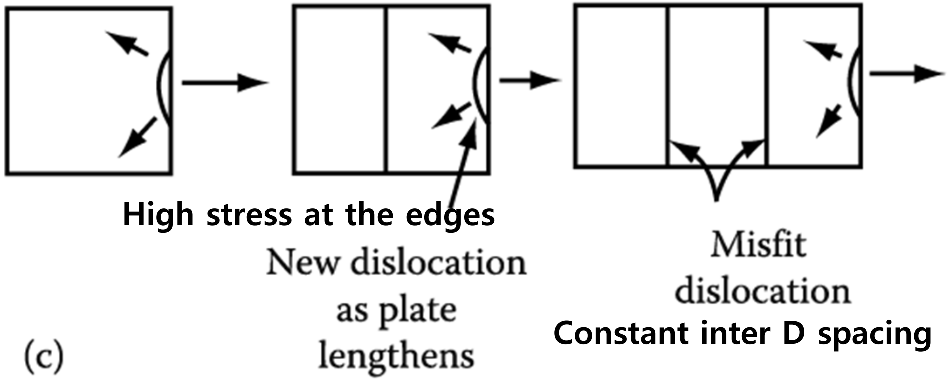


the precipitate with a suitable Burgers vector

assisted by mechanical deformation

(b) **Capture of matrix dislocation**

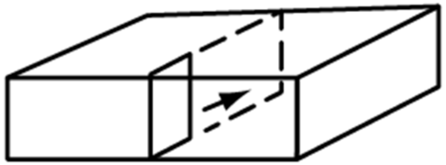
2) Plate precipitate



High stress at the edges

Constant inter D spacing

Nucleation of dislocation at the edge → maintain a roughly constant inter-dislocation spacing during plate lengthening



Prismatic dislocation loop

Nucleation of D loops within the precipitate

(d) Vacancies can be attracted to coherent interfaces and 'condense' to form a prismatic dislocation loop which can expand across the precipitate

Contents for today's class

- **Interphase Interfaces in Solid (α/β)**

  - Types of interphase interfaces in solid (α/β)

  - Second-Phase Shape

    $$\sum A_i \gamma_i + \Delta G_S = minimum$$

    Interface Energy Effects

    Coherent / Semi-coherent / incoherent

    Misfit Strain Effects

  - Coherency Loss

  - **Glissil Interfaces ⟺ Solid/Liquid Interfaces**

- **Interface migration**

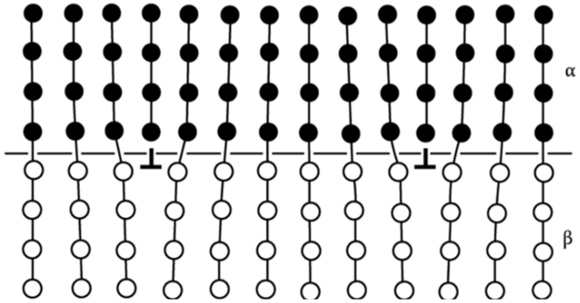
  - **Interface controlled growth ⟺ Diffusion controlled growth**

Q: What is Glissile interface?

Glissile interface → coordinated glide of the interfacial disl. → (α→β) phase transformation

Interphase Interfaces in Solid (α/β)

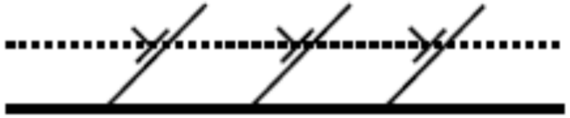
1) Glissile Interfaces (평활 이동 계면)



: epitaxial; Can't move forward or backward

(interface//burgers vector) ⟶ **Non-glissile interface**

: Glide of the interfacial disl. cannot cause the interface to advance



: Glissile; Boundary moves toward α or β

: semi-coherent interfaces which can advance by the coordinated glide of the interfacial disl.

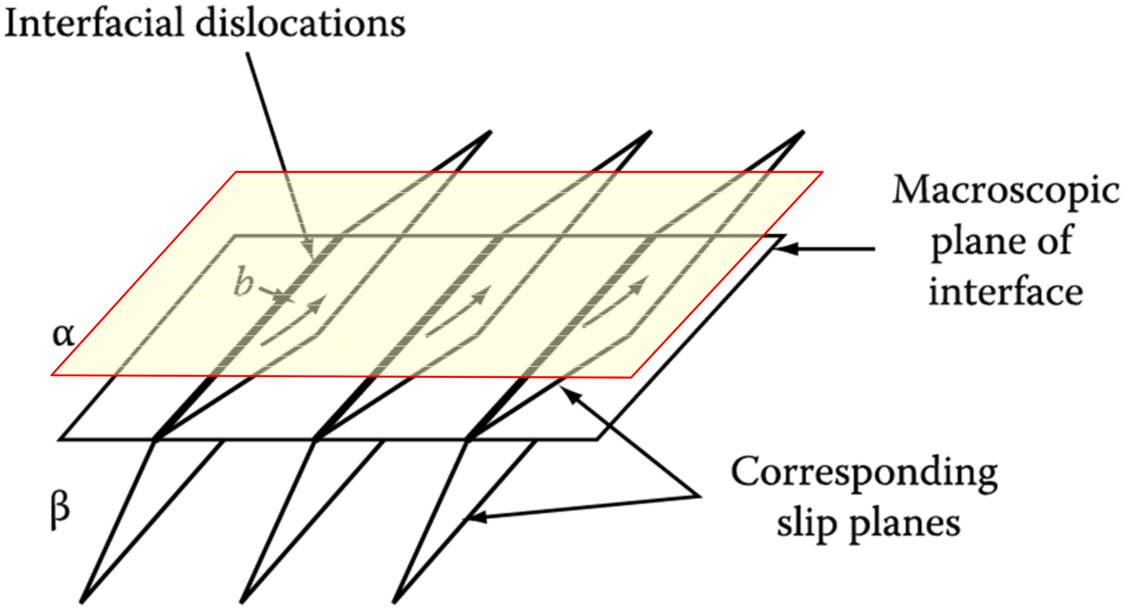


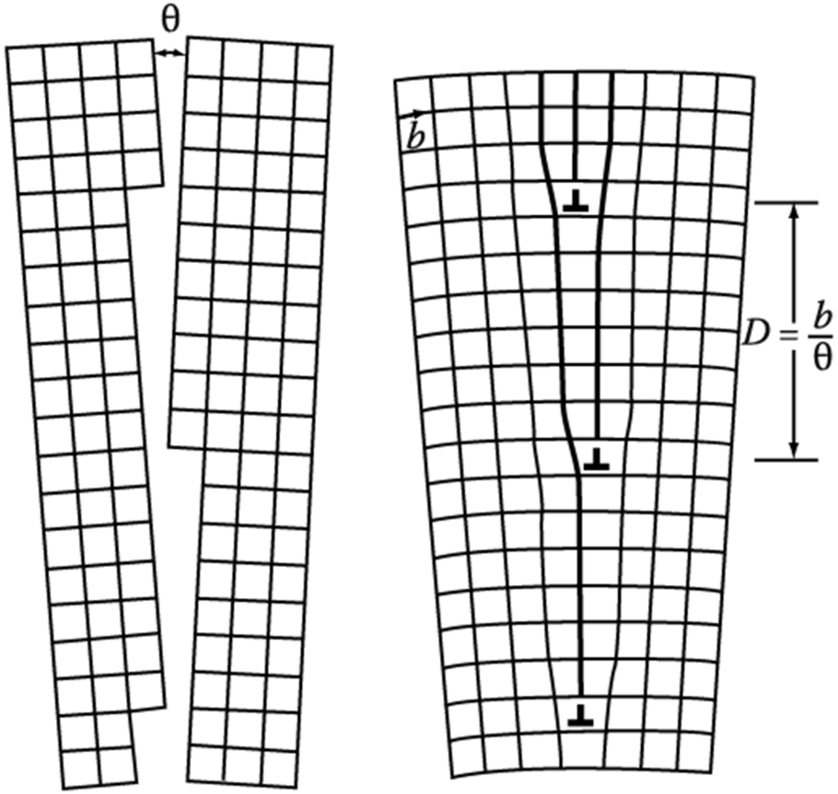
Fig. 3. 55 The nature of a glissile interface.

The dislocations have a Burgers vector that can glide on matching planes in the adjacent lattices.

Slip planes : continuous across the interface

Gliding of the dislocation : α is sheared into the β structure.

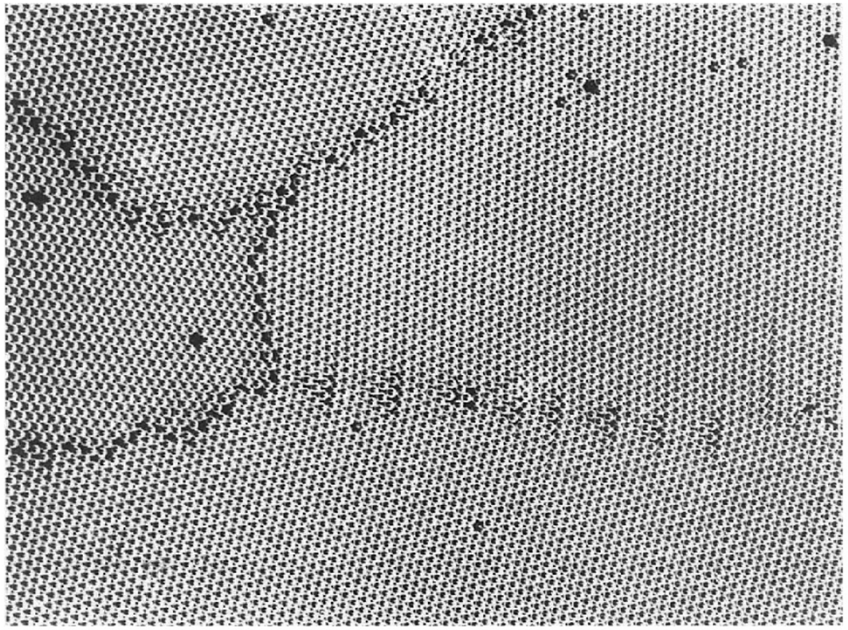
Low-Angle tilt Boundaries



Bugers vector = edge dislocation

But, this is not interphase interface.

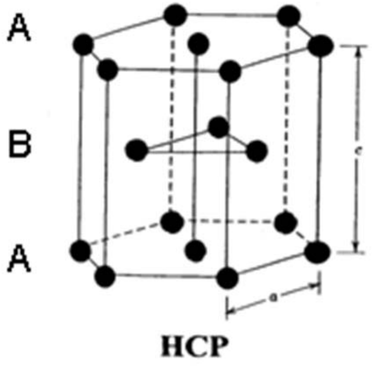
∵ crystal structure is same,
only lattice rotation

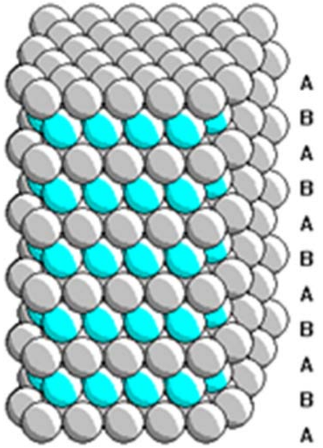


* As disl. glide at low-angle grain boundary
: no change in crystal structure,
just rotation of the lattice into the other grain

Glissile Interfaces between two lattices

Shockley partial dislocation

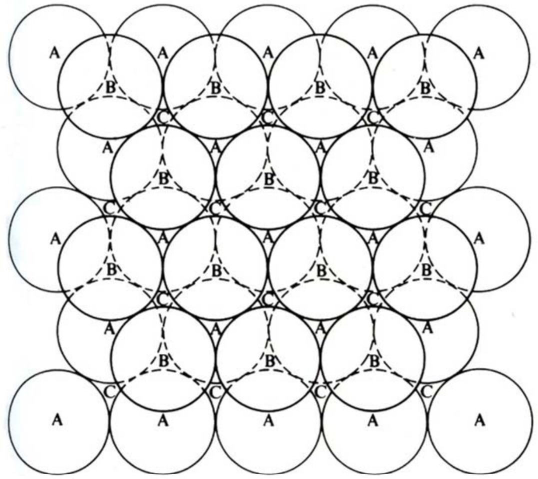




HCP: ABABABAB…

close packed plane: (0001)

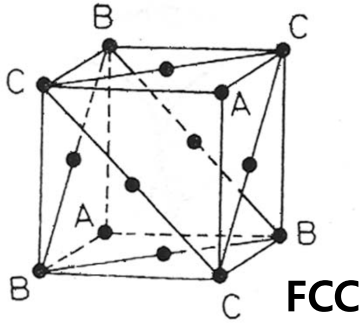
close packed directions: $<11\bar{2}0>$

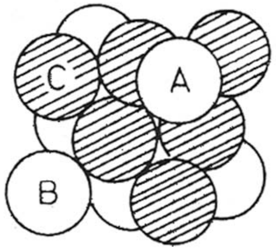


FCC: ABCABCAB…

close packed planes: {111}

close packed directions: $<110>$



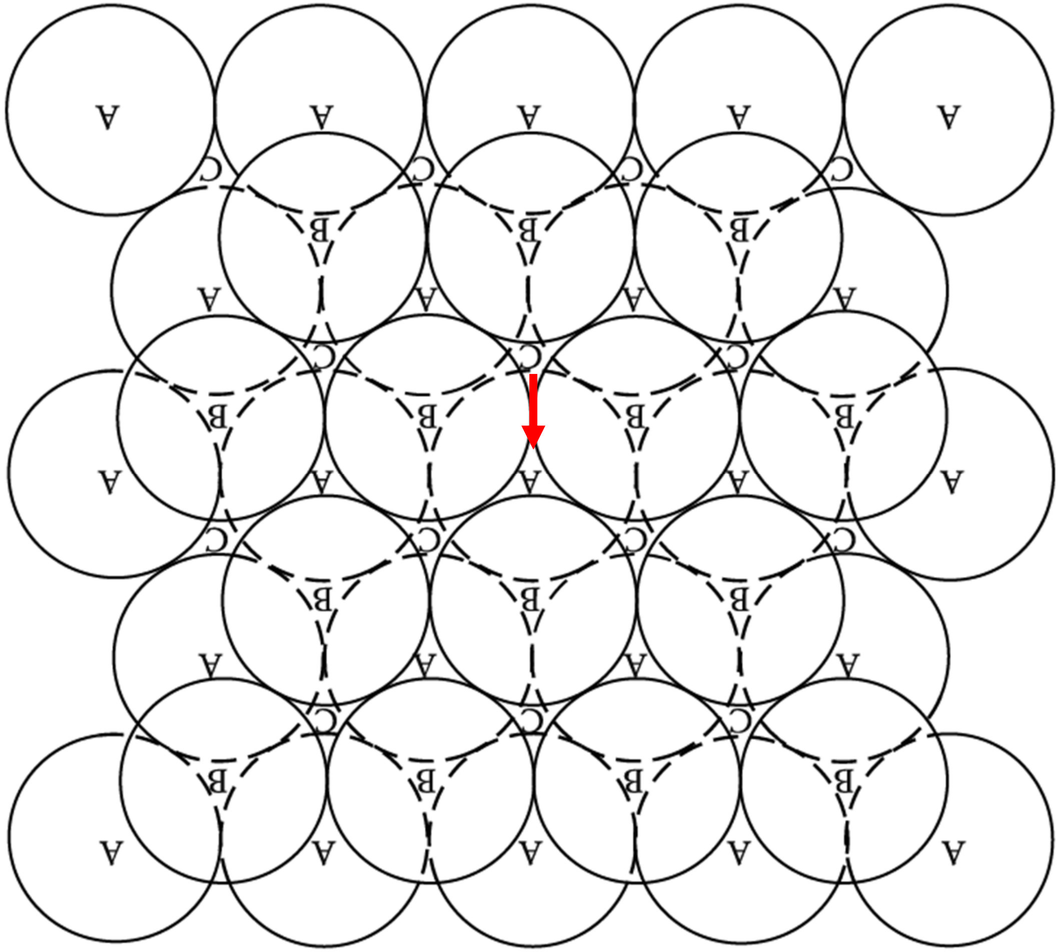


1) Perfect dislocation

$$\vec{b} = \frac{a}{2}[10\bar{1}]$$

In C layer, atoms move
C' → C"
: remain a cubic close-packed arrangement with a fcc unit cell

< FCC → FCC >



2) Shockley partial dislocation

when atoms move: C' → C"

possible to move C' → A → C"

$$\frac{a}{2}[10\bar{1}] = \boxed{\frac{a}{6}[11\bar{2}]} + \frac{a}{6}[2\bar{1}\bar{1}]$$

This burgers vector of partial disl.
is not located at lattice point.

(can't connect lattice points in the FCC structure)

$$\begin{tikzcd} & C' \arrow[dl, "\vec{b} = \frac{a}{2}[10\bar{1}]"'] \arrow[d, "\frac{a}{6}[11\bar{2}]"] \\ C'' & A \arrow[l, "\frac{a}{6}[2\bar{1}\bar{1}]"] \end{tikzcd}$$

< FCC → HCP: phase transformation by stacking fault over the area of glide plane swept by the disl.>

Gliding of Shockley partial dislocations ➡ Stacking fault region

적층결함

In thermodynamically stable FCC lattices, the stacking fault is a region of high free energy.
→gliding of partial dislocations : difficult

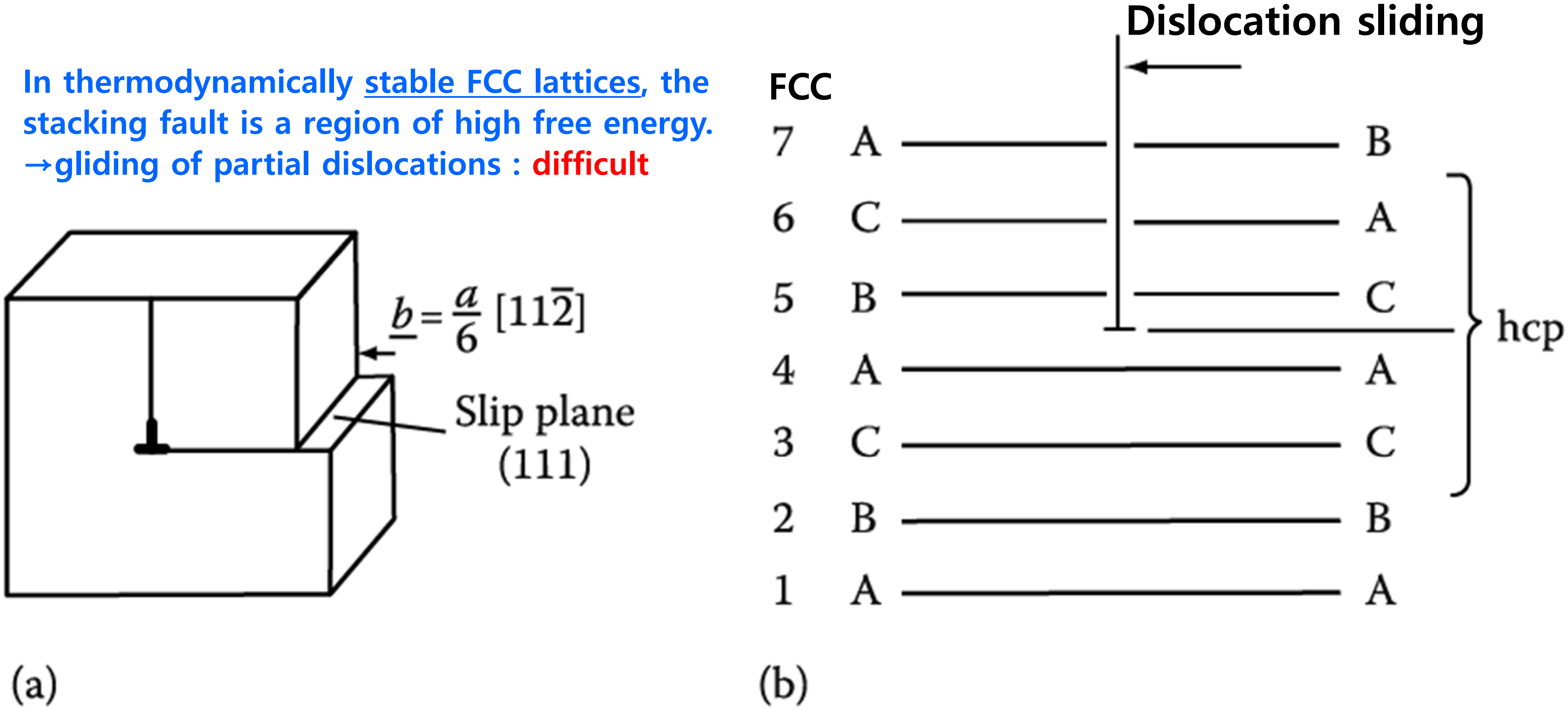
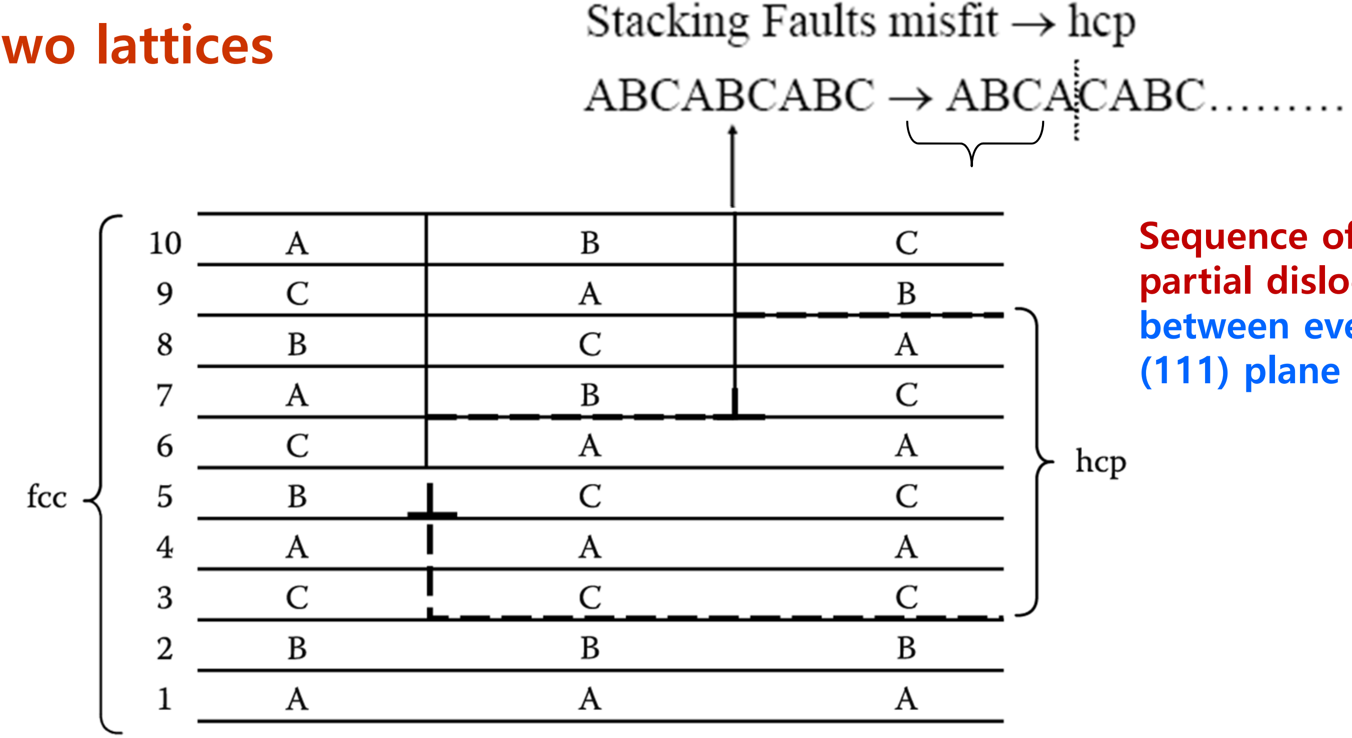


Fig. 3. 59 (a) An edge dislocation with a Burgers vector b = $\frac{a}{6}$[112] on (11$\bar{1}$). (shockley partial dislocation.)
(b) The same dislocation locally changes the stacking sequence from fcc to hcp.

If FCC lattice is only metastable with respect to the HCP structure → stacking fault E ~ effectively negative → gliding of partial dislocation: easy due to decrease the free energy of system

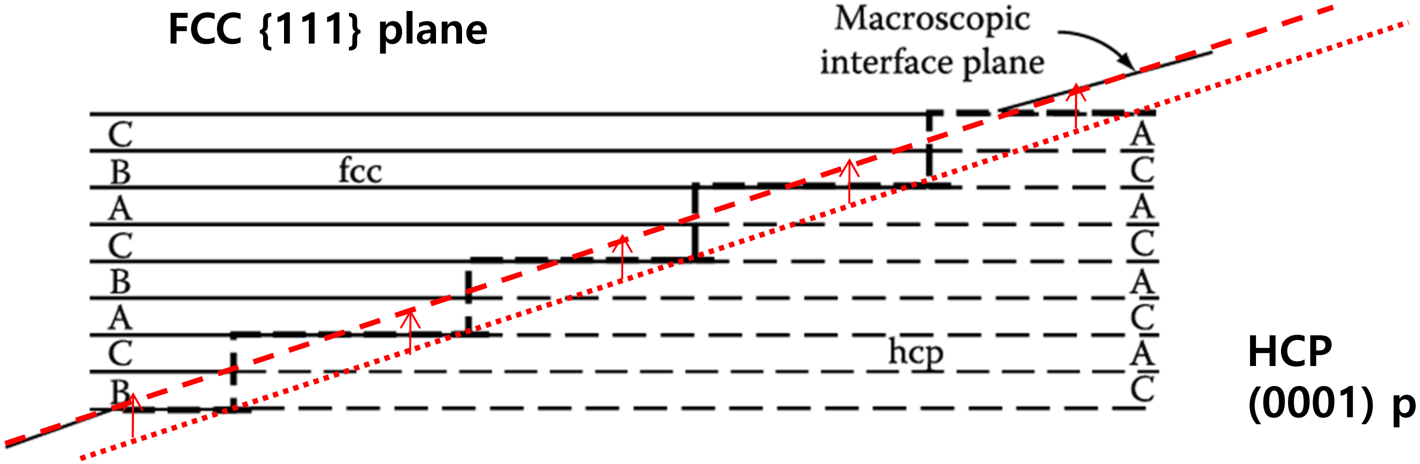
Glissile Interfaces between two lattices



Sequence of Shorkley partial dislocations between every other (111) plane

Fig. 3. 60 Two Shockley partial dislocation on alternate (111) planes create six layers of hcp stacking.

FCC {111} plane



HCP (0001) plane

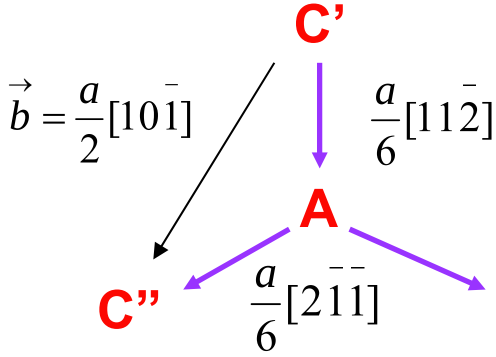
Fig. 3. 61 An array of Shockley partial dislocations forming a glissile interface between fcc and hcp crystals.

An important characteristic of glissile dislocation interfaces
→ they can produce a macroscopic shape change in the crystal.

1) Sequence of same Shorkley partial dislocations between every other (111) plane

→ Pure shear deformation
→ Fcc → Hcp
→ shape change

2) If transformation is achieved using all three partials in equal #s,



→ No overall shape change

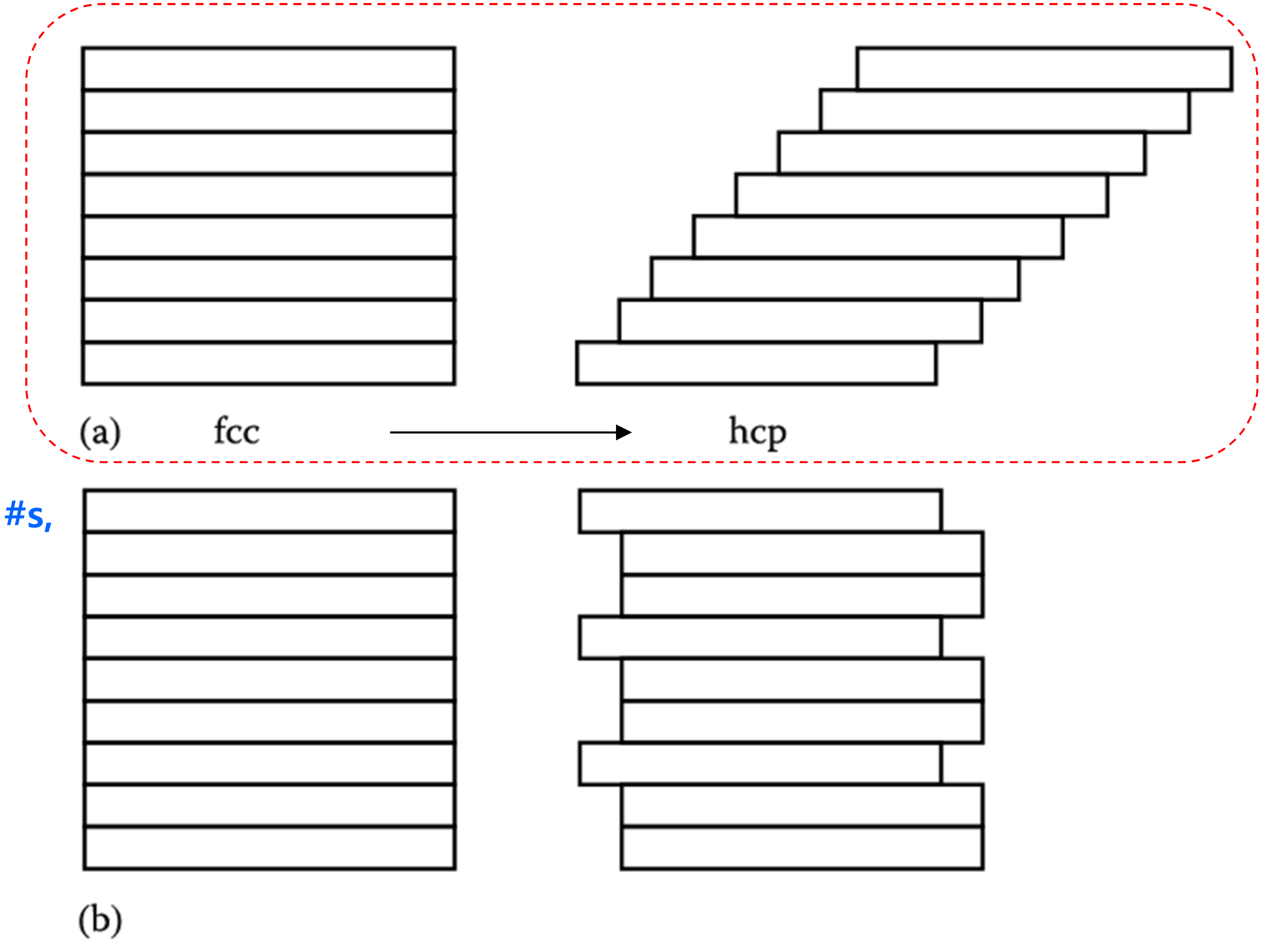


Fig. 3. 62 Schematic representation of the different ways of shearing cubic close-packed planes into hexagonal close-packed (a) Using only one Shockley partial, (b) using equal numbers of all three Shockley partials.

* **Formation of martensite in steel and other alloys:** Motion of Glissile-dislocation interface
: macroscopic shape change & no change in composition
→ more complex interface but same principles (chapter 6)

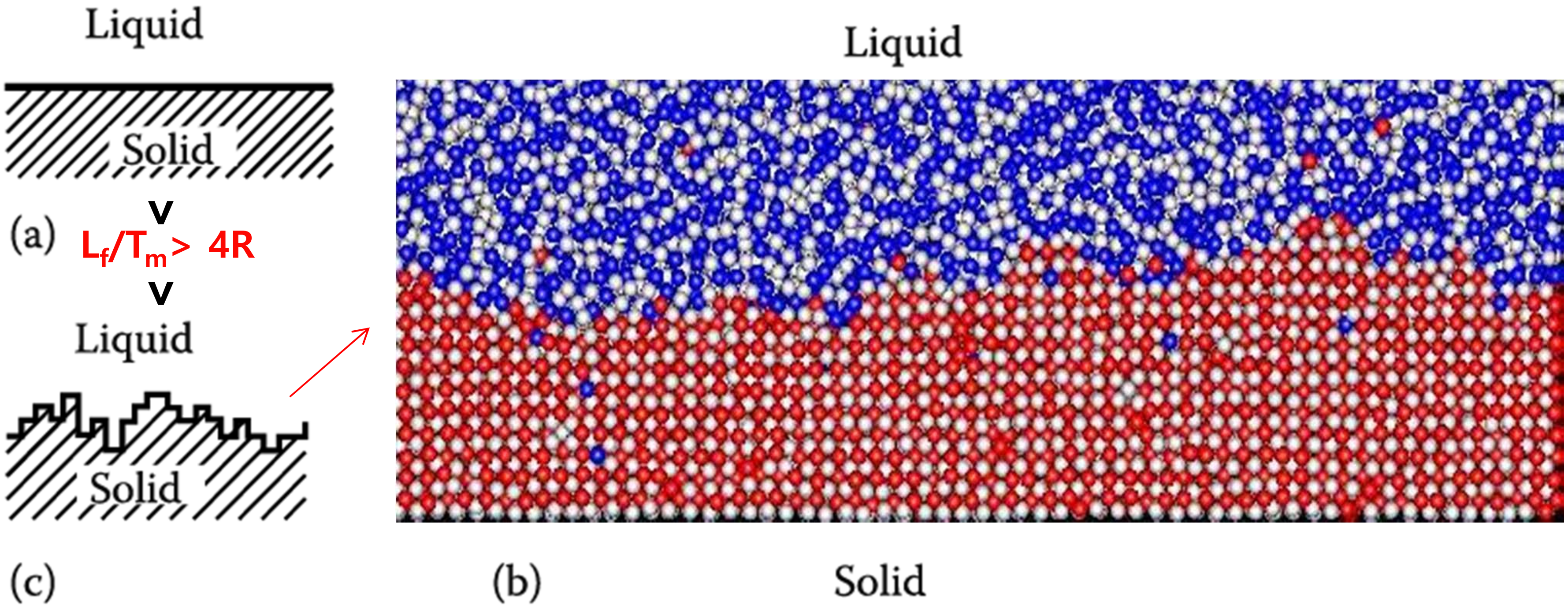
Q: What is the structure of interface between liquids and crystals?

Interphase Interfaces in Solid (α/β)

2) Solid /Liquid Interfaces: consequences for the structure and energy of the interface

Faced interface **Rather narrow transition zone** approximately one atom layer thick
~ same as solid/vapor interfaces, i.e., atomically flat close-packed interface

: some intermetallic compounds, elements such as Si, Ge, Sb, and most nonmetals

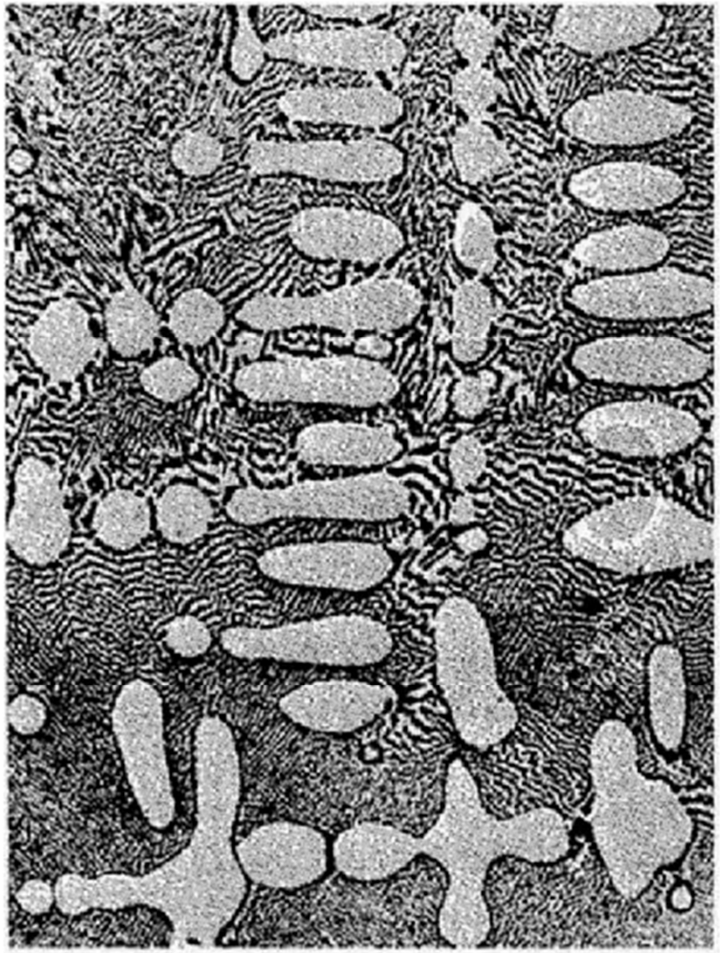


$L_f/T_m > 4R$

Fig. 3. 63 Solid/liquid interfaces: (a) atomically smooth, (b) and (c) atomically rough, or diffuse interfaces.

Diffusion interface (non-faceted) **Rather wide transition zone** over several atom layers
: most metals, L_f/T_m ~ R (gas constant) ~automatically **rough & diffuse interface**

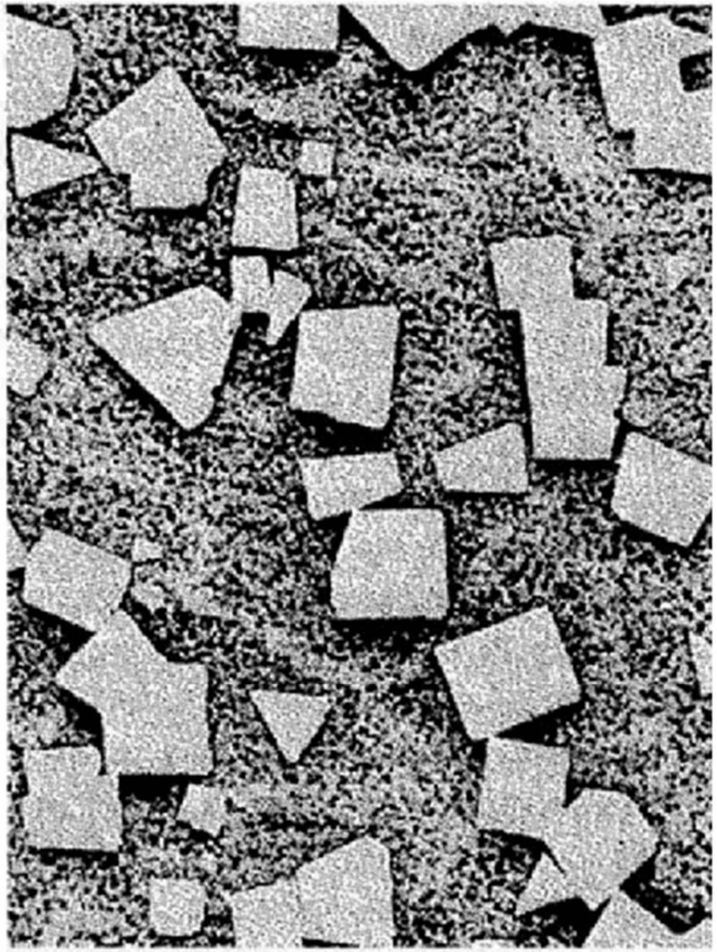
Primary Ag dendrite
in Cu-Ag eutectic matrix



(a) **Non-faceted**

- Free E ~do not vary with crystallographic orientation
- γ-plot ~ spherical

β′-SnSb intermetallic compound
in Sn(Sb) solid solution

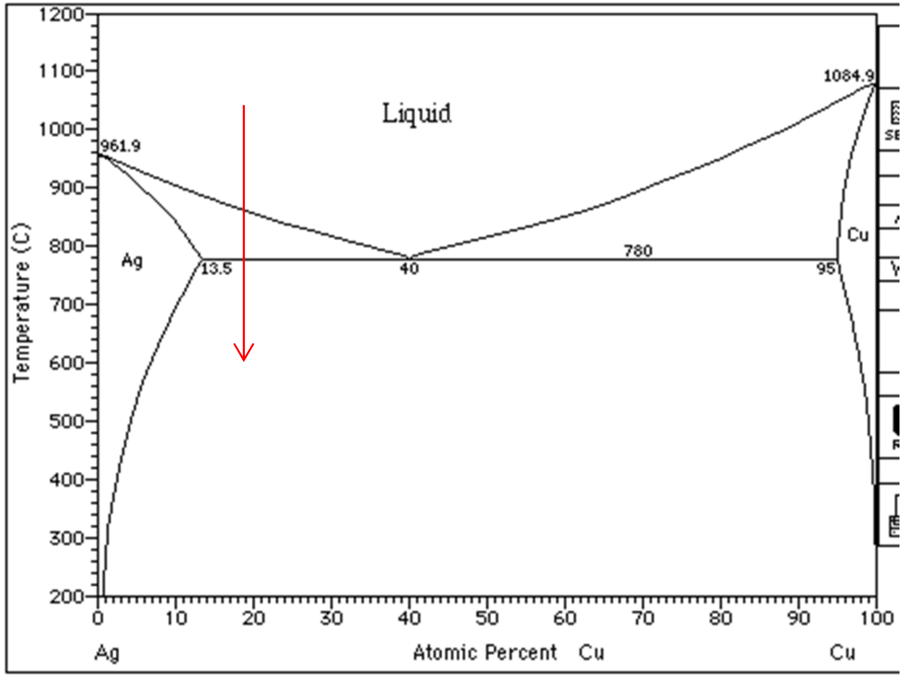


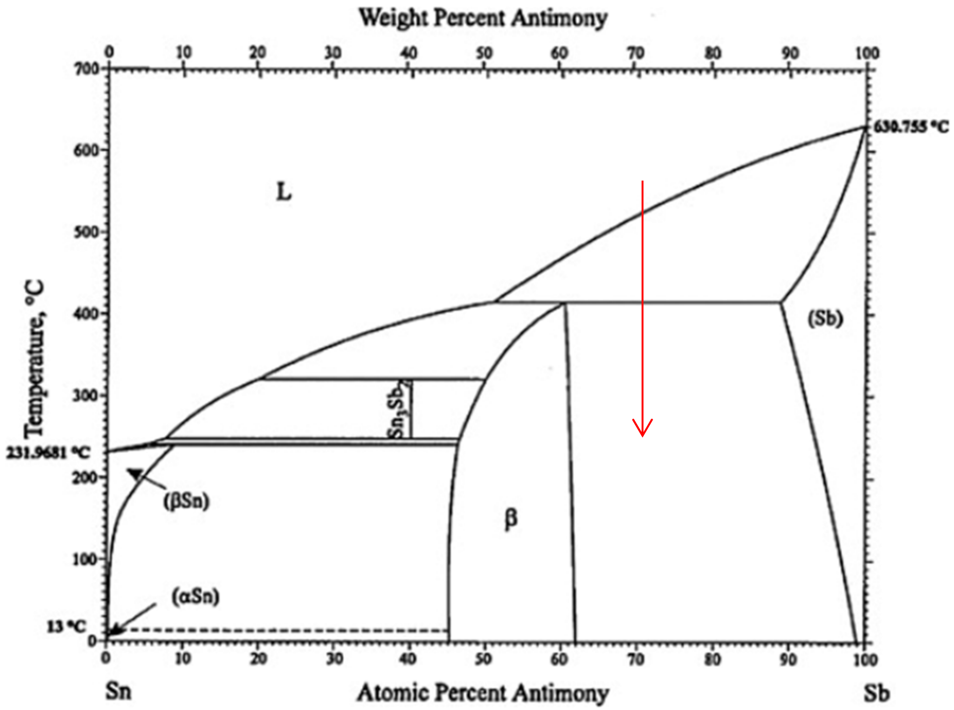
(b) **Faceted**

- Strong crystallographic effects
- Solidify with low-index close-packed facets

Primary Ag dendrite
in Cu-Ag eutectic matrix

β′-SnSb intermetallic compound
in Sn(Sb) solid solution

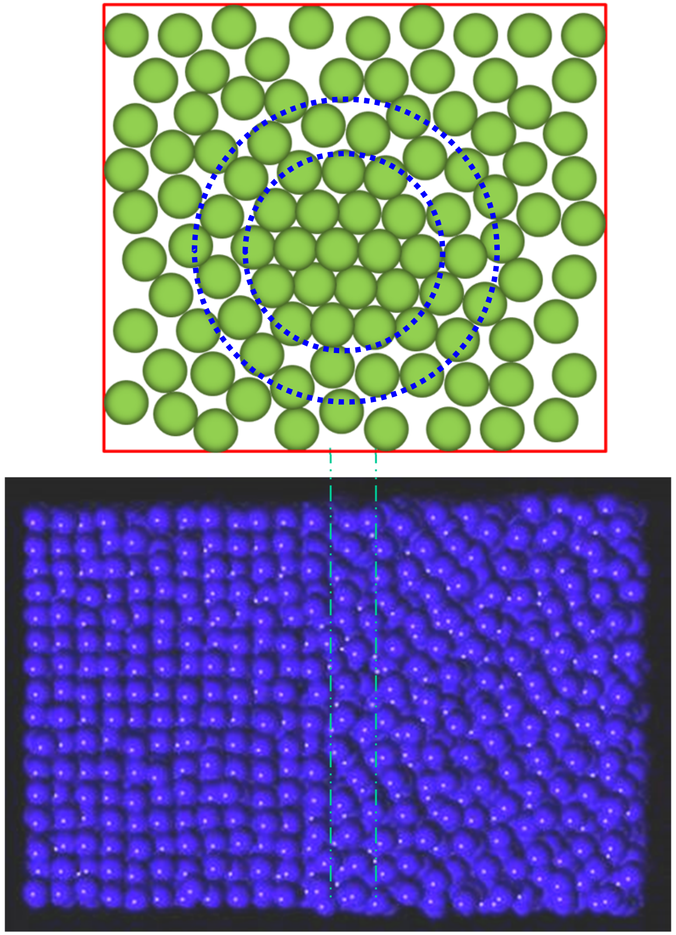


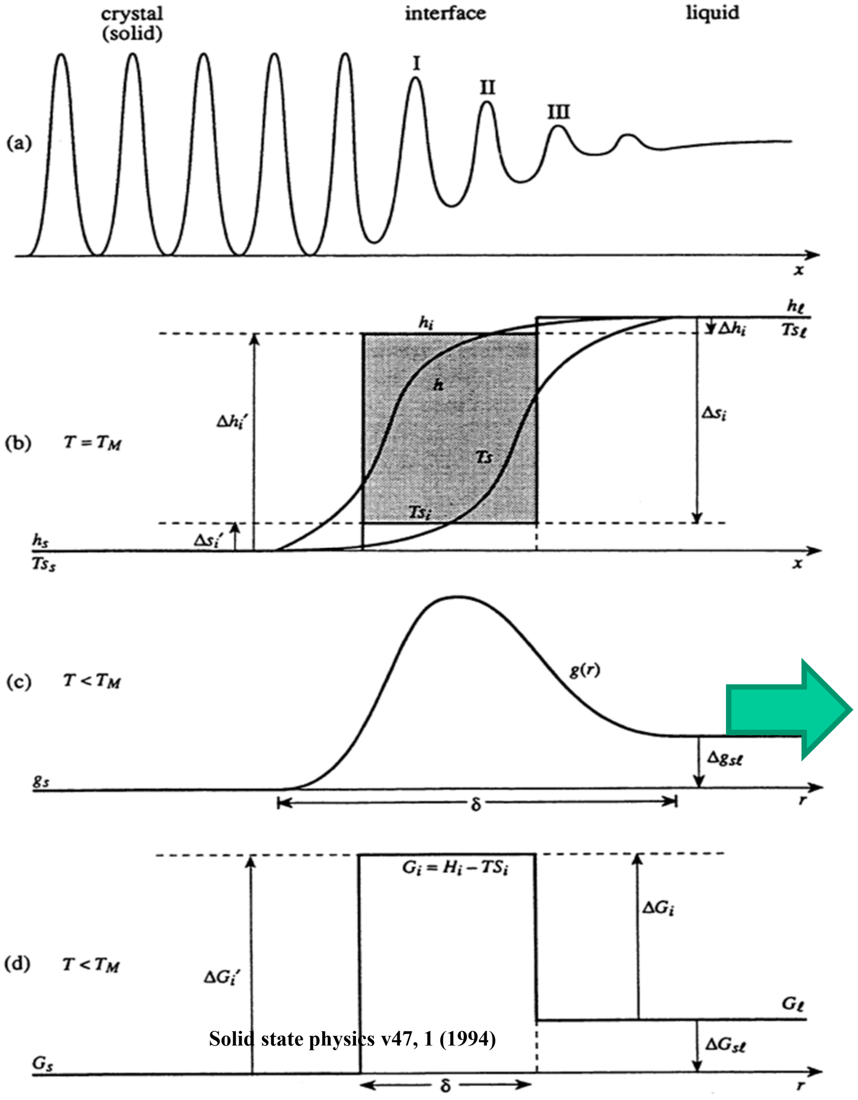


What is the structure of interface between liquids and crystals?

Turnbull’s solid-liquid interface

Turnbull’s insight: Liquid orders substantially near a crystal surface due to entropy decreasing caused by the crystal-melt interfacial tension

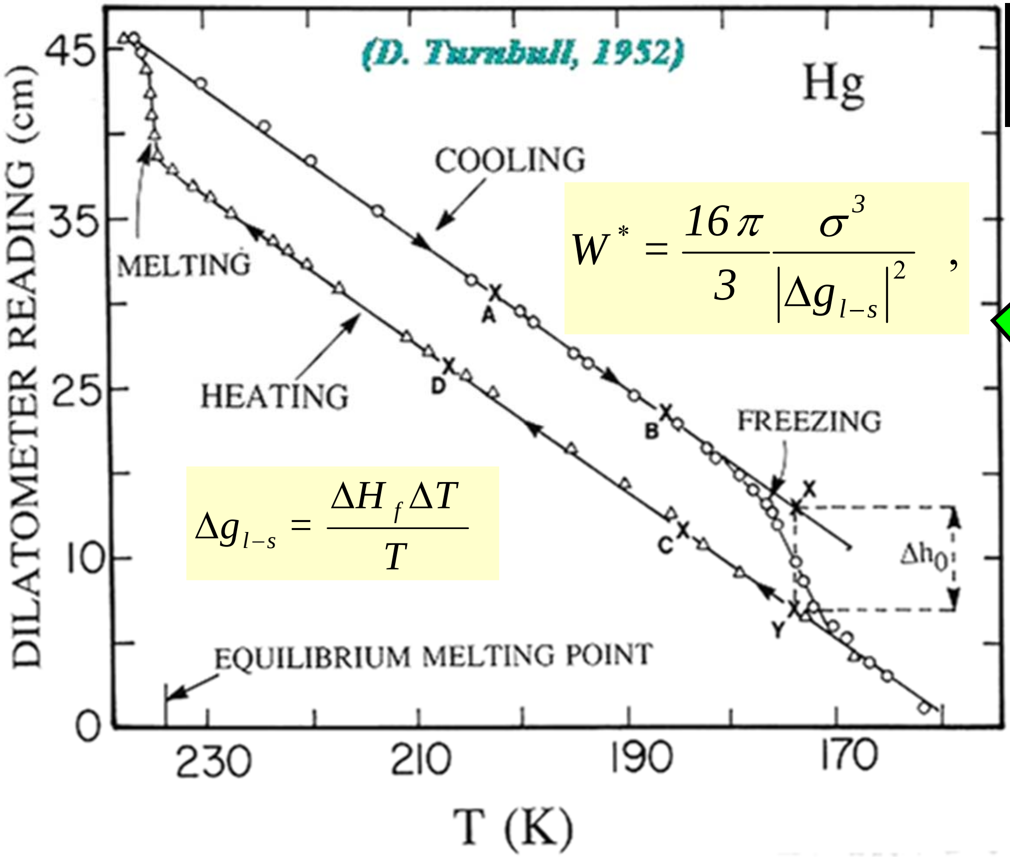




Solid state physics v47, 1 (1994)

Supercooling

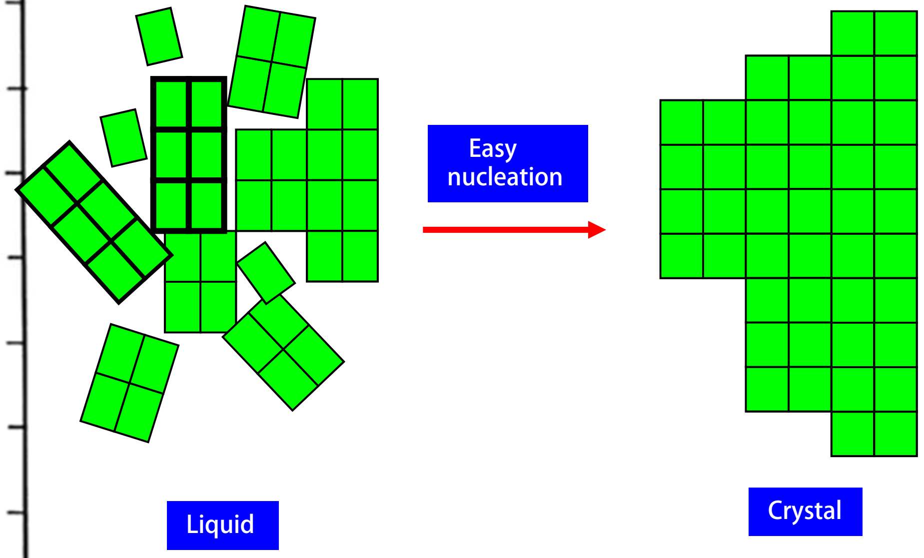
Turnbull's Insight for Supercooling



$$W^* = \frac{16\pi}{3}\frac{\sigma^3}{|\Delta g_{l-s}|^2},$$

$$\Delta g_{l-s} = \frac{\Delta H_f \Delta T}{T}$$

Microcrystalline structure for liquid metals
–*Same short range order with crystals*!



- **How does the liquid metals resist on the crystallization?**
- **What the origin of high nucleation barrier against the crystallization?**

* Broken bond model → calculation of the E of solid/ liquid interface

$0.5L_f / N_a \rightarrow 0.45L_f / N_a$ (엔트로피 효과로 감소)

Table 3.1. Relationship between Maximum Supercooling, Solid-Liquid Interfacial Energy and Heat of Fusion[a]

Metal	Interfacial Energy σ (ergs/cm²)	σ_g (cal/mole)	σ_g/L	ΔT_{MAX} (deg)
Mercury	24.4	296	0.53	77
Gallium	55.9	581	0.44	76
Tin	54.5	720	0.42	118
Bismuth	54.4	825	0.33	90
Lead	33.3	479	0.39	80
Antimony	101	1430	0.30	135
Germanium	181	2120	0.35	227
Silver	126	1240	0.46	227
Gold	132	1320	0.44	230
Copper	177	1360	0.44	236
Manganese	206	1660	0.48	308
Nickel	255	1860	0.44	319
Cobalt	234	1800	0.49	330
Iron	204	1580	0.45	295
Palladium	209	1850	0.45	332
Platinum	240	2140	0.45	370

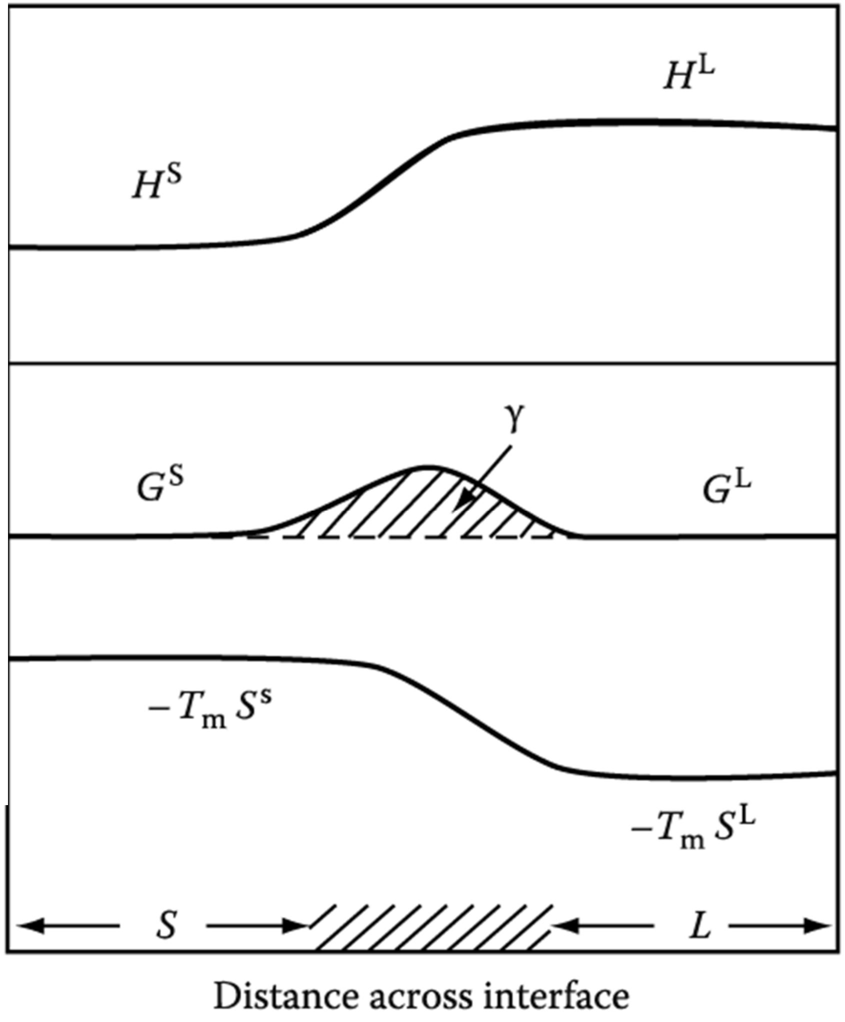
[a] Data from D. Turnbull, *J. Appl. Phys.*, **21**, 1022 (1950) and Ref. 3.

$\gamma_{SL} \approx 0.45\ \gamma_b$ $(= 0.15\gamma_{SV})$

for the most metals

$\gamma_{SV} > \gamma_{SL} + \gamma_{LV}$

at equilibrium melting temp.



Showing the origin of the solid/ liquid interfacial energy, γ

4.1.4. Nucleation of melting

Although nucleation during solidification usually requires some undercooling, melting invariably occurs at the equilibrium melting temperature even at relatively high rates of heating.

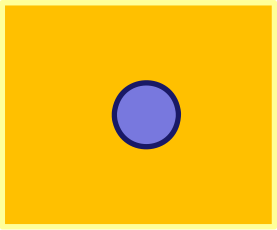
Why?

$$\boxed{\gamma_{SL} + \gamma_{LV} < \gamma_{SV}}$$ (commonly)

In general, wetting angle = 0 ⇨ No superheating required!

Melting and Crystallization are Thermodynamic Transitions

Solidification: Liquid → Solid



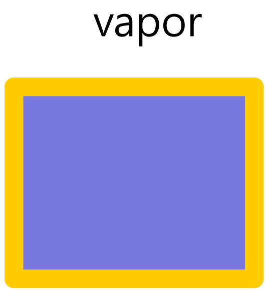
<Thermodynamic>

- Interfacial energy ⇨ ΔT_N

No superheating required!

- Interfacial energy ⇨ No ΔT_N

$$\gamma_{SL} + \gamma_{LV} < \gamma_{SV}$$



Melting: Liquid ← Solid

Q: What is the role of interface migration on phase transformation ?

Interface controlled growth ⟺ Diffusion controlled growth

3.5. Interface Migration

Phase transformation = Interface creation & Migration

Heterogeneous Transformation (general): parent and product phases during trans.

Nucleation (interface creation) + **Growth** (interface migration)

Nucleation barrier **Ex. Precipitation**

- at certain sites within metastable alpha phase → new beta phase = Nucleation
- most of transformation product is formed during the growth stage by the transfer of atoms across the moving parent/product interface.

Homogeneous Transformation: PT occurs homogeneously throughout the parent phase.

Growth-interface control

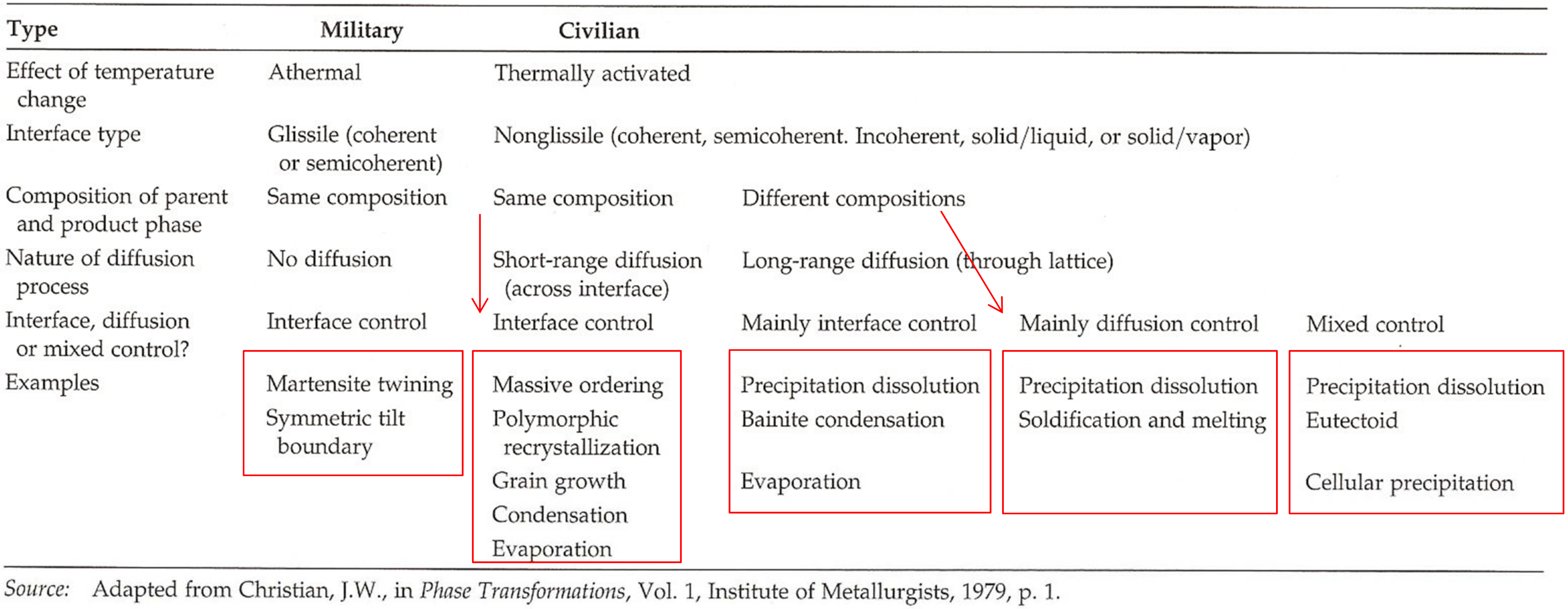
No Nucleation barrier **Ex. Spinodal decomposition (Chapter 5)**

Order-disorder transformation

*** Types of Interface** → **Types of transformation**

- **Glissile Interface: Athermal, Shape change** → **Military transformation**
  Dislocation gliding
- **Non-Glissile Interface: Thermal,** → **Civilian transformation**
  Random jump of individual atoms: extremely sensitive to temp.
  ~ similar way to the migration of a random high angle GB

Classification of Nucleation and Growth Transformation



Type	Military	Civilian			
Effect of temperature change	Athermal	Thermally activated			
Interface type	Glissile (coherent or semicoherent)	Nonglissile (coherent, semicoherent. Incoherent, solid/liquid, or solid/vapor)			
Composition of parent and product phase	Same composition	Same composition	Different compositions		
Nature of diffusion process	No diffusion	Short-range diffusion (across interface)	Long-range diffusion (through lattice)		
Interface, diffusion or mixed control?	Interface control	Interface control	Mainly interface control	Mainly diffusion control	Mixed control
Examples	Martensite twining Symmetric tilt boundary	Massive ordering Polymorphic recrystallization Grain growth Condensation Evaporation	Precipitation dissolution Bainite condensation Evaporation	Precipitation dissolution Soldification and melting	Precipitation dissolution Eutectoid Cellular precipitation

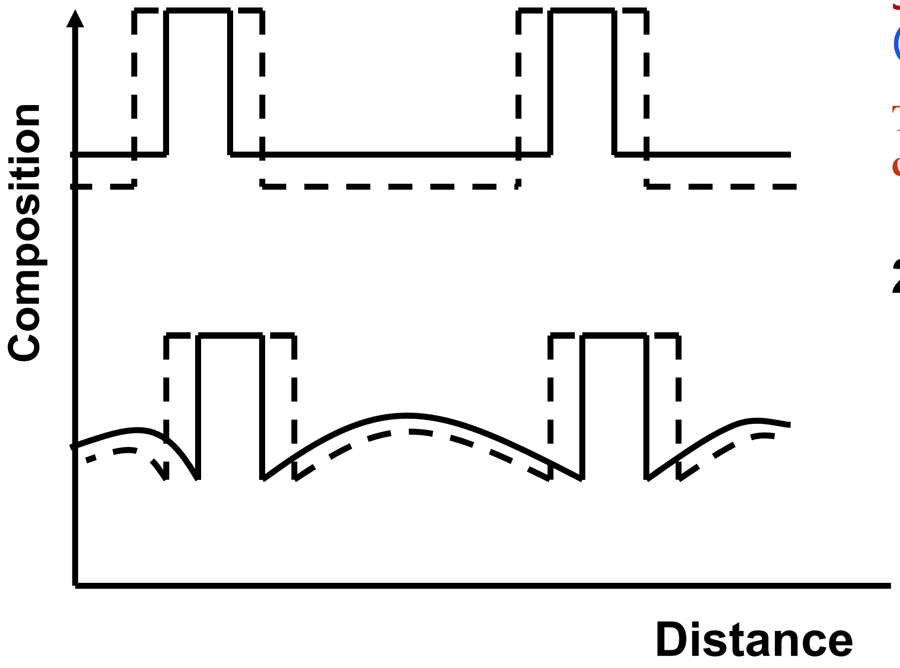
Source: Adapted from Christian, J.W., in *Phase Transformations*, Vol. 1, Institute of Metallurgists, 1979, p. 1.

exception) bainite transformation: thermally activated growth/ shape change similar to that product by the motion of a glissile interface

(need to additional research)

3.5. Interface Migration

* Civilian Transformation



1) Interface control

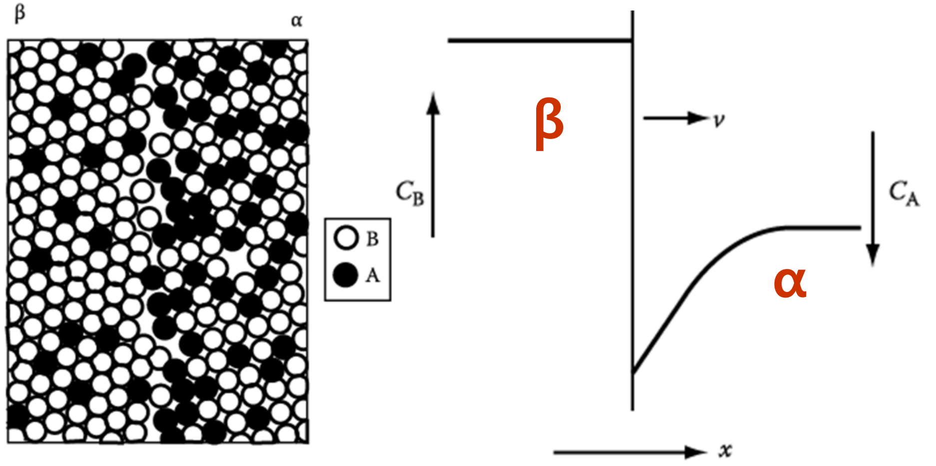
Same composition between parent and product
(Ex. α→γ transformation in pure iron)

The new phase will be able to grow as fast as the atoms can cross the interface. : interfacial reaction velocity dominant

2) Diffusion control

Different composition between parent and product
(Ex. The growth of the B-rich phase into the A-rich α-phase)

Growth of the new phase will require long-range diffusion
Growth rate: governed by the rate at which lattice diffusion can remove the excess atoms from ahead of the interface.



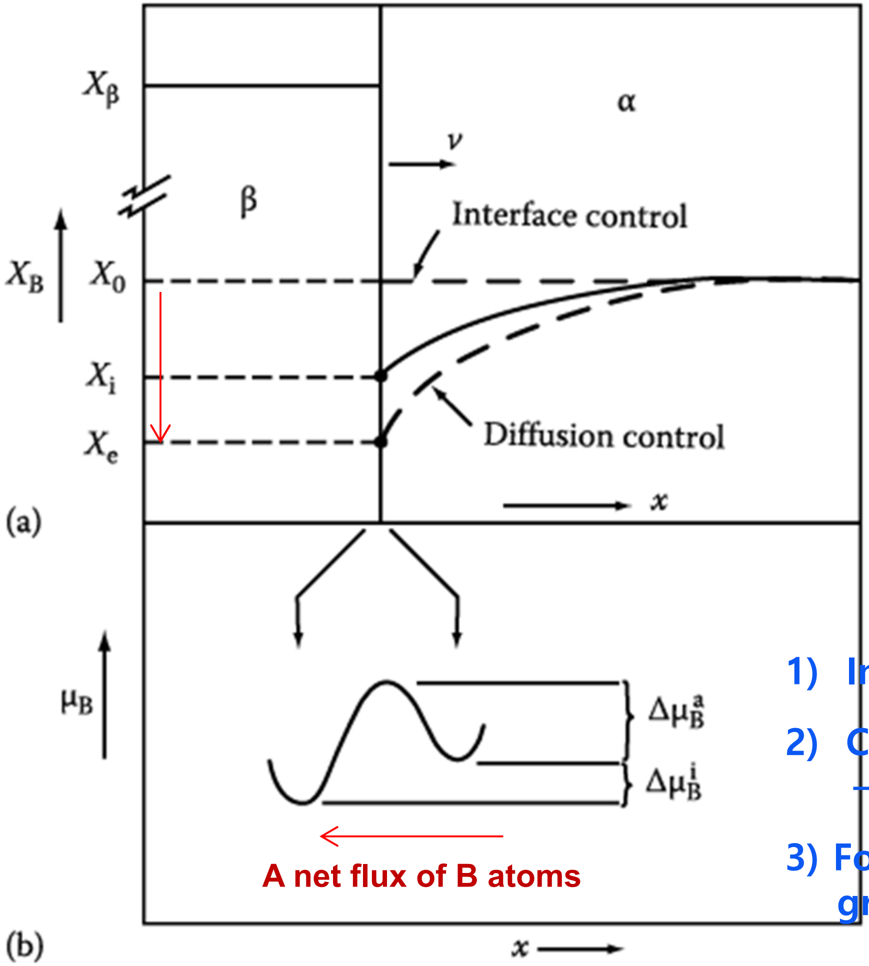
3) Mixed control: interface reaction = diffusion process

3.5.1. Diffusion-Controlled and Interface-Controlled Growth

Fig. 3.67 Interface migration with long-range diffusion

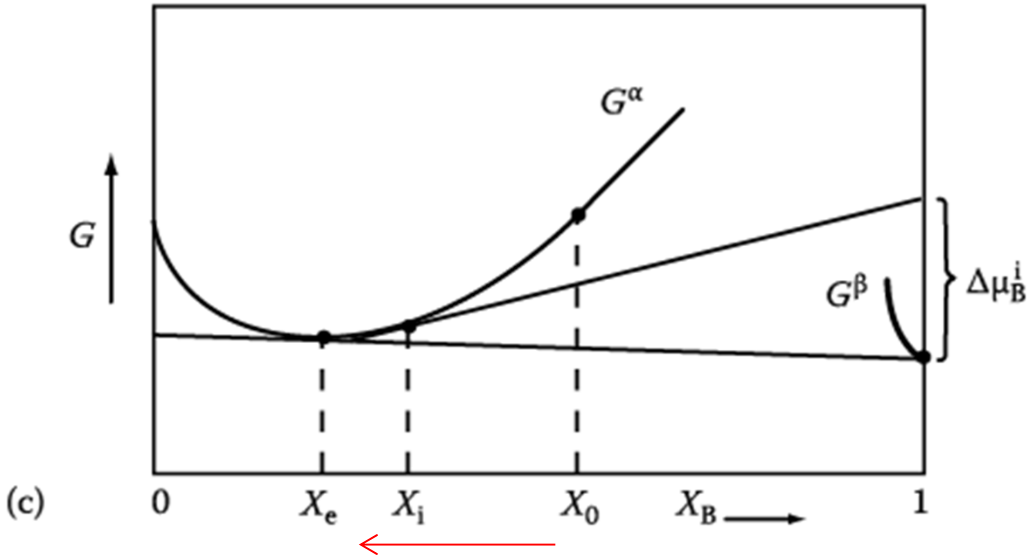
; <u>the migration of interface separating two phases of different composition</u>

(a) Composition profile across the interface



(b) The origin of the driving force for boundary migration into the α-phase $\Delta\mu_B{}^i$

(c) A schematic molar free energy diagram showing the relationship between $\Delta\mu_B{}^i$, X_i, X_0



1) Initial composition of A-rich α phase X_0

2) Concentration B in the α phase adjacent to the inteface X_i
→ B concentration in α : $X_0 \rightarrow X_i$

3) For growth to occur the interface, composition must be greater than the equilibrium concentration X_e.

Total Free Energy Decrease per Mole of Nuclei ΔG_0

: Driving force for phase transformation of system

⇕

Driving Force for Precipitate Nucleation $\boxed{\alpha \rightarrow \alpha + \beta}$ ΔG_V

$$\Delta G_1 = \mu_A^\alpha X_A^\beta + \mu_B^\alpha X_B^\beta$$

: Decrease of total free E of system
by removing a small amount of material
with the nucleus composition (X_B^β) (P point)

$$\Delta G_2 = \mu_A^\beta X_A^\beta + \mu_B^\beta X_B^\beta$$

: Increase of total free E of system
by forming β phase with composition X_B^β
(Q point)

$$\boxed{\Delta G_n = \Delta G_2 - \Delta G_1}$$ (length PQ)

$$\boxed{\Delta G_V} = \frac{\Delta G_n}{V_m} \quad \text{per unit volume of } \beta$$

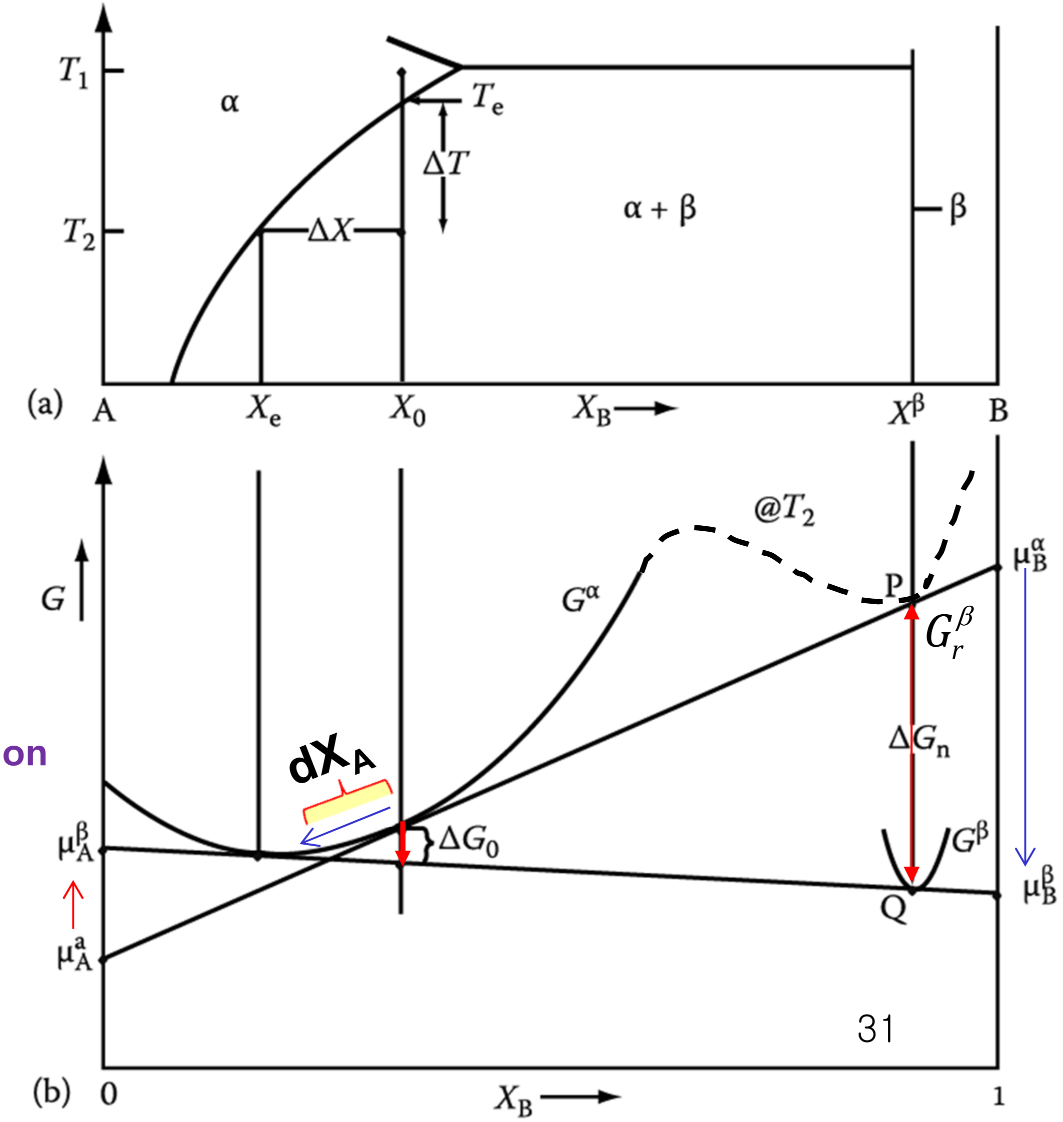
: driving force for β precipitation

For dilute solutions,

$$\Delta G_V \propto \Delta X \quad \text{where } \Delta X = X_0 - X_e$$

$$\boxed{\Delta G_V \propto \Delta X \propto \Delta T}$$

∝undercooling below T_e



3.5.1. Diffusion-Controlled and Interface-Controlled Growth

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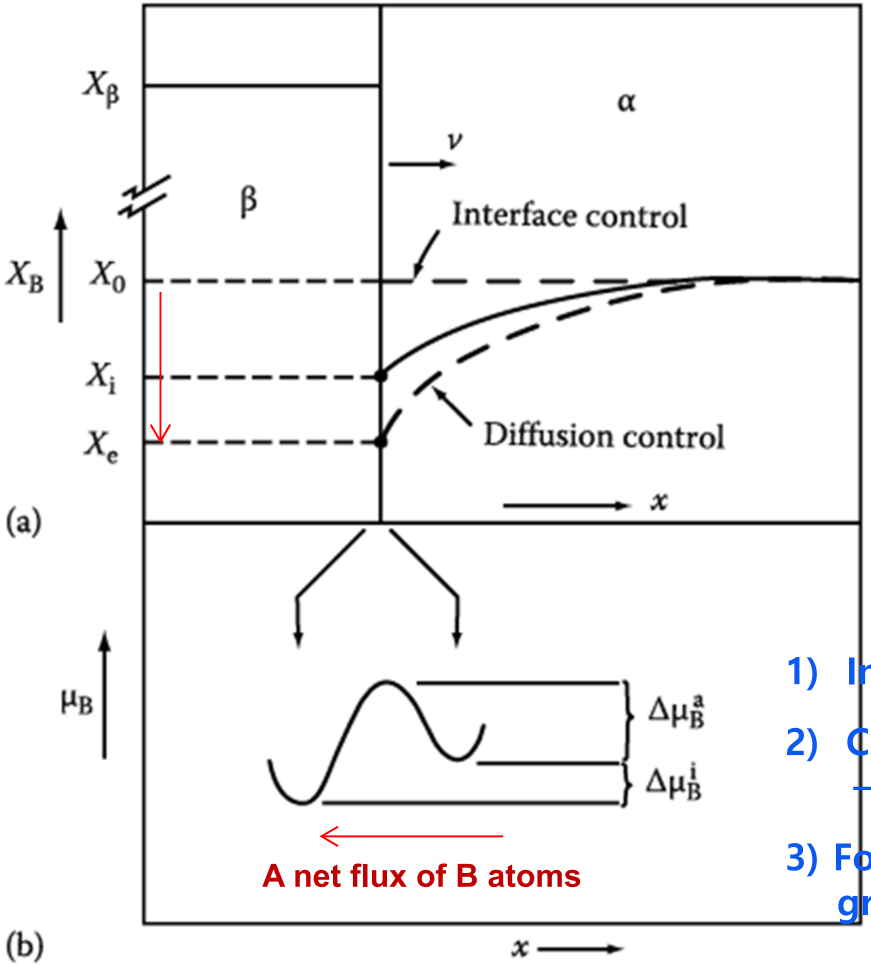
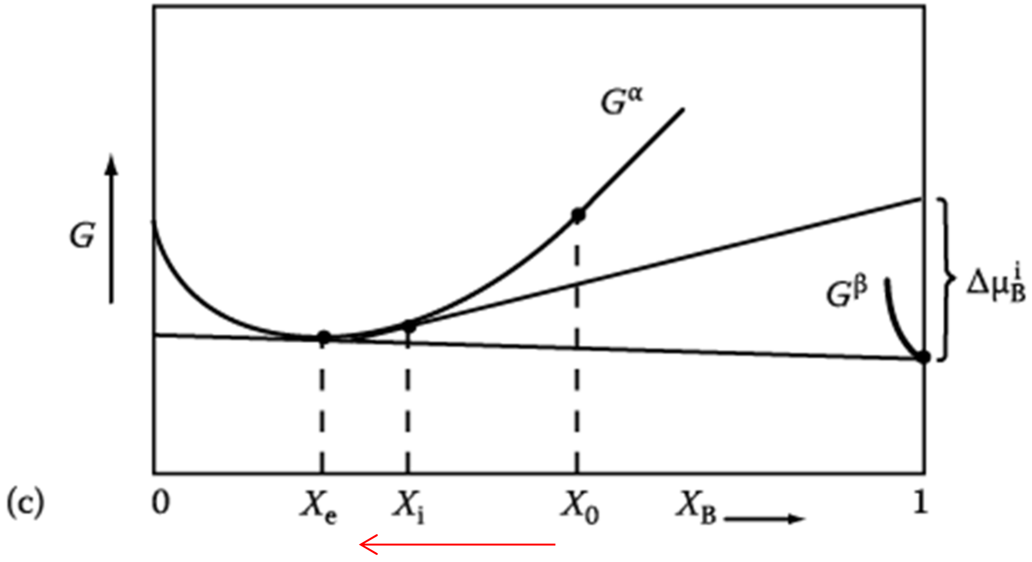


A net flux of B atoms

(b) The origin of the driving force for boundary migration into the α-phase $\Delta\mu_B{}^i$

(c) A schematic molar free energy diagram showing the relationship between $\Delta\mu_B{}^i$, X_i , X_0



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2) Concentration B in the α phase adjacent to the inteface X_i
→ B concentration in α : $X_0 \rightarrow X_i$

3) For growth to occur the interface, composition must be greater than the equilibrium concentration X_e.

By analogy with the migration of a high-angle GB (section 3.3.4), the net flux B across the interface will produce an interface velocity *v* given by

- **(section 3.3.4) Kinetics of grain growth**
  - **Grain boundary migration (*v*) by thermally activated atomic jump**

Boundary velocity

$$\boxed{v = \boxed{\frac{A_2 n_1 v_1 V_m^{\ 2}}{N_a RT}\exp\left(-\frac{\Delta G^a}{RT}\right)}\frac{\Delta G}{V_m}}$$

$v \sim \Delta G/V_m$ driving force

→ F = $\Delta G/V_m$

M : mobility = velocity under unit driving force ~ exp (-1/T)

rate of grain growth $dD/dt \sim 1/D$, exponentially increase with *T*

- Interface velocity of precipitate

$$v = M\Delta\mu_B^i / V_m$$

M **= interface mobility,**
V_m = molar volume of the β phase

- **Corresponding flux across the interface** (negative sign_negative direction of flux along the x-axis)

β ⟵ α

$$\boxed{J_B^i} = -C_B^i v_B$$

$$= -\frac{1}{V_m}\left(M\cdot\Delta\mu_B^i\cdot\frac{1}{V_m}\right)$$

$$= \boxed{J_B^\alpha} = -D\left(\frac{\partial C_B}{\partial X}\right)_{\text{interface}}$$

$$\Delta\mu_B^i\frac{1}{V_m} = \frac{Jmol^{-1}}{m^3 mol^{-1}}$$

Steady state at interface,

$$\boxed{J_B^i = J_B^\alpha}$$

A flux of B atoms toward the interface by the concentration gradient in the α phase

Corresponding flux across the interface (negative sign_negative direction of flux along the x-axis)

$$\boxed{J_B^i} = -C_B^i v_B$$

$$= -\frac{1}{V_m}(M \cdot \Delta\mu_B^i \cdot \frac{1}{V_m})$$

$$= \boxed{J_B^\alpha} = -D(\frac{\partial C_B}{\partial X})_{\text{interface}}$$

$$\Delta\mu_B^i \frac{1}{V_m} = \frac{Jmol^{-1}}{m^3 mol^{-1}}$$

Steady state at interface,

$$\boxed{J_B^i = J_B^\alpha}$$

A flux of B atoms toward the interface by the concentration gradient in the α phase

1) If the interface mobility is very high, e.g. an incoherent interface,

High Mobility: $\Delta\mu_B^i$ small, $X_i \approx X_e$, $\boxed{\text{Diffusion-Control}}$

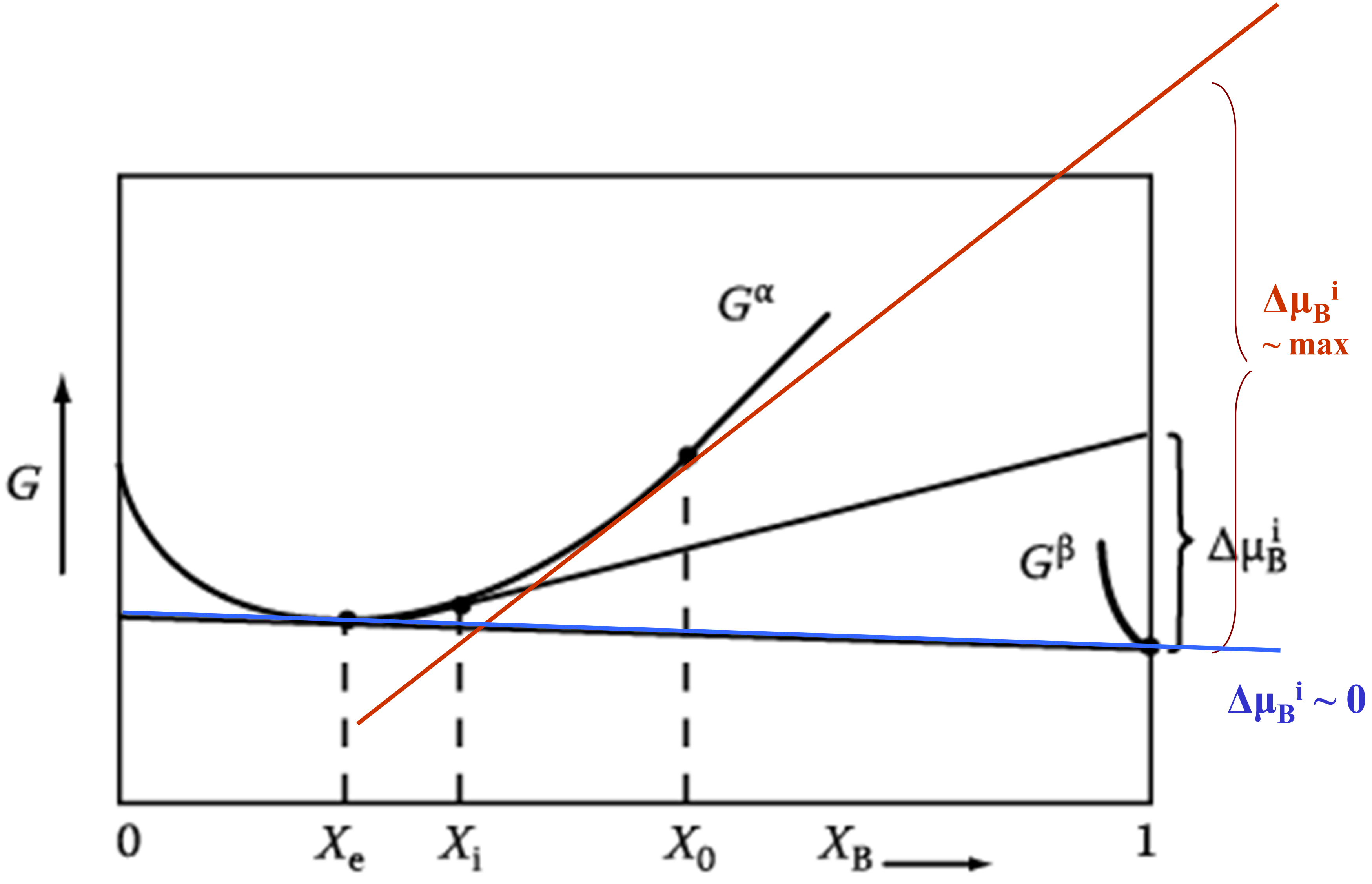
2) When the interface has a low mobility,

Low Mobility: $\Delta\mu_B^i$ must be large, $\boxed{\text{Mixed-Control}}$

3) In the limit of a very low mobility,

Very low Mobility: $Xi \approx X_o$, $\left(\frac{\partial C}{\partial X}\right)_{\text{intrface}} \approx 0$, $\boxed{\text{Interface-Control}}$

$\Delta\mu_B{}^i$ 는 최대


G
G^{α}
G^{β}
$\Delta\mu_B^i$
$\Delta\mu_B^i$ ~ max
$\Delta\mu_B^i \sim 0$
0
X_e
X_i
X_0
X_B
1

When two phases have a different composition,

In general,

the necessary long-range diffusion involves a great many atom jumps

while interface reaction essentially involves only one jump.

All interface reactions should be very rapid in comparison to lattice diffusion, i.e.,

all growth should be diffusion controlled. (next page)

Ex) Diffusion control

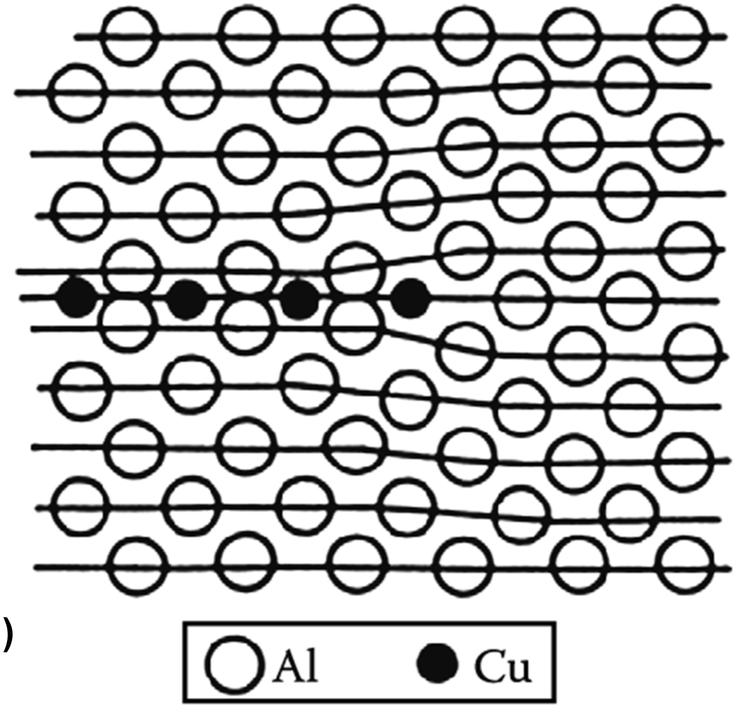
Al-Cu ppt structures

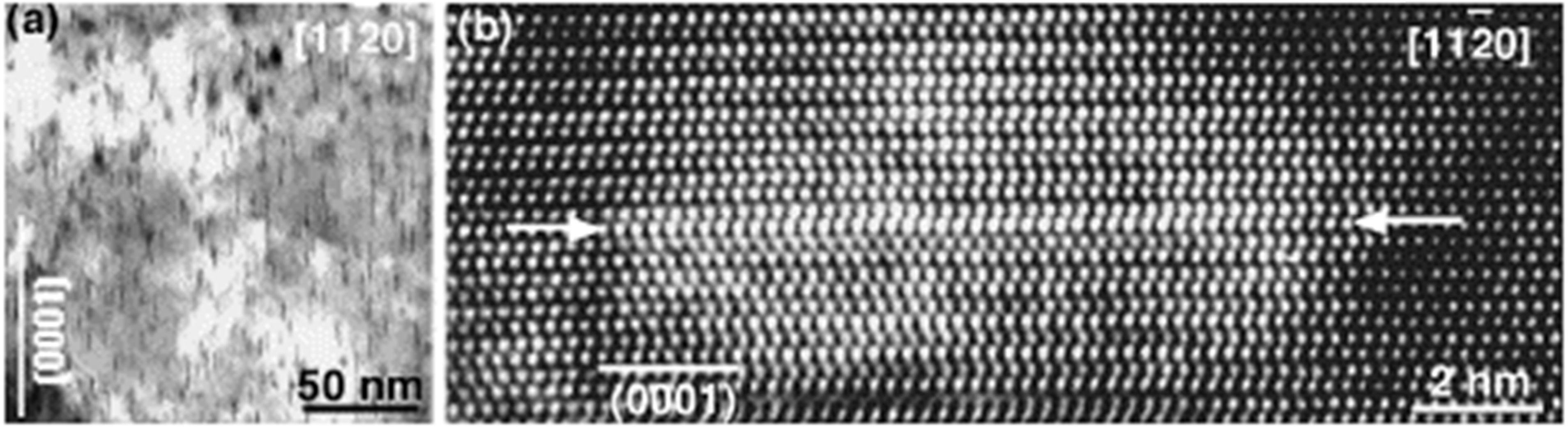
If two phases with different compositions, but the same crystal structure are separated by a coherent interface

Interface can advance by normal lattice diffusion involving vacancies.

No need for a separated interface reaction

Ex) GP zones/ semicoherent interface with misfit dislocation (vacancy creation and annihilation)





(a) Bright-field TEM image showing G.P. zones, and (b) HRTEM image of a G.P. zone formed on a single $(0\ 0\ 0\ 1)_{\alpha}$ plane. Electron beam is parallel to in both (a) and (b).

When two phases have a different composition,

In general,
the necessary long-range diffusion involves a great many atom jumps
while interface reaction essentially involves only one jump.

All interface reactions should be very rapid in comparison to lattice diffusion, i.e.,
all growth should be diffusion controlled. (next page)

➡ In many cases ~ valid, but under certain conditions ~ insufficient

Accommodation factor (A_수용인자)

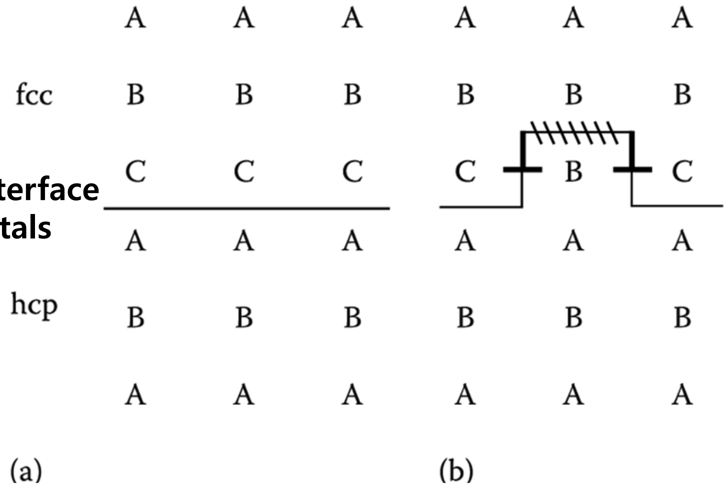
Probability that an atom crossing the boundary will accommodated on arrival at the new phase

Incoherent interfaces and diffuse interface solid/liquid interfaces,
as high-angle grain boundaries = value of A close to unity ➡ "diffusion control"

Coherent or semicoherent interfaces as well as
smooth solid/liquid interfaces = low values of A ➡ "Some degree of interface control"

Ex) Interface control

1) When two phases forming a coherent or semicoherent interface have different crystal structures.



Coherent close-packed interface between fcc and hcp crystals

A loop of Shorkley partial disl.

~ high E/ unstable configuration
→ force back to its original position

Fig. 3. 68 Problems associated with the continuous growth of coherent interfaces between phases with different crystal structures.

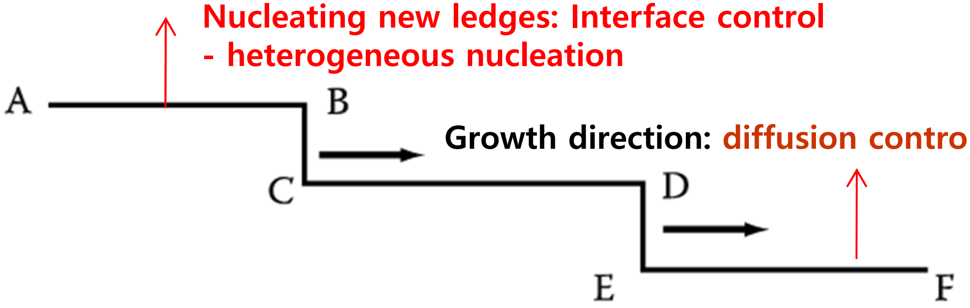
hcp phase growth by individual atomic jumps (i.e., so-called continuous growth) ~ very difficult
(very low accommodation factors and low mobility)

2) Solid/vapor as well as smooth solid/liquid interfaces ~ similar manner

*** A way of avoiding the difficulties of continuous growth**

"Growth by ledge mechanism"

Facets: AB, CD, EF
Ledge: BC, DE



Nucleating new ledges: Interface control
- heterogeneous nucleation

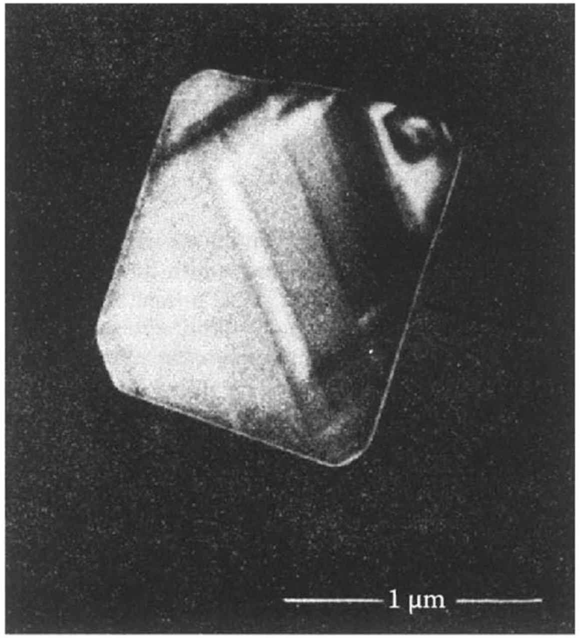
Growth direction: diffusion control

Fig. 3. 69 The ledge mechanism.

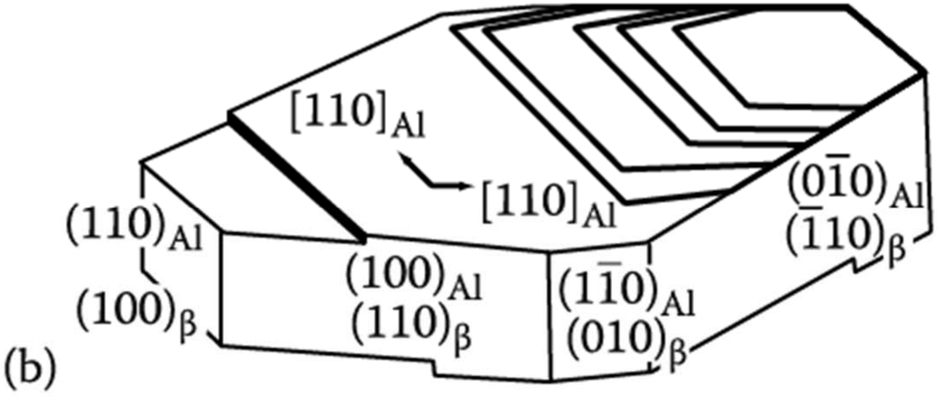
→ Problem of nucleation new ledges may often lead to a degree of interface control on the overall rate.

Fig. 3. 70 (a) Growth ledges at an Mg_2Si plate in Al-1.5 wt% Mg_2Si, solution treated and aged 2h at 350 ℃. Dark field micrograph.

(b) Schematic diagram of (a) showing ledges on Mg_2Si plate.



(a)



Note that growth ledges are usually hundreds of atoms layers high.

성장 돌출맥의 층 두께는 보통 수백 원자층 두께

The mechanism of interface migration can have important effects on the shape of second-phase inclusions. (section 3.4.2)

- **if absence of strain E effect, equilibrium shape of a precipitate should be determined by**
  - **a) the relative energies of the bounding interfaces (dominant)** **γ-plot** 계면 E의 상대적 비
    - **ex) a partially coherent precipitate~disk or plate shape with an aspect ratio of $\gamma i / \gamma c$**
  - **b) (in practice) "relative rates" at which the coherent and incoherent interface can migrate**

정합/ 부정합 계면의 상대적인 이동속도 차에 의해 형상 변화

*** Homework 3 : Exercises 3 (pages 186-188)**

until 13st November (before class)

Good Luck!!

Contents in Phase Transformation

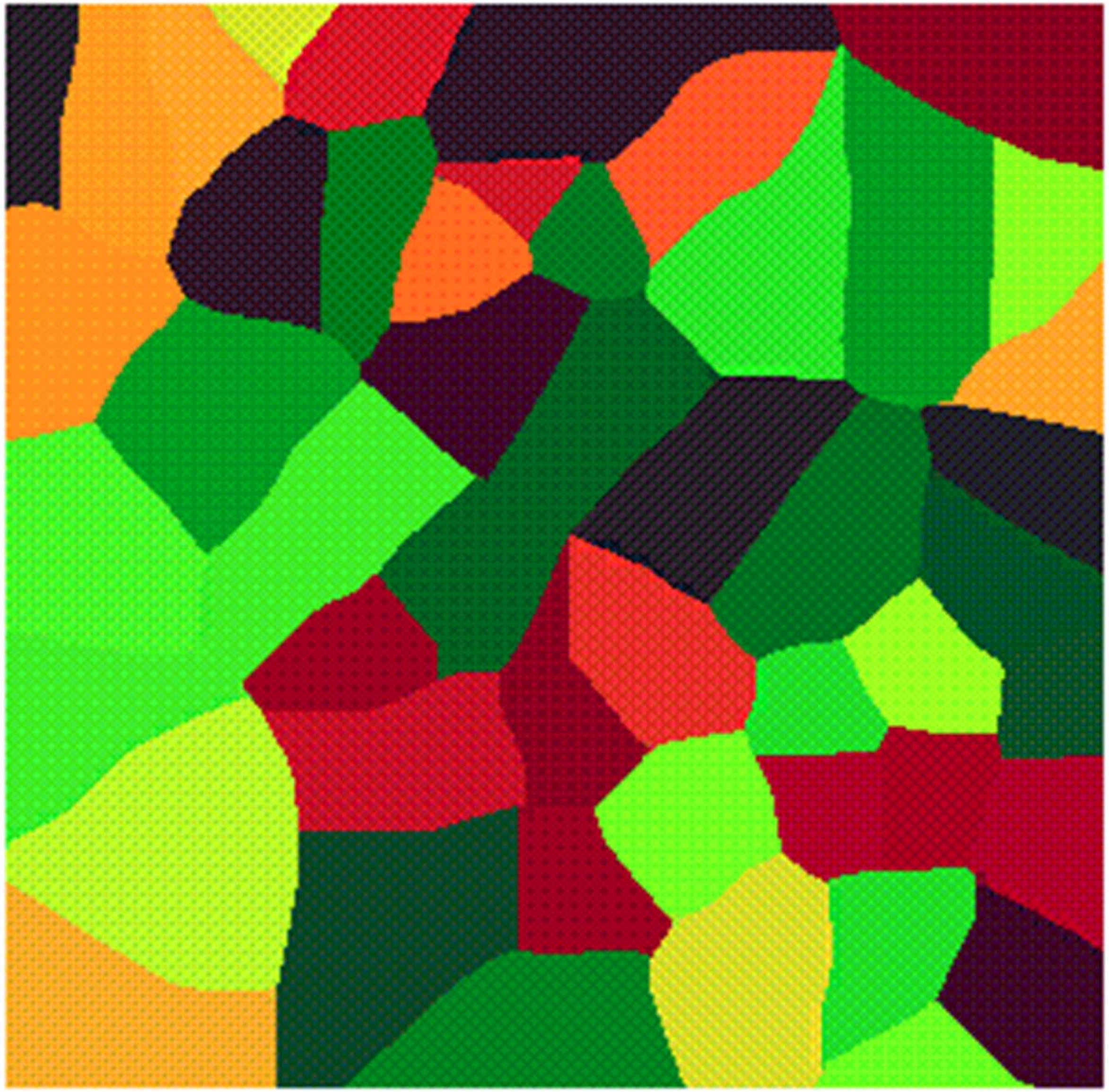
Background to understand phase transformation

- (Ch1) Thermodynamics and Phase Diagrams
- (Ch2) Diffusion: Kinetics
- (Ch3) Crystal Interface and Microstructure

Representative Phase transformation

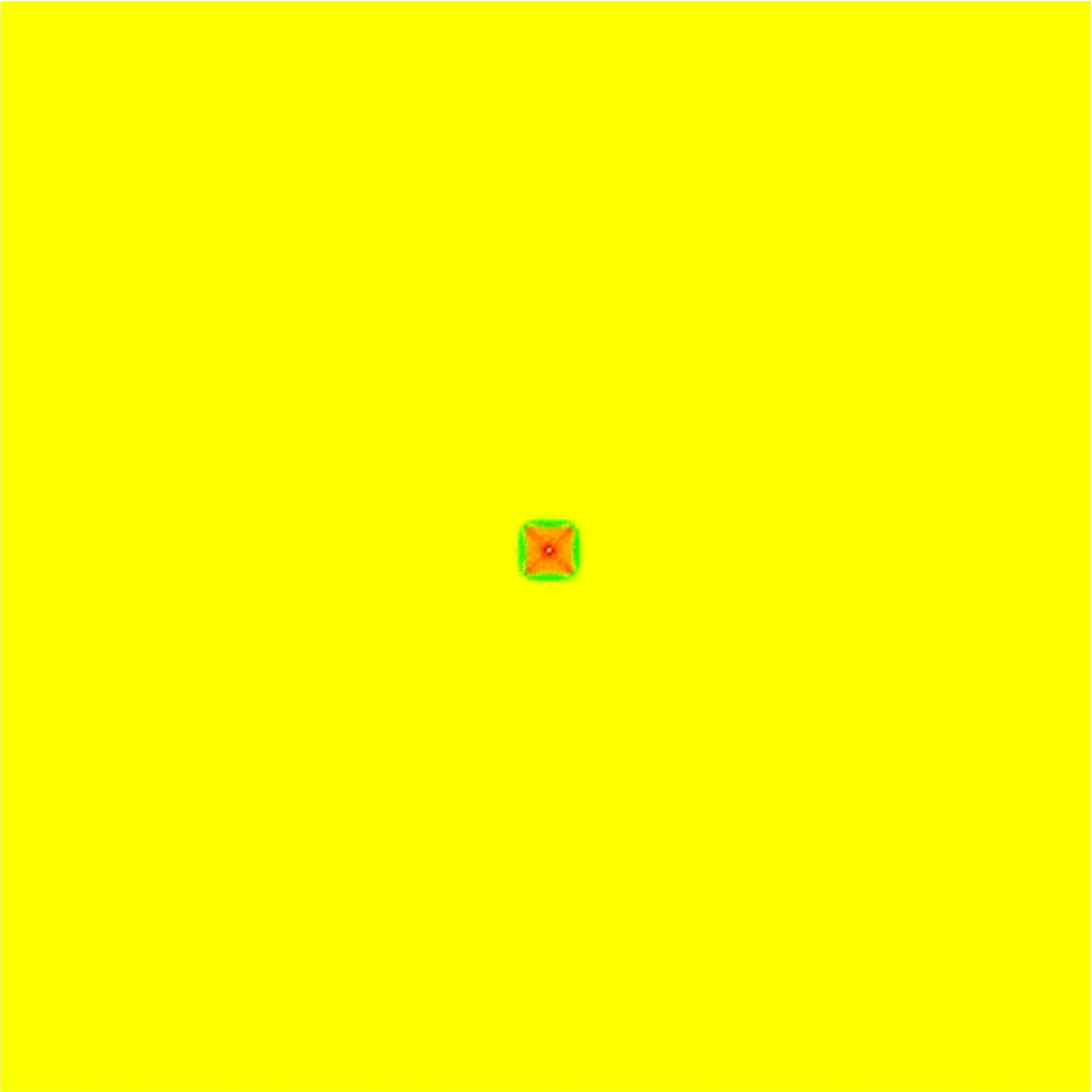
- (Ch4) Solidification: Liquid → Solid
- (Ch5) Diffusional Transformations in Solid: Solid → Solid
- (Ch6) Diffusionless Transformations: Solid → Solid

Solidification: Liquid → Solid



4 Fold Anisotropic Surface Energy/2 Fold Kinetics, Many Seeds

Solidification: Liquid ⟶ Solid



4 Fold Symmetric Dendrite Array

Solidification: Liquid → Solid

- casting & welding
- single crystal growth
- directional solidification
- rapid solidification

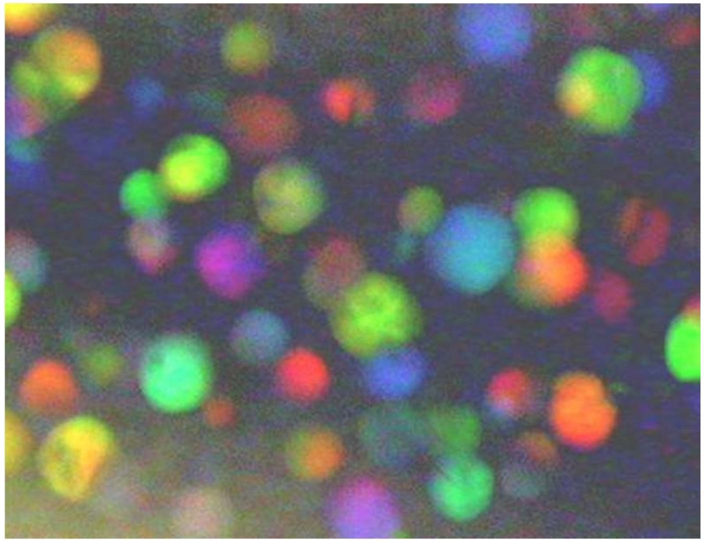
4.1. Nucleation in Pure Metals

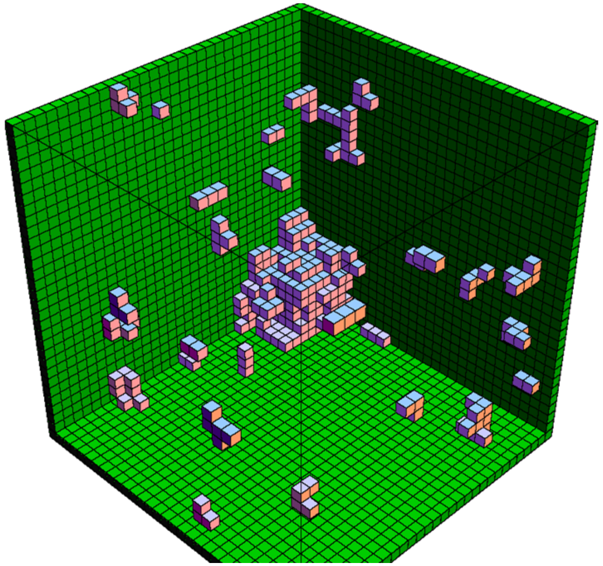
T_m : $G_L = G_S$

- Undercooling (supercooling) for nucleation: 250 K ~ 1 K

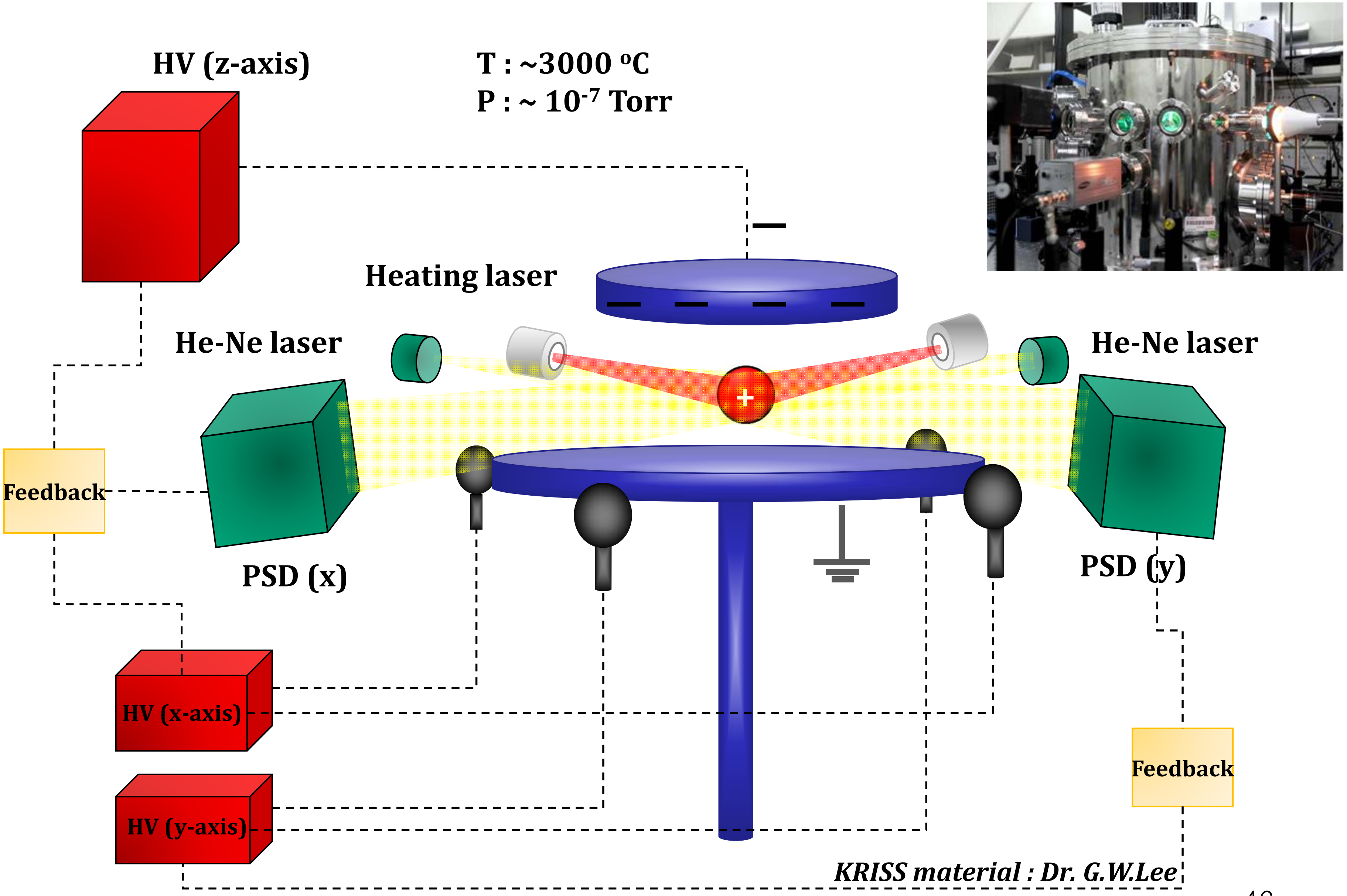
<Types of nucleation>

- Homogeneous nucleation
- Heterogeneous nucleation



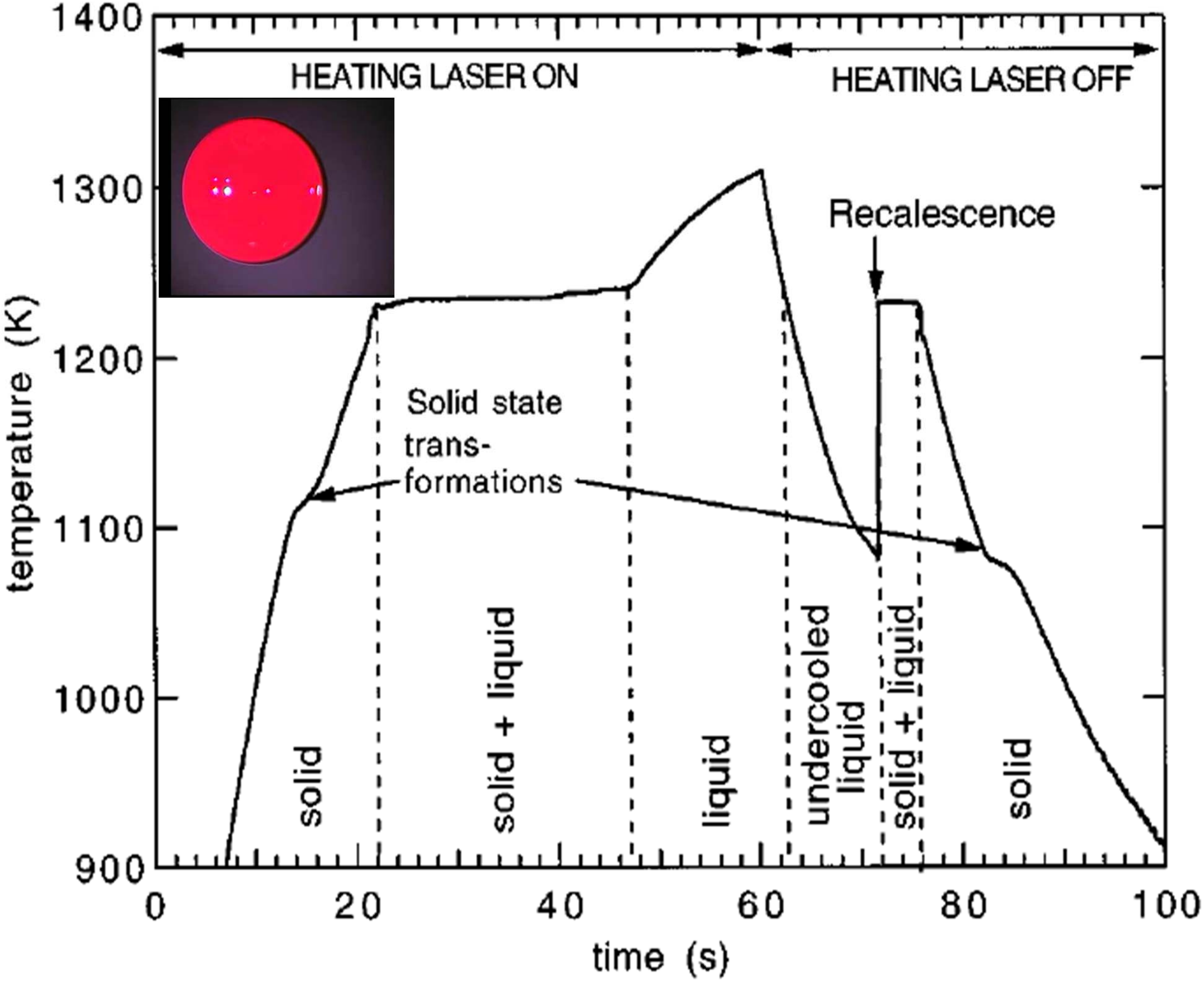


Electrostatic levitation in KRISS



T : ~3000 °C
P : ~ 10^{-7} Torr

KRISS material : Dr. G.W.Lee


HEATING LASER ON
HEATING LASER OFF
Recalescence
Solid state
trans-
formations
solid
solid + liquid
liquid
undercooled
liquid
solid + liquid
solid
1400
1300
1200
1100
1000
900
temperature (K)
0
20
40
60
80
100
time (s)

Electrostatic Levitation: cooling curve of Vitreloy 1 system