

2016 Spring

**“Advanced Physical Metallurgy”
- Bulk Metallic Glasses -**

05.04.2016

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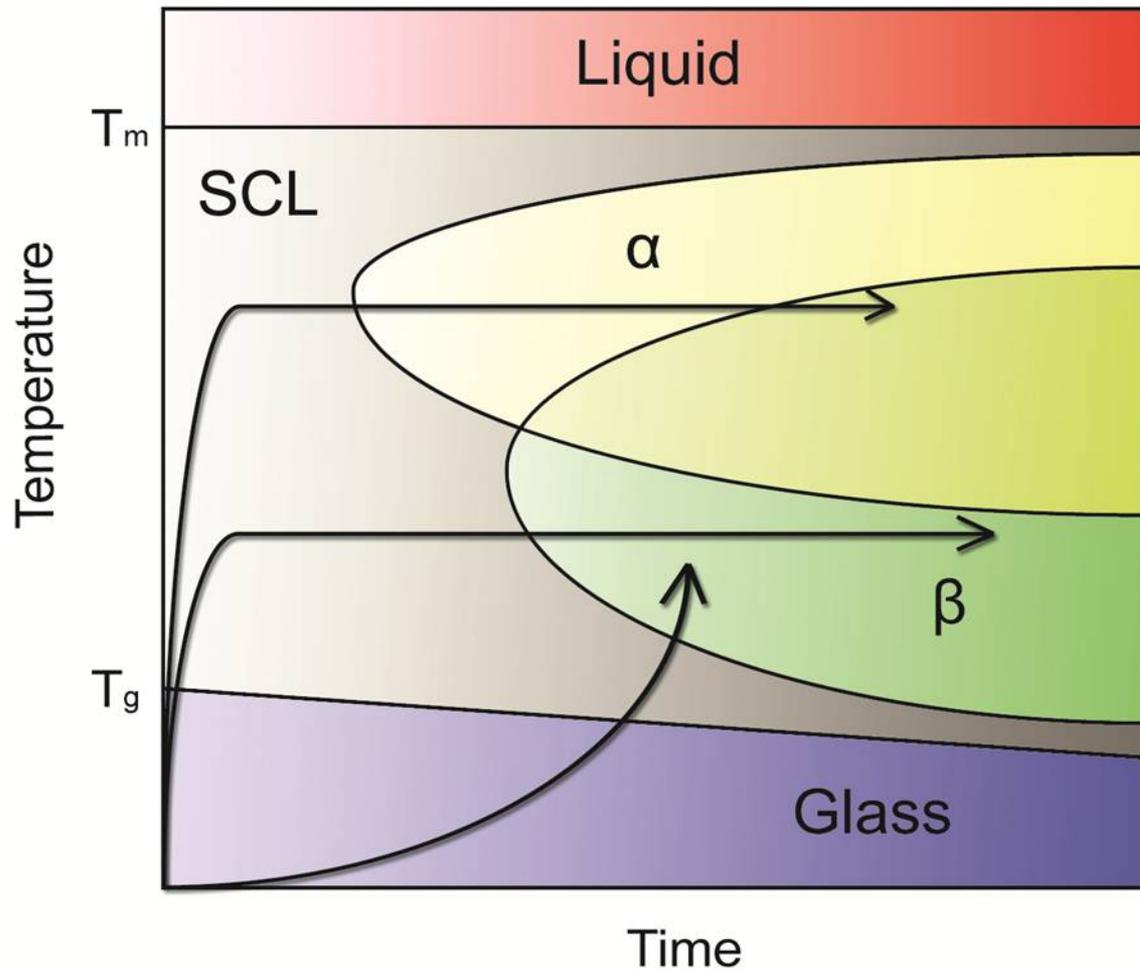
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5 Crystallization Behavior

Crystallization to Equilibrium or Non-equilibrium Phase.



α : Equilibrium phase

β : Non-Equilibrium phase

Cooling Process.

Only α phase

Heating Process.

Low T Crystallize to β

High T Crystallize to α

5.2.3 Structural Details

- **Amorphous vs Nanocrystalline**

1) *Microstructural observation*

XRD, (HR)TEM, EXAFS ...

2) *Thermal analysis*

DSC (Differential Scanning Calorimetry)

: Measure heat absorbed or liberated during heating or cooling

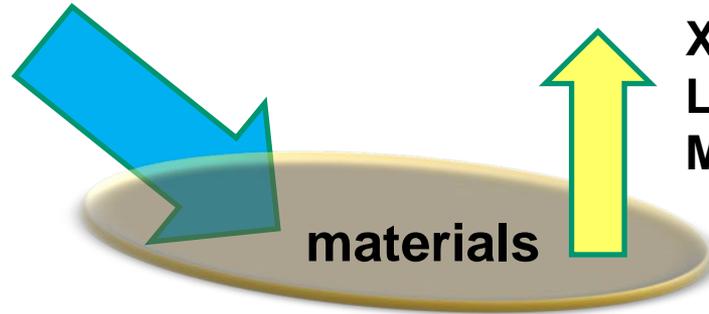
cf) a) *glass* → *nucleation & growth*
(perfect random)

b) *local clustering: quenched-in nuclei* → *only growth*

c) *Nanocrystalline* → *growth*

Simple idea of analytical tools

Electron
X-ray
Laser
Light
Shockwave
Mechanical

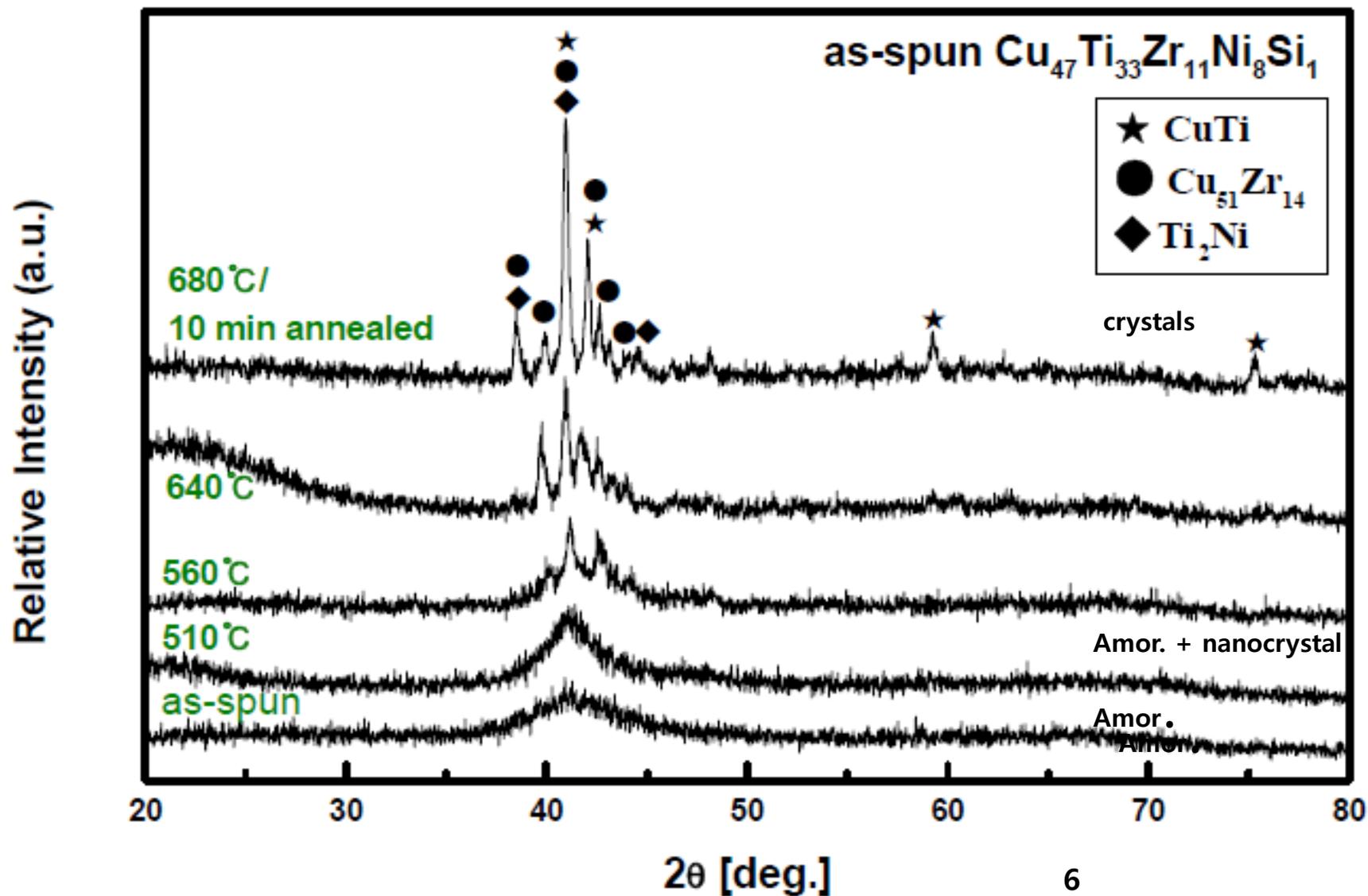


Electron
X-ray
Light
Mechanical

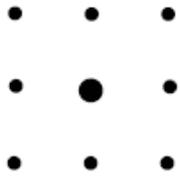
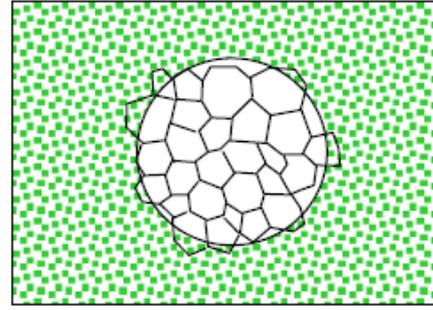
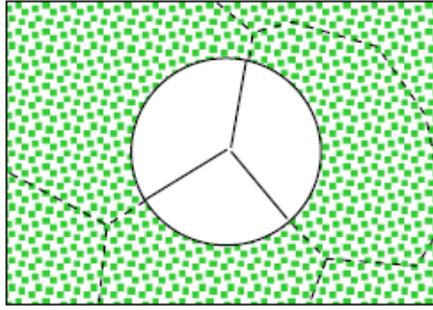
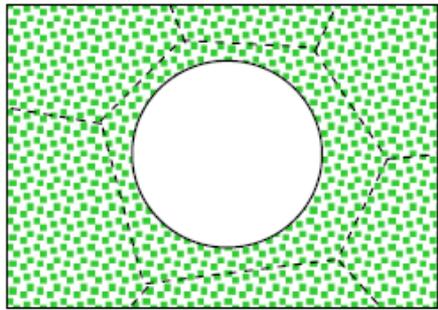
Analytical tool	Abbreviation	Source	Signal	Main Analysis
X-ray diffraction	XRD	X-ray	X-ray	Structure
Transmission Electron Microscopy Scanning Electron Microscopy	TEM SEM	Electron	Electron, Photon (X-ray, Light)	Structure/ Chemistry
X-ray Photoelectron Spectroscopy	XPS	X-ray	Electron	Surface chemistry/ bonding
Auger Electron Spectroscopy	AES	Electron	Electron	Surface chemistry
Energy Dispersive Spectroscopy Wavelength Dispersive Spectroscopy	EDS WDS	Electron	X-ray	Chemistry
Electron BackScattered Diffraction	EBSD	Electron	Electron	Structure/ chemistry

< X-ray diffraction >

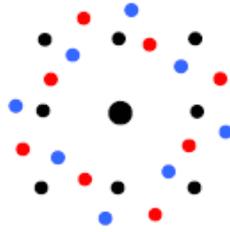
Crystallization after annealing



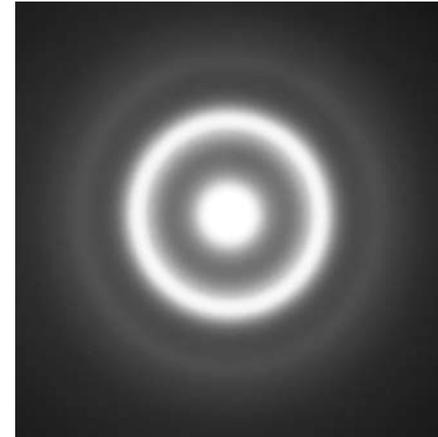
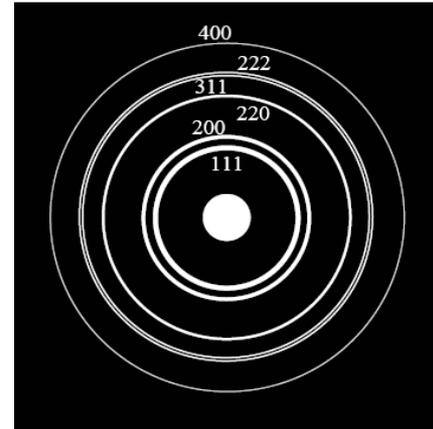
Electron Diffraction Pattern--Spot to Ring



(a)



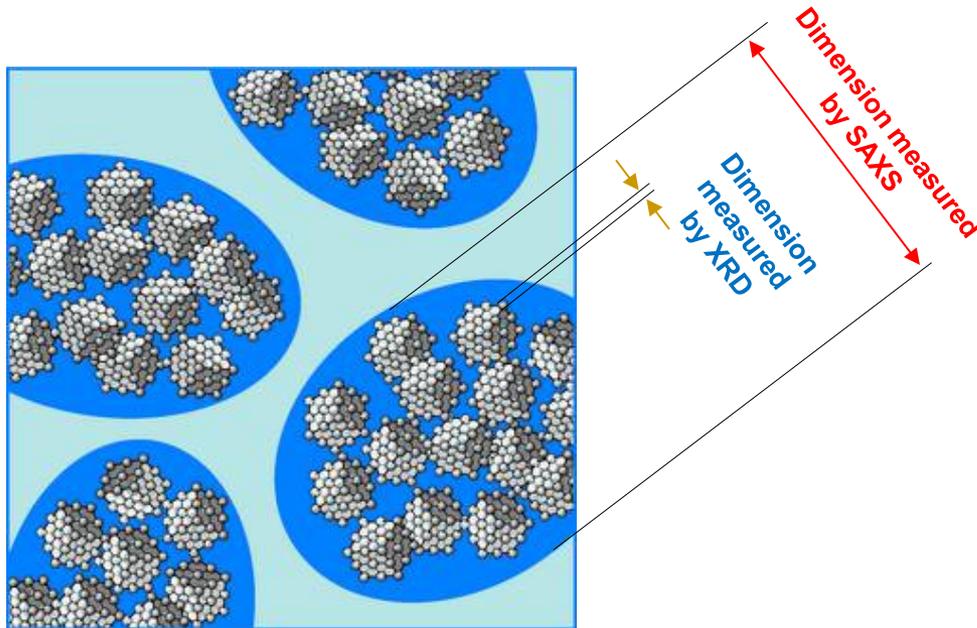
(b)



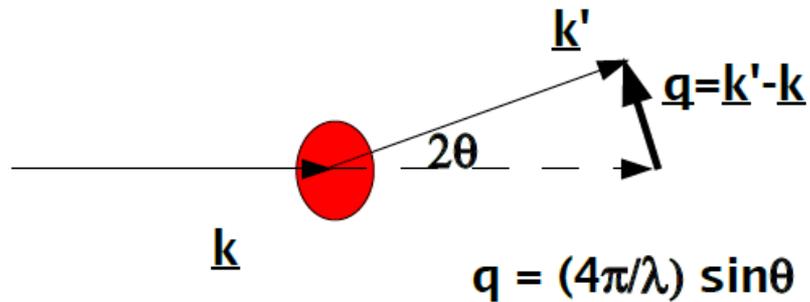
Angle range of Small angle scattering

- Length scale of small angle scattering : 1 – 1000 nm
- Information on relatively large r is contained in $I(q)$ at relatively small q ($=4\pi\sin\theta/\lambda$)
- Bragg's law : $\sin\theta=\lambda/2d$

$d = \text{few } \text{\AA}$	$\lambda = 1 \text{ \AA}$	$2\theta = 20$
$d = 100 \text{ \AA}$	$\lambda = 1 \text{ \AA}$	$2\theta = 0.6$
- Sample contains a scattering length density inhomogeneity of dimension larger than 1 nm, scattering becomes observable in small angle region ($0 \sim 4^\circ$)



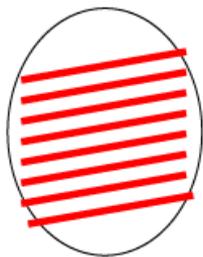
Diffraction v.s. Small angle scattering



x-rays scatter from *electrons*

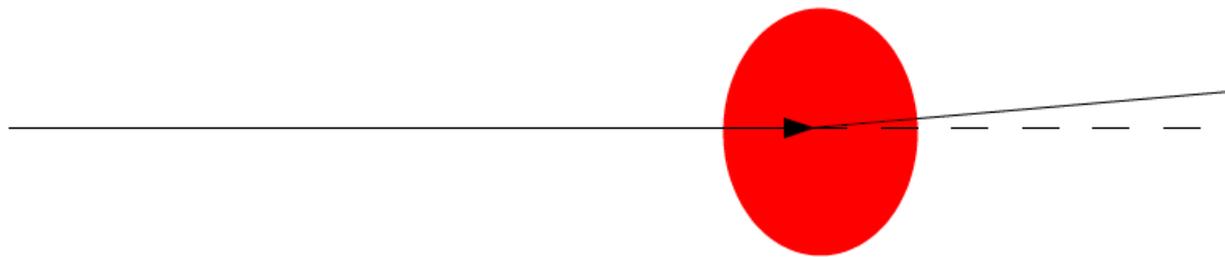
neutrons scatter from *nuclei*
and *magnetic moments*

diffraction



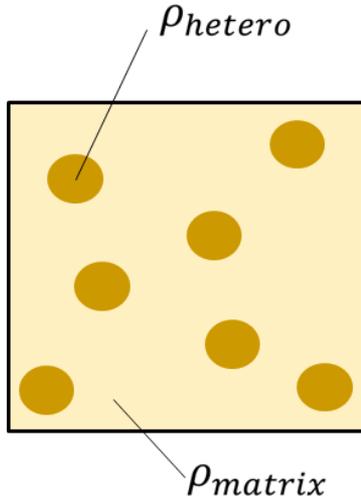
- scattering from *atomic structures*
- size of objects $\sim \lambda$
- small length scale \rightarrow large angle

small-angle scattering



- scattering from *interfaces*
- size of objects $\gg \lambda$
- large length scale \rightarrow small angle

Scattering length density (of heterogeneities)



$$\rho = \frac{\sum_j^n b_j}{\bar{V}}$$

$$\Delta\rho = \rho_{hetero} - \rho_{matrix}$$

$$\bar{V} = \frac{M_w}{N_A \cdot \rho_{mass}}$$

Intensity in small angle scattering

$$I(q) = \Delta\rho^2 d_N \int_0^R N(r) [V(r)F(q, r)]^2 dr$$

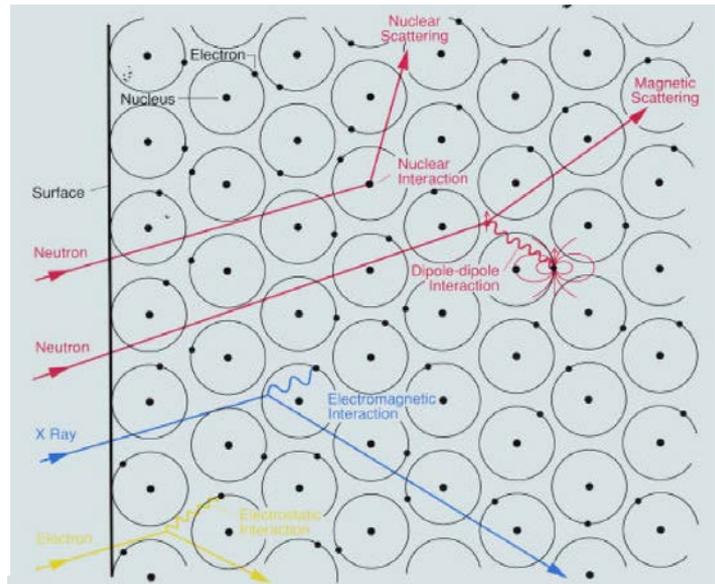
d_N : number density factor
 $N(r)$: normalized size distribution
 $V(r)$: Volume
 $F(q, r)$: Form factor of particles

$$\Delta\rho = \rho_{hetero} - \rho_{matrix}$$

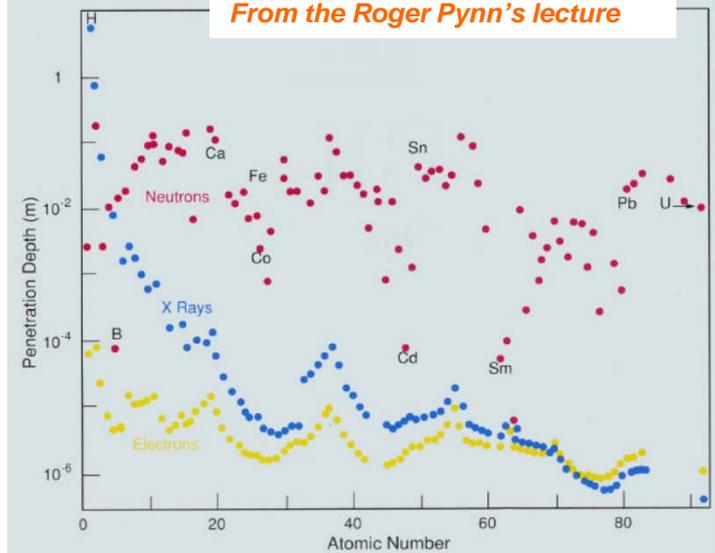
**Common factors in
 both SANS and
 SAXS**

Coherent and Incoherent Scattering

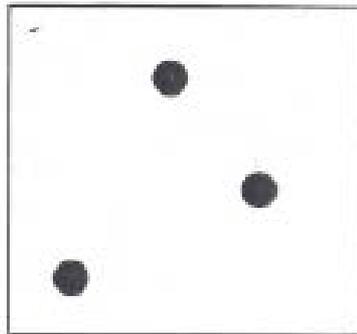
items	X-ray	neutron
source	collision of electrons with target metals(Cu, Mo, W...) acceleration of charged particles	nuclear reactor spallation neutron source (accelerator)
scattered by	electrons	atomic nuclei, unpaired spins
interaction	EM(electromagnetic)	Nuclear(strong int.) EM
scattering amplitude	linearly depend on Z	nearly indep. on Z
sample amount	$\mu\text{g} \sim \text{mg}$	$\sim\text{g}$
meas. time	10^{1-2}min (step scan: $\sim\text{hr}$)	10^{0-2}hr
hard to see	relatively light elements (H, Li, B, C, O ...)	highly abs. nuclei (Gd, Sm, Eu, Cd, B...)



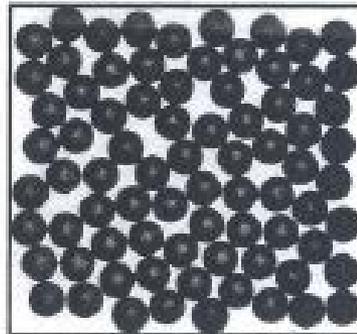
From the Roger Pynn's lecture



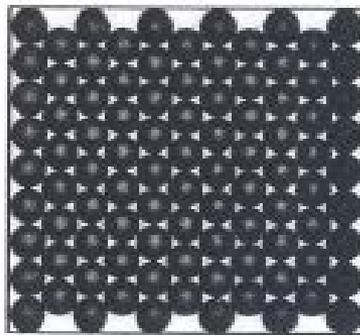
Characterizing the structure - radial distribution function, also called pair distribution function



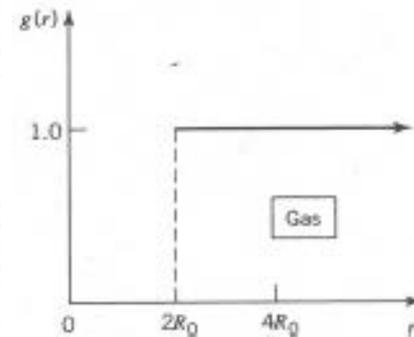
(a)



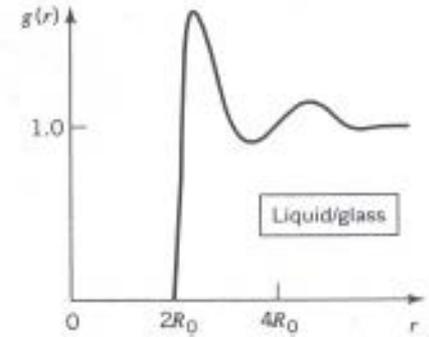
(b)



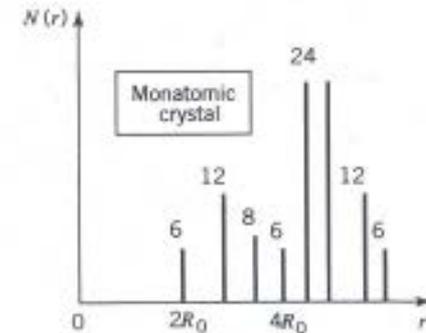
(c)



(a)



(b)



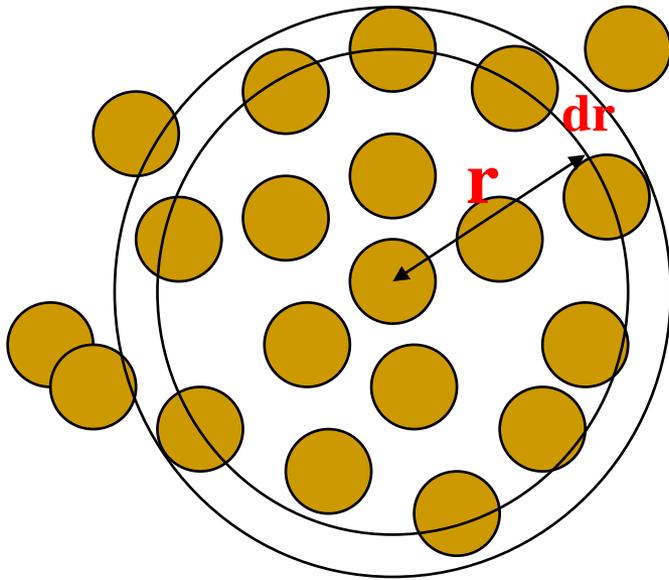
(c)

Figure 2.4 Hard-sphere model of (a) gas, (b) liquid/glass, and (c) crystalline solid.

Figure 2.5 Pair-distribution functions for (a) a gas and (b) liquid or glass. (c) The radial dependence of the number of neighbors $N(r)$ for a primitive cubic crystal with one atom per lattice site.

Gas, amorphous/liquid and crystal structures have very different radial distribution function

Radial distribution function - definition



$$g(r) = \frac{1}{\langle \rho \rangle} \frac{dn(r, r + dr)}{dv(r, r + dr)}$$

1. Carve a shell of size r and $r + dr$ around a center of an atom.

The volume of the shell is

$$dv = 4\pi r^2 dr$$

1. Count number of atoms with centers within the shell (dn)
2. Average over all atoms in the system
3. Divide by the average atomic density $\langle \rho \rangle$

RDF: count the neighbors

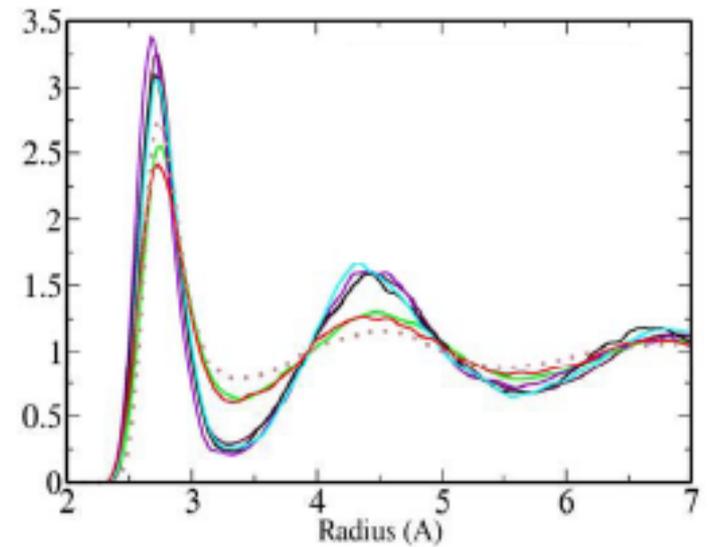
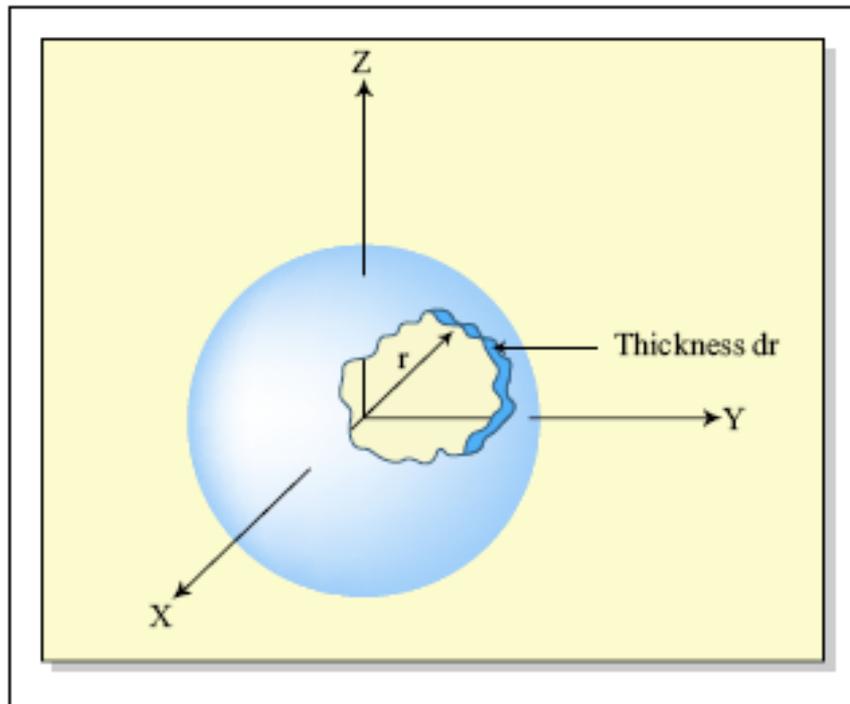


Figure by MIT OCW.

Properties of the radial distribution function

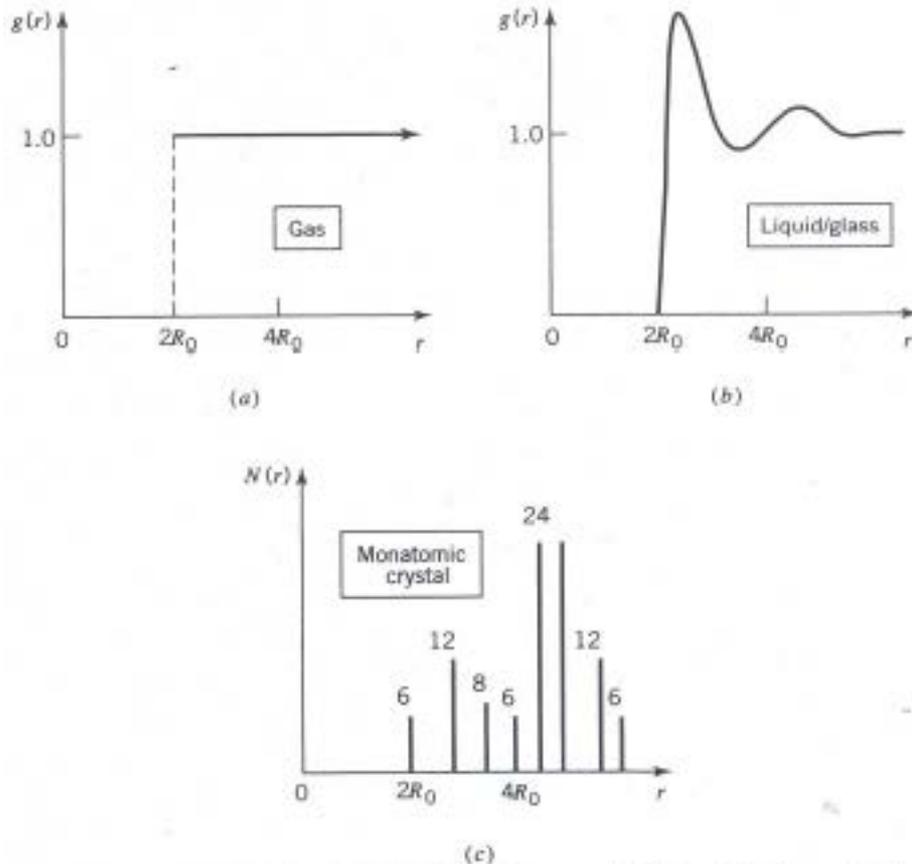


Figure 2.5 Pair-distribution functions for (a) a gas and (b) liquid or glass. (c) The radial dependence of the number of neighbors $N(r)$ for a primitive cubic crystal with one atom per lattice site.

For gases, liquids and amorphous solids $g(r)$ becomes **unity** for large enough r .

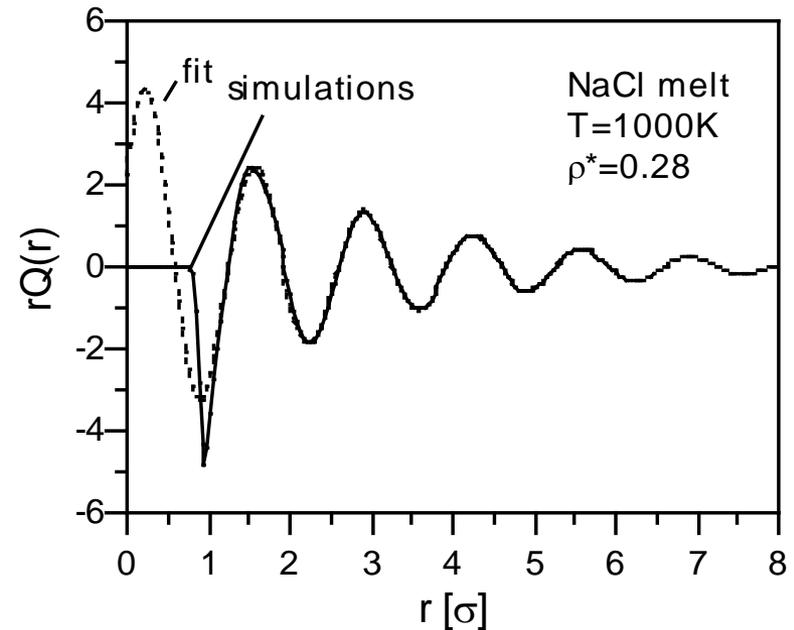
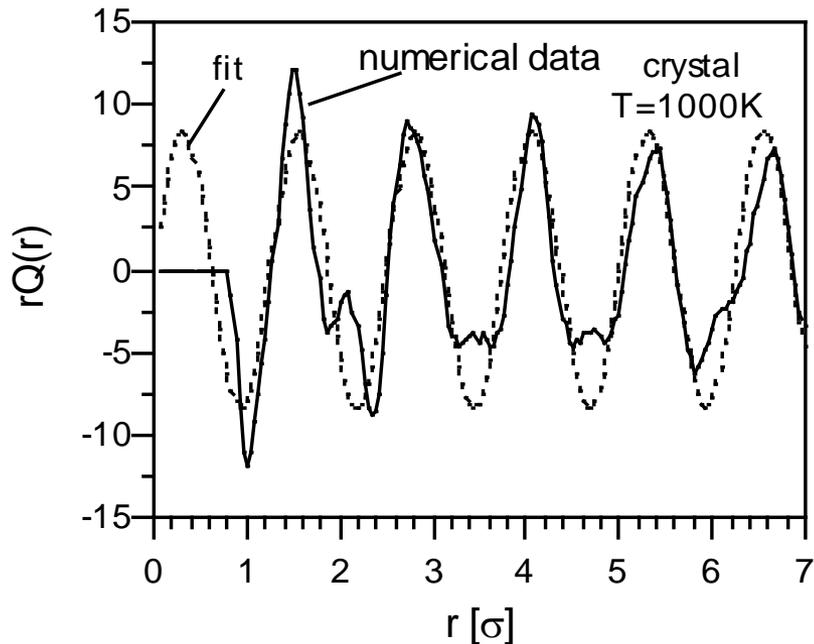
Features in $g(r)$ for liquids and amorphous solids are due to packing (exclude volume) and possibly bonding characteristics.

The distance over which $g(r)$ becomes unity is called the correlation distance which is a measure of the extent of so-called **short range order (SRO)**

The first peak corresponds to an average nearest neighbor distance.

Radial Distribution Function - Crystal and Liquid

$$Q(r) = g(r) - 1 \sim \frac{1}{r} \sin(r/d + \varphi) \exp(r/\lambda)$$



- Liquid/amorphous $g(r)$, for large r exhibit oscillatory exponential decay.
- Crystal $g(r)$ does not exhibit an exponential decay ($\lambda \rightarrow \infty$).

Radial distribution functions and the structure factor

- **The structure factor, $S(k)$** , which can be measured experimentally (e.g. by X-rays) is given by the **Fourier transform of the radial distribution function** and vice versa.

$$S(k) = 1 + \frac{4\pi\langle\rho\rangle}{k} \int_0^{\infty} r[g(r) - 1]\sin(kr)dr$$

Radial distribution functions can be obtained from experiment and compared with that from the structural model.

➔ More detailed structural characterization - Voronoi Polyhedra

Your Assignment 6: study and summary for Voronoi Polyhedra and submit as a ppt file (under 5 pages)

5.2 Methodology

5.2.1 Transformation Temperatures

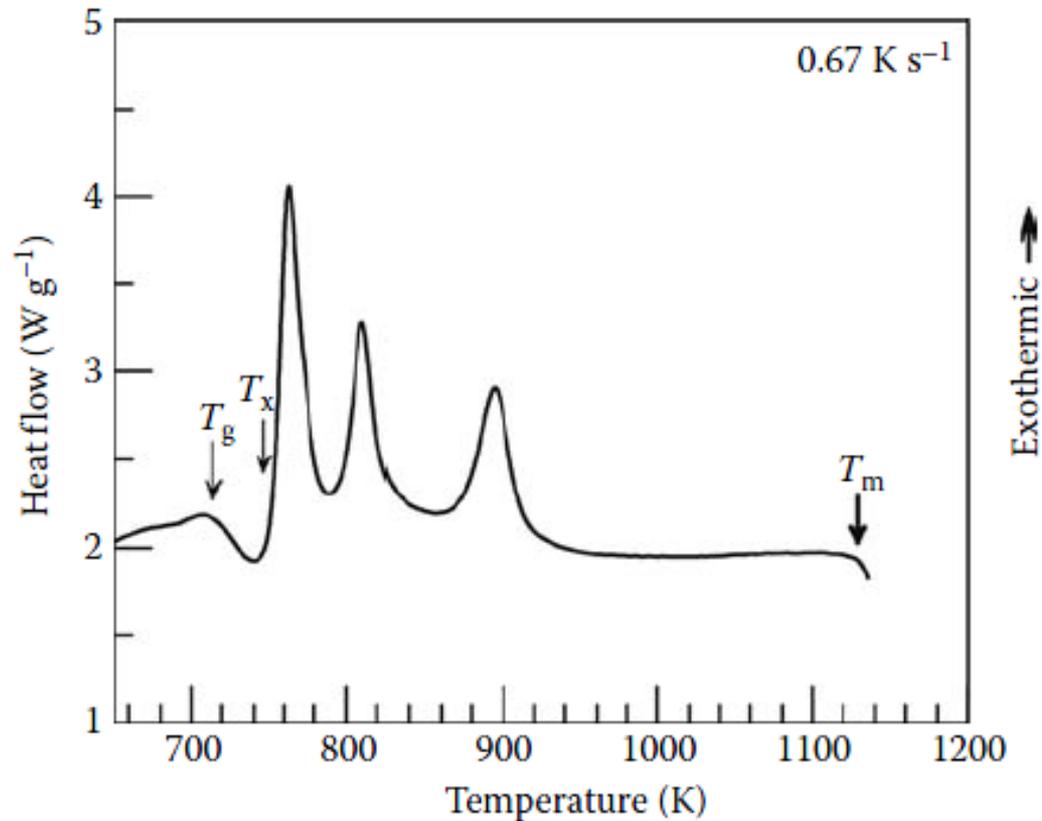
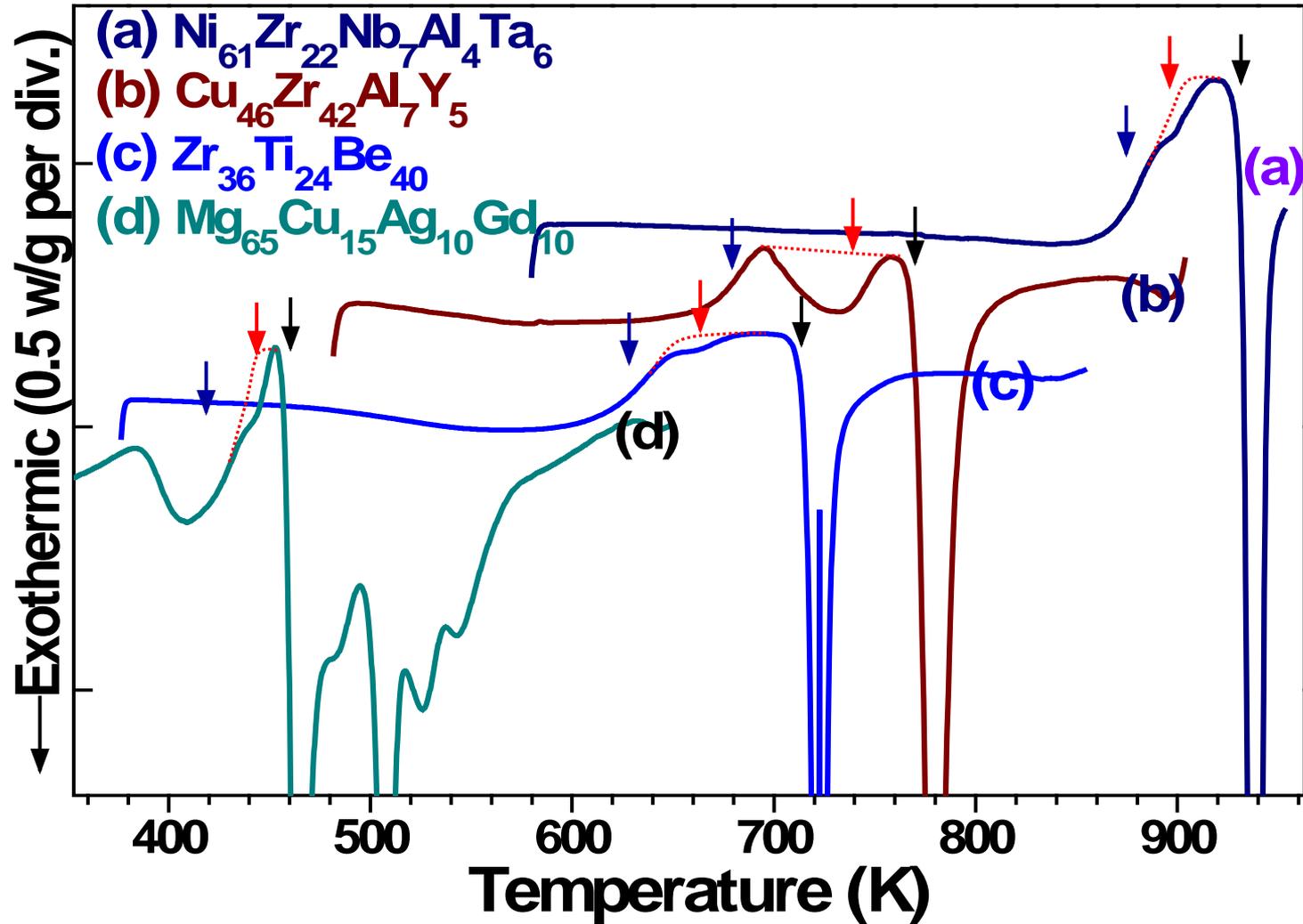
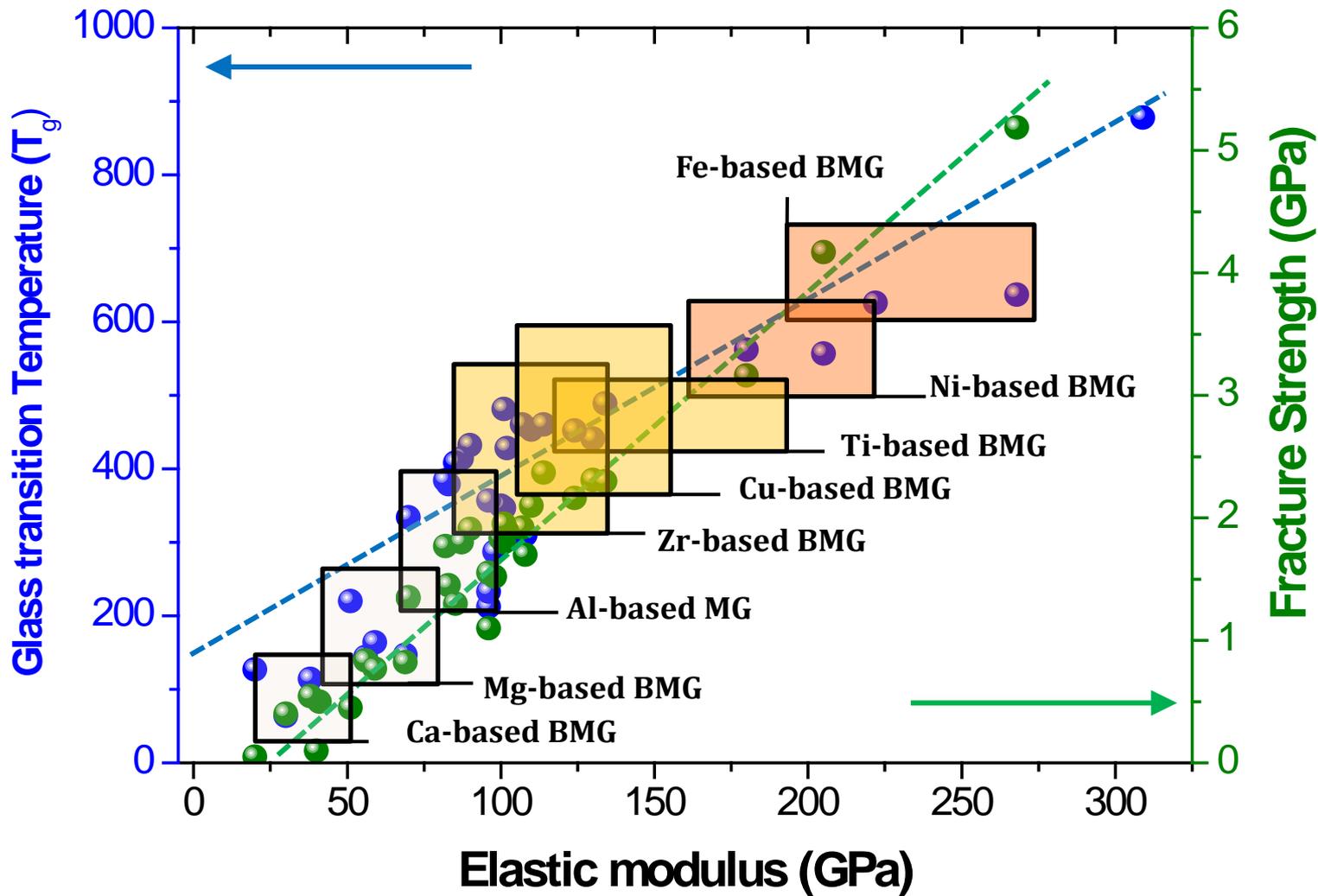


FIGURE 5.1

Schematic of a typical differential scanning calorimeter (DSC) curve obtained on heating a BMG alloy from room temperature to high temperatures at a constant heating rate of 40 K min⁻¹. Note that the curve displays three important temperatures—the glass transition temperature, T_g , the crystallization temperature, T_x , and the melting temperature, T_m . In some cases, there may be more than one crystallization temperature, depending upon the number of stages in which the glass or the supercooled liquid transforms into the crystalline phase(s).

Variation of T_g depending on alloy compositions → Broken Bonds





→ Almost all of the BMGs exhibit T_g . But, there are a few exceptions to this. For example, Nd-Fe-Al and Pr-Fe-Al glassy alloys did not exhibit any T_g , even though $D_{\max} > 10$ mm.

TABLE 5.1

Transformation Temperatures Determined for Some Typical BMG and Melt-Spun Glassy Alloys

Composition	Glass-Forming Technique		T_g (K)	T_x (K)	ΔT_x ($= T_x - T_g$) (K)	Heating Rate (K s ⁻¹) ^a
Au ₅₅ Cu ₂₅ Si ₂₀	Cu-mold casting	low	348	383	35	0.33
Ca _{66.4} Al _{33.6}	Cu-mold casting		528	540	12	0.33
Ca ₆₀ Al ₃₀ Ag ₁₀	Cu-mold casting		483	531	48	0.33
Ca ₅₈ Al ₃₂ Mg ₁₀	Cu-mold casting		513	539	26	0.33
Ca ₆₅ Mg ₁₅ Zn ₂₀	Cone-shaped Cu-mold casting		374	412	38	0.67
Ce ₆₀ Al ₁₀ Ni ₁₀ Cu ₂₀	Suction casting		374	441	67	0.167
Co ₄₃ Fe ₂₀ Ta _{5.5} B _{31.5}	Cu-mold casting	high	910	982	72	0.67
Cu ₅₀ Zr ₅₀	Cu-mold casting		675	732	57	0.67
Cu ₅₀ Zr ₅₀	Melt spinning		686	744	58	0.67
Cu ₆₀ Zr ₃₀ Ti ₁₀	Cu-mold casting (2.5 mm dia rod)		714	758	44	0.67
Cu ₆₀ Zr ₃₀ Ti ₁₀	Melt spinning		711	754	43	0.67
Cu ₄₆ Zr ₄₂ Al ₇ Y ₅	Injection casting into Cu-mold (10 mm dia rod)		672	772	100	0.33
Fe ₆₄ Mo ₁₄ C ₁₅ B ₇	Injection casting into Cu-mold (2.5 mm dia rod)		793	843	50	0.33
Hf _{52.5} Cu _{17.9} Ni _{14.6} Al ₁₀ Ti ₅	Suction casting		767	820	53	0.167
La ₅₅ Al ₂₅ Ni ₁₀ Cu ₁₀	High-pressure die casting (9 mm dia rod)		460	527	67	0.67
La ₅₅ Al ₂₅ Ni ₁₀ Cu ₁₀	Melt spinning		460	550	90	0.67
Mg ₆₅ Cu _{7.5} Ni _{7.5} Zn ₅ Ag ₅ Y ₁₀	Melt spinning		430	459	29	0.67
Ni _{60.25} Nb _{39.75}	Injection casting into Cu-mold (1 mm dia rod)		891	923	32	0.33
Ni ₆₂ Nb ₃₃ Zr ₅	Injection molding (3 mm dia rod)		877	917	40	0.33
Pd ₇₉ Ag _{4.5} Si _{16.5}	Splat cooling		640	672	32	0.33
Pd ₇₈ Ag _{5.5} Si _{16.5}	Dropping a molten droplet onto metal substrate		642	683	41	0.33
Pd ₈₀ Au _{3.5} Si _{16.5}	Splat cooling		644	675	31	0.33
Pd ₇₈ Au ₄ Si ₁₈	Roller quenching		656	696	40	0.67

TABLE 5.1

Transformation Temperatures Determined for Some Typical BMG and Melt-Spun Glassy Alloys

Composition	Glass-Forming Technique	T_g (K)	T_x (K)	ΔT_x ($= T_x - T_g$) (K)	Heating Rate ($K s^{-1}$) ^a
Pd ₄₃ Cu ₂₇ Ni ₁₀ P ₂₀ (fluxed)	Water quenching	585	716	largest 131	0.67
Pd ₄₃ Cu ₂₇ Ni ₁₀ P ₂₀ (foam)	Water quenching	577	667	90	0.33
Pd ₄₀ Cu ₃₀ Ni ₁₀ P ₂₀ (unfluxed)	Melt spinning	572	663	91	0.33
Pd ₄₀ Cu ₃₀ Ni ₁₀ P ₂₀ (fluxed)	Melt spinning	572	670	98	0.33
Pd _{77.5} Cu ₆ Si _{16.5}	Dropping a molten droplet onto metal substrate	646	686	40	0.33
Pd ₄₀ Ni ₄₀ P ₂₀	Centrifugal spinning	583	650	67	0.33
Pd ₄₀ Ni ₄₀ P ₂₀ (fluxed)	Water quenching (7 mm dia rod)	576	678	102	0.33
Pd ₄₀ Ni ₄₀ P ₂₀ (unfluxed)	Melt spinning	580	643	63	0.33
Pd ₄₀ Ni ₄₀ P ₂₀ (fluxed)	Melt spinning	590	671	91	0.33
Pd ₈₁ Si ₁₉ (fluxed)	Air cooling	638	696	58	0.33
Pd ₈₁ Si ₁₉ (fluxed)	Melt spinning	633	675	42	0.33
Pd ₈₀ Si ₂₀	Splat cooling	655	667	12	0.33
Pr ₆₀ Cu ₂₀ Ni ₁₀ Al ₁₀	Suction casting	409	452	43	0.167
Pt _{57.5} Cu _{14.7} Ni _{5.3} P _{22.5} (fluxed)	Water quenching (16 mm dia rod)	508	606	98	0.33
Sm ₅₆ Al ₂₂ Ni ₂₂	Suction casting	544	582	38	0.33
Ti ₃₂ Hf ₁₈ Ni ₃₅ Cu ₁₅	Planar flow casting	722	766	44	0.167
Ti ₅₀ Ni ₂₄ Cu ₂₀ B ₁ Si ₂ Sn ₃	Cu-mold casting	726	800	74	0.67
Ti ₄₀ Zr ₂₅ Ni ₂ Cu ₁₃ Be ₂₀	Cu-mold casting	599	644	45	0.33
Y ₃₆ Sc ₂₀ Al ₂₄ Co ₂₀	—	645	760	115	0.67
Zr ₆₅ Al _{7.5} Ni ₁₀ Cu _{17.5}	Water quenching (16 mm dia rod)	625	750	125	0.67
Zr ₆₅ Al _{7.5} Ni ₁₀ Cu _{17.5}	Melt spinning	622	749	127	0.67
Zr _{41.2} Ti _{13.8} Cu _{12.5} Ni ₁₀ Be _{22.5}	Cu-mold casting	625	705	80	0.33

^a 0.167K s⁻¹ = 10 K min⁻¹; 0.33K s⁻¹ = 20K min⁻¹; and 0.67K s⁻¹ = 40 K min⁻¹.

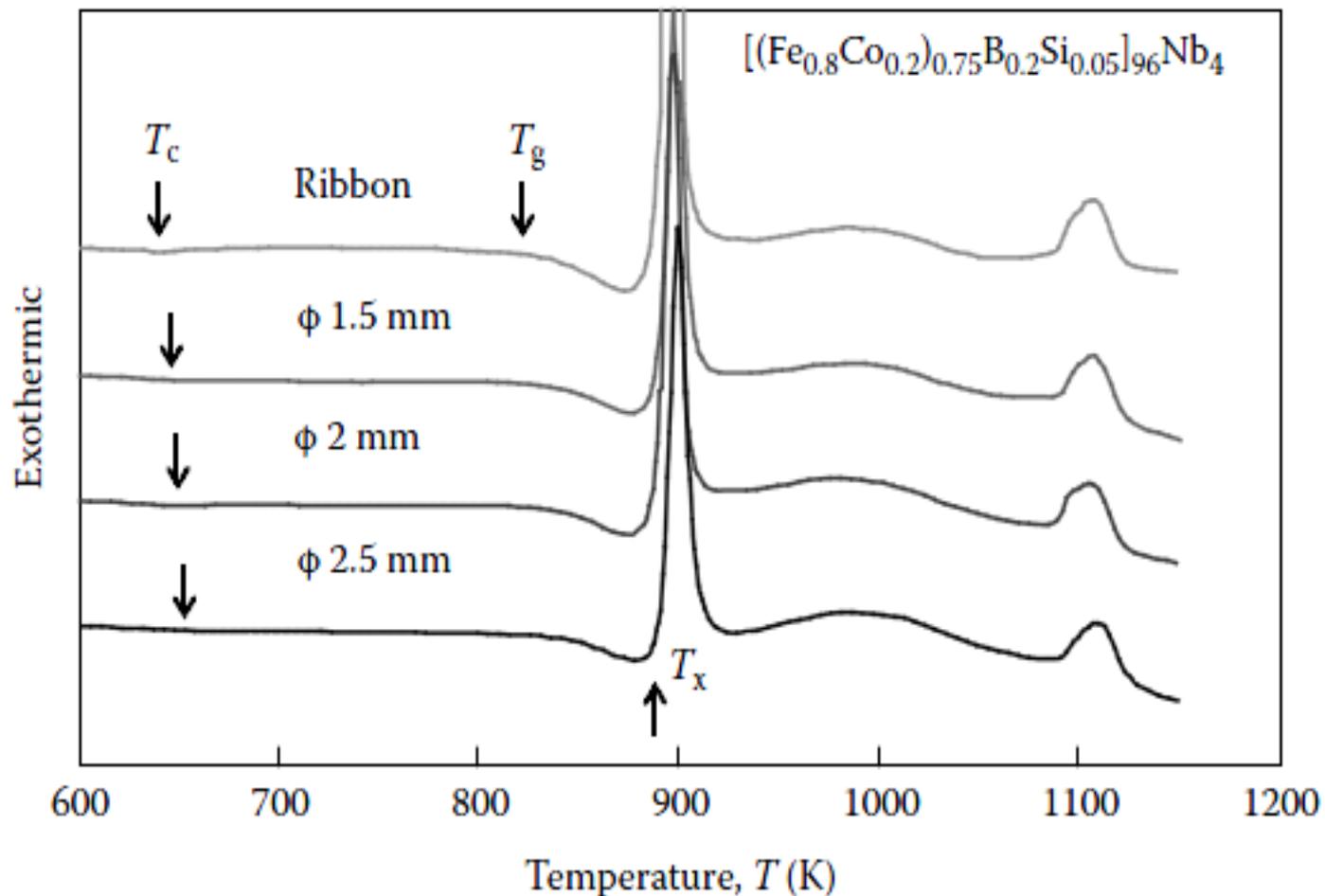
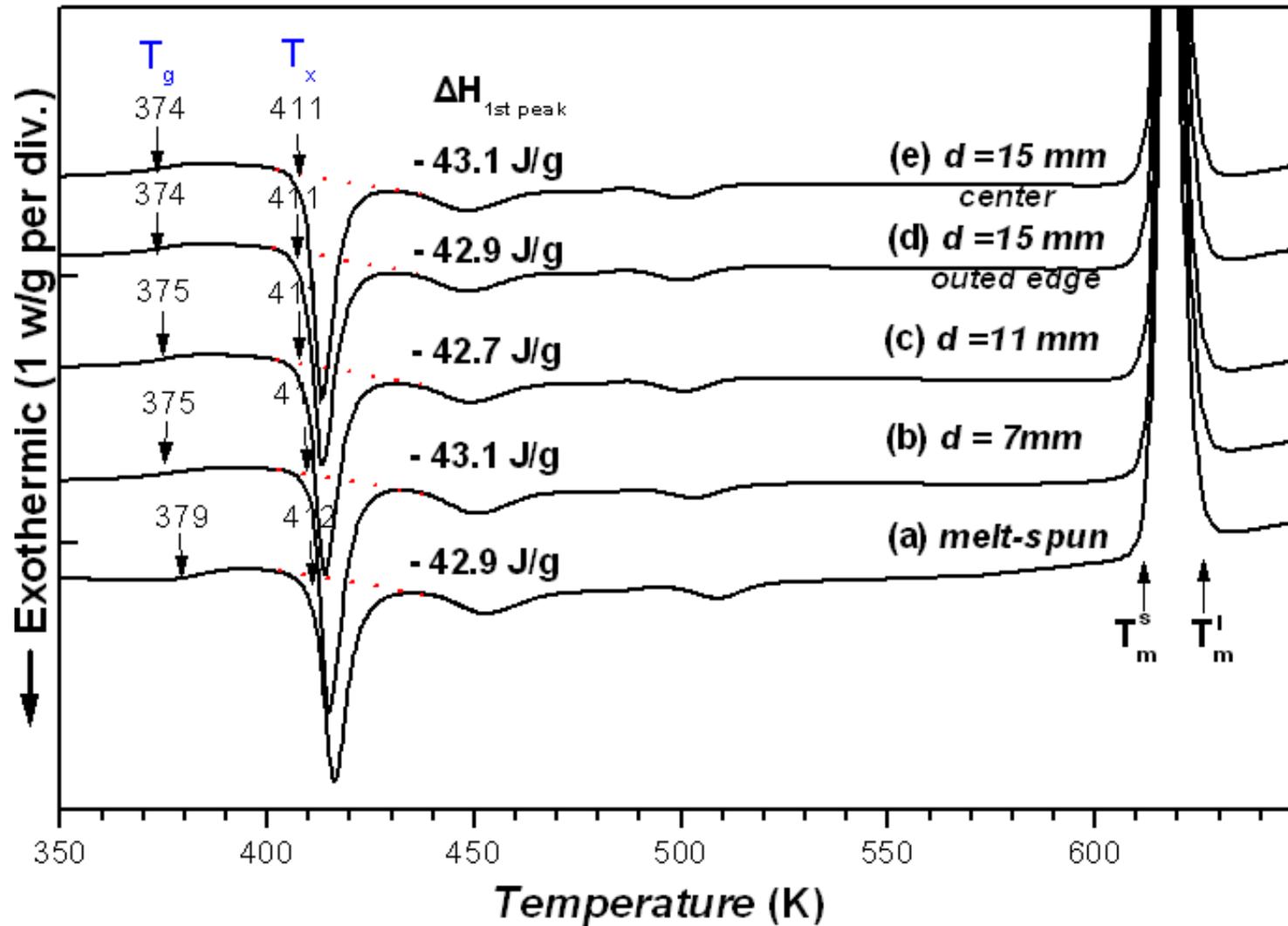


FIGURE 5.2

DSC curves of bulk glassy $[(\text{Fe}_{0.8}\text{Co}_{0.2})_{0.75}\text{B}_{0.2}\text{Si}_{0.05}]_{96}\text{Nb}_4$ alloy of different diameters (1.5, 2.0, and 2.5 mm) and melt-spun ribbon of the same composition. These curves clearly demonstrate that the transformation temperatures are identical for all the samples and that the transformation temperatures do not depend upon the diameter of the rod or the thickness of the ribbon. (Reprinted from Inoue, A. et al., *Acta Mater.*, 52, 4093, 2004. With permission.)

*** Typically T_g is ~ 50-60% of the melting point.**



* J Mater Res, 19 (2004) 685.

The glass transition temperature T_g is a kinetic parameter and its value depends on the cooling rate at which the glass is formed (and also on the heating rate at which the glassy sample is reheated). It was also noted that T_g was lower when the glass had formed at lower cooling rates. Therefore, it would be possible to assume that the T_g for the melt-spun ribbon and BMG rod will be different. But, this is not true. The reason is that T_g , the temperature at which the glass is formed is estimated during the cooling of the molten alloy. On the other hand, T_g is usually measured experimentally during the heating of the glassy alloy that has already formed. Once the glass is heated from room temperature to higher temperatures, it is structurally relaxed and, therefore, it does not matter how the glass had initially formed. Accordingly, both types of glasses will have the same T_g and T_x temperatures, when measured at the same heating rate. That is, there is no difference between the T_g values of glasses prepared by RSP or slow solidification methods.

5.4 Differences in the Crystallization Behavior between Melt-Spun Ribbons and Bulk Metallic Glasses

- (a) The melt-spun metallic glass ribbons **solidified at higher cooling rates** are farther from equilibrium than the BMGs. → **a larger decrease in density and higher energy stored** in the melt-spun ribbons → One would expect that, **due to the larger departure from equilibrium, the kinetics of crystallization in melt-spun glassy ribbons would be faster than that in BMGs. But this is not necessarily true.**
- (b) Once the glass is heated to a temperature above T_g , the glass **becomes a supercooled liquid** (but still exists in the form of a solid). At this stage there is **no difference in the “structure”** between the BMG and the melt-spun metallic glass that was obtained directly by rapidly solidifying the metallic melt, **except that the extent of structural relaxation would be different in the two glasses.** → Therefore, once the BMG has been heated to **above T_g , the crystallization behavior of BMGs and melt-spun metallic glassy ribbons will be identical** (assuming that both the glasses have the same chemical composition).

5.2.2 Activation energy for crystallization

Two different methods: (a) Kissinger method, (b) Ozawa method

(a) Kissinger method $\ln\left(\frac{\beta}{T_p^2}\right) = \left(-\frac{Q}{RT_p}\right) + A$ where
A is a constant
R is the universal gas constant

Thus, by plotting $\ln(\beta/T_p^2)$ against $1/T_p$, one obtains a straight line whose slope is $-Q/R$, from which the activation energy for the transformation, Q can be calculated (Figure 5.3).

- Could get the required data during continuous heating in a DSC
- Possible to evaluate the individual activation energies for the nucleation and growth stages of the transformation
- May not be useful in all studies of decomposition

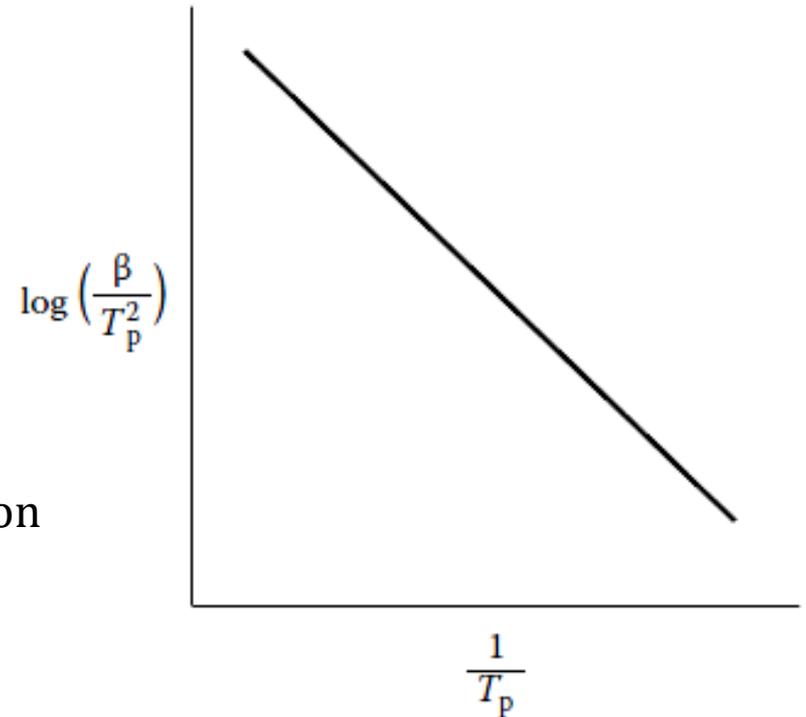


FIGURE 5.3

Kissinger plot in which $\ln(\beta/T_p^2)$ is plotted against $1/T_p$, when a straight line is obtained. The activation energy for crystallization can be calculated from the slope of this straight line.

- Chen & Spaepen (Harvard, 1988)

a) *glass* → *nucleation & growth*
(perfect random)

→ Isothermal annealing
: rapid heating + maintain the temp.



- *Glass* :

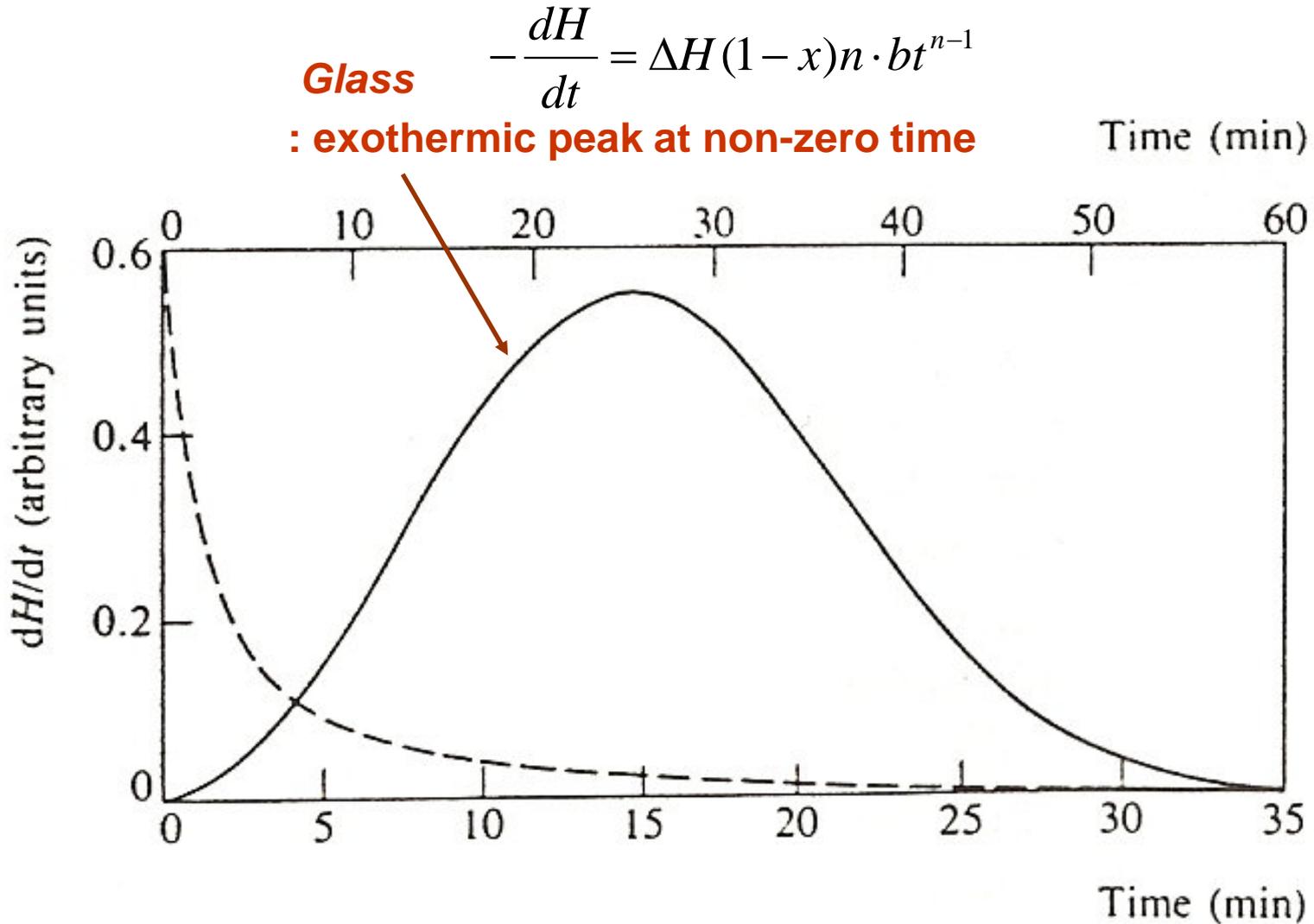
J-M-A Eq. $x = 1 - \exp(-bt^n)$ (n : 2~4, nucleation mechanism)
crystallized volume fraction after time t

→ *Corresponding heat release*

$$-\frac{dH}{dt} = \Delta H (1-x)n \cdot bt^{n-1}$$

(ΔH : total transformation enthalpy)

Fig. 1.4 Isothermal enthalpy release rates for crystallite nucleation and growth (solid line) and crystallite grain-coarsening mechanisms (dashed line)



- **c) Nanocrystalline → grain growth**

$$\rightarrow \frac{dr}{dt} = M \cdot \frac{\gamma}{r^m}$$

(M: atomic mobility, γ : interfacial surface tension)

→ corresponding heat release

$$-\frac{dH}{dt} = H(0) \cdot r(0) \cdot M\gamma / r^{m+2}$$

(H(0): zertime enthalpy of a grain size of r (0))

→ Monotonically decreasing curve

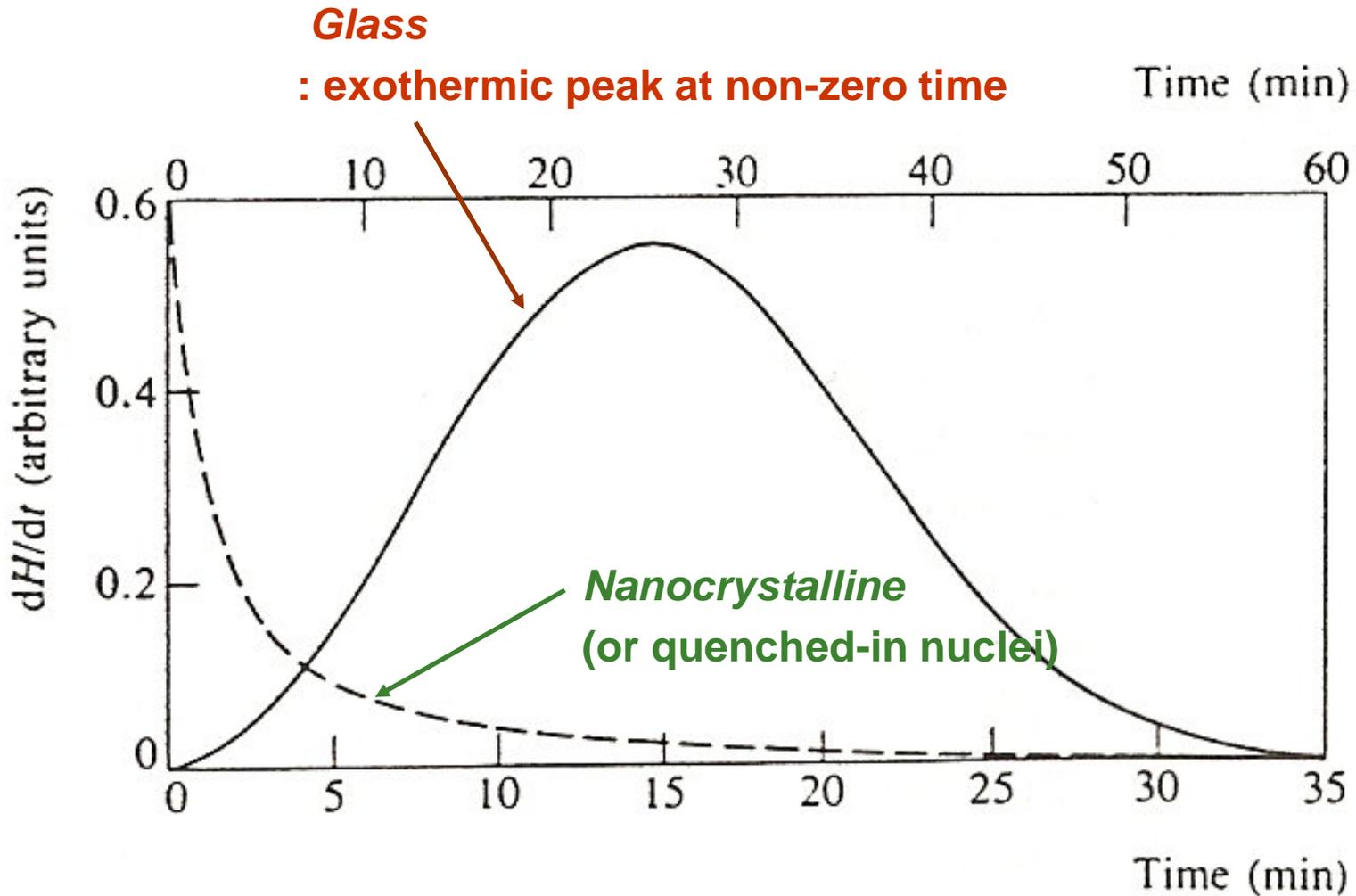
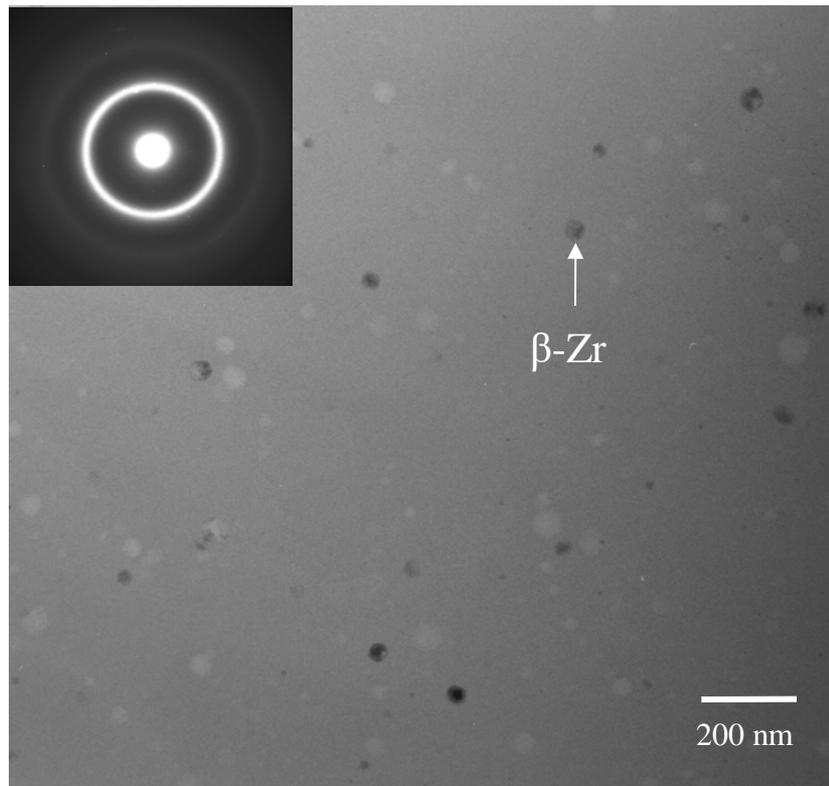
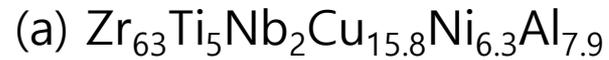


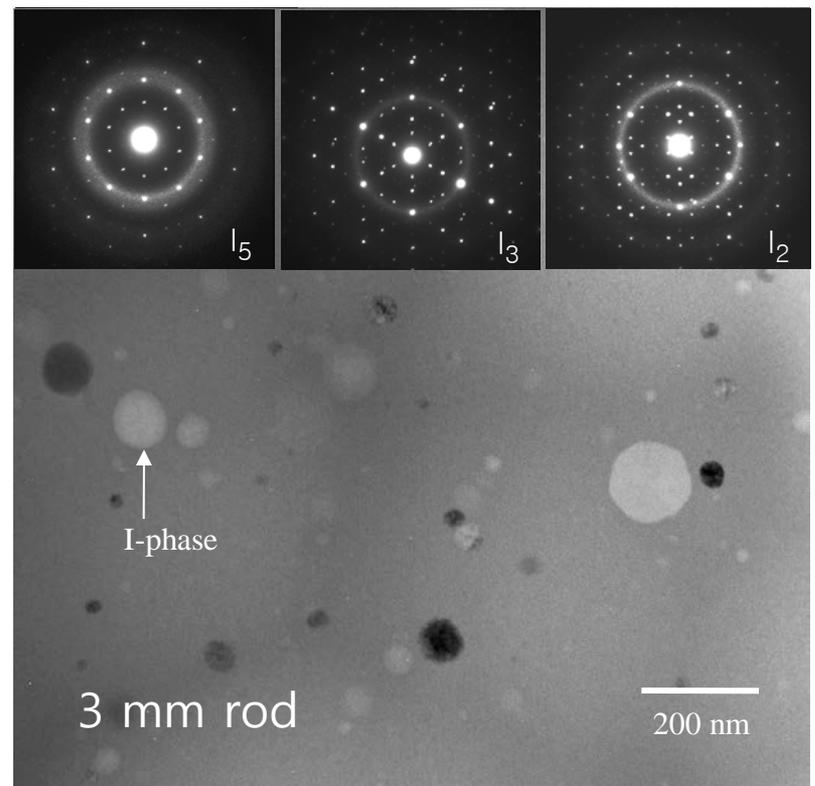
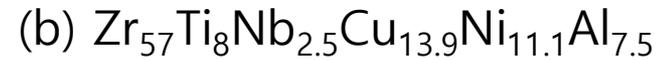
Fig. 1.4 Isothermal enthalpy release rates for crystallite nucleation and growth (solid line) and crystallite grain-coarsening mechanisms (dashed line)

Effect of quenched-in quasicrystal nuclei

2 mm rod

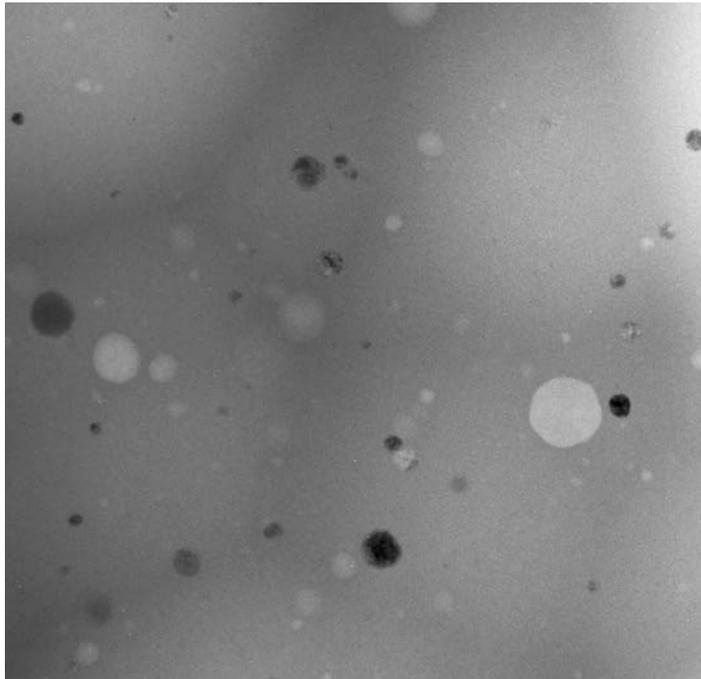


$\beta\text{-Zr}$ particle (~70 nm) in amorphous matrix

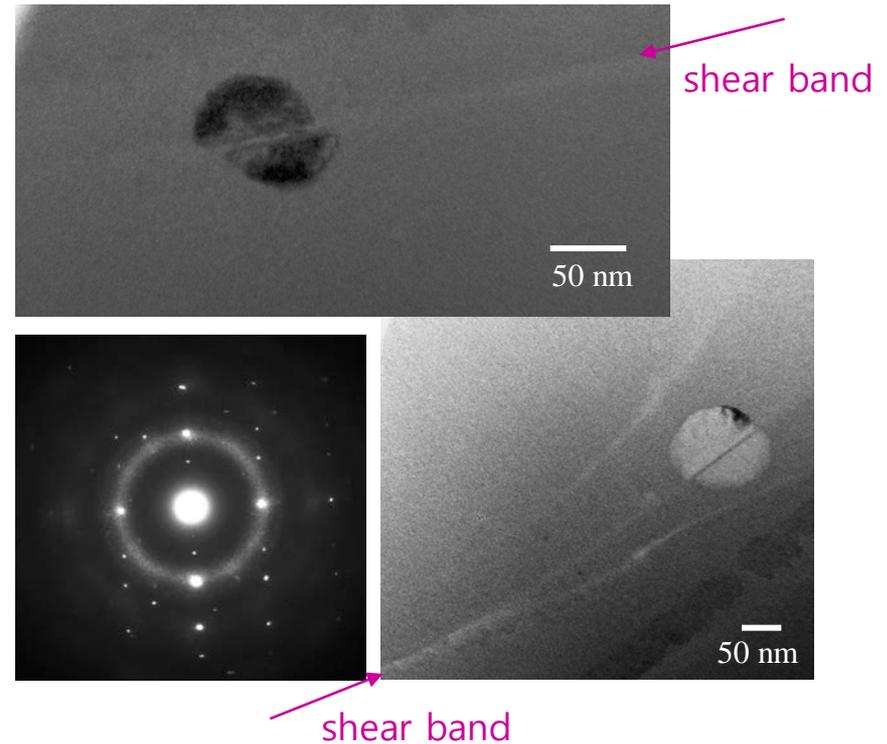


I-phase particle in amorphous matrix

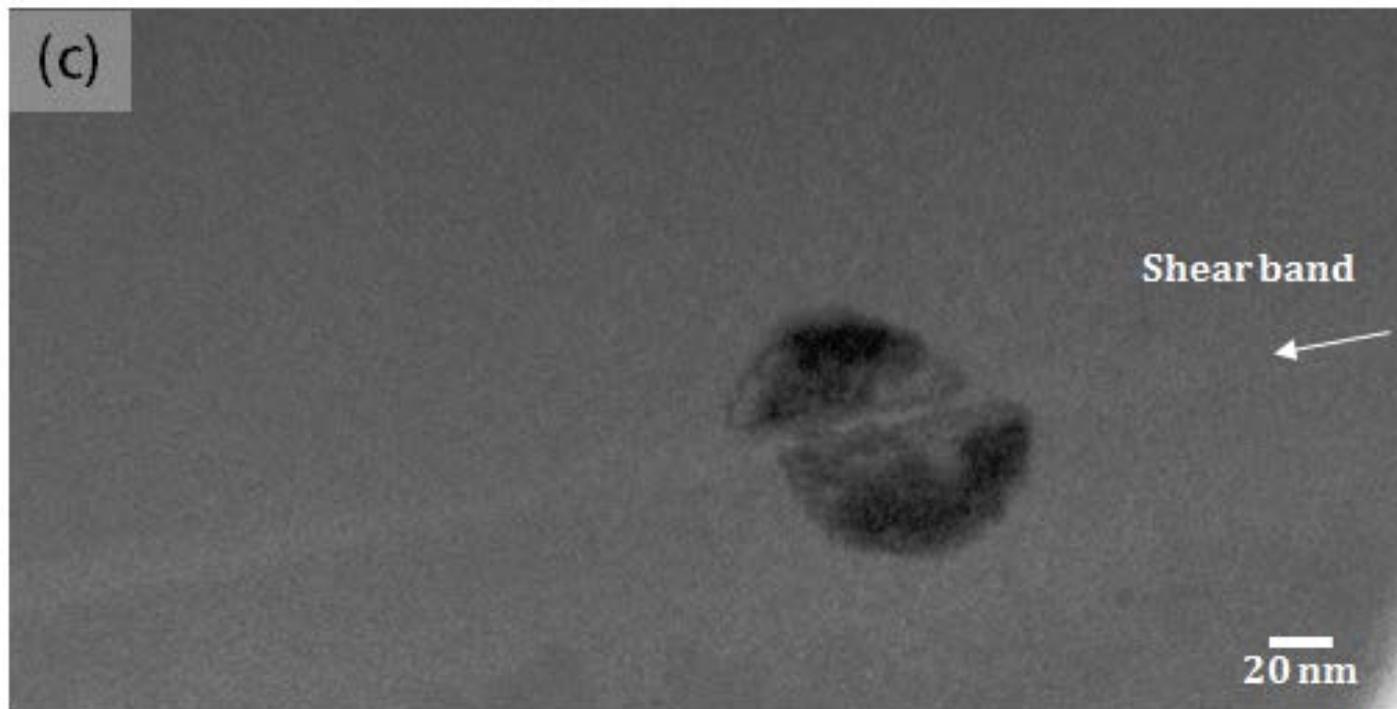
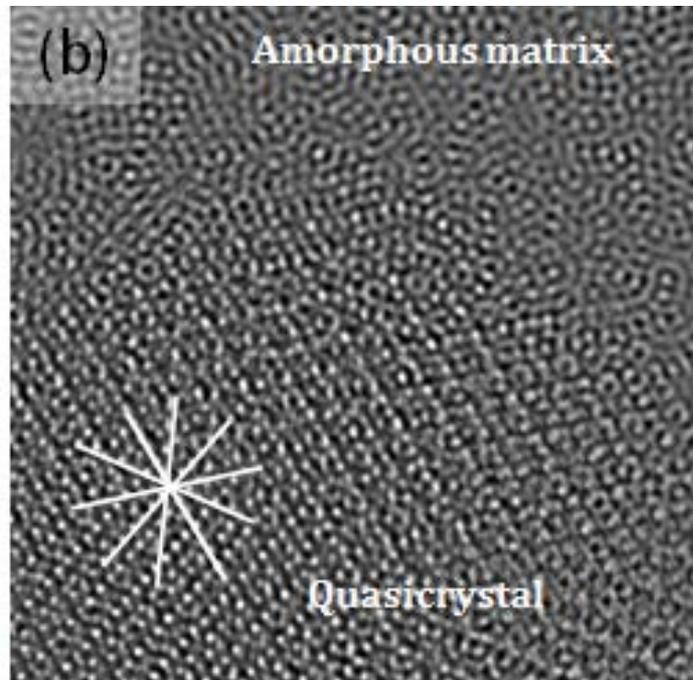
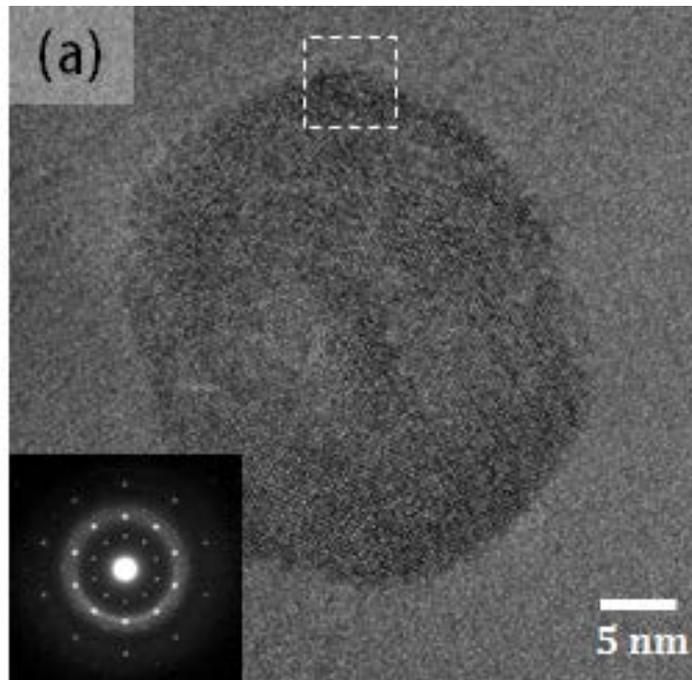
Before deformation



After deformation

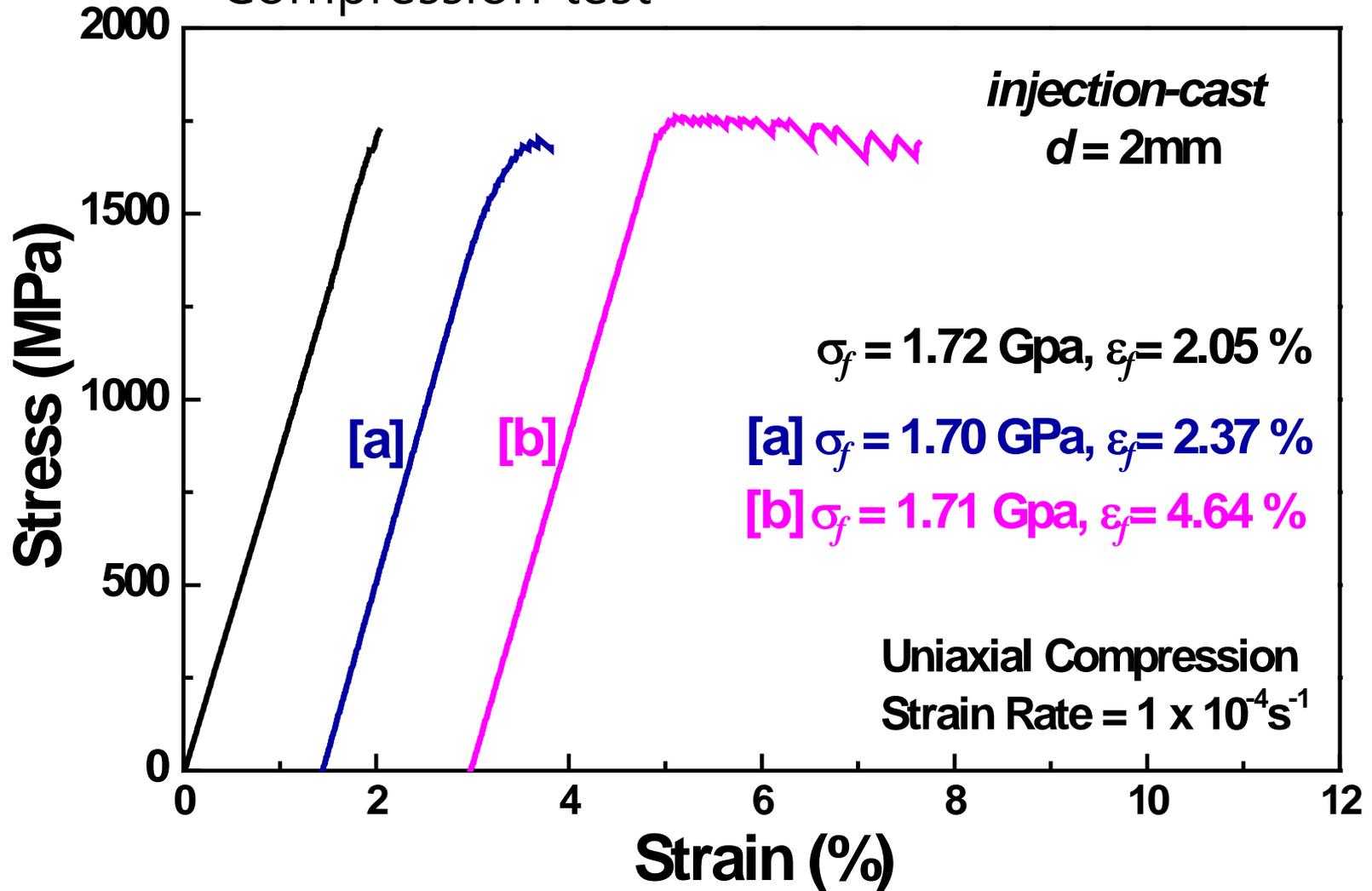


- No distribution of icosahedral particle to blocking the propagation of shear band.
- No enhancement of plasticity in MGMC with icosahedral particle

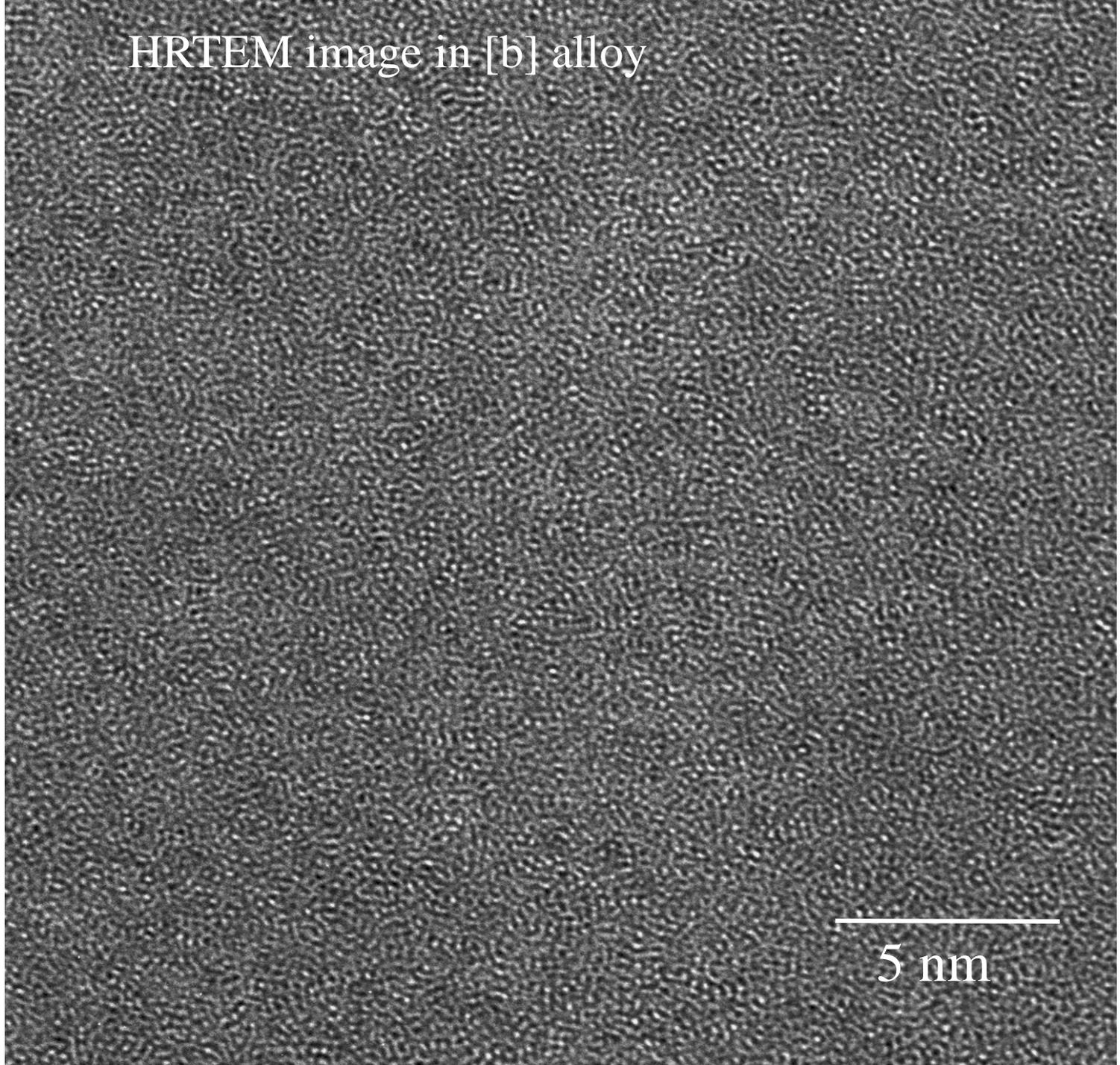


• Compression test

* unpublished (2010)

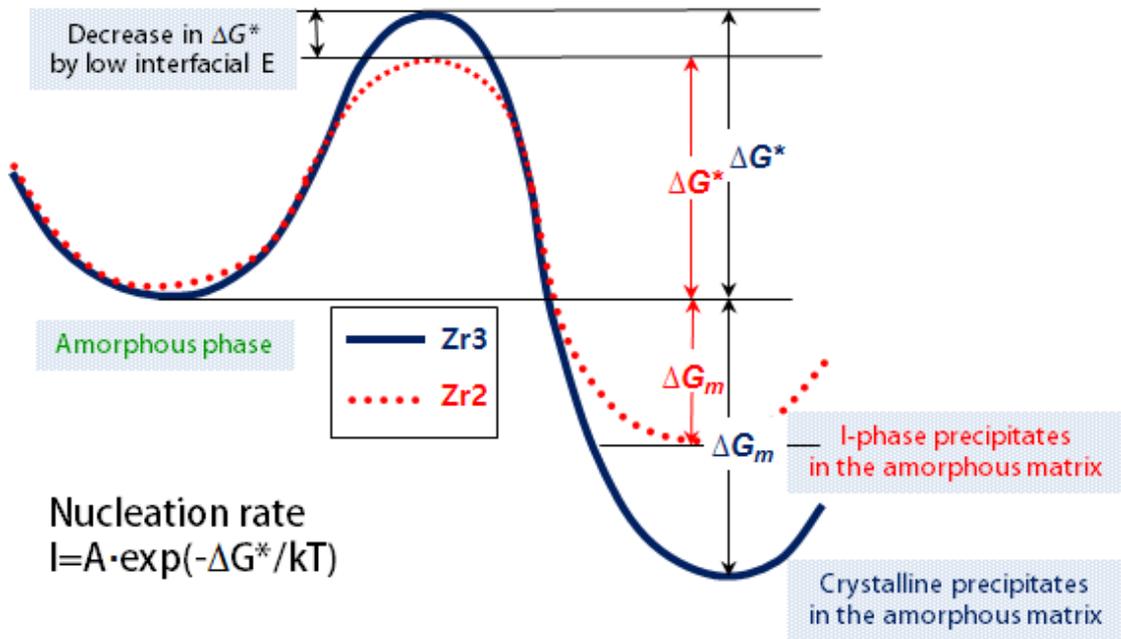
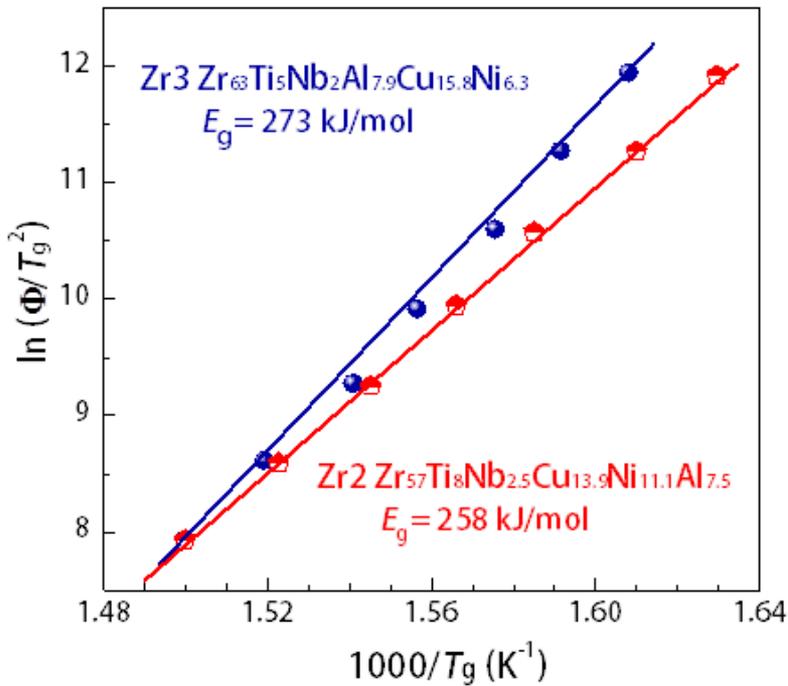


HRTEM image in [b] alloy



5 nm

- Activation E : driving force for nucleation



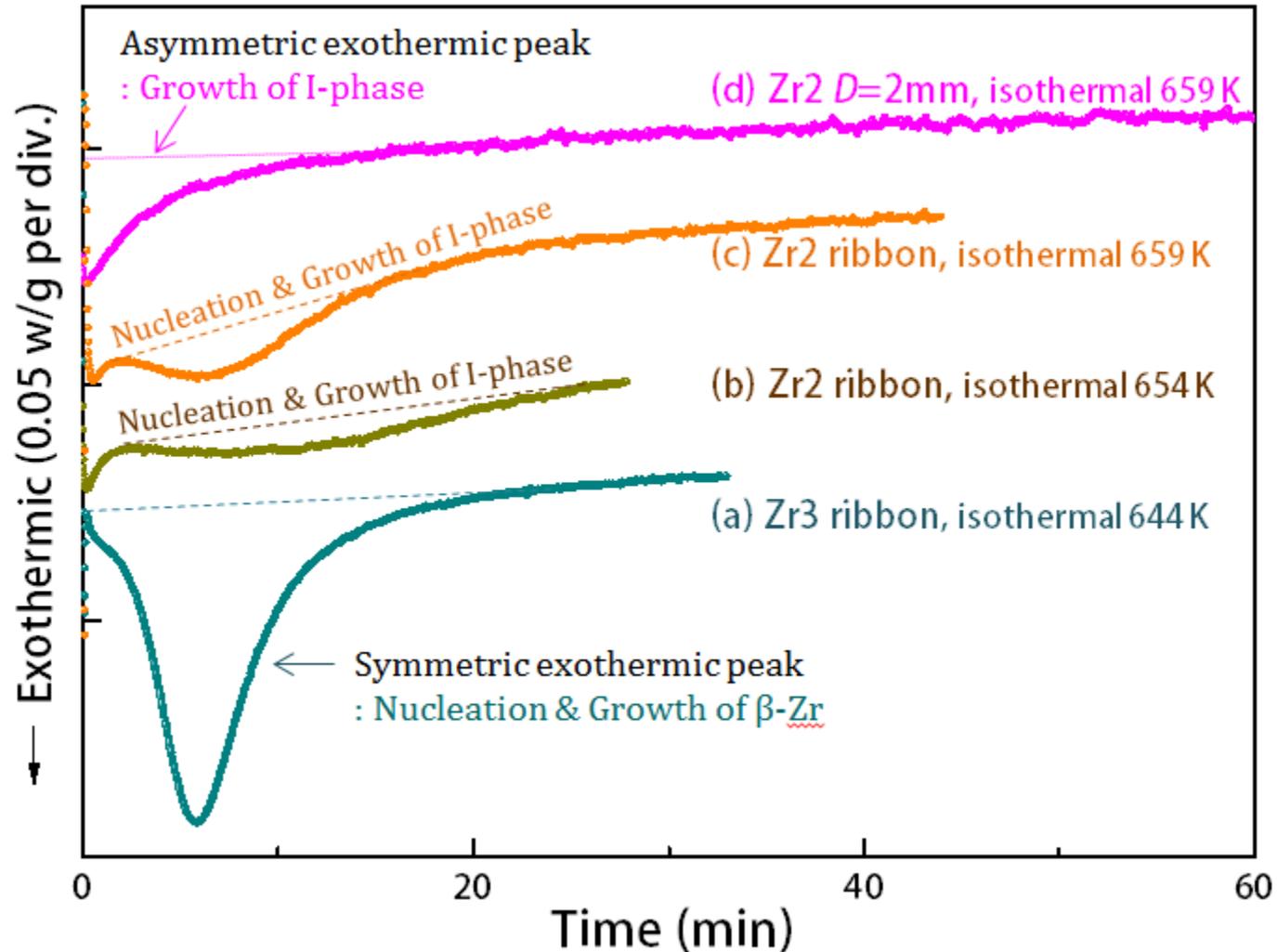
Nucleation rate
 $I = A \cdot \exp(-\Delta G^*/kT)$

Kissinger's equation
 $\ln(\Phi / T_g^2) = -Q / RT_g + const.$

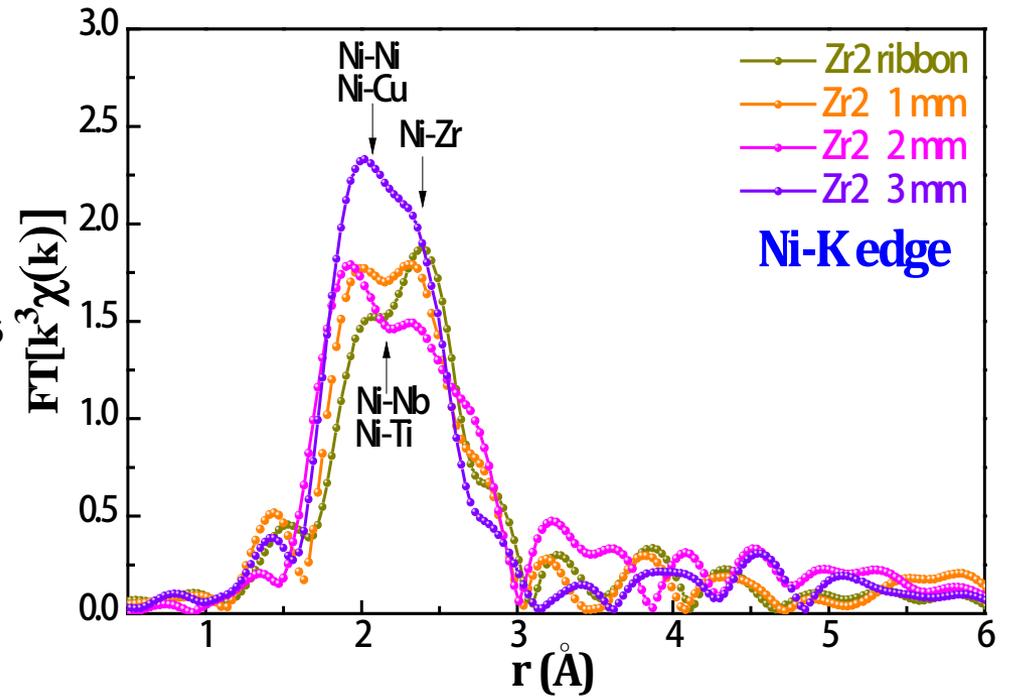
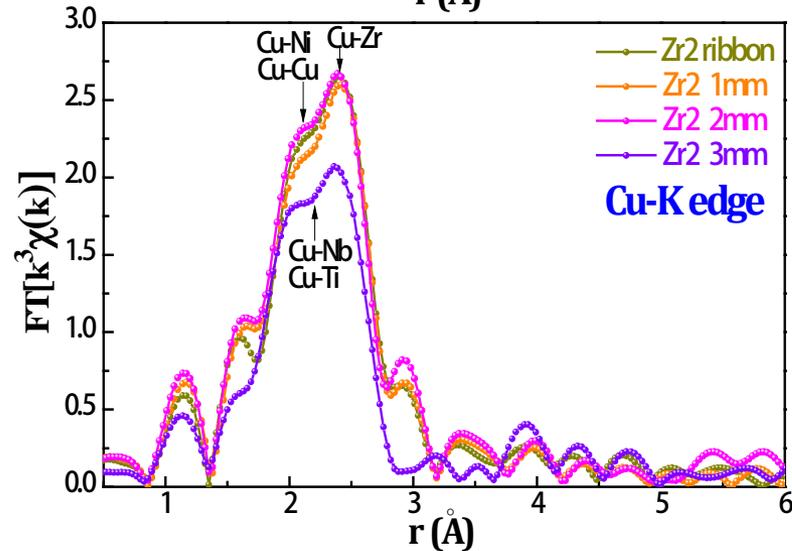
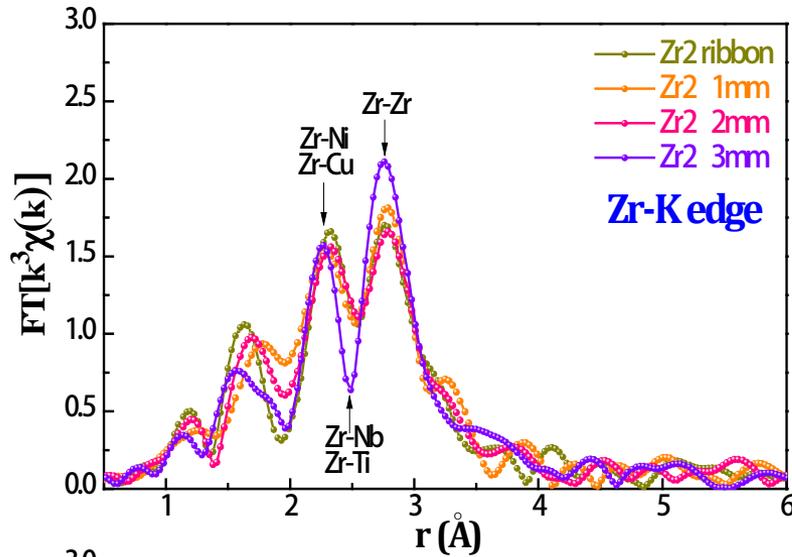
Effect of quenched-in quasicrystal nuclei

● Isotherm in DSC

Isothermal annealing



Characterizing the structure – Measurement of radial distribution function, also called pair distribution function by EXAFS analysis



5.2.3 Structural Details

- **Amorphous vs Nanocrystalline**

1) *Microstructural observation*

XRD, (HR)TEM, EXAFS ...

2) *Thermal analysis*

DSC (Differential Scanning Calorimetry)

: Measure heat absorbed or liberated during heating or cooling

cf) a) *glass* → *nucleation & growth*
(perfect random)

b) *local clustering: quenched-in nuclei* → *only growth*

c) *Nanocrystalline* → *growth*

5.3 Crystallization Modes in Melt-Spun Ribbons

Variables: solid solubility, number of stable & metastable intermetallic phases, composition

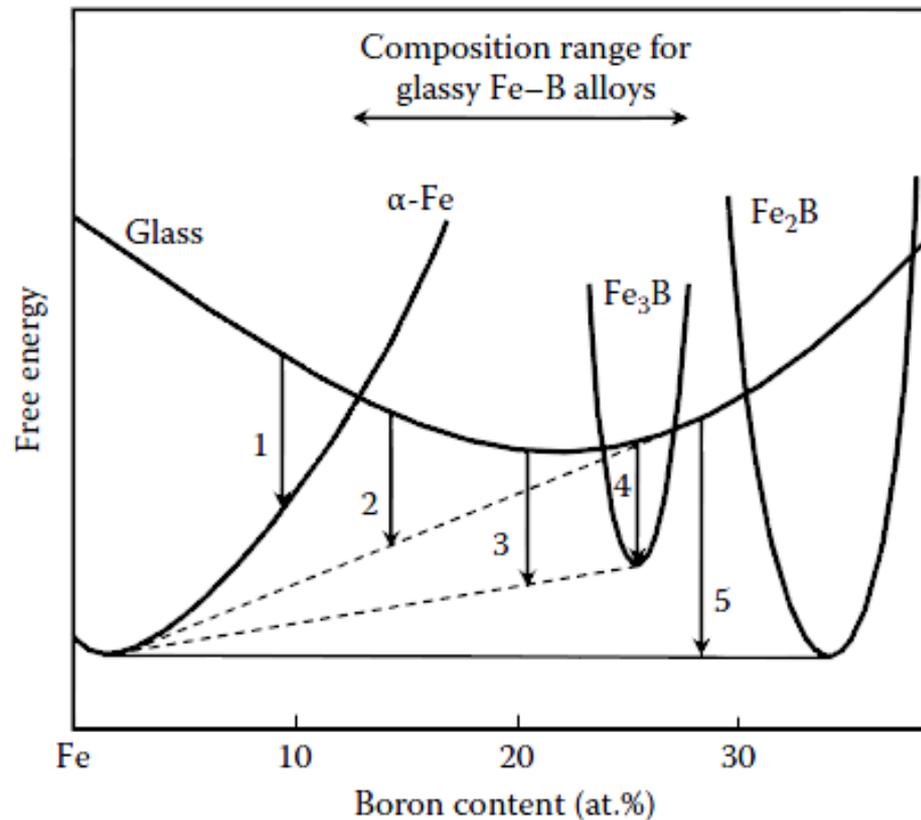


FIGURE 5.5

Hypothetical free energy vs. composition diagram for the Fe-rich Fe-B alloy system. The variation of free energy with composition is represented for the equilibrium α -Fe solid solution and the Fe_2B phases and the metastable Fe_3B phase and the glassy phase. The use of the common tangent approach will help in determining the compositions of the individual phases. The solid common tangent line represents the stable equilibrium between α -Fe and Fe_2B phases, while the dotted common tangent lines represent the metastable equilibrium between α -Fe and Fe_3B phases and α -Fe and glassy phases.

THERMODYNAMICS OF CRYSTALLIZATION

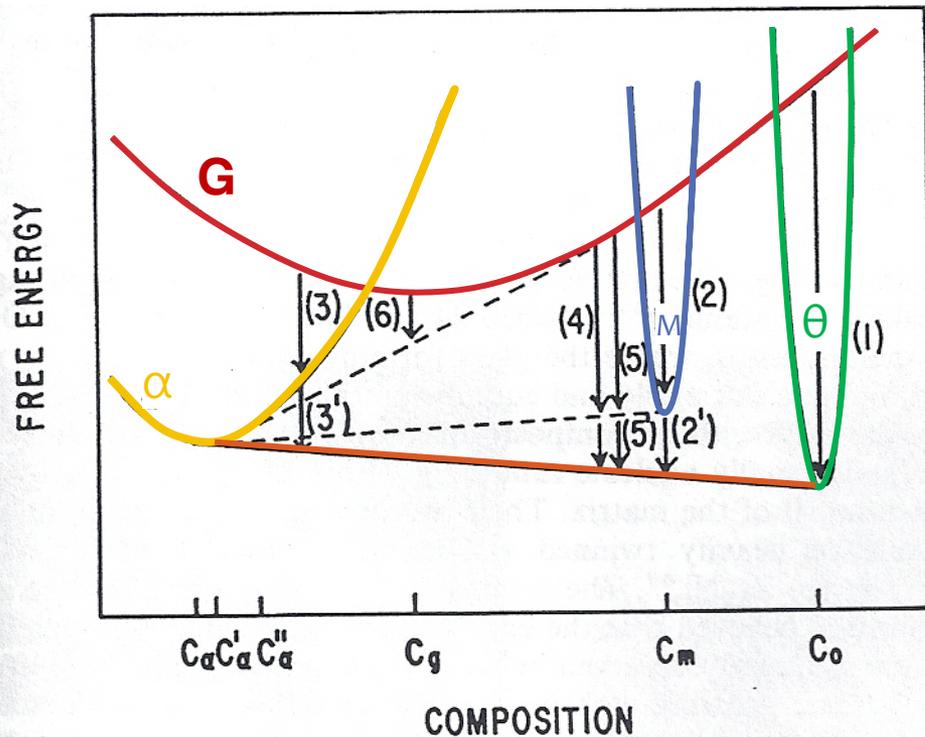


Figure 10.7 Hypothetical free energy diagram to illustrate the crystallization of a metallic glass. G, α , θ , M are respectively the free energy curves of the glass, a terminal solid solution, a stable inter-metallic phase, and a metastable phase. Stable equilibrium is indicated by the solid line; metastable equilibrium by the broken lines. The numbered arrows refer to the devitrification processes described in the text

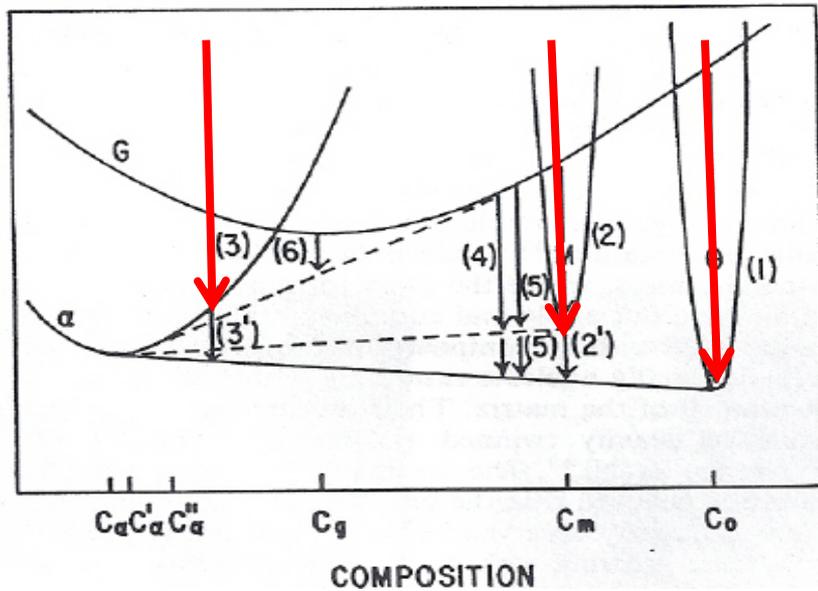
Crystallization mechanisms

(a) Polymorphous transformation of the glass to a crystalline phase of the **same composition**.

The product may be either θ (1) or M(2) or a supersaturated solid solution α (3).

In the latter two cases subsequent decomposition can occur to the **equilibrium mixture of α and θ (2' and 3')**

1. Polymorphous Crystallization: single crystalline phase without any change in composition

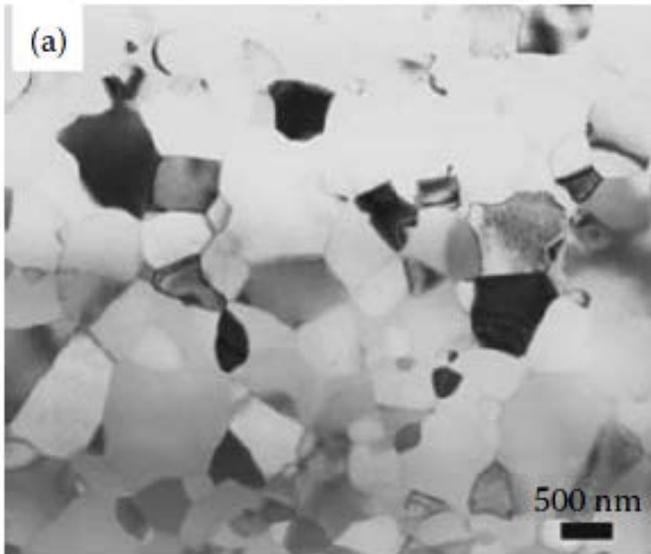


Growth rates and morphologies

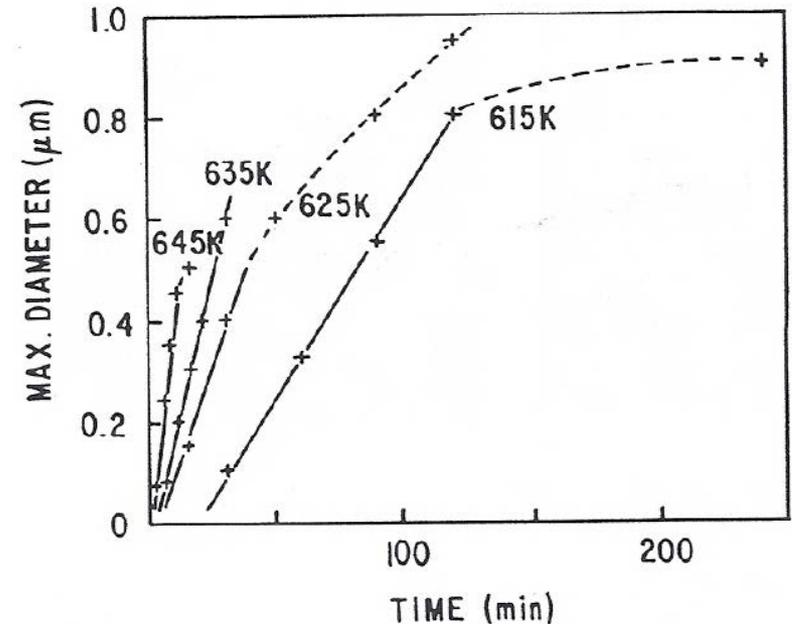
$$u = a_0 v_0 \left\{ \exp \left[\frac{-\Delta F_a}{kT} \right] \right\} \left\{ 1 - \exp \left[\frac{-\Delta F_v}{kT} \right] \right\}$$

ΔF_a = activation energy for an atom to leave the matrix and attach itself to the growing phase

ΔF_v = The molar free energy difference btw C and G



Polymorphous crystallization in a $\text{Ti}_{50}\text{Ni}_{25}\text{Cu}_{25}$ BMG alloy on annealing for 28 min at 709 K.



Growth kinetics of Zr_2Ni crystals in glass of same composition. The broken lines indicate crystal impingement.

THERMODYNAMICS OF CRYSTALLIZATION

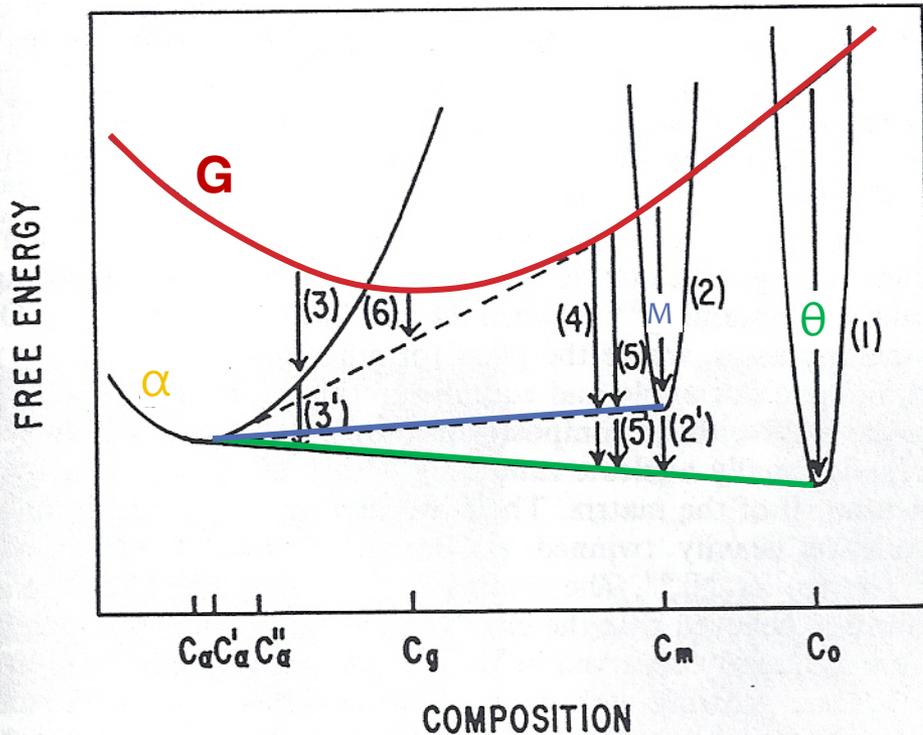


Figure 10.7 Hypothetical free energy diagram to illustrate the crystallization of a metallic glass. G , α , θ , M are respectively the free energy curves of the glass, a terminal solid solution, a stable inter-metallic phase, and a metastable phase. Stable equilibrium is indicated by the solid line; metastable equilibrium by the broken lines. The numbered arrows refer to the devitrification processes described in the text

Crystallization mechanisms

(b) Eutectic crystallization of liquids

The glass can reduce its free energy to a point on the **common tangent** between either α and θ (4) or α and M (5).

In the case of the metastable eutectic between α and M subsequent further decomposition to α and θ can occur. (4' and 5')

THERMODYNAMICS OF CRYSTALLIZATION

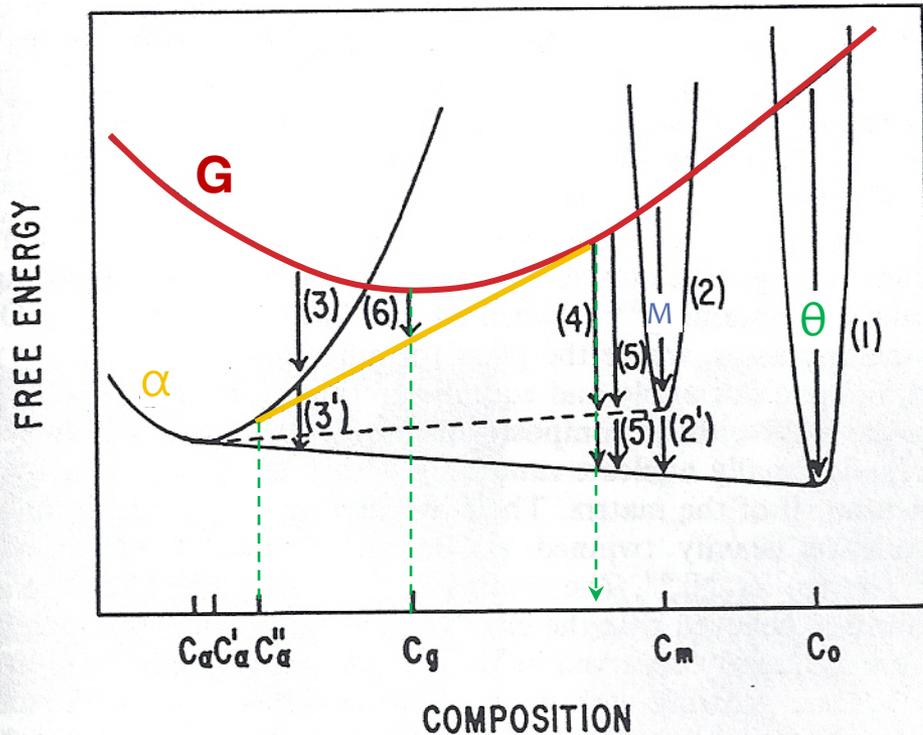


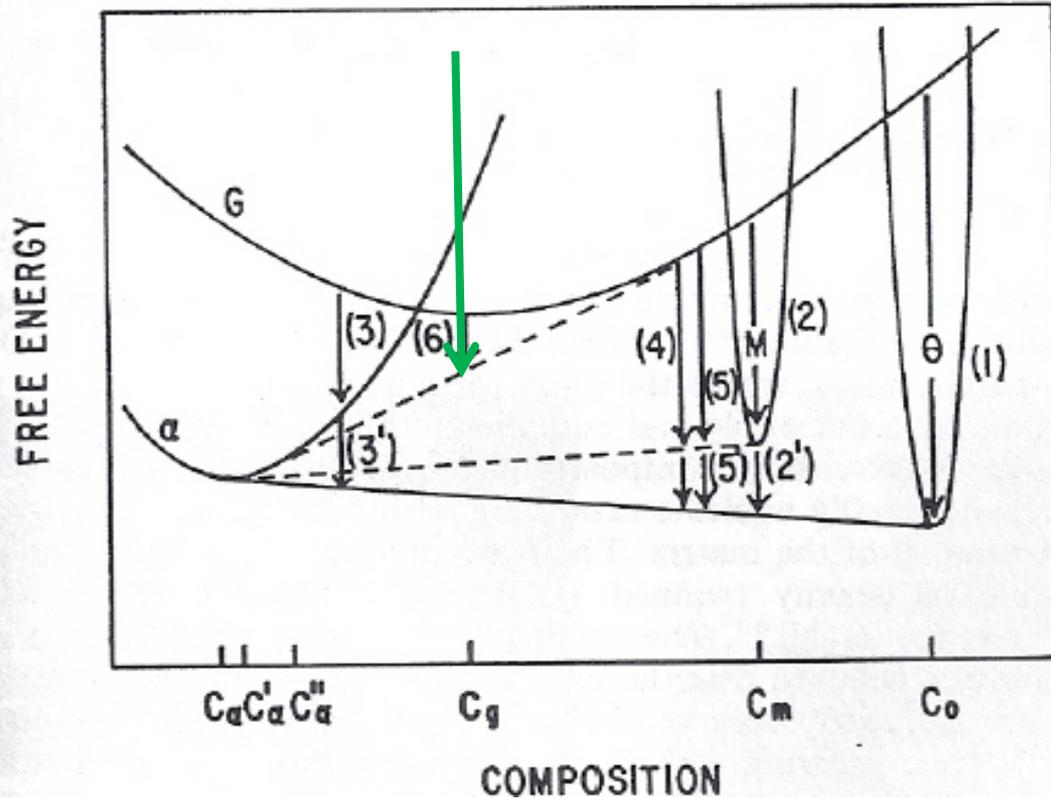
Figure 10.7 Hypothetical free energy diagram to illustrate the crystallization of a metallic glass. G, α , θ , M are respectively the free energy curves of the glass, a terminal solid solution, a stable inter-metallic phase, and a metastable phase. Stable equilibrium is indicated by the solid line; metastable equilibrium by the broken lines. The numbered arrows refer to the devitrification processes described in the text

Crystallization mechanisms

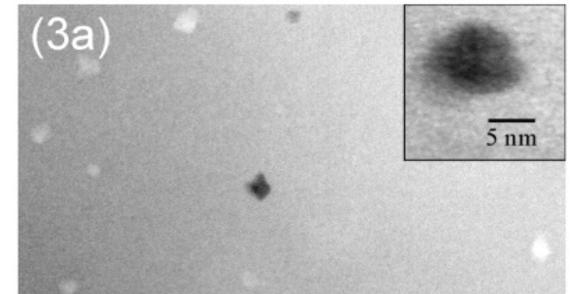
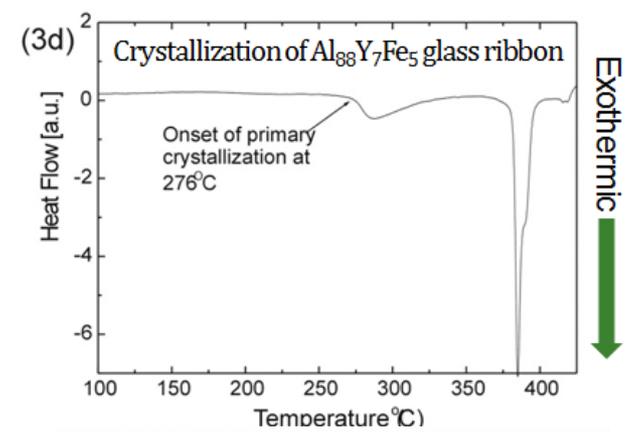
(c) Primary crystallization of supersaturated solid solution (6)

Since the α has a composition c_α which is less than that of the glass c_g solute is rejected from the growing crystals into the glass (4). Ultimately the untransformed, enriched glass (4) transforms by one of the other mechanisms discussed above.

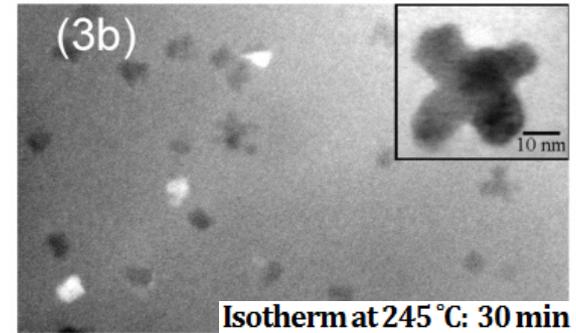
3. Primary Crystallization



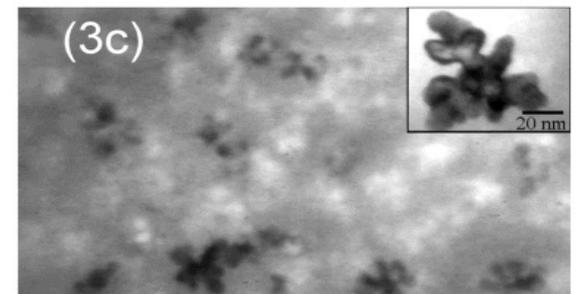
- Forms first from the glass phase
- Supersaturated solid solution
- Since the concentration of the solute in the α -Fe phase is lower than that in the glassy phase, the solute (boron) atoms are rejected into the glassy phase and consequently the remaining glass phase becomes enriched in B until further crystallization is stopped.



Isotherm at 245 °C: 10 min



Isotherm at 245 °C: 30 min



Isotherm at 245 °C: 100 min