2021 Fall

“Phase Transformation *in* Materials”

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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

Contents for today's class

< Phase Transformation in Solids >

2) Diffusionless Transformation

Q1: What is a martensitic transformation?

Q2: Microstructural characteristics of martensite?

Q3: Driving Forces of Martensitic transformation?

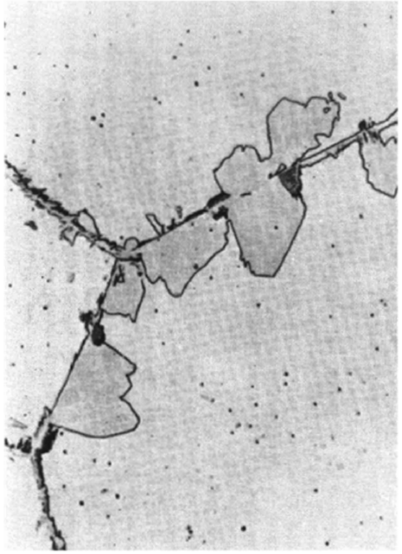
Q4: Why tetragonal Fe-C martensite?

Q5: Martensite crystallography (Orientation btw M & γ)

Q6: Mechanisms for martensitic transformations?

Massive vs. Martensitic Transformations

- There are two basic types of *diffusionless* transformations.
- One is the ***massive transformation***. In this type, a diffusionless transformation takes place ① without a definite orientation relationship. The interphase boundary (between parent and product phases) migrates so as to allow the new phase to grow. It is, however, a ② civilian transformation because the atoms move individually.

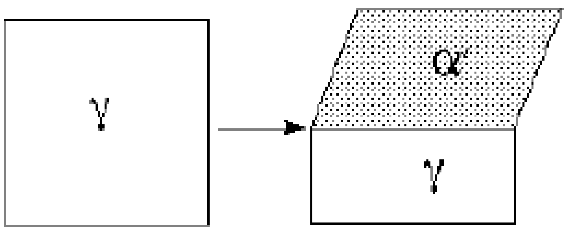


- The other is the ***martensitic transformation***. In this type, the change in phase involves a ① definite orientation relationship because the atoms have to ② move in a coordinated manner. (Military transformation) There is always a ③ change in shape which means that there is a strain associated with the transformation.

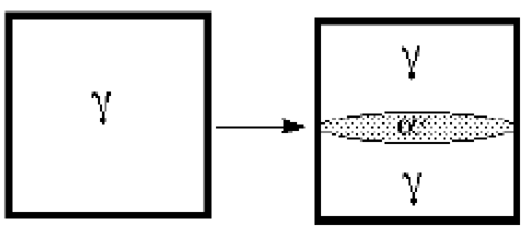
Microstructure of Martensite

- The microstructural characteristics of martensite are:
  - **the product (martensite) phase has a <u>well defined crystallographic relationship with the parent (matrix)</u>.**

1) martensite(designated α') forms as platelets within grains.



Unconstrained transformation



Constrained transformation

"Lens shape" **Plate density: independent of grain size**

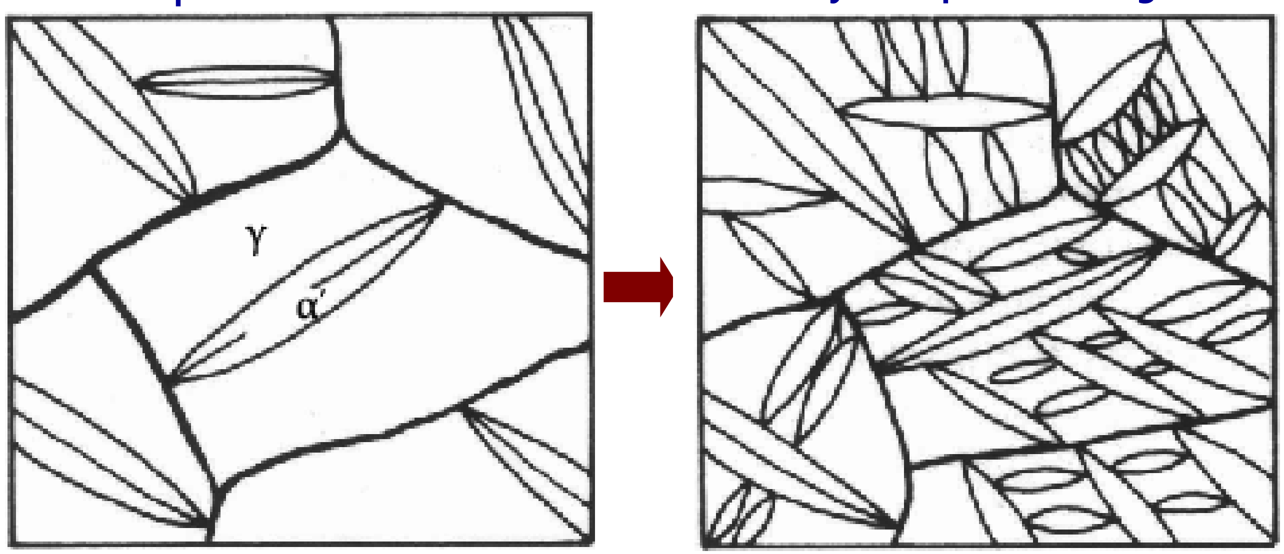


Fig. 6.1 Growth of martensite with increasing cooling below Ms.

<u>→ Martensite formation rarely goes to completion</u>

Microstructure of Martensite

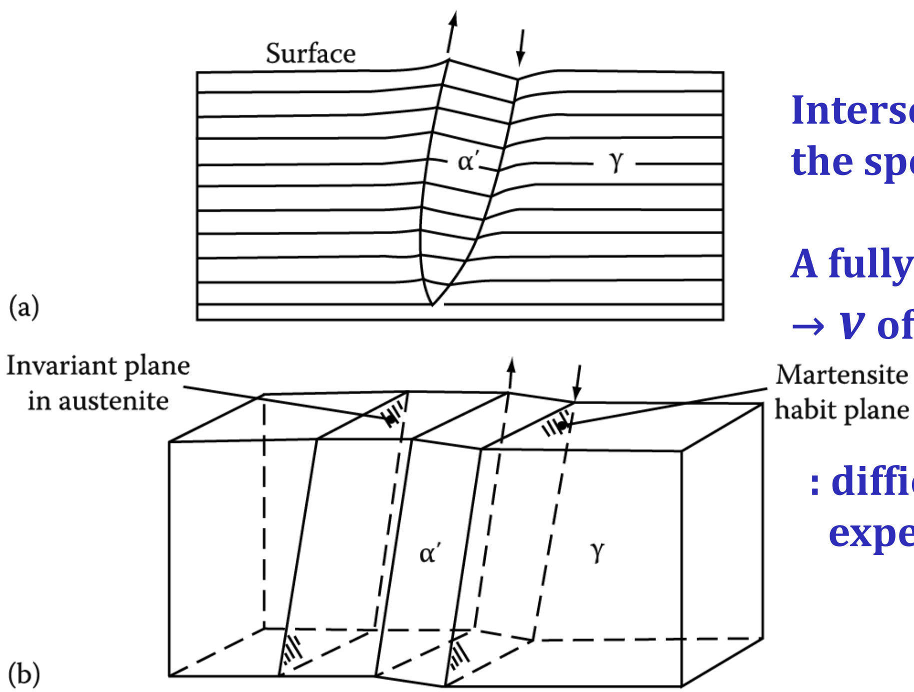
- The microstructural characteristics of martensite are:

  2) each platelet is accompanied by a shape change.

  - the shape change appears to be a simple shear parallel to a habit plane (the common, coherent plane between the phases) and a "uniaxial expansion (dilatation) normal to the habit plane".

strain associated with the transformation

Polished surface_elastic deformation or tilting → but, remain continuous after the transformation



Intersection of the lenses with the surface of the specimen does not result in any discontinuity.

A fully grown plate spanning a whole grain ~10^{-7} sec
→ v of α'/γ interface ∝ speed of sound in solid

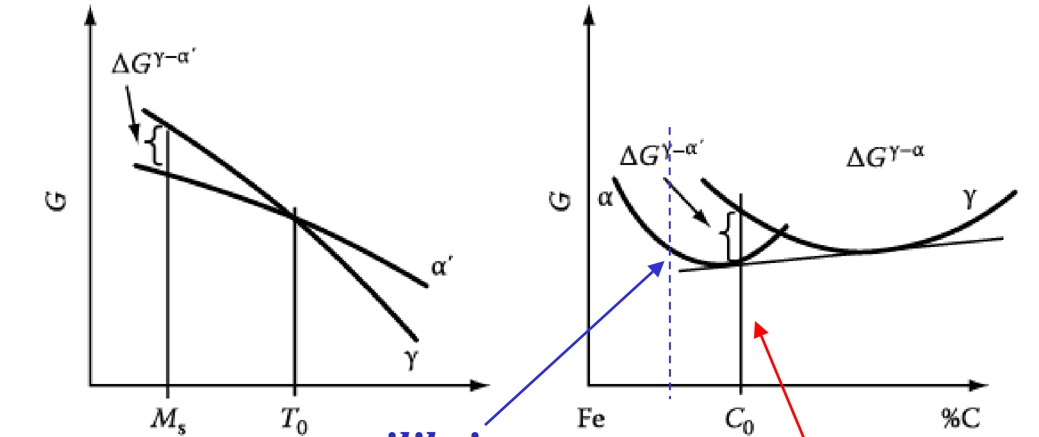
: difficult process to study M nucleation and growth experimentally

Fig. 6.2 Illustrating how a martensite plate remains macroscopically coherent with the surrounding austenite and even the surface it intersects.

Q3: Driving Forces of Martensitic transformation?

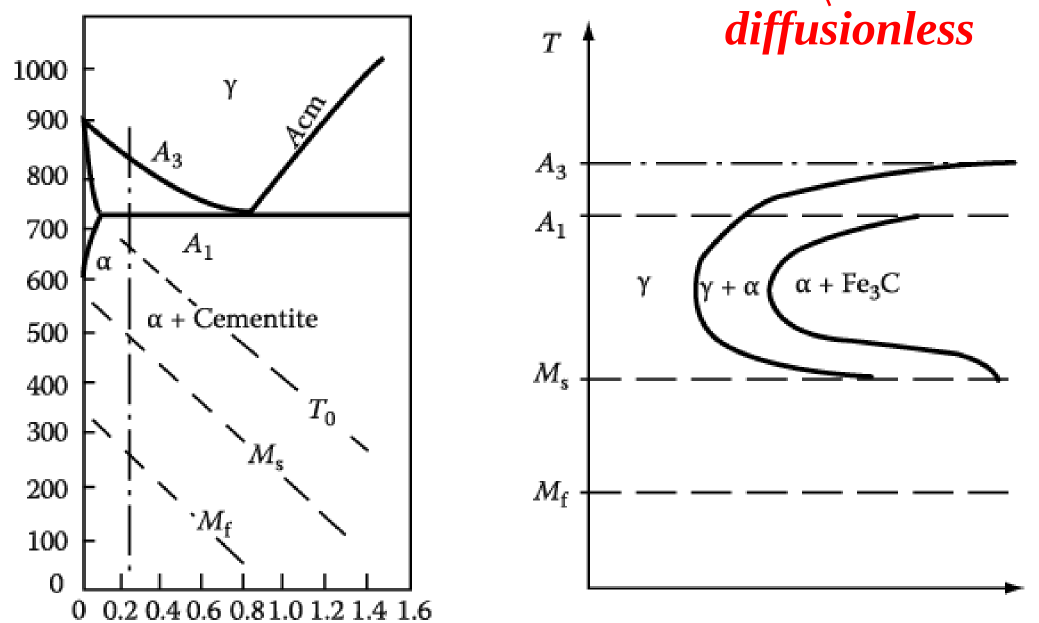
Various ways of

showing the martensite transformation



(a) G-T diagram *equilibrium* (b) G-X diagram for C_0 at M_s

diffusionless



(c) C_0 C (%) (d) Log time

Fe-C phase diagram
Variation of $T_0/M_s/M_f$

TTT diagram
for alloy C_0 in (c)

Note that the M_s line is horizontal in the TTT diagram; also, the M_f line.

Some retained austenite can be left even below M_f. In particular, as much as 10%-15% retained austenite is a common feature of especially the higher C content alloys such as those used for ball bearing steels.

Q4: Why tetragonal Fe-C martensite?

Interstitial sites
for C in Fe

fcc:
carbon occupies the octahedral sites

bcc:
carbon occupies the octahedral sites

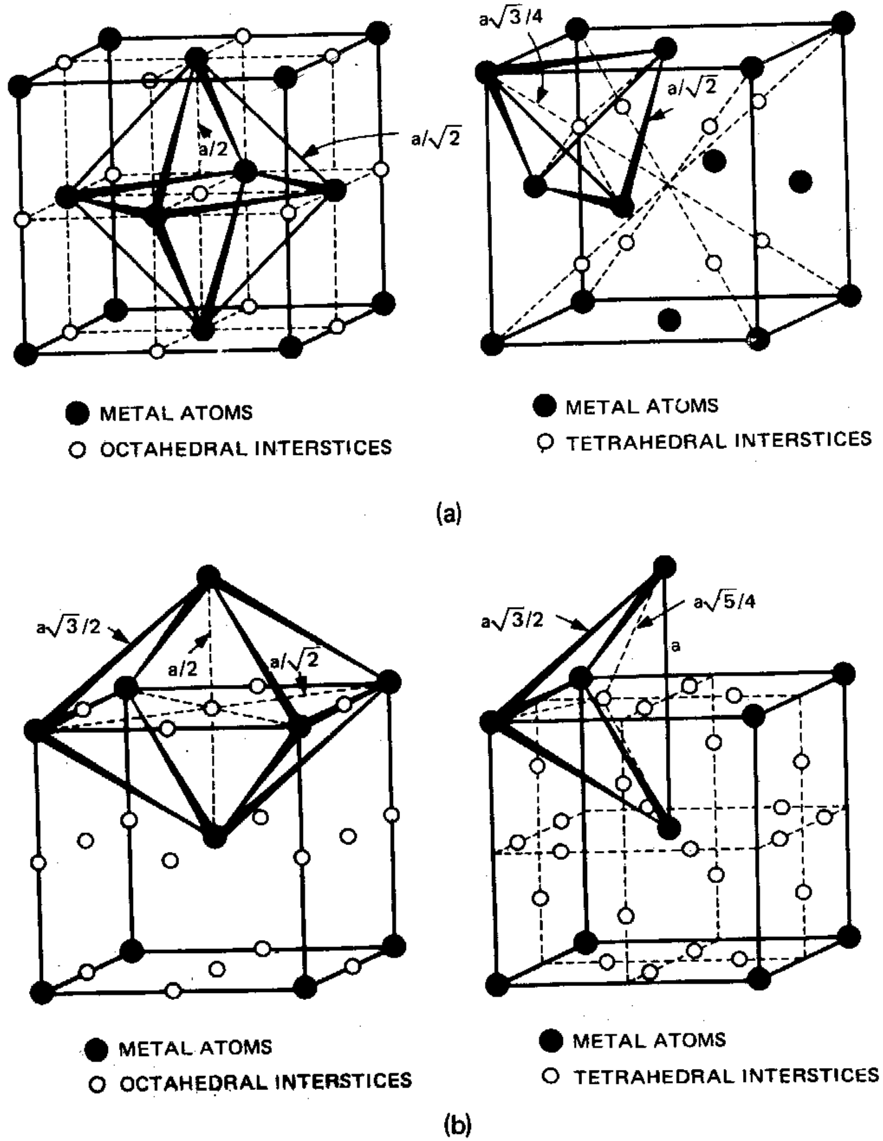


Figure II-1. Interstitial voids in iron. (*a*) Interstitial voids in the fcc structure, octahedral (1) and tetrahedral (2). (*b*) Interstitial voids in the bcc structure; octahedral (1) and tetrahedral (2). (From C.S. Barrett and T.B. Massalski, *Structure of Metals*, 3d ed., copyright 1966, used with the permission of McGraw-Hill Book Co., New York.)

[Leslie]

Carbon in BCC α ferrite

[P&E]

- One consequence of the occupation of the octahedral site in ferrite is that the carbon atom has <u>only two nearest neighbors</u>.
- Each carbon atom therefore distorts the iron lattice in its vicinity.
- The distortion is a *<u>tetragonal distortion</u>*.
- If all the carbon atoms occupy the *same type of site* then the entire lattice becomes <u>tetragonal, as in the martensitic structure</u>.
- Switching of the carbon atom between adjacent sites leads to <u>strong internal friction peaks</u> at characteristic temperatures and frequencies.

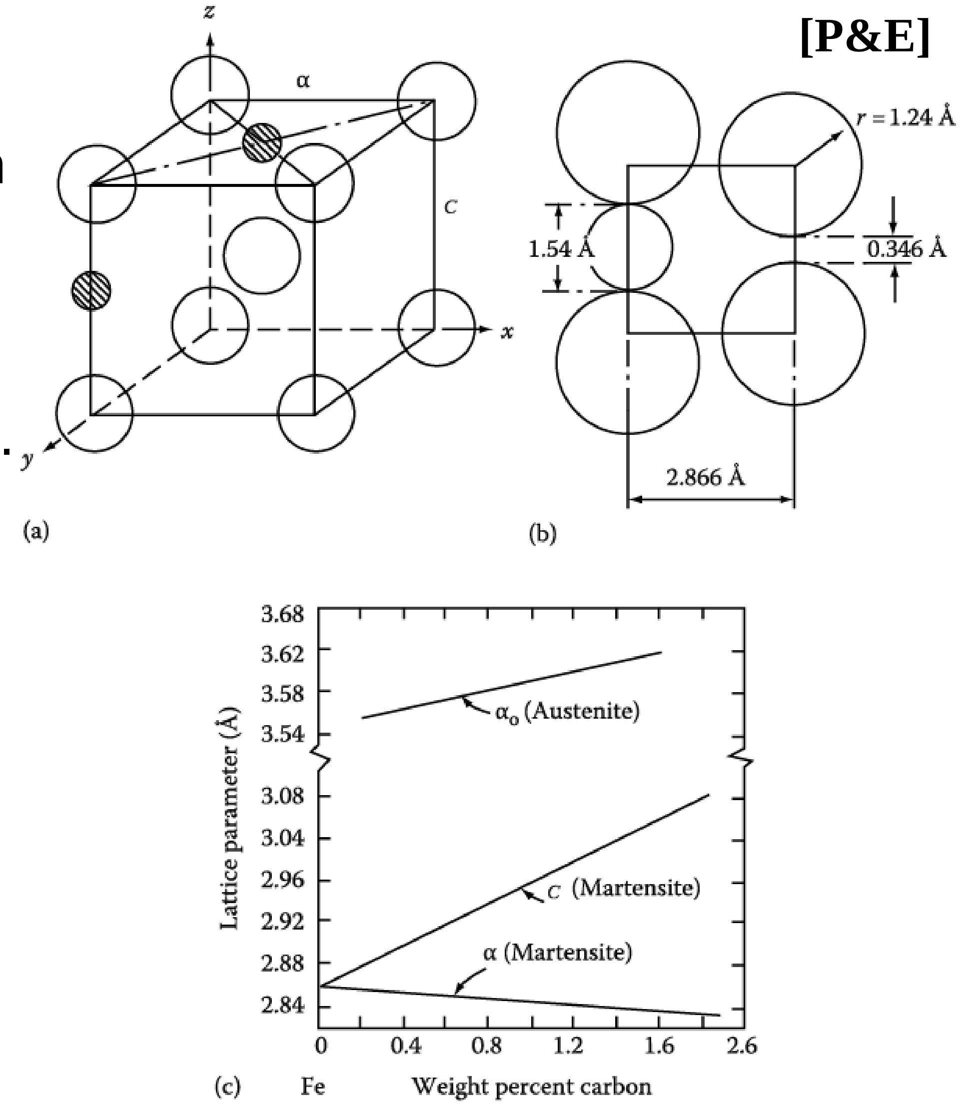


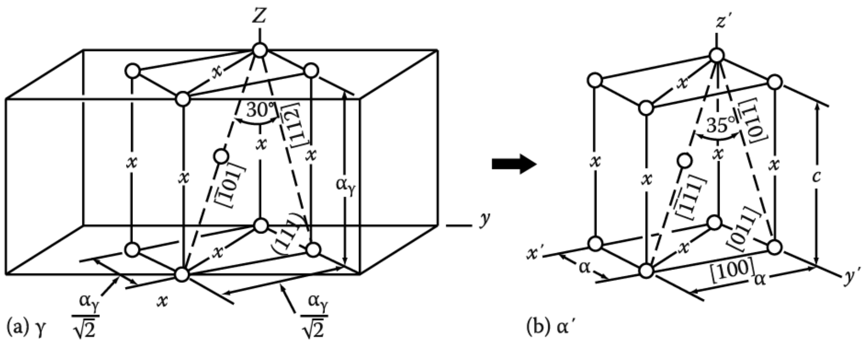
Fig. 6.5 Illustrating (a) possible sites for interstitial atoms in <u>bcc lattice</u>, and (b) the <u>large distortion necessary</u> to accommodate a carbon atom (1.54 Å diameter) compared with the space available (0.346 Å). (c) Variation of a and c as a function of carbon content.

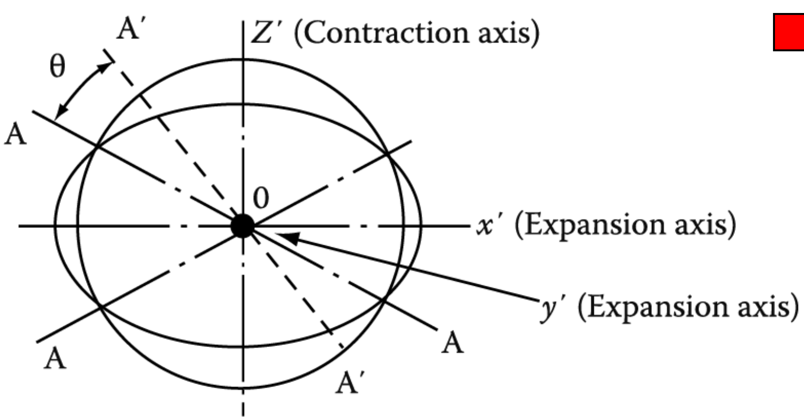
Q5. Martensite crystallography (Orientation btw M & γ) 6.2.절

$\gamma \rightarrow \alpha'$: (1) Habit plane of M: not distorted by transformation

(2) A homogeneous shear (s) parallel to the habit plane

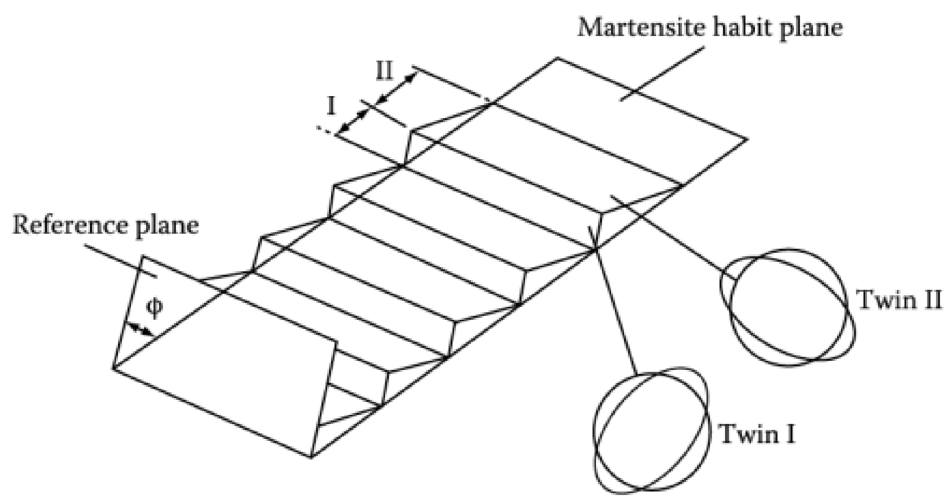
(3) ~4% expansion_dilatation normal to the habit plain (lens)





Bain Model for martensite

Applying the twinning analogy to the Bain model,



Twins in Martensite

may be self-accommodating and reduce energy by having alternate regions of the austenite undergo the Bain strain along different axes.

Q6:

Mechanisms for martensitic transformations?

- The mechanisms of military transformations are <u>not entirely clear.</u>

- **Why does martensite require <u>heterogeneous nucleation</u>?**

  The reason is the <u>large critical free energy for nucleation</u> outlined above

- <u>Possible mechanisms for martensitic transformations</u> include

  (a) dislocation based
  (b) shear based

(a) • ***Dislocations* in the parent phase (austenite) clearly provide sites for heterogeneous nucleation.**

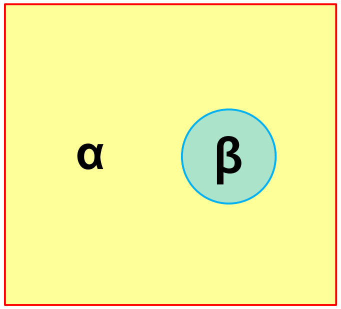
- **Dislocation mechanisms are thought to be important for *propagation/growth of martensite platelets or laths*.**

(b) • Martensitic transformations strongly constrained by crystallography of the parent and product phases.

- This is analogous to slip (dislocation glide) and <u>twinning</u>, especially the latter.

6.3.1 Formation of Coherent Nuclei of Martensite (Homogenous nucleation)

Free Energy Change Associated with the Nucleation



Negative and Positive Contributions to ΔG?

1) Volume Free Energy : $-V\Delta G_V$

2) Interface Energy : $A\gamma$

3) Misfit Strain Energy : $V\Delta G_S$

$$\Delta G = \boxed{-V\Delta G_V} + \boxed{A\gamma \underset{<}{+} V\Delta G_S}$$

This expression does not account for possible additional energies (e.g. thermal stresses during cooling, externally applied stresses, and stresses produced ahead of rapidly growing plates, etc).

In M transformations, the strain energy (ΔG_s) of the coherent nucleus is much more Important than the surface energy, since the shear component of the pure Bain strain is as high as S = 0.32 which produces large strains in the surrounding austenite. However, the interfacial (surface) energy (γ) of a fully coherent nucleus is relatively small.

(If homogenous nucleation) 6.3.1 Formation of Coherent Nuclei of Martensite

for thin ellipsoidal nucleus (radius a, semi thickness c and volume V),

$$\Delta G = A\gamma + V\Delta G_s - V\Delta G_v$$

Assumption: 1) Nucleation does not necessarily occur at grain boundaries.
2) Nucleation occurs homogeneously without the aid of any other types of lattice defects.

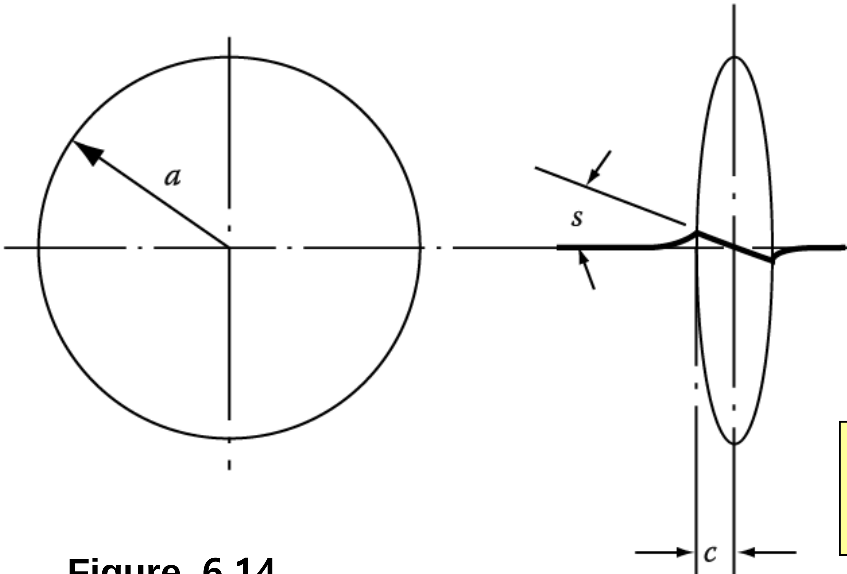


Figure. 6.14
Schematic representation of a M nucleus.

→ The Nucleus forms by a simple shear, S, parallel to the plane of the disc, and complete coherency is maintained at the interface.

$$\Delta G = 2\pi a^2\gamma + 2\mu V (s/2)^2 \frac{2(2-\nu)}{8(1-\nu)} \pi c/a - \frac{4}{3}\pi a^2 c \cdot \Delta G_v$$

If ν=1/3,

$$\Delta G = \underbrace{2\pi a^2\gamma}_{\text{Surface E}} + \underbrace{\frac{16\pi}{3}(s/2)^2 \mu a c^2}_{\text{Elastic E (shear component of strain only)}} - \underbrace{\frac{4\pi}{3} a^2 c \cdot \boxed{\Delta G_v}}_{\text{Volume E}}$$ Eq. (6.7)

6.3.1 Formation of Coherent Nuclei of Martensite

By differentiating Eq. (6.7) with respect to a and c, respectively

→ Min. free energy barrier to nucleation: extremely sensitive to "γ, ΔG_v and s"

$$\boxed{\Delta G^* = \frac{512}{3} \cdot \frac{\gamma^3}{(\Delta G_v)^4} \cdot (s/2)^4 \mu^2 \pi}$$ **(joules/nucleus)**

→ Critical nucleus size (c* and a*): highly dependent to "γ, ΔG_v and s"

$$\boxed{c^* = \frac{2\gamma}{\Delta G_v}} \qquad \boxed{a^* = \frac{16\gamma\mu(s/2)^2}{(\Delta G_v)^2}}$$ **Eq. (6.9) & (6.10)**

For steel, 1) typically $\Delta G_v = 174$ MJm^{-3}, and

2) s (varies according to whether the net shear of a whole plate (e.g. as measured from surface markings) or shear of fully coherent plate (as measured from lattice fringe micrographs)
= 0.2 (macroscopic shear strain in steel)

3) γ = 20 mJm^{-2} (fully coherent nucleus)

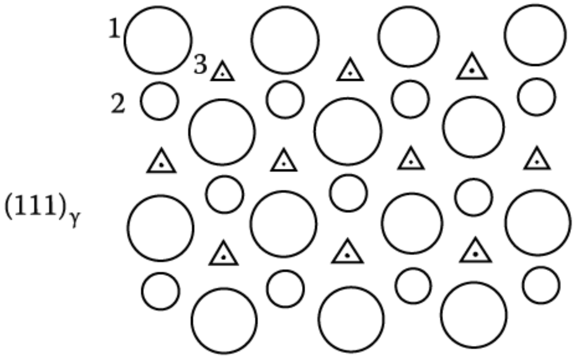
→ c*/a* ~ 1/40, ΔG* ~ 20 eV : <u>too high for thermal fluctuation alone to overcome</u> (at 700 K, kT = 0.06 eV)

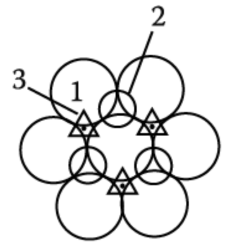
→ "M nucleation = heterogeneous process" : possibly in dislocation

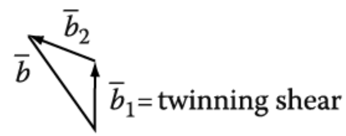
(#= 10^5 per 1 mm^2)

6.3.2 Role of Dislocations in Martensite Nucleation
: based on ① atomic shuffles within the dislocation core

1) Zener: demonstrated how the movement of <112>$_\gamma$ partial dislocations during twinning could generate in thin bcc region of lattice from an fcc one.



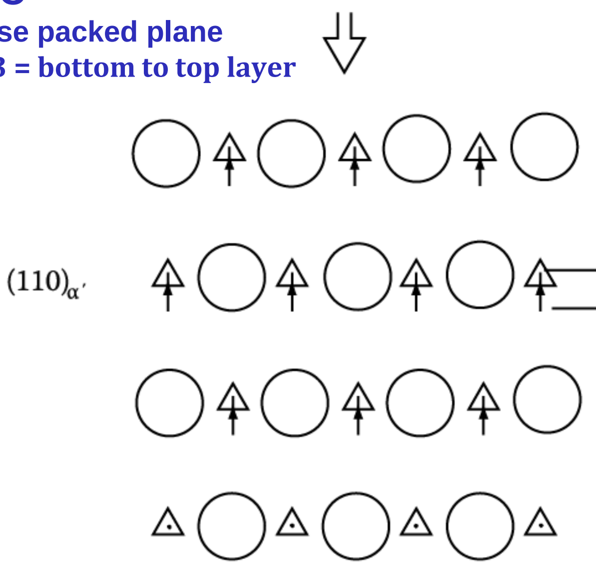


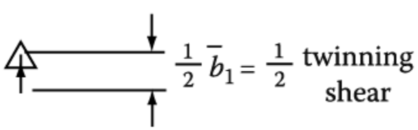


$$\bar{b} = \bar{b}_1 + \bar{b}_2$$

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

FCC
Close packed plane
1→3 = bottom to top layer





In order to generate the bcc structure it requires that all the 'triangular' ① (Level 3) atoms jumps forward by

$$\boxed{\frac{1}{2}\bar{b}_1 = \frac{a}{12}[\bar{2}11]}$$

In fact, the lattice produced is only two atom layer thickness and not quite the bcc one after this shear, but requires an ② additional dilatation to bring about the correct lattice spacings.

The habit plane of the M corresponds to the glide planes of austenite.

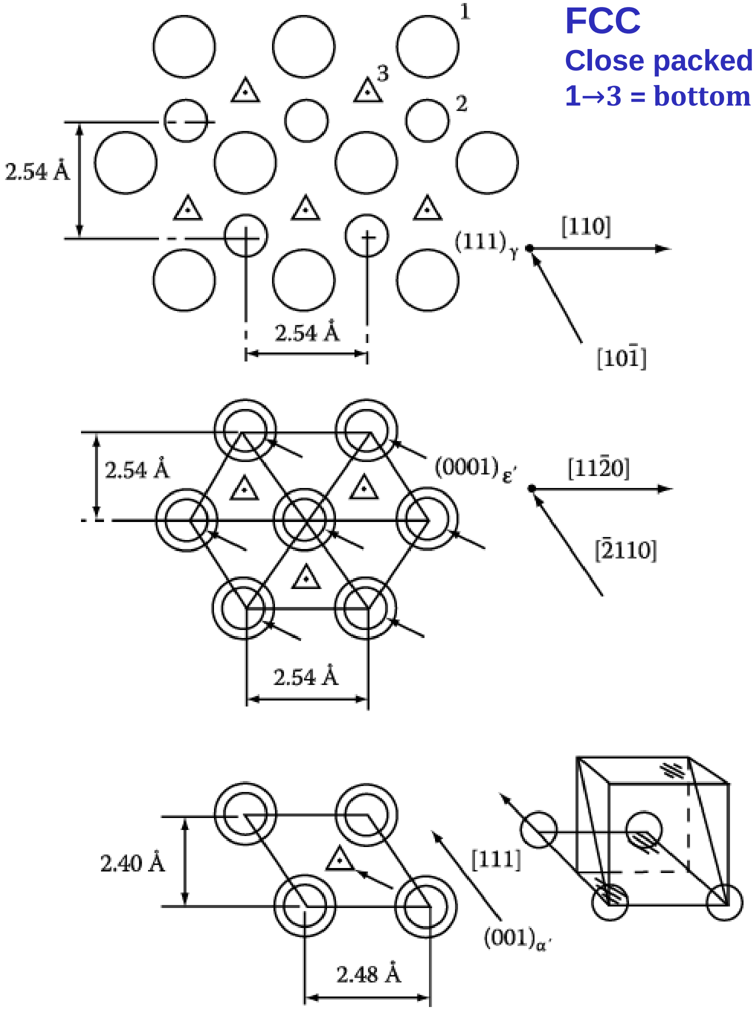
Region with dislocation pile-ups → possible to form thicker M nuclei

Figure. 6.15 Zener's model of the generation of two-atom-thick martensite by a half-twinning shear (①+②)

6.3.2 Role of Dislocations in Martensite Nucleation

2) Venables: M transformation induced by half-twinning shear in fcc mater.

a. in the case of alloys of low stacking fault energy (e.g. steel, etc)



FCC
Close packed plane
1→3 = bottom to top layer

$$\gamma\ (\text{fcc}) \rightarrow\ \varepsilon'\ (\text{hcp}) \rightarrow\ \alpha'\ (\text{bcc})$$

ε'-martensite structure thickens by inhomogenous half-twining shears of $\frac{a}{12}[\bar{2}11]$ on every other {111}γ plane.

→ Indeed, α' regions have been observed to form in conjunction with M.

→ But, no direct evidence of the ε'→α' transition

→ possible γ→ε' and γ→α' by different mechanism

Figure. 6.16 Venables's model for the γ→ε'→α'

6.3.2 Role of Dislocations in Martensite Nucleation

Understanding so far…

- **It is thus seen that some types of M can form directly by the systematic generation and movement of extended dislocations.**

  (M transformation induced by half-twinning shear in fcc mater related to ① atomic shuffles within the dislocation core)

  → M_s temperature : a transition from positive to negative SFE

Limitation…

- **However, 1) this transition type cannot occur in ① high SFE nor in ② thermoelastic martensites**

  2) this transition is also difficult to understand ③ twinned martensite, merely on the basis of dislocation core changes.

→ need to consider alternative way in which dislocations can nucleate martensite other than by changes at their cores.

6.3.3 ② Dislocation strain energy assisted transformation
: help of the elastic strain field of a dislocation for M nucleation

- **Assumption: coherent nuclei are generated by a pure Bain strain, as in the classical theories of nucleation**

The stain field associated with a dislocation can in certain cases provide a favorable interaction with the strain field of the martensite nucleus, such that one of the components of the Bain strain is neutralized thereby reducing the total energy of nucleation.

→ the dilatation associated with the a)extra half plane of the dislocation contributes to the Bain strain.

→ Alternatively, the shear component of the dislocation could be utilized for M transformation.

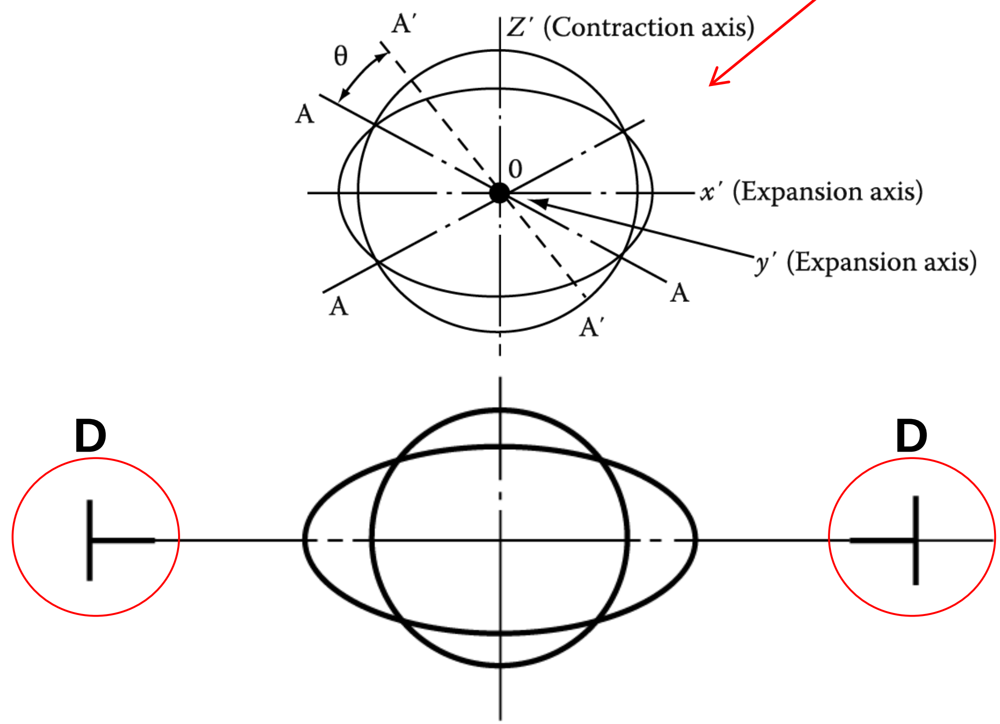


Figure. 6.19 Illustrating how one of the strain components of the Bain deformation may be compensated for by the strain field of a dislocation which in this case is tending to push atom planes together.

6.3.3 ② Dislocation strain energy assisted transformation

: help of the elastic strain field of a dislocation for M nucleation

$$\Delta G = A\gamma + V\Delta G_s - V\Delta G_v - \Delta G_d$$

Creation of nucleus~destruction of a defect($-\Delta G_d$)
→ Dislocation interaction energy which reduces the nucleation energy barrier

$$\Delta G_d = 2\mu s\pi \cdot ac \cdot \bar{b}$$

where $\bar{b}$ = Burgers vector of the dislocation,
s = shear strain of the nucleus

$$\Delta G = 2\pi a^2\gamma + \frac{16\pi}{3}(s/2)^2\mu ac^2 - \frac{4\pi}{3}a^2c \cdot \Delta G_v - 2\mu s\pi ac \cdot \bar{b}$$ **Eq. (6.16)**

전단변형과 bain 변형 포함

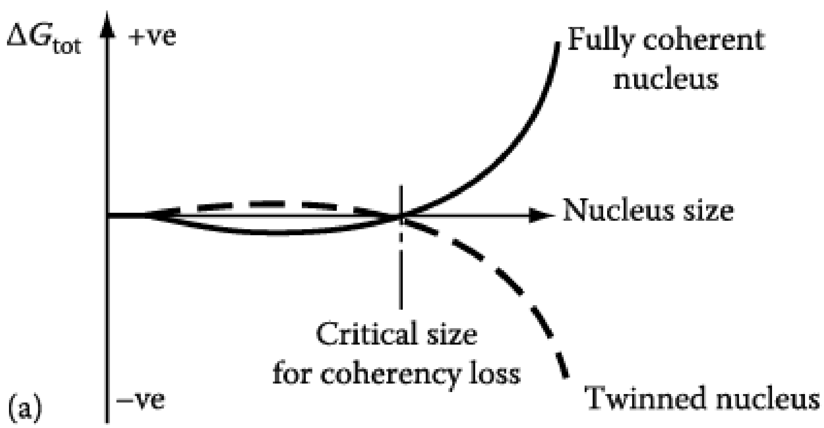


Figure. 6.20 (a) schematic diagram based on Eq. 6.16, illustrating the need for the nucleus to twin if it is to grow beyond a certain critical size.

→ Total energy of martensite nucleus:
as a function of 1) diameter and thickness (a, c)
(whether it is twinned or not (this affect "s"))
2) Degree of assistance from the strain field of a dislocation (or group of dislocations)

e.g. A fully coherent nucleus from partial interaction with the strain field of a dislocation ~ 20 nm dia. & 2-3 atoms in thickness → further growth need to twin and slip formation

6.3.3 ② Dislocation strain energy assisted transformation

: help of the elastic strain field of a dislocation for M nucleation

Burst phenomenon

: auto-catalytic process of rapid, successive M plate formation occurs over a small temperature range , e.g. Fe-Ni alloys (Large elastic stresses set up ahead of a growing M plate → Elastic strain field of the M plate act as the interaction term of elastic strain field of dislocation in Eq. (6.16) → reduces the M nucleation energy barrier)

In summary,

- we have not dealt with all the theories of martensite nucleation in this section as recorded in the literature, or even with all alloys exhibiting martensitic transformations.

- Instead we have attempted to "illustrate some of the difficulties associated with explaining a complex event which occurs at such great speeds as to exclude experimental observation."

- A general, all embracing theory of martensite nucleation has still evaded us, and may not even be feasible.

Q7_6.4 Martensite Growth

- **Once the nucleation barrier has been overcome, the chemical volume free energy term (ΔG_v) is so large that the martensite plate grows rapidly until it hits a barrier such as another plate or a high angle grain boundary.**

- **High speed of M growth** → interface btw austenite and M must be a **glissile semi-coherent boundary** consisting of a set of parallel dislocations or twins with Burgers vector common to both phases, i.e. transformation dislocations → **dislocation motion** brings about **required lattice invariant shear transformation (may or may not generate an irrational habit plane)**

- **Increased alloying lowers the Ms temperature and that it is the temp. of transformation that dictates the mode of lattice invariant shear.**
  → Slip (고온)-twinning (저온) transition in a crystal at low temperatures:
  increased difficulty of nucleating whole dislocations needed for slip, but
  1) not so temp dependence (as the Peierls stress for a perfect dislocation) of critical stress needed for the nucleation of a partial twinning dislocation &
  2) chemical energy for transformation ~ independent of M_s temp.

→ Slip (고온)-twinning (저온) transition in a crystal at low temperatures:

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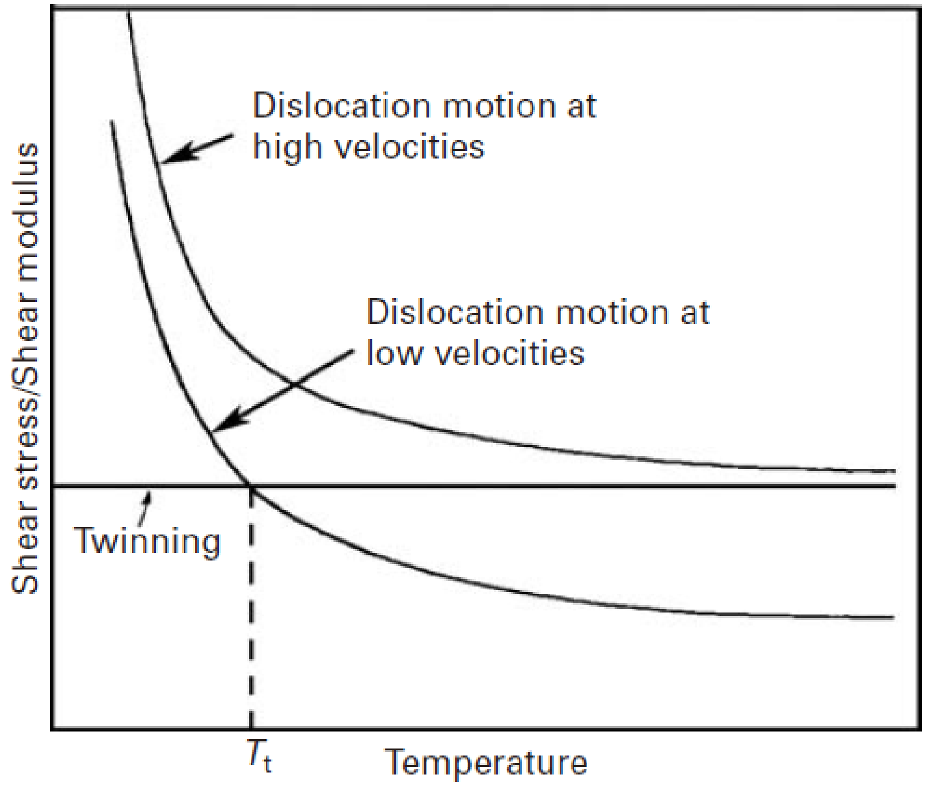
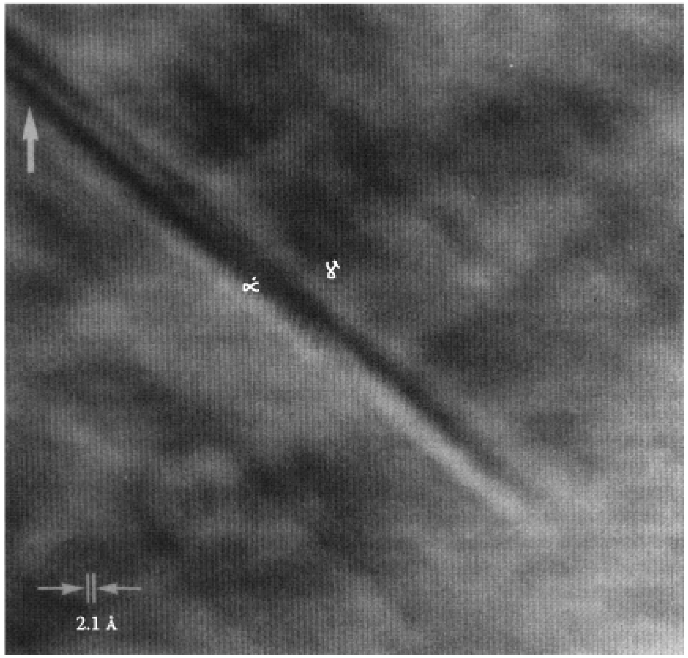
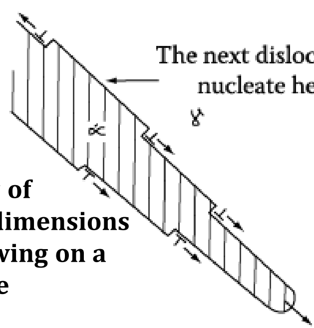


→ When M_s temperature is lowered, the mechanism of M transformation chosen is governed ① by the growth process having least energy. Other factor affecting mode of growth = ② how the nucleus forms

* Two main cases of rational (lath) and irrational (plate) M growth in ***steel***

Q8 6.4.1 Growth of Lath Martensite





Morphology of a lath with dimensions a>b>>c growing on a {111}γ plane

(b)

Figure. 6.20 (b) Lattice image of the tip of a martensite plate in a Ti-Ni alloy. The first interfacial dislocation behind the growing front is indicated.

- Morphology of a lath with dimensions a>b>>c growing on a {111}γ plane → thickening mechanism involving the nucleation and glide of transformation dislocations moving on discrete ledges behind the growing front, e.g. NiTi M and steel M

- Due to the large misfit between the bct and fcc, 1) lattices dislocations could be self-nucleated at the lath interface. → the stress at the interface exceeds the theoretical strength of the material.

- Eshelby's approach: for thin ellipsoidal plate (a>>c)

Maximum shear stress at the interface btw M and γ due to shear transformation

$$\boxed{\sigma \cong 2\mu sc / a}$$ μ= shear modulus of γ

~Sensitive to ❶ shape (c/a) and ❷ angle of shear (*s*)

: Of course in practice it is very difficult to define the morphology of M in such simple c/a terms, but this gives us at least a qualitative idea of what may be involved in the growth kinetics of M.

6.4.1 Growth of Lath Martensite

- **Lath M growth by shear loop nucleation (∵ σ/μ > threshold stress) :**
  by nucleating dislocations at the highly strained interface of the laths
  → the misfit energy reduced and the lath M can continue to grow into γ

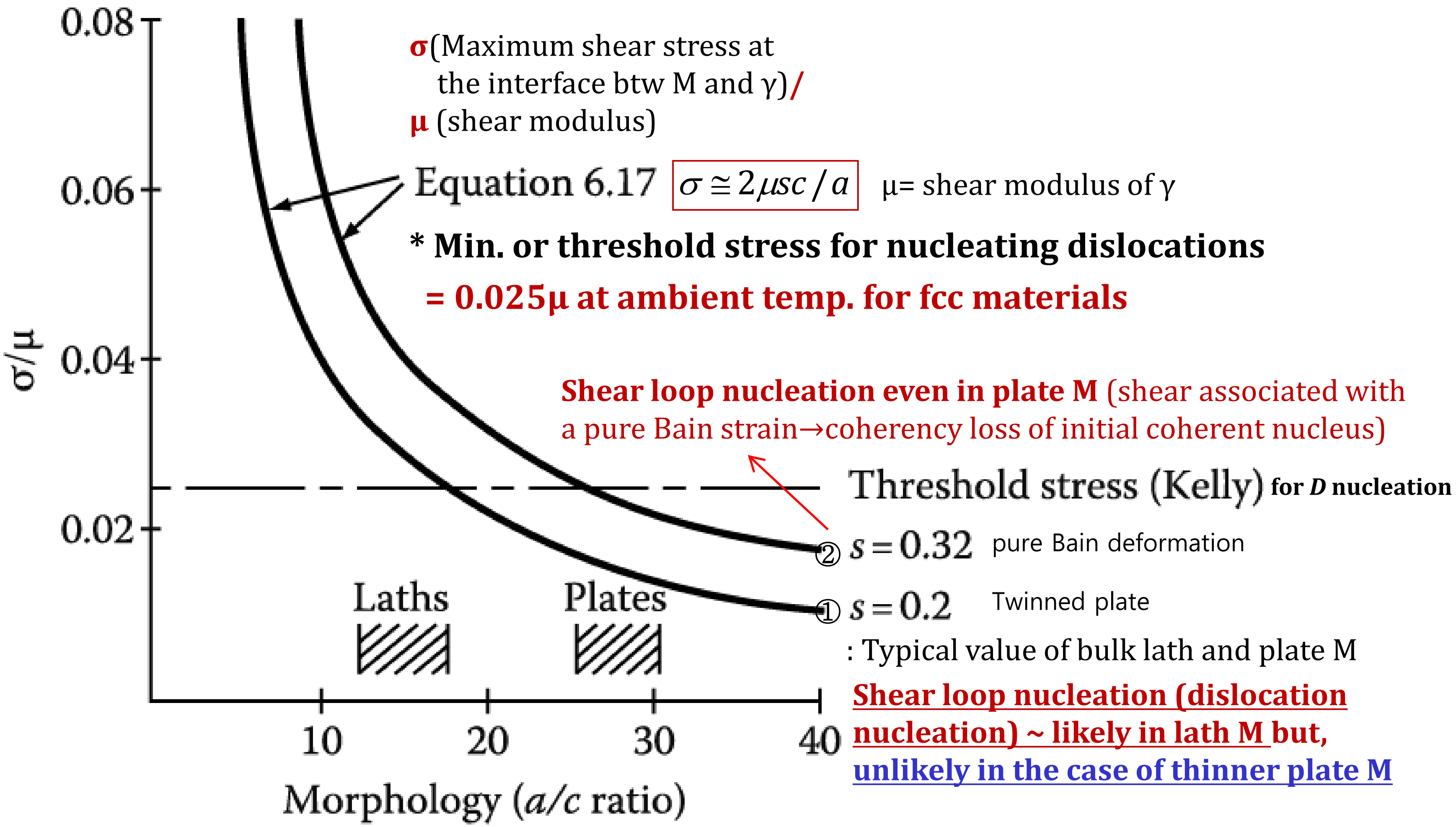


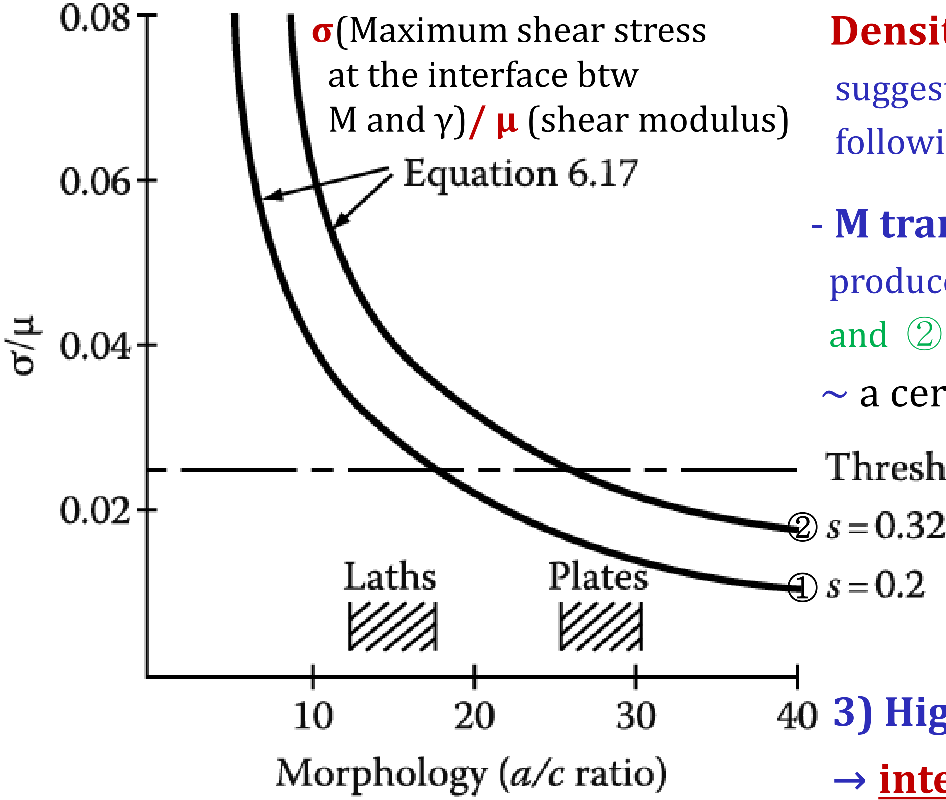
Figure. 6.21 **Eq. 6.17 plotted for two values of shear corresponding to a pure Bain deformation (s=0.32) and a twinned plate (s=0.2)**

6.4.1 Growth of Lath Martensite

- Lath M growth by shear loop nucleation (∵ σ/μ > threshold stress) :
by nucleating dislocations at the highly strained interface of the laths
→ the misfit energy reduced and the lath M can continue to grow into γ

- By internal friction measurements,
  Density of carbon in lath M : cell walls > within cell
  suggesting that 2) **limited diffusion of carbon takes place** following or during the transformation

- **M transformation (at least at higher Ms like lath M)→** produce **adiabatic heating** which may affect ① diffusion of carbon and ② dislocation recovery (by dislocation climb and cell formation).
  ~ a certain relationship between lower bainite and M



for *D* nucleation
Shear loop nucleation lath M and plate M
Shear loop nucleation in lath M

3) High growth speed of lath M
→ interface of predominantly screw dislocation
& volume of retained γ ~relatively small in lath M
(important to the mechanical properties of low-carbon steel)
due to sideways growth of screw dislocation not too difficult

*** Figure. 6.21 Eq. 6.17 plotted for two values of shear corresponding to a pure Bain deformation (s=0.32) and a twinned plate (s=0.2)**

Q9 6.4.2 Growth of Plate Martensite

- **In medium and high carbon steels, or high nickel**
  Morphology: Lath M → Plate M (due to lower M_s temp. and more retained γ)
  : much thinner than lath M or bainite

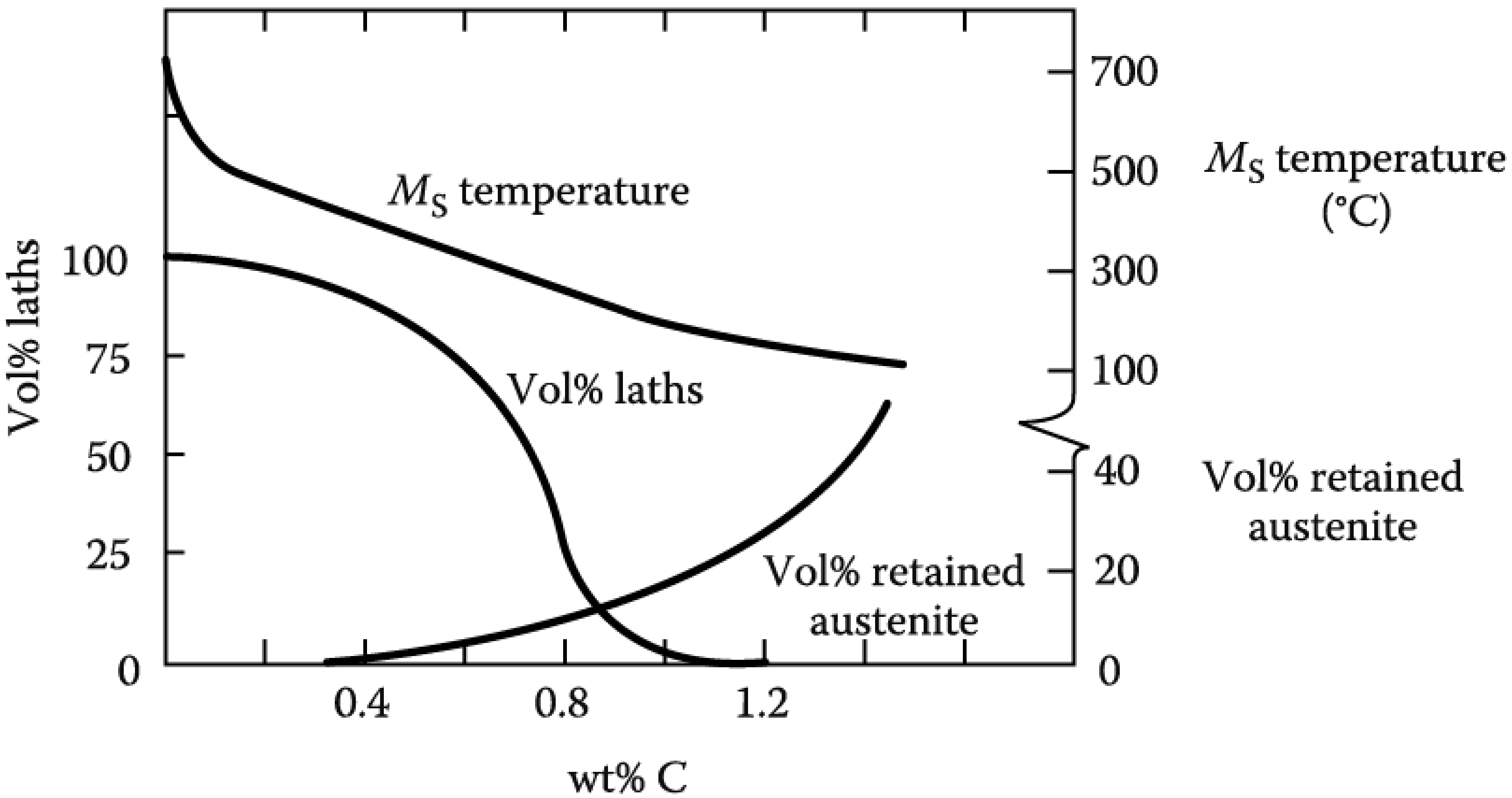


Figure. 6.22 Approximate relative percentages of lath martensite and retained austenite as function of carbon content in steels.

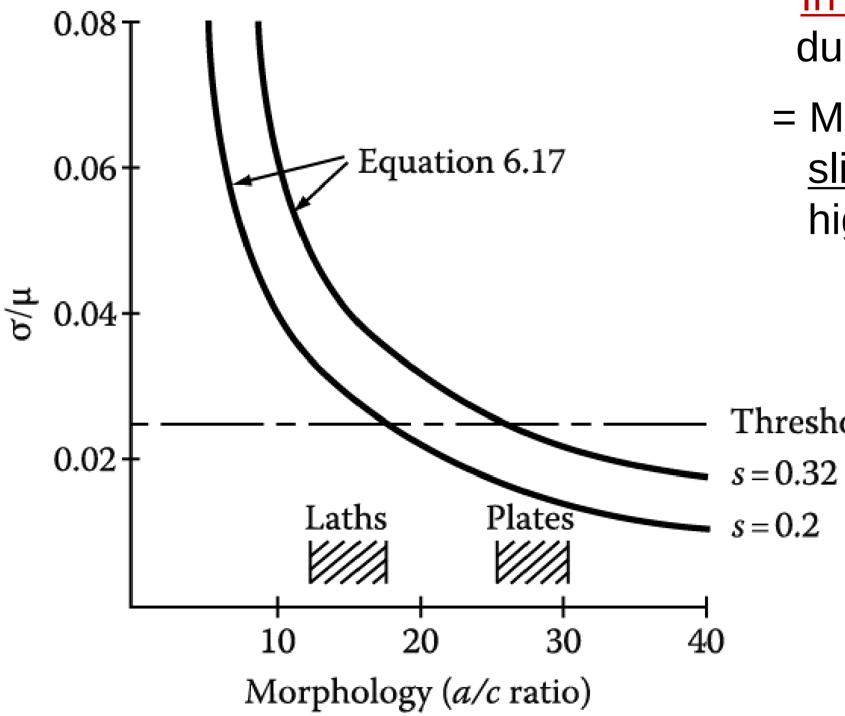
6.4.2 Plate Martensite

- **In medium and high carbon steels, or high nickel**
  Morphology: Lath M → Plate M (due to lower M_s temp. and more retained γ)
  : much thinner than lath M or bainite

- **Transition from plates from growing on $\{225\}_\gamma$ planes to $\{259\}_\gamma$ planes with increasing alloy contents** (carbon 함량 증가시 habit 면 변화)

In lower carbon or nickel, $\{225\}_\gamma$ M = plates with a central twinned 'midrib', the outer (dislocation) regions of the plate being free of twins

In high carbon and nickel, $\{259\}_\gamma$ M = completely twined & less scattered habit plane



* In Midrib M, transition from twinning → dislocations due to a change in growth rate after the midrib forms

= M formed at higher temp. or slower rates grows by a slip mechanism, while M formed at lower temp and higher growth rates grows by a twining mode.

Why? much thinner than lath M or bainite

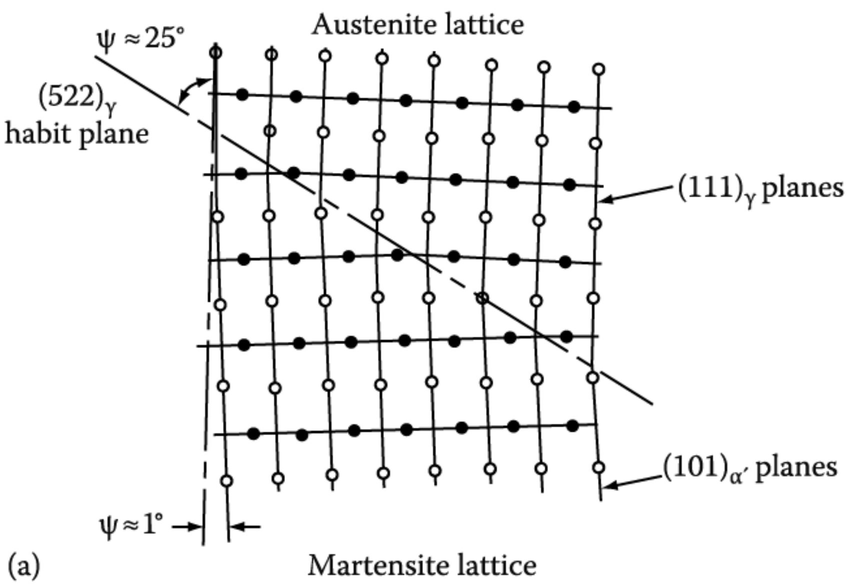
1) s=2, problem in nucleating whole dislocations in the case of growing plate M, but partial twining dislocations evidently can nucleate.

→ Once nucleated, twinned M grows extremely rapidly, but the mechanism is unclear.

6.4.2 Plate Martensite

빠른 M growth 설명: 정상탄성파가 쌍정전위를 생성시키고 이로 인해 판상의 빠른 성장

* **Dislocation generated {225}γ M (Frank)**



Close-packed plane

: slight misfit along the $[01\bar{1}]_\gamma$ & $[11\bar{1}]_{\alpha'}$

= M lattice parameter is ~2% less than that of γ

→ Insertion of an array of screw dislocations

with a spacing of six atom planes in the interface

* In terms of the min. shear stress criterion (Fig. 6.21), when the midrib reaches some critical a/c ratio further expansion and thickening of a {225}γ twinned midrib by a Frank dislocation interface could occur.

. → "No detailed explanation"

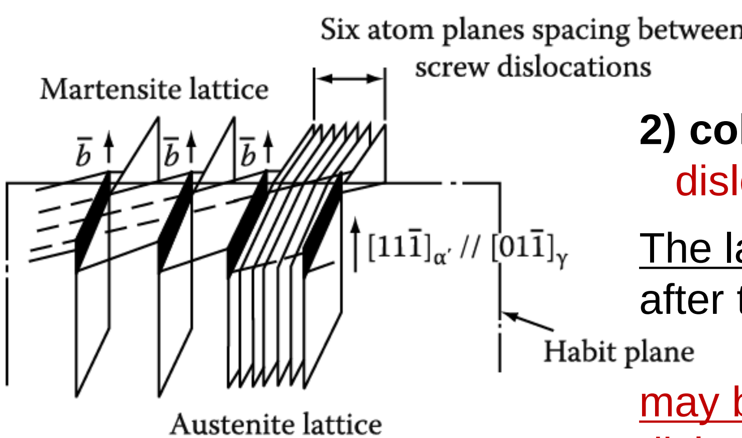


Figure. 6.23 Model for the {225}γ habit austenite-martensite interface in steel.

2) coherent nucleus with s=0.32: possible for dislocation nucleation to occur to relieve coherency.

The larger amount of chemical free energy, available after the critical size for growth has been exceeded,

may be sufficient to homogeneously nucleate dislocations particularly in the presence of the large strain energy of the rapidly growing plate.

Q10 * Factors for affecting the growth of M:

❶ shape (c/a), ❷ angle of shear (s)
❸ phenomenon of stabilization,
❹ external stresses, and ❺ grain size

6.4.3 Stabilization

* In **intermittent cooling between M_s and M_f**, transformation does not immediately continue, and the total amount of transformed M is less than obtained by continuous cooling throughout the transformation range.

6.4.4 Effect of External Stresses

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

* In view of the dependence of M growth on dislocation nucleation, it is expected that an externally applied stress (**ES**) will aid **the generation of dislocations** and hence **the growth of M**.

a) ES lowers the nucleation barrier for coherency loss of second phase precipitates.
b) ES aid M nucleation if the ① external elastic strain components contribute to the Bain strain.
  → **M_s** temperature can be raised ↑. But, if plastic deformation occurs, there is an upper limiting value of **M_s** defined as "the **M_d** temperature".
c) ② Under hydrostatic compression, M_s temperature can be suppressed to lower temp ↓.
  (P ↑ → stabilizes the phase with the smaller atomic volume (close-packed austenite) → lowering the driving force ΔG_v for the transformation to M)
d) ③ large magnetic field can raise the Ms temperature↑on the grounds that it favors the formation of the ferromagnetic phase.
e) Plastic deformation of samples can aid both nucleation and growth of M, but too much plastic deformation may in some cases suppress the transformation (nucleation ↑ & nuclei growth ↓).

* **Ausforming process** : plastically deforming the austenite prior to transformation → number of nucleation sites and hence refining M plate size → High strength (fine M plate size + solution hardening (due to carbon) and dislocation hardening)

* **Factors for affecting the growth of M:**

> ❶ **shape (c/a),** ❷ **angle of shear (*s*)**
> ❸ **phenomenon of stabilization,**
> ❹ **external stresses, and** ❺ **grain size**

6.4.5 Role of Grain Size

- Martensite growth ~ maintaining a certain coherency with the surrounding austenite

→ high-angle grain boundary is an effective barrier to plate growth.

→ While grain size does not affect the number of M nuclei in a given volume,

the **1) final M plate size** is a function of the grain size.

2) Degree of residual stress after transformation is completed.

- In large grain sized materials: dilatation strain associated with the transformation
  → Large residual stresses to built up btw adjacent grains
  → **GB rupture (quench cracking)** and substantially increase of dislocation density in M

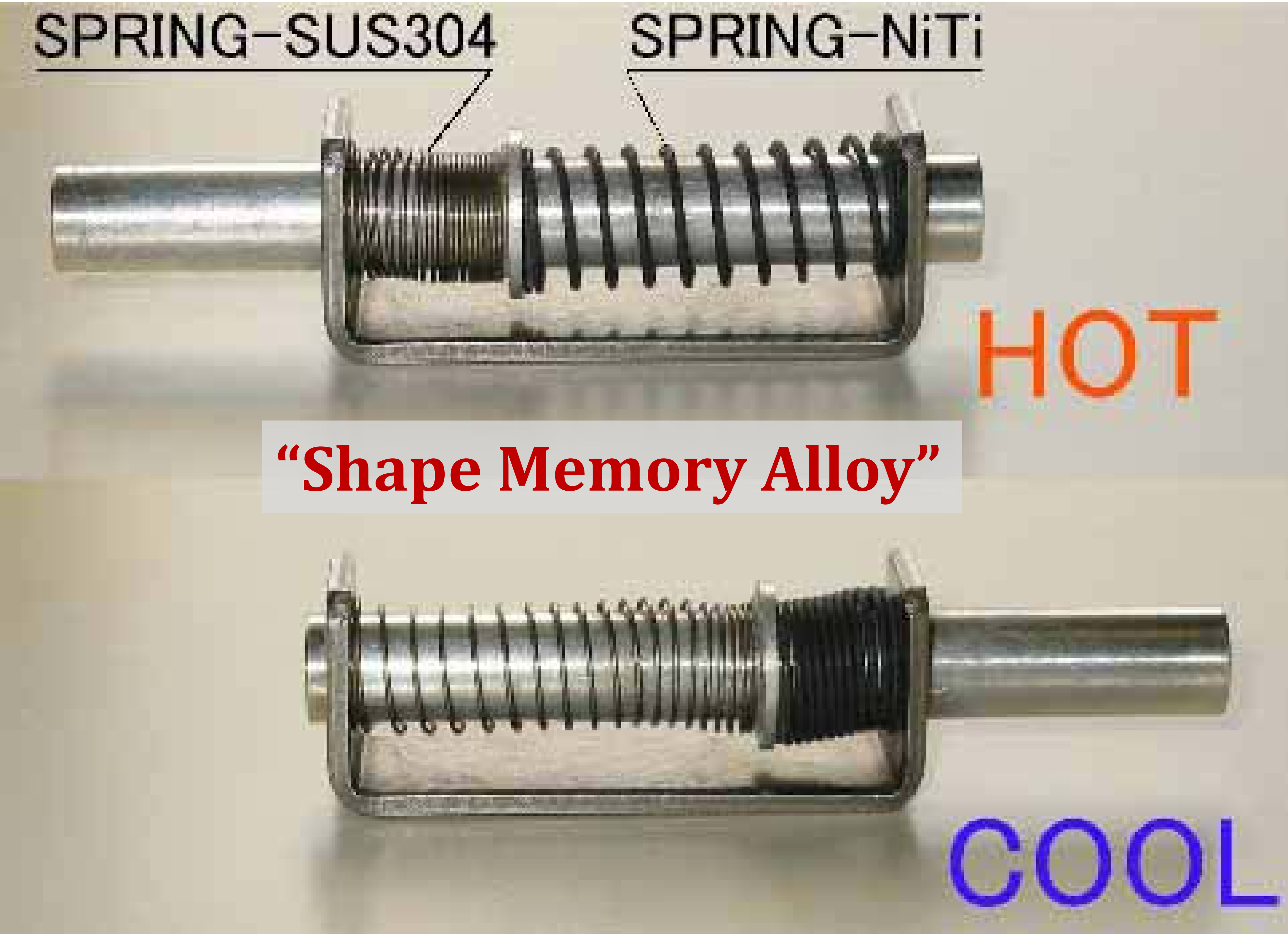
- In fine grain-sized metals: dilatation strain associated with the transformation
  → more self-accommodating & smaller M plate size
  → **stronger & tougher material**

* In summary, **theories of M nucleation and growth are far from developed to a state where they can be used in any practical way** – such as helping to control the fine structure of the finished product. It does appear that **nucleation is closely associated with the presence of dislocations (dislocation density)** and the process of ausforming (deforming the austenite prior to transformation) could possibly be influenced by this feature if we know more of the mechanism of nucleation. However, **growth mechanisms, particularly by twining, are still far from clarified.**

6.5, 6.6 & 6.7 Skip

IH: Summarize the pre-martensite phenomena and the tempering behavior of Ferrous martensite. (before final exam)

*** Homework 6 : Exercises 6 (pages 504-508) until 20th December (before exam)**


SPRING-SUS304
SPRING-NiTi
HOT
"Shape Memory Alloy"
COOL

Representative Diffusionless Transformation

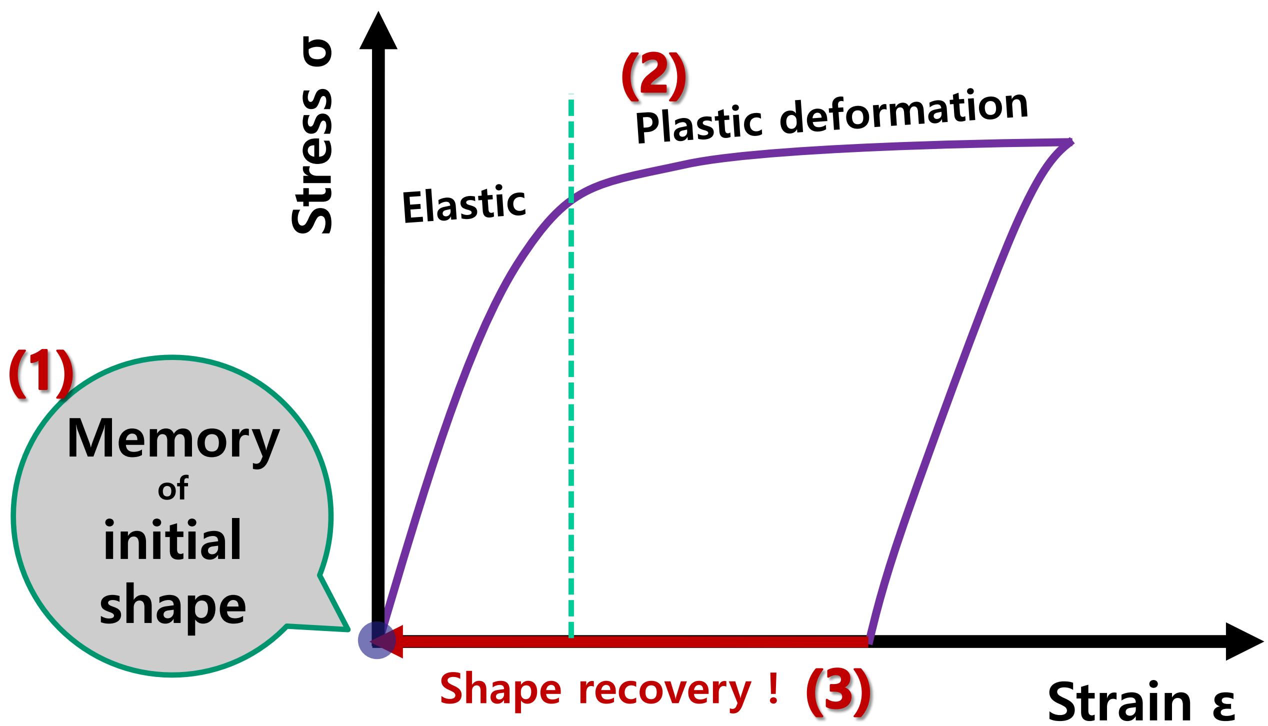
Martensitic transformation in Ni-Ti alloy ;

55~55.5 wt%Ni - 44.5~45 wt%Ti ("Nitinol")

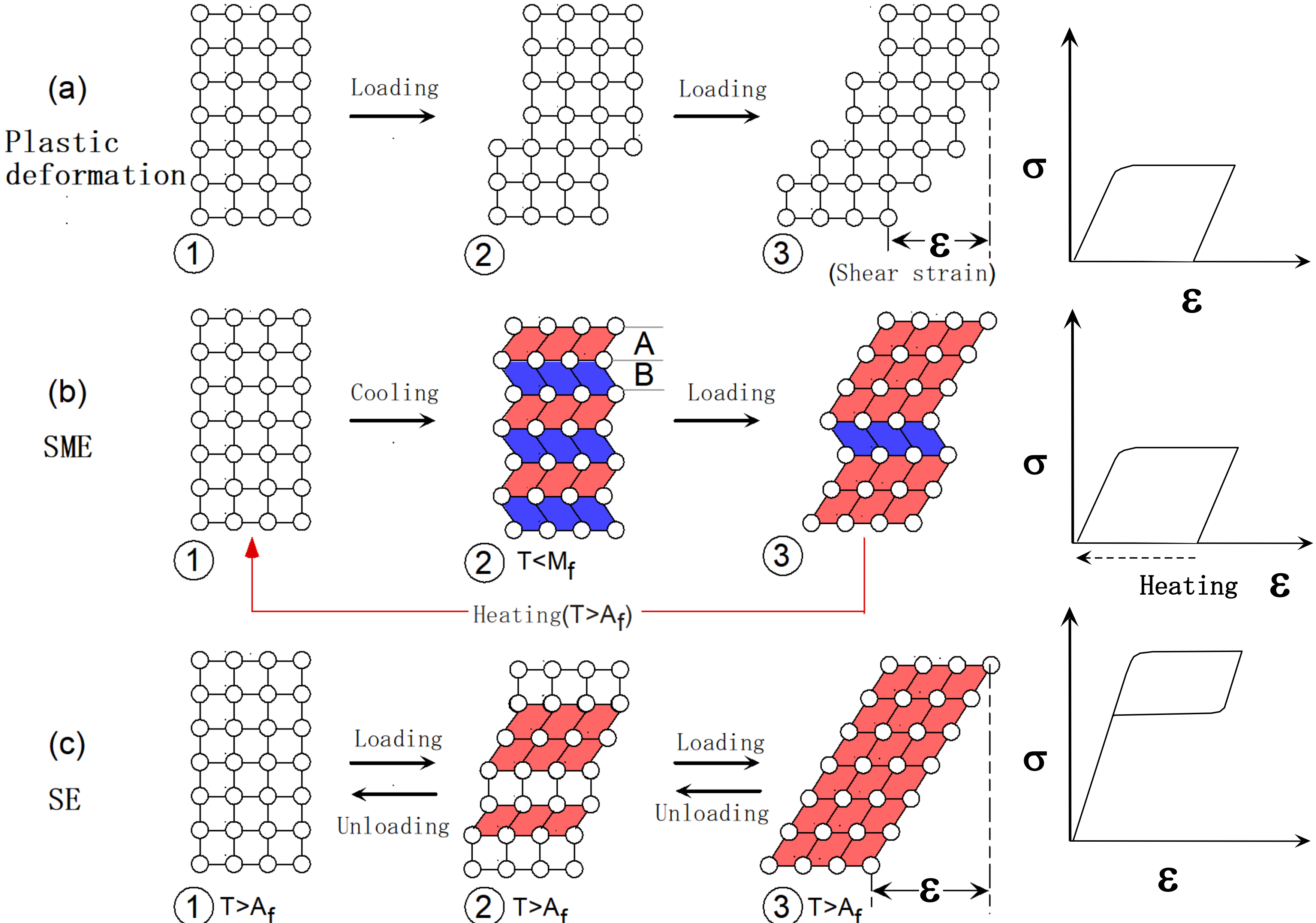
Ex) Shape memory alloy

Introduction - Stress-Strain curve

Introduction - Shape memory effect



	Elastic Deformation	Plastic Deformation	Transformation Deformation
Ceramics	○	×	×
Conventional Metals, Alloys & Plastics	○	○	×
Shape Memory Alloys	○	○	○
	Recoverable Small Deformation ↓ Elasticity	Permanent Large Deformation ↓ Plasticity	Recoverable Large Deformation ↓ Shape Memory Effect Superelasticity (Pseudoelasticity)


(a)
Plastic deformation
Loading
Loading
1
2
3
ε
(Shear strain)
σ
ε
(b)
SME
Cooling
A
B
Loading
1
2 T<M_f
3
Heating(T>A_f)
σ
Heating
ε
(c)
SE
Loading
Unloading
Loading
Unloading
1 T>A_f
2 T>A_f
3 T>A_f
ε
σ
ε

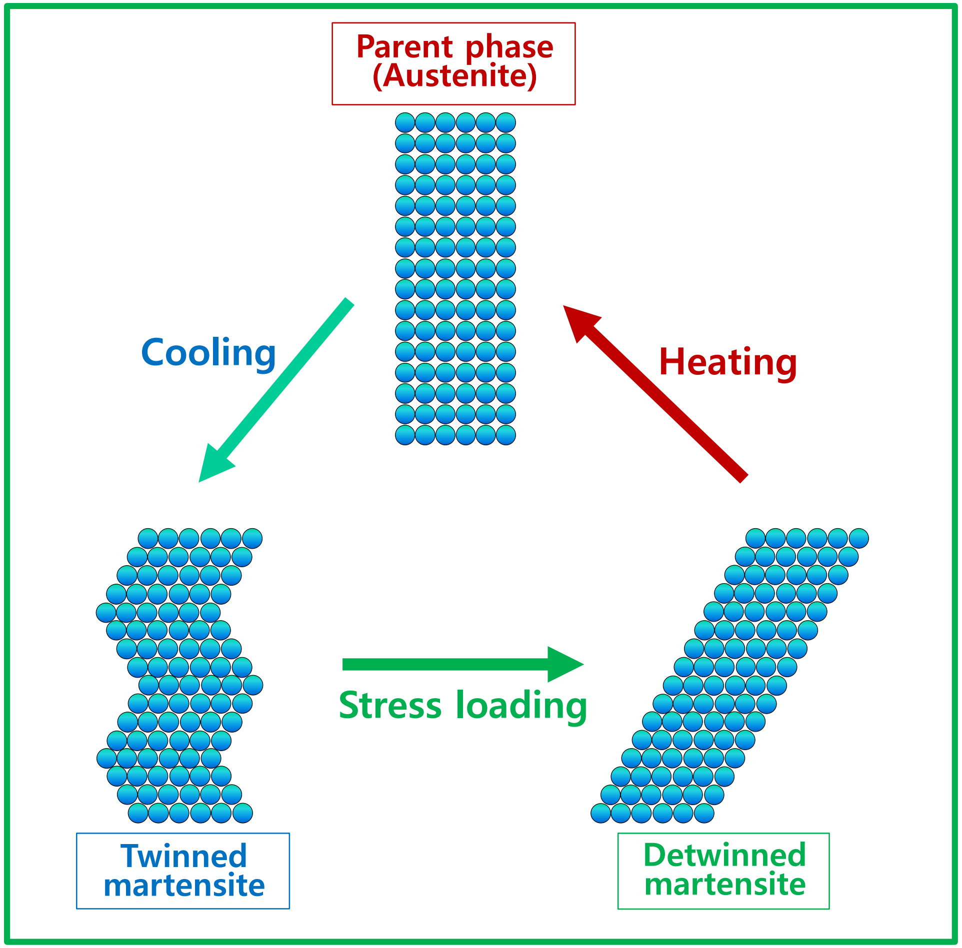
Principles

How can shape memory effect occur?

Principles

How can shape memory effect occur?

Principles- Shape memory process



1. A_f 이상의 온도로 열처리를 통해
   Austenite 상에서 형상 기억

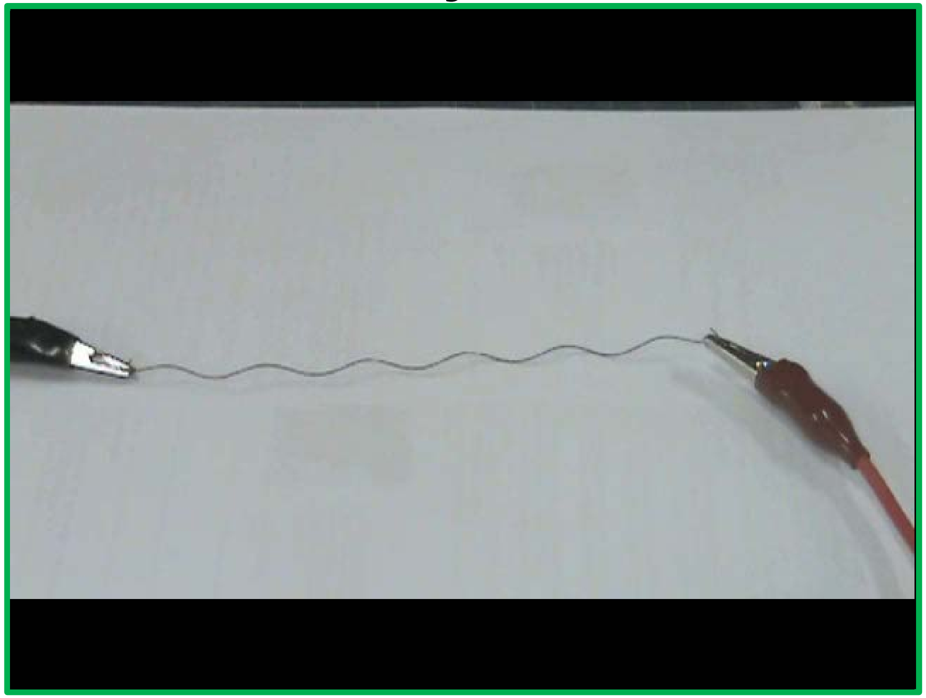
2. M_s 이하의 온도로 냉각시
   Twinned martensite 생성

3. 항복강도 이상의 응력을 가하면 Twin
   boundary의 이동에 의한 소성 변형

4. A_f 이상으로 가열해주면 martensite
   에서 다시 Austenite로 변태
   ➡ 기억된 형상으로 회복

* One-way / Two-way shape memory effect

▼ One-way SME

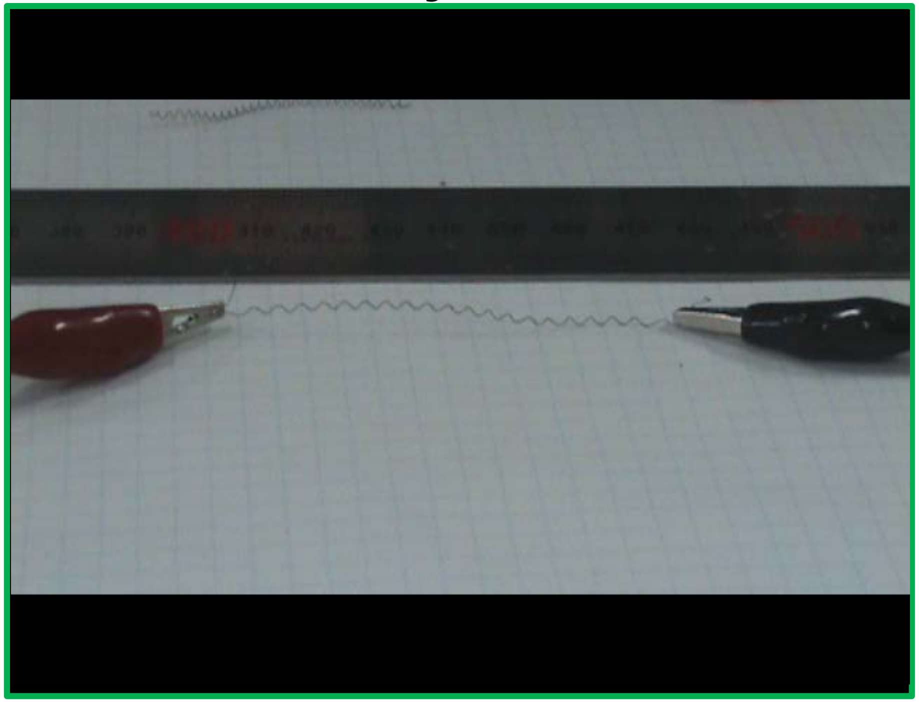


↳ A_f 이상의 고온 형상만을 기억

➡ 저온(< M_f)에서 소성변형 후
A_f 이상의 고온으로 가열

➡ 기억된 고온 형상으로 회복

▼ Two-way SME



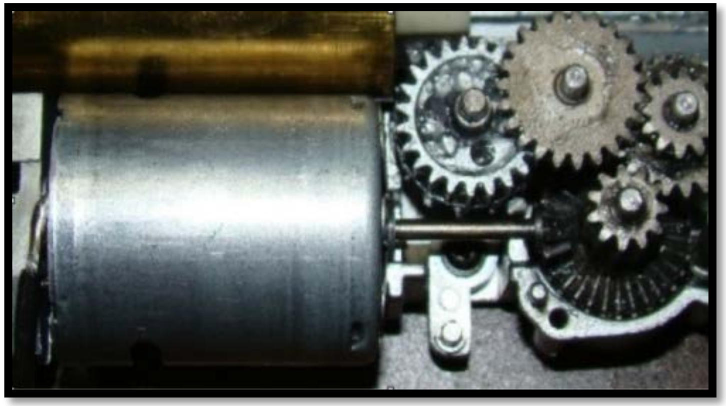
↳ 고온(> A_f) 형상과 저온(< M_f) 형상을 모두 기억

➡ 반복적인 변형으로 인한 형상기억합금 내
전위 밀도의 상승 & 특정방향 응력장의 형성

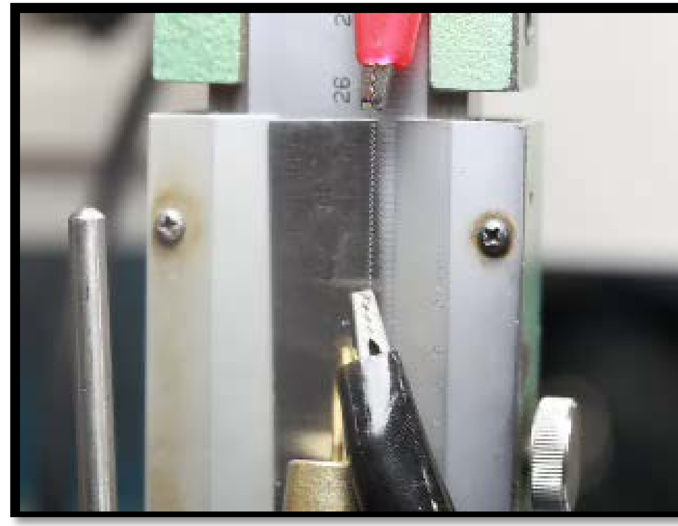
➡ 저온에서 반복소성변형 방향으로 회복

* SMA Actuator

▶ 액츄에이터(Actuator) : 전기 에너지, 열에너지 등의 에너지원을 운동에너지로 전환하여 기계장치를 움직이도록 하는 구동소자

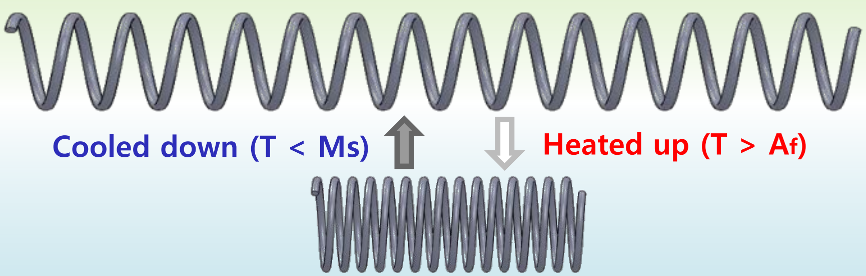


▲ 기존의 매크로 스케일 액츄에이터 (모터-기어 방식)



▲ SMA 스프링 액츄에이터

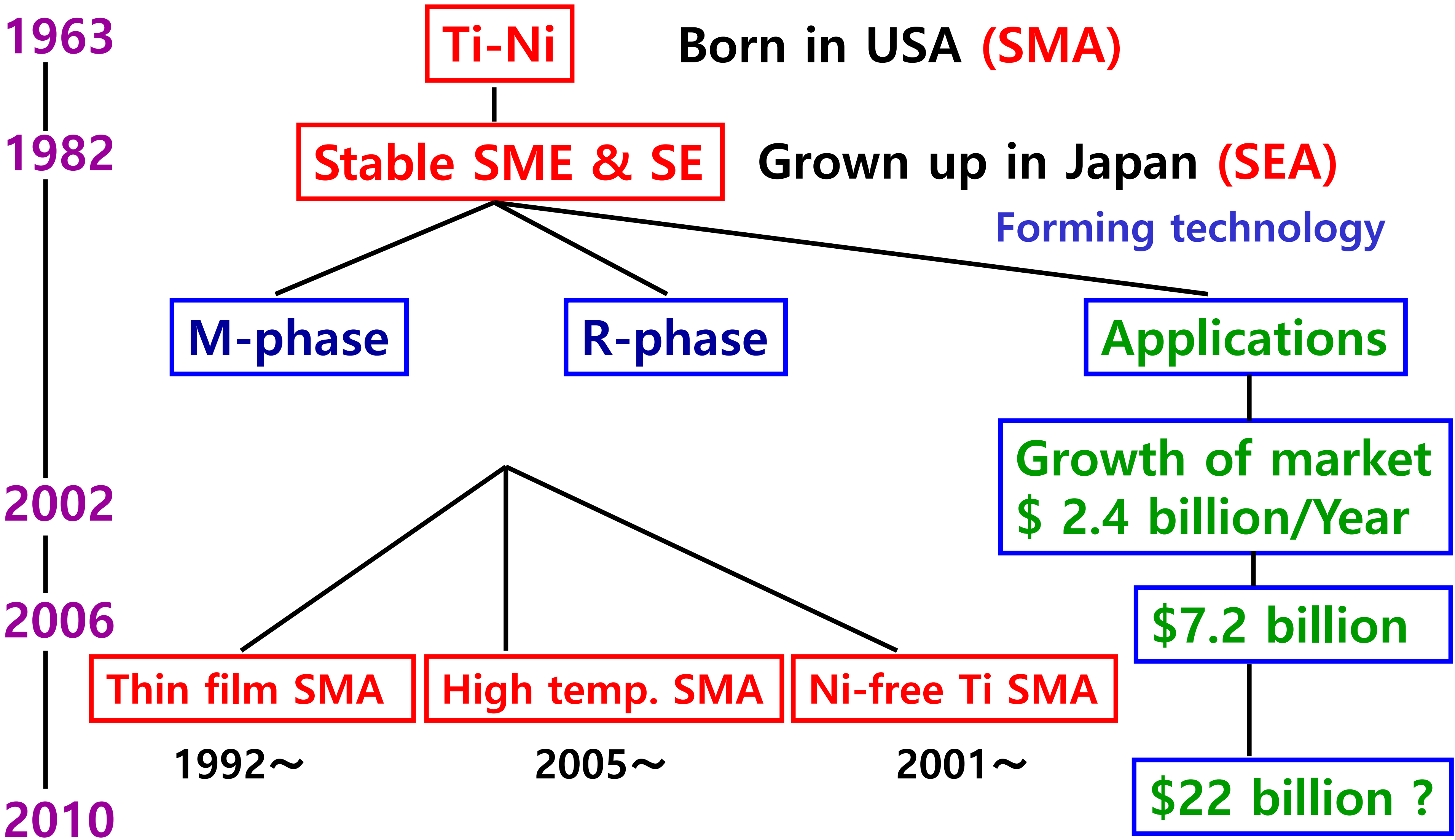
SMA Spring Actuator



재료의 수축과 신장을 통하여 기계적인 동작을 가능하게 함.

1. 단위 체적당 출력이 높음
2. 모터 구동에 비해 매우 단순한 구조
3. 온도에 의한 제어가 용이
4. 소형화가 쉬움.

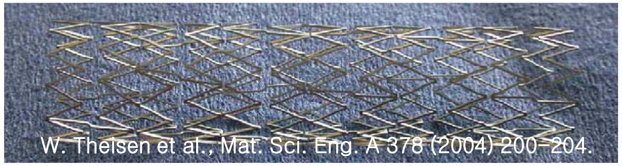
Summary

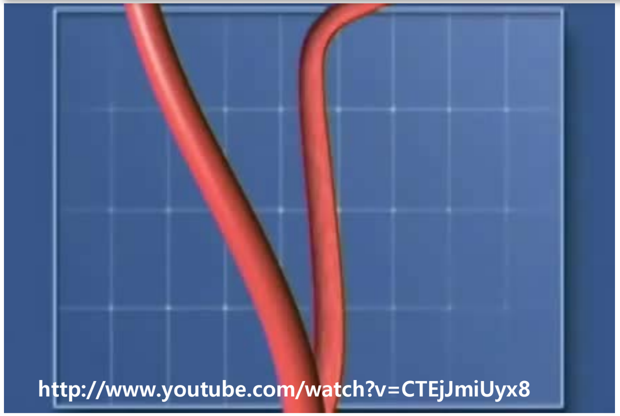


1963
Ti-Ni
Born in USA (SMA)
1982
Stable SME & SE
Grown up in Japan (SEA)
Forming technology
M-phase
R-phase
Applications
Growth of market
$ 2.4 billion/Year
2002
2006
$7.2 billion
Thin film SMA
High temp. SMA
Ni-free Ti SMA
1992～
2005～
2001～
$22 billion ?
2010

* Application of SMAs

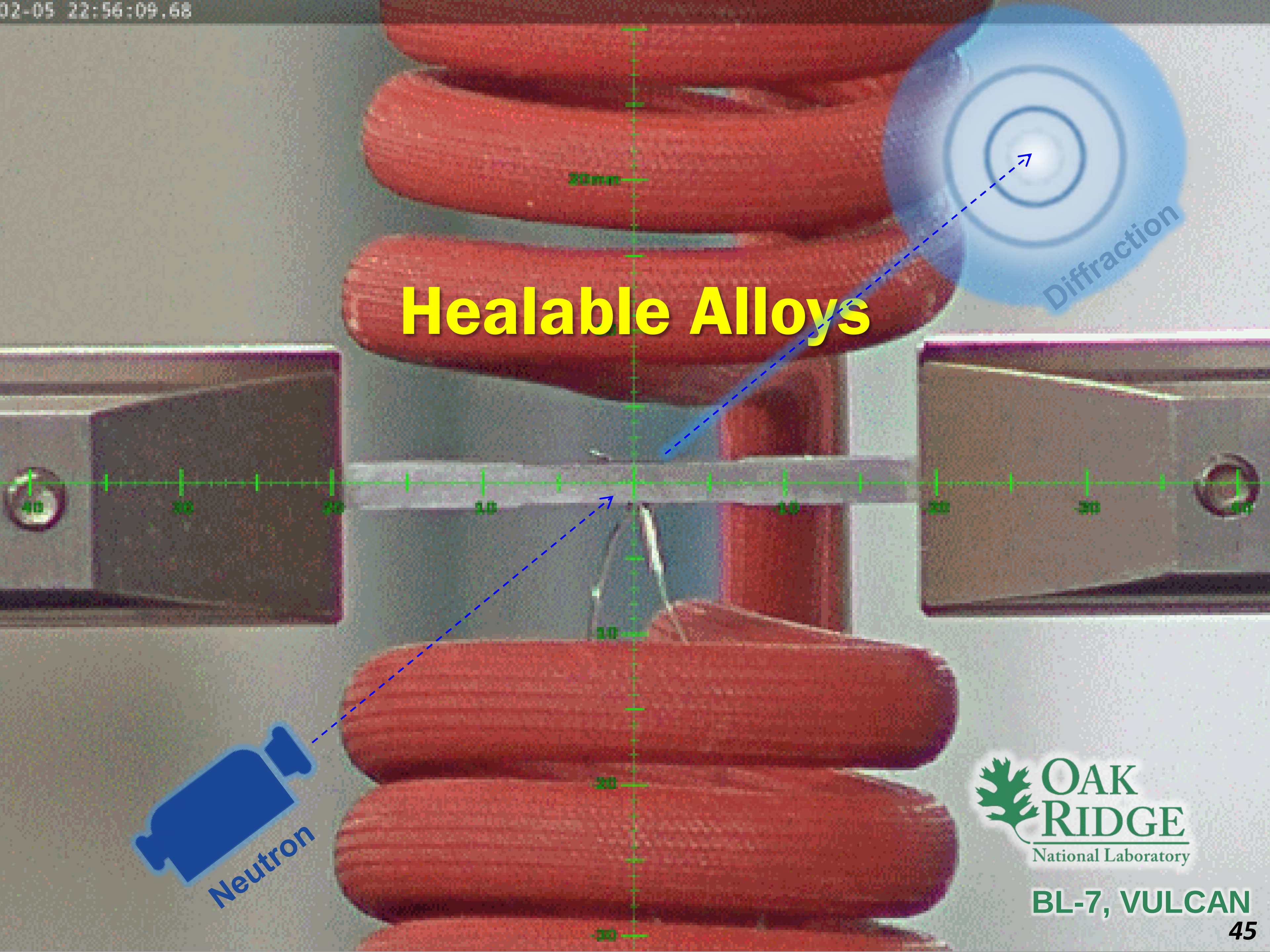
▼ 산업 부문: 부품소재 (파이프 이음, 스위치소자나 온도제어용 장치 등)





▲ 생체의료 부문: 첨단의료재료
(stent, 치열교정용 강선 등)

▲ 심해저/우주항공 부문: 극지재료
(잠수함, 태양전지판 등)



02-05 22:56:09.68
Healable Alloys
Diffraction
Neutron
OAK RIDGE
National Laboratory
BL-7, VULCAN

Materials design for reuse

Damage process is incremental, and often local → repair opportunity

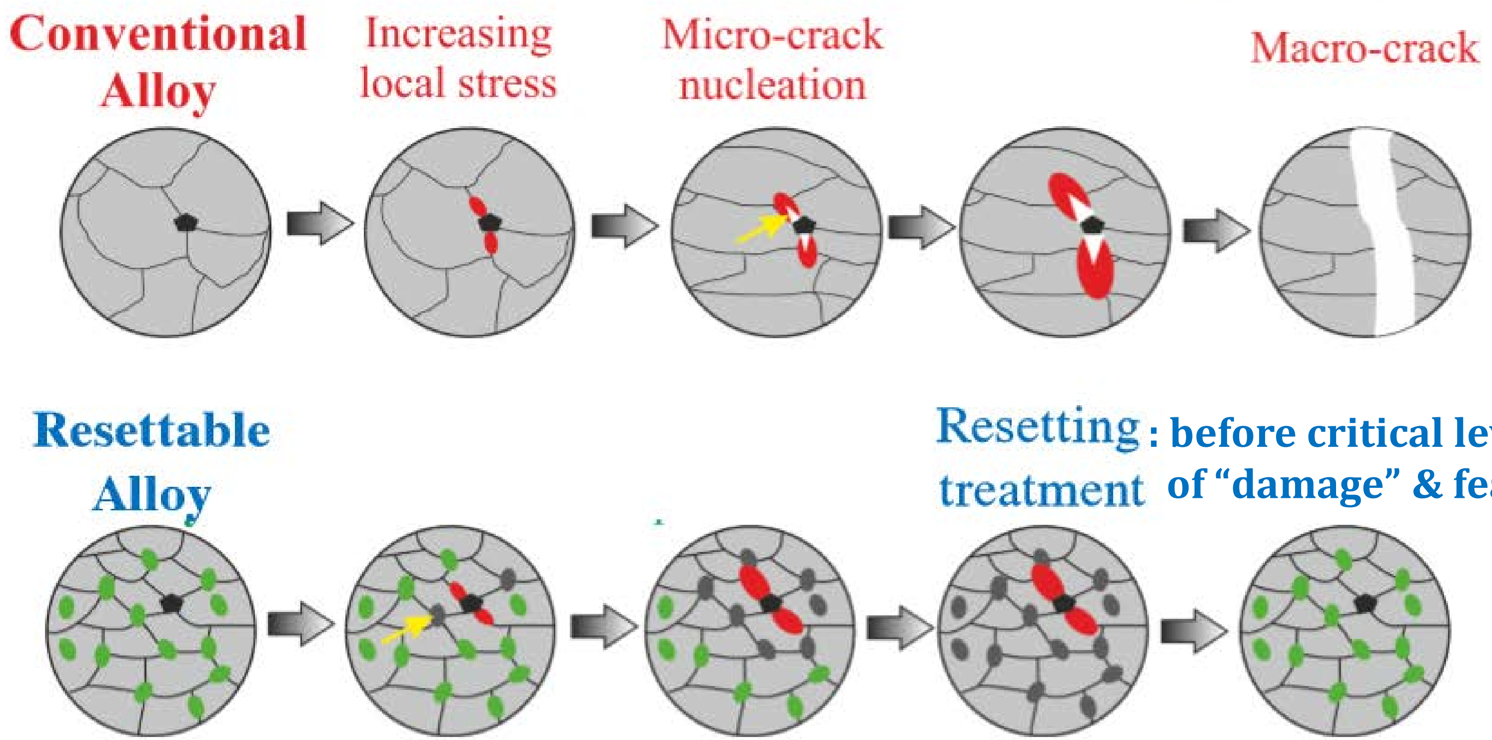
Two damage repair options possible:

- The metal autonomously repair damage → ***Self-healing***

- Damage is repaired by an external treatment → ***Resetting***

Self-healing metals vs Resettable alloys

- self-healing: "*autonomic closure of micro-cracks*"
- resetting: "*non-autonomic retrieval of crack-arresting ability*"



: before critical level of "damage" & feasible

: revert to the original microsturcture

Different failure mechanisms require different resetting strategies

서 울 대 학 교
SEOUL NATIONAL UNIVERSITY

한국연구재단

준정적 가역상변화 기반 무한수명 특성복귀합금 개발

2018년 선정 도전형 소재기술개발 프로그램

2018 년 – 2022 년 (수행중)

박 은 수

서울대학교 재료공학부

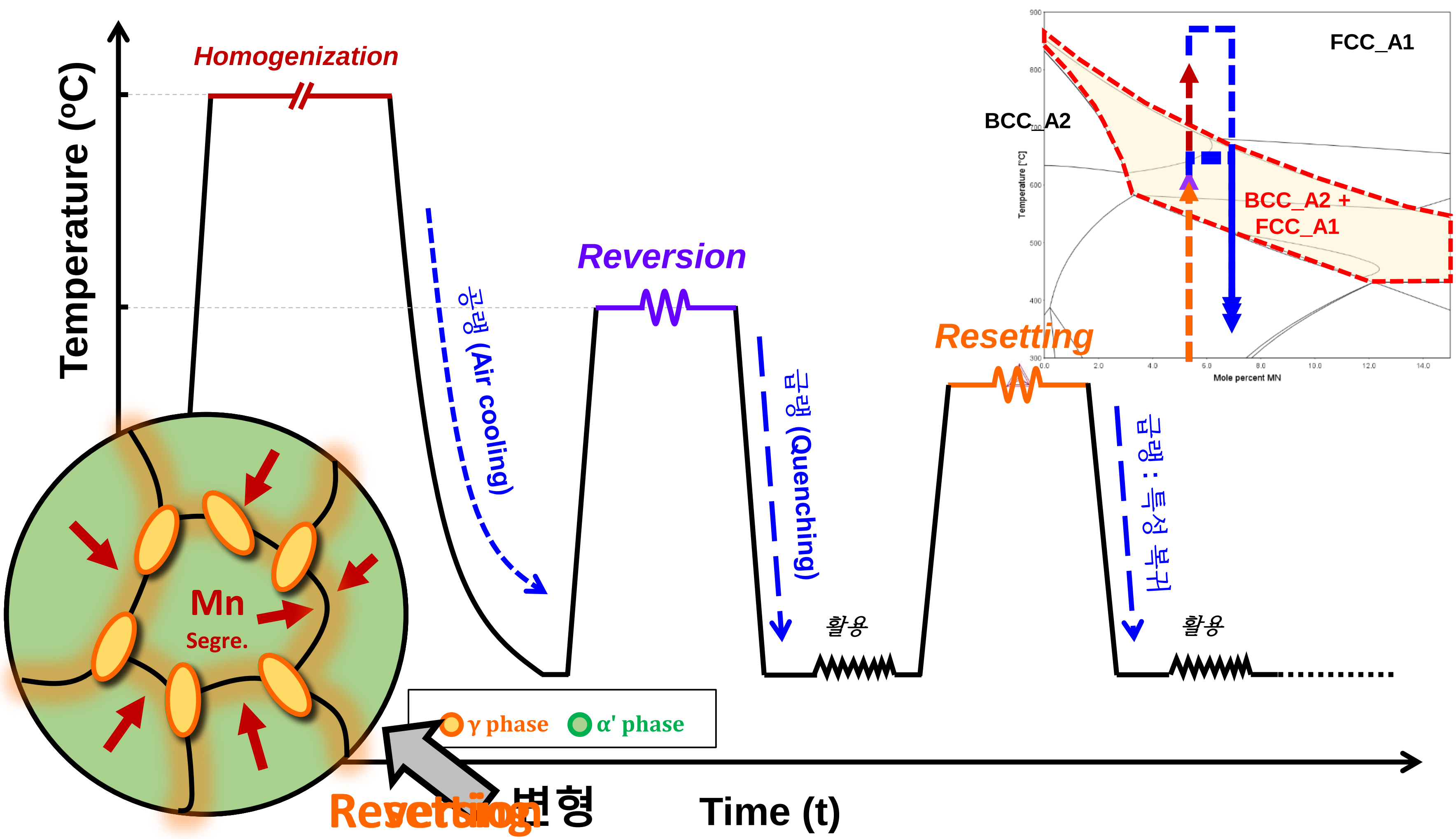
Homepage : http://espark.snu.ac.kr

ESPark Research Group

e-mail : espark@snu.ac.kr

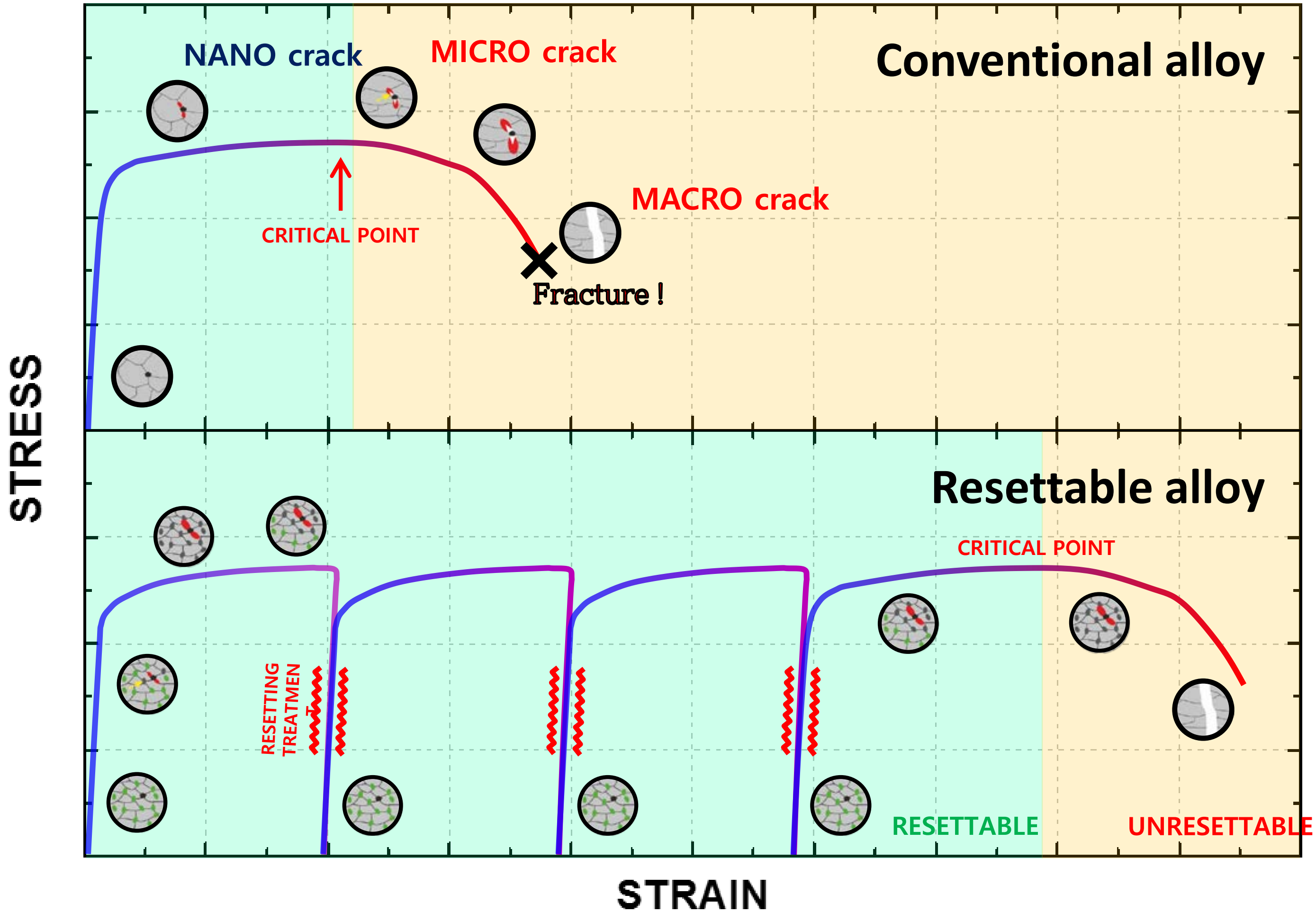
New challenges : *Resettable alloys!*

응력유기 변태 가능 A 상- M 상 Nano-laminate 구조 합금 조성 최적화



Resetting treatment 를 통해 초기 미세구조로 회복 가능한 Resettable alloy!

New challenges : *Resettable alloys!*

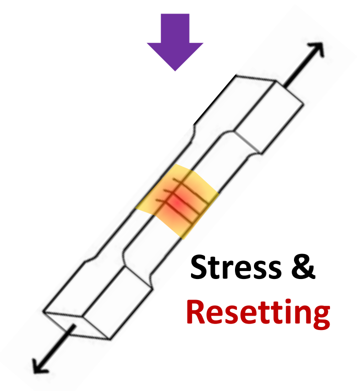


Resetting treatment 를 통해 초기 미세구조로 회복 가능한 Resettable alloy!

(3) 실험결과 I : 준정적 가역 상변화 가능 특성복귀 합금 개발

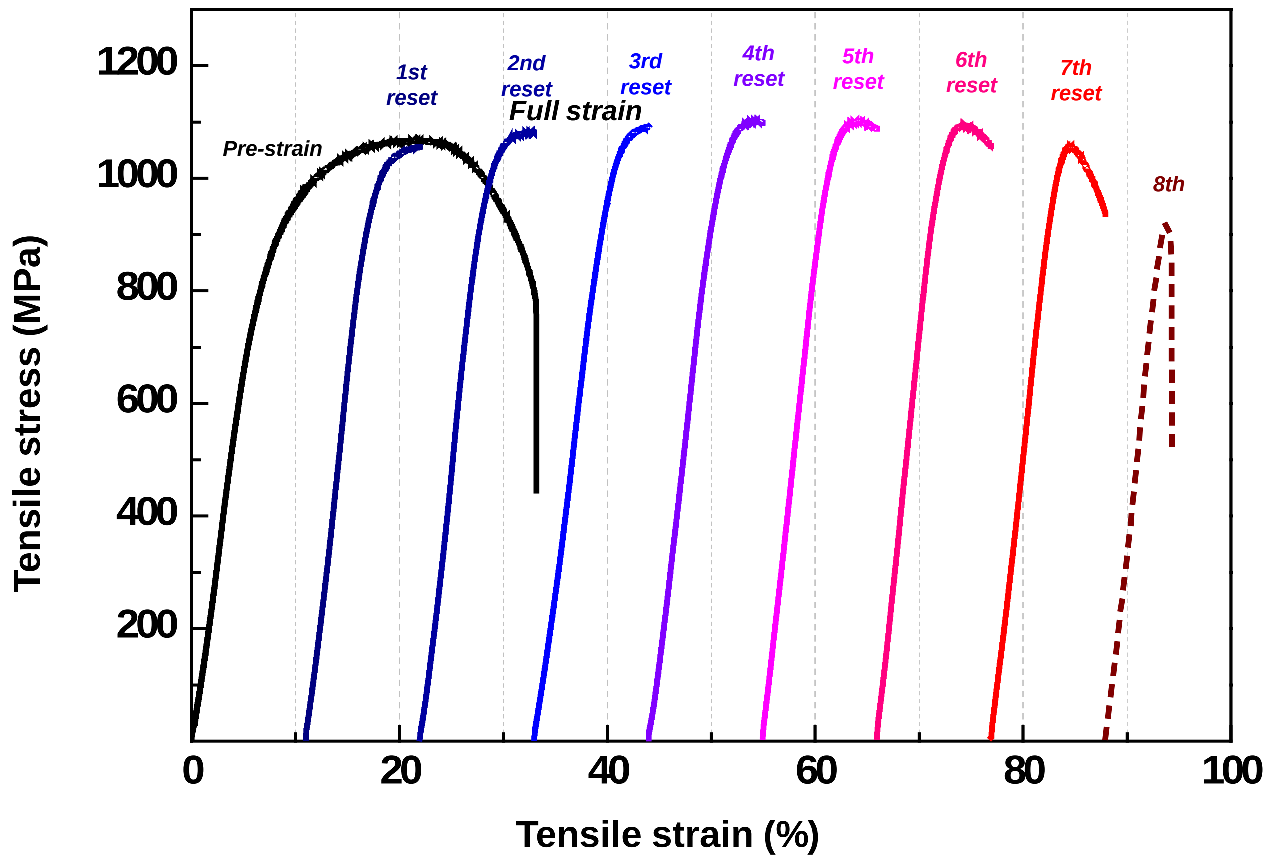
c. (조성)-(결함)-(Resetting 공정) 최적화 기반 변형 Informatics 구축

Air cooling
After homogenization
at 980 °C for 8h
and Reversion
at 550 °C for 20 hour



(1) Pre-strain (12%)
by tensile test
for crack formation

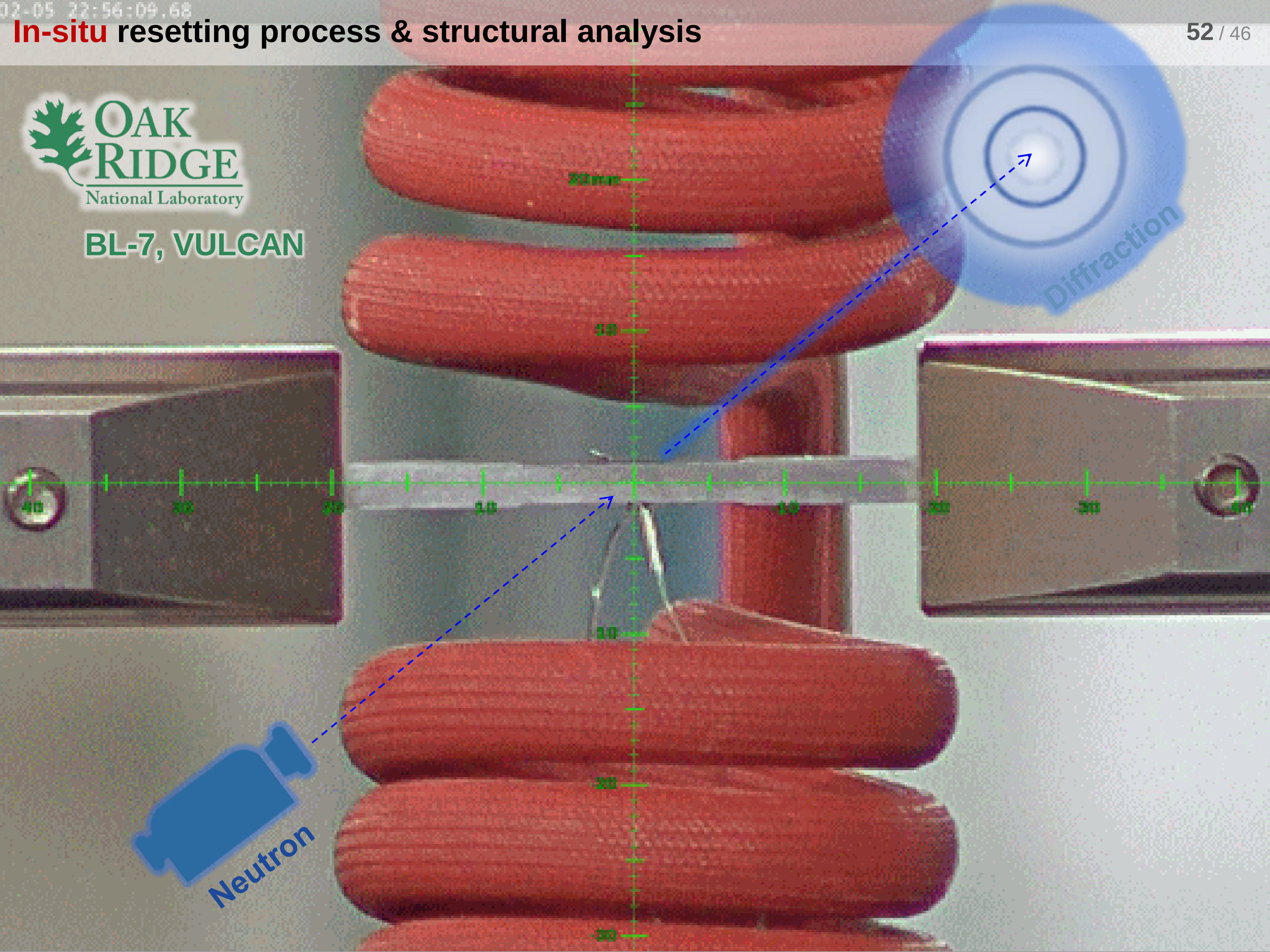
(2) Reset Annealing
at 550 °C for 2h



- Resetting Process를 통해, 300% 이상의 연신율 회복

<u>리셋 공정 통해 특성 향상 가능 특성 복귀 합금 성공적 개발!</u>

In-situ resetting process & structural analysis

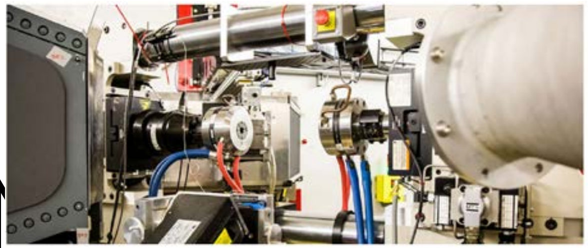


(3) 실험결과 II : 리셋 공정 최적화 변형 Informatics 구축

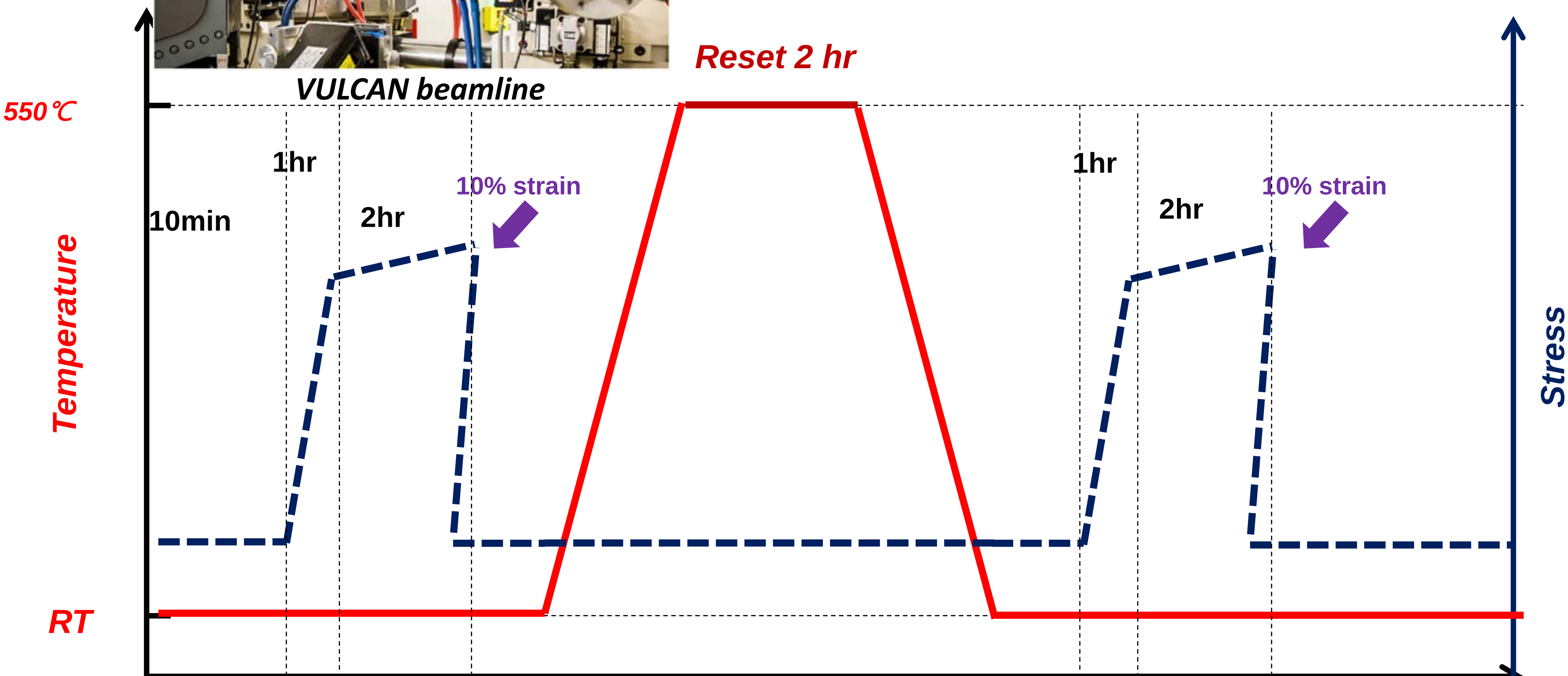
결함 정량화 Smart Monitoring 기반 Deformation Informatics 구축

② 비파괴 검사 통한 사용환경 중 변형 Smart Monitoring 기술 개발

: (손상)-(회복)-(손상) cycle 시의 A 상분율 정량 분석을 통한 손상 정량화



VULCAN beamline



(3) 실험결과 II : 리셋 공정 최적화 변형 Informatics 구축

결함 정량화 Smart Monitoring 기반 Deformation Informatics 구축

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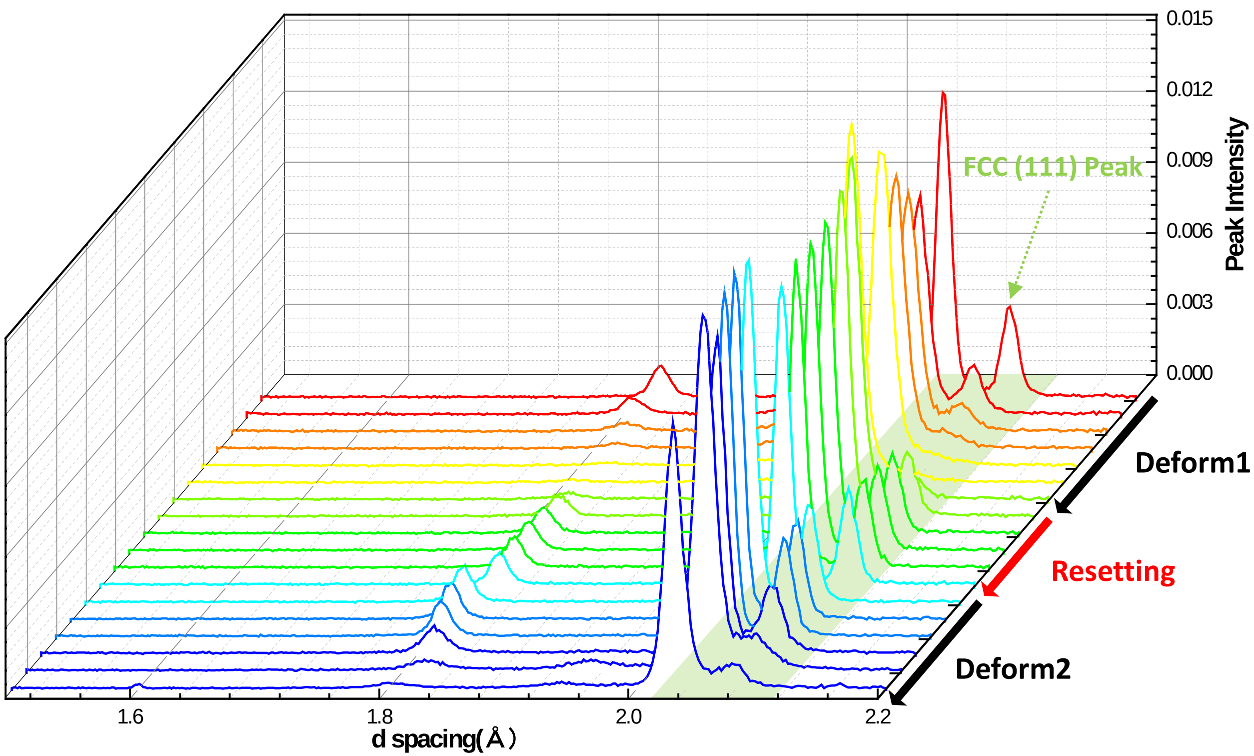
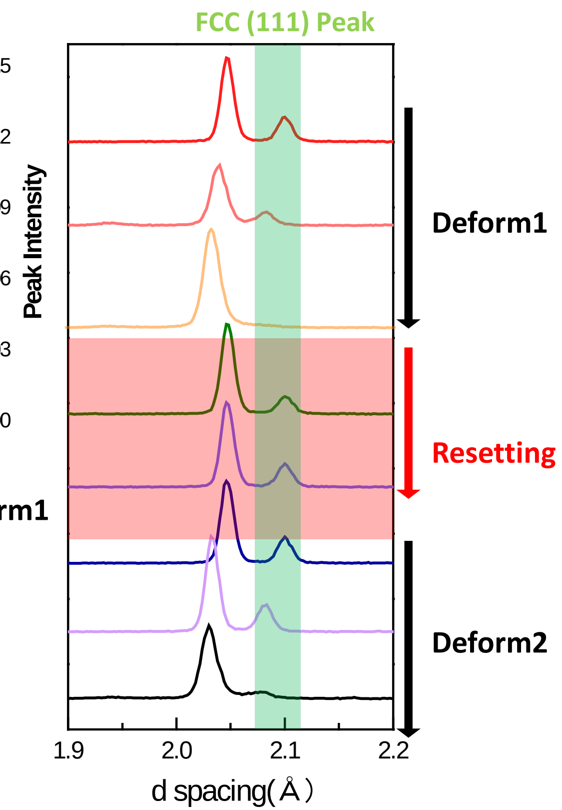
: (손상)-(회복)-(손상) cycle 시의 A 상분율 정량 분석을 통한 손상 정량화

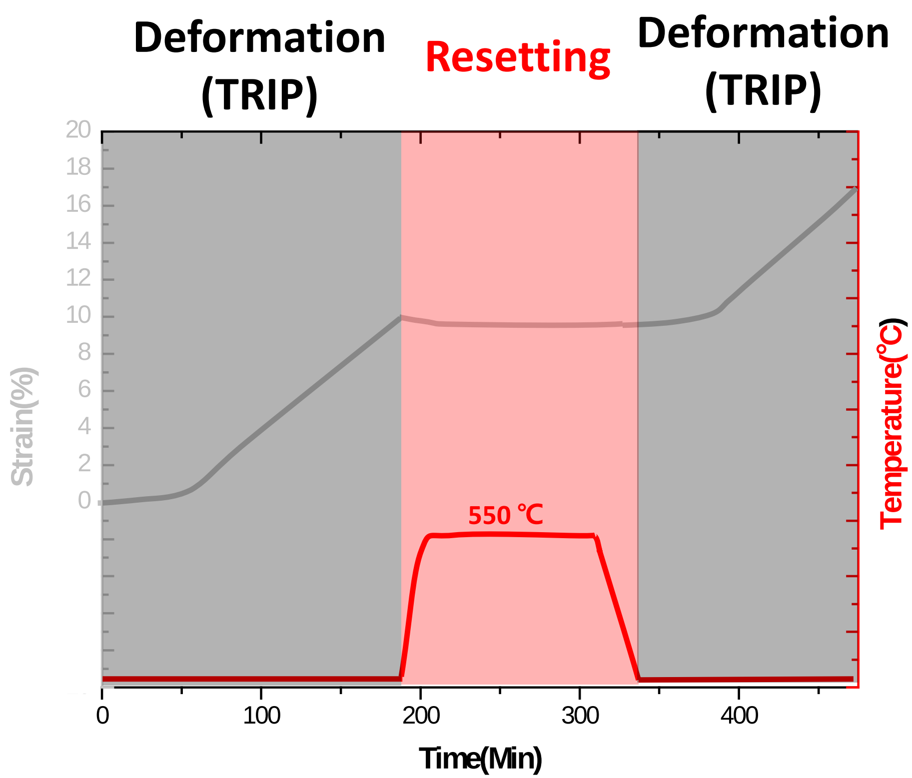


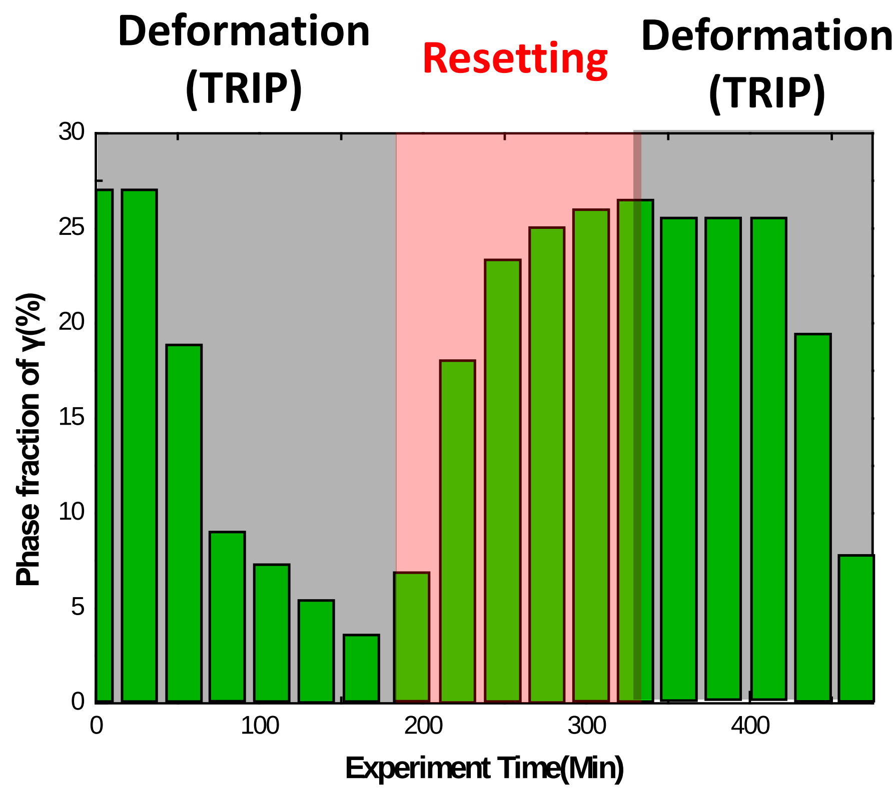


(3) 실험결과 II : 리셋 공정 최적화 변형 Informatics 구축

결함 정량화 Smart Monitoring 기반 Deformation Informatics 구축

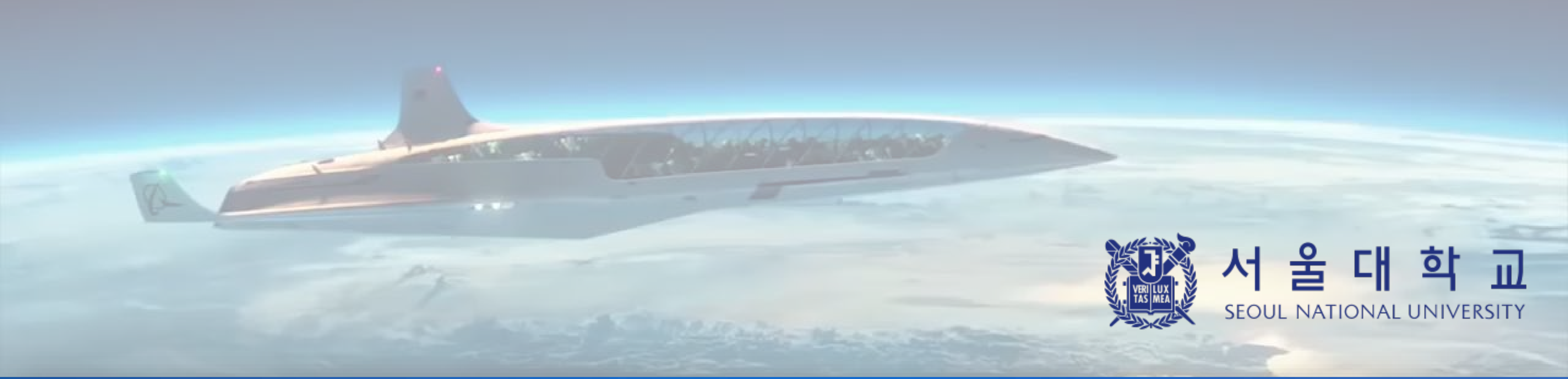
② 비파괴 검사 통한 사용환경 중 변형 Smart Monitoring 기술 개발





▶ (손상) – (회복) 에 따른 상분율 정량 변화 분석을 통한 Deformation Informatics 구축!

2019년도 미래소재디스커버리사업 선정 2019년- 2025년 (수행중)



지능형 자가변환 기반 자가치유 센테니얼 합금 개발

연구책임자: 박은수 (서울대학교 재료공학부)

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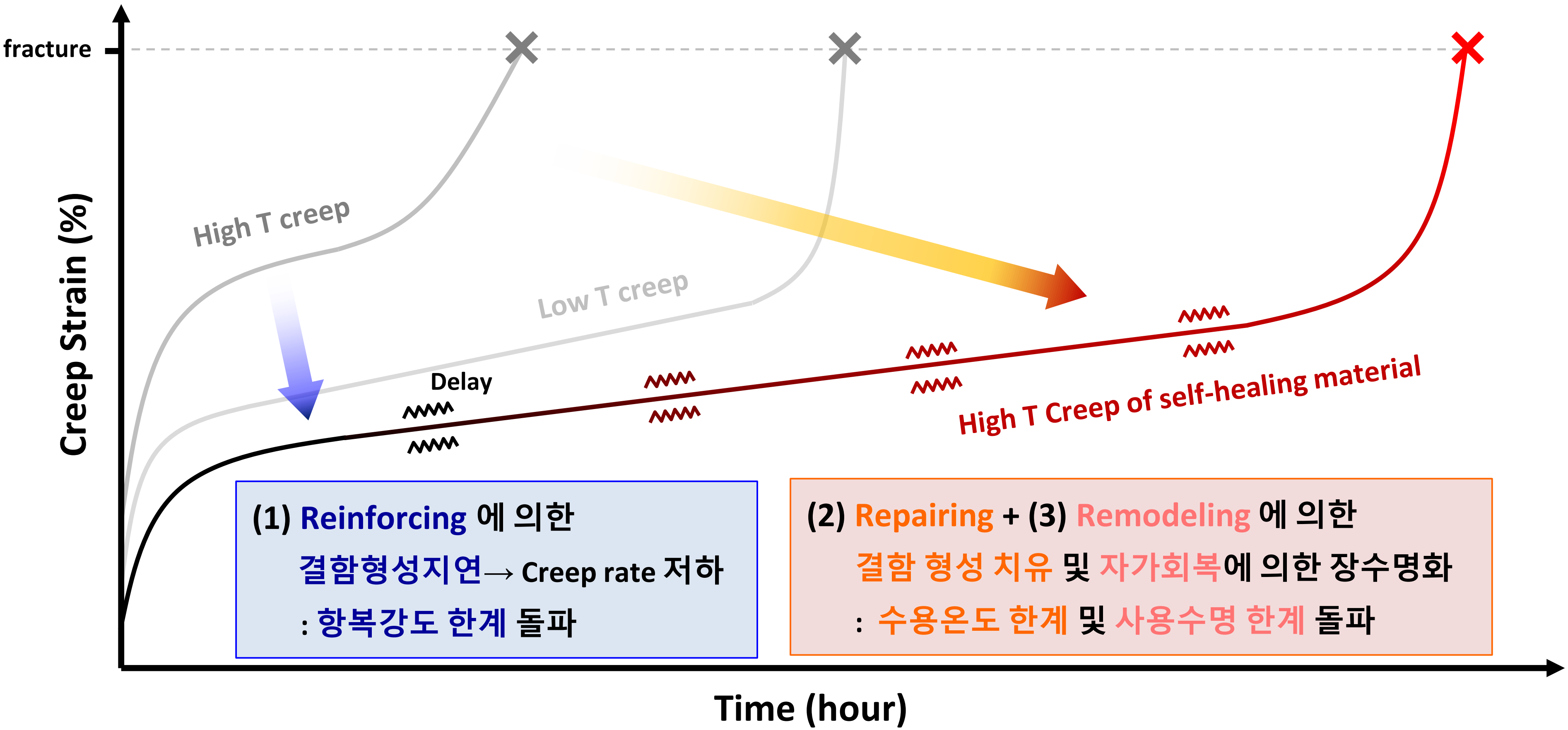
New Challenges ! = 지능형 자가치유 센테니얼 합금

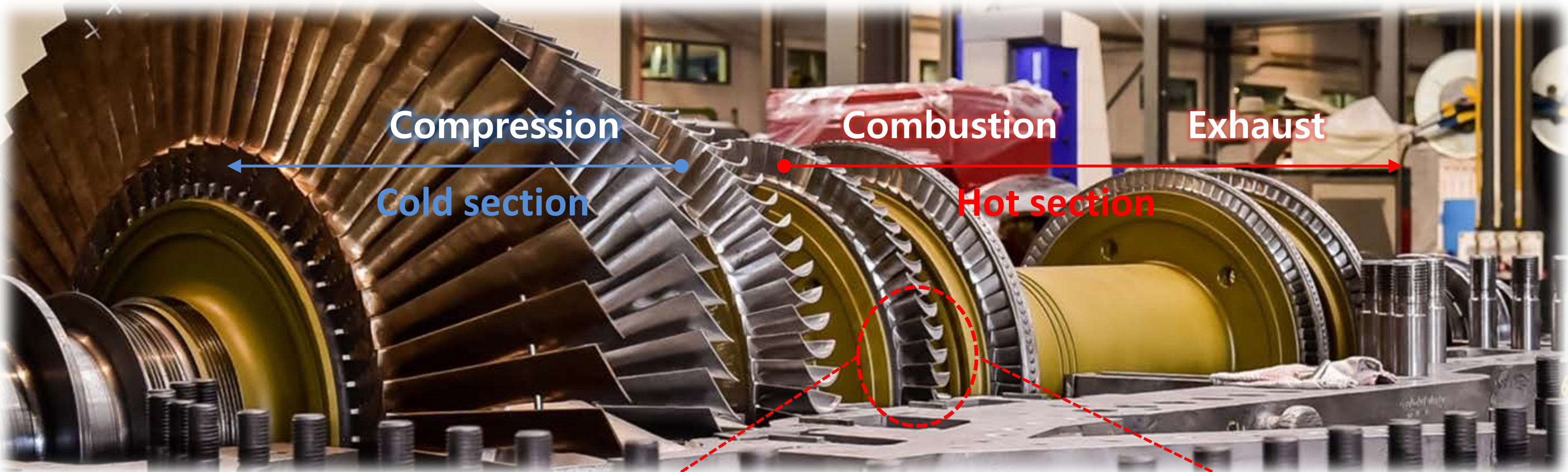
Creep def. of Centennial alloy (high T or high σ)



(항복강도 · 수용온도 · 사용수명) 한계 돌파 센테니얼 합금

맞춤형 구조제어 직접디지털제조 공정기술 개발

Target 부품 : 가스복합화력발전용 H-class (TIT 1500℃) gas turbine * Siemens SGT6-8000 H



고효율 균일냉각
Complex cooling channel

고온 크리프 특성 향상
자가치유 + 미세구조 제어

고온 강도 향상
Precipitate control

3D 프린팅 기반의 초고효율 블레이드 부품개발

❖ 발전부품 3D 프린팅 현재 국내 기술수준: 인코넬 이용 E/F-class (TIT 1300°C 이하) vane 개발 진행 중

Contents in Phase Transformation

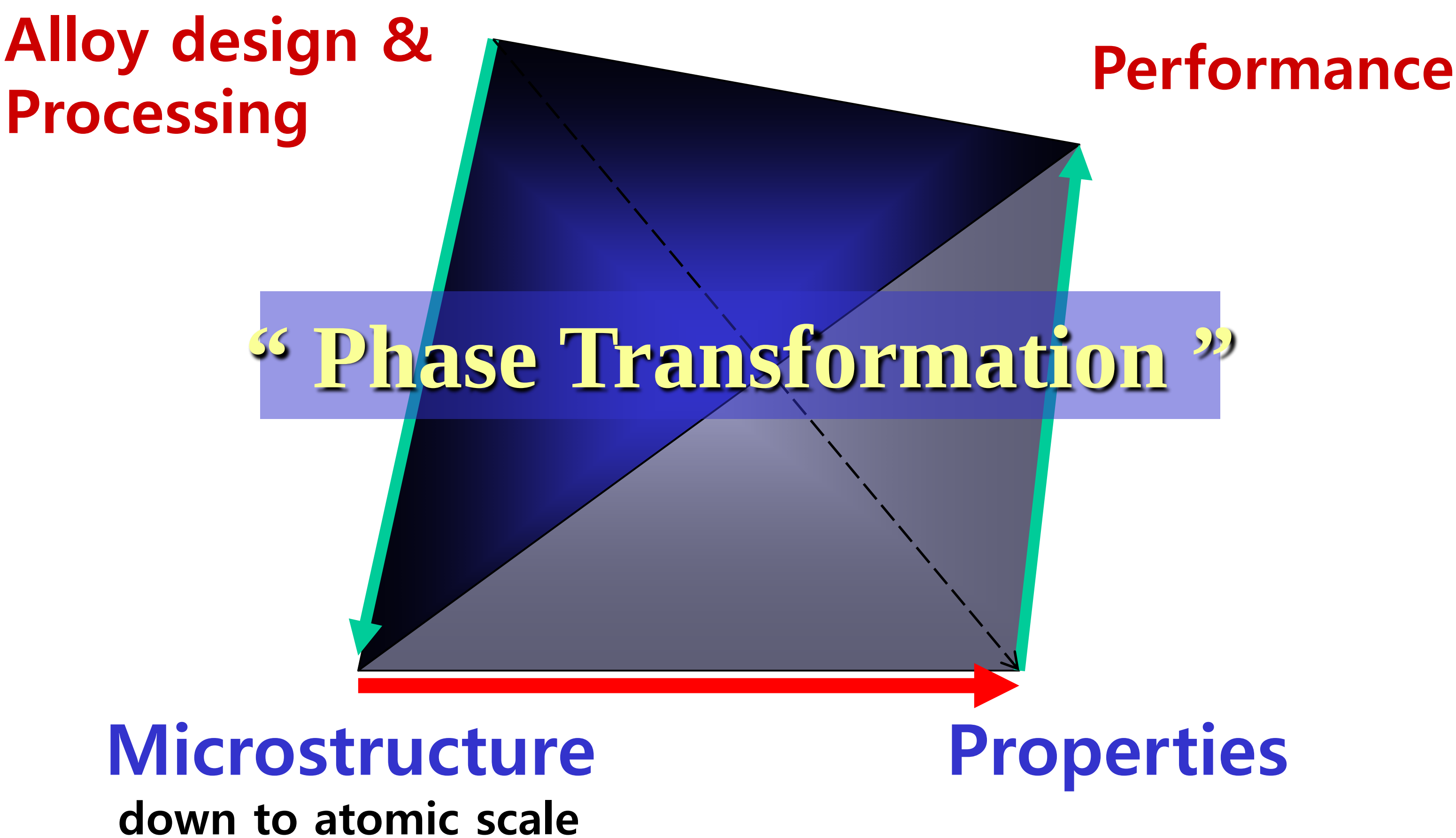
Background to understand phase transformation

- (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

Microstructure-Properties Relationships



"Tailor-made Materials Design"