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6. Diffusionless Transf.

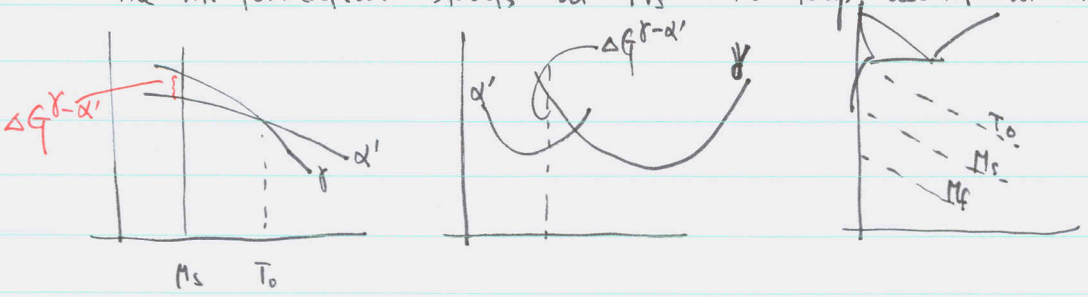
6.1. Charac. of Diffusionless Transf.

- the formatim of M appears to be a random process
- M(α') in the shape of a lens and spans an entire grain
  the ρ of plates ≠ f (γ grain size)
  ① the transformed region coherent w/ the surrounding γ.
  ② the intersectiom of α' w/ the surface — continuous
  ③ happens quick (10^{-7} s to span in a grain)
  α'/γ interface speed ~ sound. in the solid.

- Fig 6.2

cf. M of Fe-Ni exhibits isothermal growth. growth: indep. of thermal activa

- The M. formatim starts at M_s : the temp. assoc'td w/ a d.f.



- Increasing C contents decreases the M_s temp. M_f : finish
  Some retained γ (10-15%) may be not 100% mart.
  due to the high elastic stressed int'n M. plates.

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Like in eq.(1.11) the driving force for the nucl. of M at Ms.

$$\Delta G^{\gamma-\alpha'} = \Delta H^{\gamma-\alpha'} \frac{(T_o - Ms)}{T_o}$$

Table 6.1 → ΔH information. 340 J/m

The ordered alloys show a rel. small ΔT. (Fe - Pt 24 % ordered)

disordered Fe - Pt 2400 J/m.

6.1.1. The solid Sol'n of Carbon in Fe.

- In an fcc lattice struc. two possible positions for interst'l atoms.
  tetrahedral, octahedral.

Fig. 6.4. $d_a = 0.225\,D$ ($0.568\,Å$) $d_b = 0.414\,D$ ($1.044\,Å$) D: parent Fe dia. 2.52Å

+ the dia. of carbon 1.54 Å → considerable distortion
  ↳ octahedral sites occupied ⊥

- In a bcc struc. 3 possible octahedral sites ($\frac{1}{2}[100], \frac{1}{2}[010], \frac{1}{2}[00$
  & 6 " tetra " in each unit cell.

$d_a = 0.291\,D$ $d_b = 0.155\,D$ D: 2.866Å.

note! ① although there's more 'free' space in bcc than the c-packed lattices, the larger # of possible interst'l sites (9) cause less space available / interst'l than for fcc.

② $d_b < d_a$, still C, N prefer octa. sites very few ($\frac{1}{2}00$) sites
→ causes considerable distortion.

distortion from ~~bcc~~ fcc to bct (body centered tet.)

at -100°C the c/a ratio of the bct

Fig 6.5 (c) $c/a = 1.005 + 0.045$ (w/o C)

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6.2 Martensite Xtallography

- Xtall·phic dependence, M grows, within a grain, in a limited # of orientations.
  ex. Fe-alloys, the orientation variants and plate morphology chosen
  - dep. on. alloy content (C or Ni)

- The irrational nature of the growth planes of high C or Ni Martensites: Subject of much discussion. Reasons?

  ① if M grows at speeds ~ that of sound, it needs highly mobile disl. interface.

  a. then, have to explain why the high mobility of an interface moving on γ planes (which are not always ass'd w/ ⊥ glide.

  b. the habit plane of M is observed to be macroscopically undistorted
  ↳ a plane common to both γ & M.

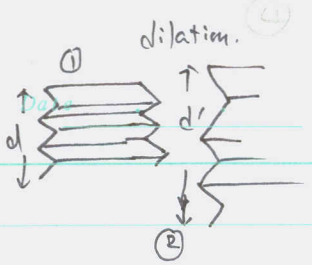
  c. → the absence of plastic def in the form of discontinuity at the surface indicates that the shape strain does not cause any noticeable rotation of the habit plane.

  → Coherence bet'n M and γ, this would cause additional displacement

  1 For the habit plane undistorted, the M. transf. by a homo. shear // to the h.p.

  2 γ → α' entails ΔV ↑. (4% expansion), ∴ the dilatation must occur normal to the habit plane (normal to the lens)

  3 Then, can the bcf M. lattice struc. be generated by simple shear?



the undistorted habit plane similar to twinning

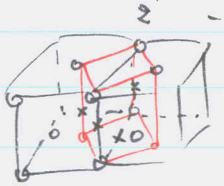
Assume that ① the equivalent macro. shape change in M formation is a twinning shear // to the h.p. (twining p)

② + a simple uniaxial tensile dilatation ⊥ to the h.p. – strain called an invariant plane strain

$\varepsilon_2 = 0$ → in twining plan

6.2-1. The Bain model of the fcc → bct transf.

– Bain showed $\gamma \rightarrow \alpha'$(bct) w/ the min. of atom movement & ε in the parent lattice.



z' – an elongated unit cell of the bcc struc from two fcc cell.

α'(bcc) can be achieved by

a) 20% contraction in the z direction

b) 12% expansion in the x, y dine.

carbon sites in z' of the bcc at $\frac{1}{2}\langle 100 \rangle$.

→ not exactly match w/ the position in fcc octa sites: the position change of C during transf.

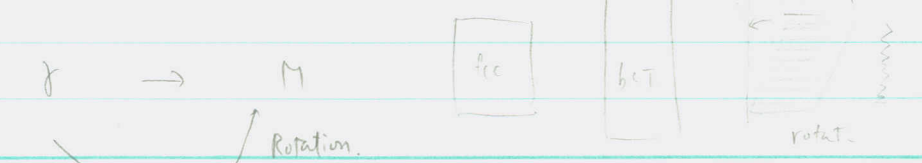
– Min. atom movement : Bain distortion. results in

$(111)_\gamma \rightarrow (011)_{\alpha'}$

$[\bar{1}01]_\gamma \rightarrow [\bar{1}\bar{1}1]_{\alpha'}$ (K-S relation) } orientation 5° off.

$[\bar{1}\bar{1}0]_\gamma \rightarrow [100]_{\alpha'}$ (N-W relation) }

$[11\bar{2}]_\gamma \rightarrow [01\bar{1}]_{\alpha'}$



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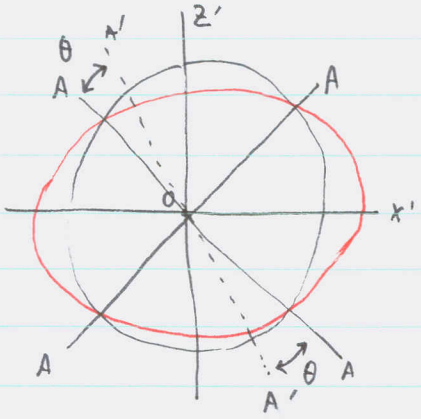
- Invariant Plane in Bain Distortion. : not moving atoms
just distance change



Sphere – represent the fcc.
ellipsoid. – " the struc. after Bain D. (B.D)

x'. & y' axis 12% expasion
z' axis 20% contraction
after B.D, the only vectors (line) not shortened = OA or OA'

How about plane? the vector to be a plane (undistorted)
there shouldn't be distortion in OY' axis, but if is not so.
(cylinder)

- the theory of Mart. Transf.
requires an additional distortion. (postulation)
that reduce the actual expansion along OY' to zero.
– in the form of ⊥ slip or twinning

In Th fcc - fct.

– an internally twinned martensite plate

– the habit plane of the M: a plane w/ no net distortion for γ.
distort. avg. over many twins ≃ 0.
∴ strain energy w/ α'/γ interface of the twins exists
if the M plate is thin (a few atomic spacings). ε → small

Slip required

slip or twin on $\langle 111 \rangle \{11\bar{2}\}_{\alpha'}$ equivalent to $\langle \bar{1}\bar{1}0 \rangle \{110\}_{\gamma}$
common slip plane in bcc slip or twinning.

6.2.2. Comparison of Xtallographic Theory w/ Exp. Results.

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- Alloying addition have a great effect on the habit plane.
  - As C content ↑ in Fe-C habit plane {111} → {225} → {259}$_\gamma$
    - {259}$_\gamma$: twinning plates, lens morphology
    - associated w/ a high ± ρ, lath morphology & bundles of needles on {111}$_\gamma$
  - Morphology quite irregular
    Twinning predmominant at high C or Ni {259}$_\gamma$
    stainless steel {11$\bar{2}$}
  - difficult to predict habit plane based on lattice para., dilatation.

- Theory only phenomenological. : no kinetic. use.

6.3 Theories of M. Nucleation.

- growth speed : 10^{-5} to 10^{-7}/s to grow its full size (800–1000 m/s
  measured using resistivity.
- nucleation important in determining final q.s. of M
  related to toughness and strength
  # of nuclei ↑ → q.struc. finer.!!

- the initial M. nucleus is coherent w/ the parent γ.

6.3.1. Formation of Coherent Nuclei of M.

(6.5) $\Delta G = A\gamma + V\Delta G_s - V\Delta G_v$. V: vol. of nucleus

No consideration of the energy due to thermal stress during cooling
applied stress etc

→ strain energy important ∵ the shear comp. of Bain strain ≃ 0.32

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the interfacial (surface) energy, relatively small ∵ coh. nature.

- Nucl. of a thin ellipsoidal nucleus (radius: a thickness: c vol.: V

Assumptions: nucl. in the matrix. homogeneous nucl.

the nucleus forms by a simple shear, s, // to the plane of the disc. (complete coherency at the interfac

(6.6) from (6.5) $\Delta G = 2\pi a^2\gamma + V\cdot 2\mu\left(\frac{s}{2}\right)^2 \frac{2(2-\nu)}{9(1-\nu)}\pi\frac{c}{a} - \frac{4}{3}\pi a^2 c\cdot\Delta G_v$

μ = shear modulus

$\nu = \frac{1}{3}$ for metal

Fig. 6.3(a)

(6.7) $\Delta G = 2\pi a^2\gamma + \frac{16\pi}{3}ac^2\left(\frac{s}{2}\right)^2 - \frac{4}{3}\pi a^2 c\cdot\Delta G_v$ ($\gamma \to \alpha'$)

only shear energy w/o ε due to the dilatation. $\perp$ to the disc.

→ the most favorable nucl. path if habit plane = invariat plane (exact)

- min. free energy barrier to nucl.

differentiat wrt a & c.

(6.8) $\Delta G^* = \frac{512}{3}\frac{\gamma^3}{(\Delta G_v)^4}\cdot\left(\frac{s}{2}\right)^4 \pi^2$ J/nucleus.

↳ sensitive to γ, ΔG_v, s

(6.9)(6.10) $c^* = \frac{2\gamma}{\Delta G_v}$ $\quad a^* = \frac{16\gamma\mu\left(\frac{s}{2}\right)^2}{(\Delta G_v)^2}$

$\Delta G_v = 174\ MJ/m^3$, $\gamma = \sim 20$ mJ. $c^*/a^* \simeq 1/40$, then $\Delta G^* = 20$ eV

→ high for thermal fluctuation to overcome (at 700k, $kT = 0.06$ eV)

∴

→ M. nucl.: a heterogeneous process

exp. w/ small droplets (Submicron to millimeter)

evidences ① Not all particles transformed even at 4°K w/ $\Delta T = 300°C$

② # of nuclei less than that expected from homo. ($\sim 10^4/mm^3$).

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③ # of nuclei ↑ as ΔT ↑ # of nuclei indep of (g.s) (g.b) simple. polycrystalline

④ the surface . not preferred nucl. sites. Dislocation is it!