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2.3 Substitutional Diffusion.

~~In dilute~~ I.D. - simple in a dilute sol'n ∵ surrounded by vac. sites.
only energy needed for passing the barrier.
S.D - need a vac. site to jump.

1. Self - Diffusion. (in a pure metal) → similar to the case of I.D.
using isotopes (or (radioactive) chemically identical at various T's
Γ (jump freq) almost identical
successful jump/sec

(2.27) $\cancel{D_A^*} = \boxed{D_A = \frac{1}{6}\alpha^2\Gamma}$

Random Walk: the assumption that each atomic jump is unrelated to the previous one
→ good for I.D. but less valid for S.D.

this eq. more ~~appropriate~~ reasonable!
→ ∵ $\underset{(S.D)}{D_A^*} = \underset{(S.D)}{f D_A} = f \cdot \alpha^2\Gamma/6$ (≥1)
└ correlation factor ($f < 1$)

for S.D. (불편한 jump의 next one will jump back to its original position. (most likely)
for I.D → only a fraction of jump will go back to the orig. site

the probability of a successful jump.

(2.28) $\boxed{\Gamma = \nu Z X_v \exp(-\Delta G_m/RT)}$

nearest neighbor
↳ probability that any one site is vac. (mole fraction)

if $X_v = X_v^e$ vac. conc in equil$^{\underline{m}}$

(1.57) = (2.29) $X_v^e = \exp\left(-\frac{\Delta G_v}{RT}\right)$. formation of vac

(2.30) $D_A = \frac{1}{6}\alpha^2 Z \nu \exp\frac{-(\Delta G_m + \Delta G_v)}{RT}$

(2.31) $= \underbrace{\frac{1}{6}\alpha^2 Z \nu \exp\frac{\Delta S_m + \Delta S_v}{R}}_{D_0} \exp\left[-\frac{\Delta H_m + \Delta H_v}{RT}\right]$

(2.32, 33) $D_A = D_0 \exp\left[-\frac{Q_{S.D}}{RT}\right]$

Richard's Rule. $\Delta S = \frac{\Delta H_m}{T_m} = \frac{L}{T_m} \simeq R$.

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· Table 2.2. for self-Diffusion.

for a given Xtal struc. and bond type $Q/RT_m \simeq$ const.

(i.e. $\Delta H_m + \Delta H_v$ = const)

$Q/RT_m \cong$ const means $Q \propto T_m$. & $D(T_m)$, $D_0 \simeq$ consts.
within each class

ex. for fcc & hcp $Q/RT_m = \sim 18$ & $D(T_m)$ 10^{-12} m^2/sec.

Fig 2.13 → D_0 of all mat'ls w/ a given Xtal struc. & bond type.
homolopous Temp.
will have $D(T/T_m)$ = const. behavior.
↳ indep. of pressure.

ex. the effect of Temp. on self-diffusion of Cu.

at 800°c in Table 2.2 $D_{Cu} = 5 \times 10^{-9}$ mm^2/sec $\alpha = 0.25nm$ $= \frac{a}{\sqrt{2}}$

$D_B = \frac{1}{6}\alpha^2\Gamma$

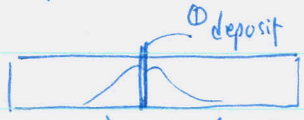
Eq. 2.3 → $\Gamma_{Cu} = 5 \times 10^5$ jumps/sec. 1 hr → $\sqrt{Dt} \sim 4\mu m$.

→ extrapolating to 20°c $D_{Cu} \sim 10^{-34}$ [mm^2/s] $\Gamma = 10^{-20}$ jump/s.

12 yrs/jump.

Measurement of $D_{S.D.}$

Fig 2.14



① deposit a known quantity (M) of a radioactive isotope A^*

② 'joined' ③ slicing the bar.

(2.18) $\frac{\partial C_B}{\partial t} = D_B \frac{\partial^2 C_B}{\partial x^2}$ → soln $C = \frac{M}{2\sqrt{\pi D t}} \exp\left(-\frac{x^2}{4Dt}\right)$ grams/cm^2

2. Vacancy Diffusion.

atoms jumping to vac sites $\overset{\text{equivalent to}}{\Longleftrightarrow}$ vac jumping to atom sites

'But' a vacancy can jump to any atom sites (similar to int. atoms) like C.

since it's surrounded by atoms. (an atom can jump only to vac.) limited !!

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∴ vacancy is regarded to have its own D_o ($\Rightarrow D_v$)

i.e.,

(2.36) $$D_v = \frac{1}{6}\alpha^2 \Gamma_v$$

(2.37) ~~(2.8)~~ Like (2.8) $$D_v = \frac{1}{6}\alpha^2 z v \exp\frac{\Delta S_m}{R} \exp\frac{-\Delta H_m}{RT}$$

migration of vacancy.

ΔS_m, ΔH_m for the migration of a vac. = subl atom.

∴ from (2.37) and (2.31) $\boxed{D_v = D_A / X_v^e}$

주인공 atom이 migration에 관여하므로 ∴ $D_v \gg D_A$

Quiz # 10/7/03

3. Diffusion in "Subst'l Alloys."

① for Self-diffusion P(finding vac. adj to any atom) = P(atom make a jump to vac)
↳ leads to a relationship bet'n Γ and D_o.

② In binary subst'l alloys → complex. w/ solvent (A) & solute (B).
intrinsic diff. coeff. D_A & D_B. are diff. (diff tendency to jump into a vac

- when the A. & B atoms jump at diff rates, the presence of conc. grad. induce a "movement of lattice"

- D_A, D_B diffusion relative to the lattice (moving)

$$J_A = -D_A \frac{\partial C_A}{\partial x}, \quad J_B = -D_B \frac{\partial C_B}{\partial x}$$

J_A, J_B across a given lattice plane (moving)

cf. In the case of interstitial diffuse no need to say
∵ the lattice plane of parent atoms not moving by diffusion of the interstitial.

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Fig 2.15 a Assume that total # of atoms / unit vol. = const, C_0

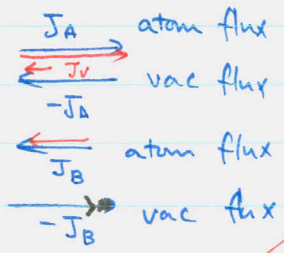
(2.41) indep. of comp. then $C_0 = C_A + C_B$

(2.42) and $\frac{\partial C_A}{\partial x} = -\frac{\partial C_B}{\partial x}$

(2.43) the fluxes of A & B rel. to the lattice $J_A = -D_A \frac{\partial C_A}{\partial x}$, $J_B = +D_B \frac{\partial C_A}{\partial x}$

- for the case $D_A > D_B$ $|J_A| > |J_B|$. diffusion couple.

Fig 2.15 a



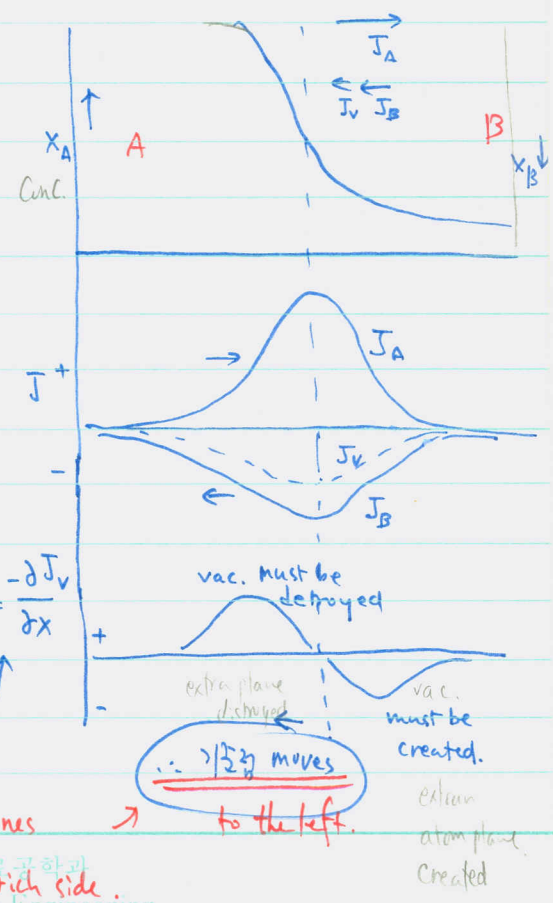
(2.44) $\therefore J_v = -J_A - J_B$.

($\because J_A = -(J_v + J_B)$).

반대 방향.

(2.45) $\therefore J_v = (D_A - D_B)\frac{\partial C_A}{\partial x}$

$\because$ a variation in J_v across the diff. couple.



In order to maintain the vac. conc. near equi^{m} vac must be created & destroyed in B, A sides, respectively. at these rates $\leftarrow \frac{\partial C_v}{\partial t} = -\frac{\partial J_v}{\partial x}$

$\therefore$ 경계 moves to the left.

jopped edge ⊥

extra atom planes in B side, whole planes of atoms will be disappear on the A-rich side.

if the velocity of the lattice plane, v, related to the flux of vac. crossing it.

then, vol. swept out $Av \cdot \delta t$, $Av \cdot \delta t \cdot C_o$ atoms.

(total ~~unit~~ area crossing; # atoms/vol; plane vel.; $v \cdot A \cdot \delta t$; vacant area)

= total # of vac. crossing the plane $= J_v A \cdot \delta t$

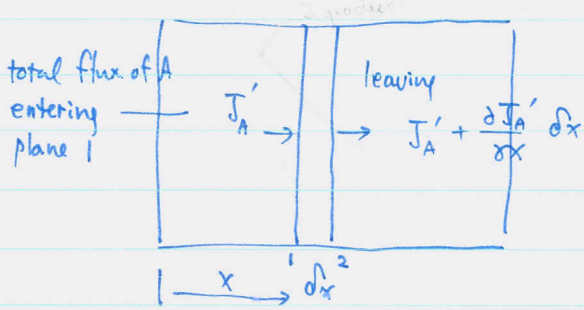
(2.46) $\therefore \quad \boxed{J_v = C_o v}$, "$v$ varies like J_v as a fn of x."

$= (D_A - D_B)\frac{\partial C_A}{\partial x}$

(2.47) (2.45) → $\boxed{v = (D_A - D_B)\frac{\partial X_A}{\partial x}}$ $\left(X_A = C_A/C_o = \text{mole frac. of } A\right)$

— Practical questions are about ① the length of time for homogenization.
& ② the change in composition at a fixed position relative to the ends of a specimen.

∴ Fick's 2nd law for Sub.'l Alloys. (needed)



or. (2.16) $\frac{\partial C_A}{\partial t} = -\frac{\partial J_A'}{\partial x}$ (2.48) w.r.t. the end of specimen

the total flux of A atoms across a stationary plane w.r.t. the ends of a spe (w.r.t. lattice)

= the sum of i) a diffusive flux $J_A = -D_A \frac{\partial C_A}{\partial x}$

& ii) a flux $v \cdot C_A$ (plane velocity × local con

(2.49) $\therefore \quad J_A' = -D_A \frac{\partial C_A}{\partial x} + v C_A$ (J_A w.r.t. the end of specimen; <0; >0)

(Atoms (A & B) → extra plane 형성; vac from extra plane; B atom; C_A)

(2.47) + (2.49) $\quad J_A' = -D_A \frac{\partial C_A}{\partial x} + C_A (D_A - D_B)\frac{\partial X_A}{\partial x}$ $\left(\partial X_A = \frac{\partial C_A}{C_o}\right)$ ($\frac{1}{C_o}\frac{\partial C_A}{\partial x}$; neg 부호를 곱함)

$= -D_A \frac{\partial C_A}{\partial x} + (X_A D_A - X_A D_B)\frac{\partial C_A}{\partial x}$ ($X_B + X_A$)

(2.49) $= -(X_B D_A + X_A D_B)\frac{\partial C_A}{\partial x}$ ← Fick's 1st law.

(2.51) Interdiffusion Coeff.: $\boxed{\tilde{D} = X_B D_A + X_A D_B}$ → $J_A' = -\tilde{D}\frac{\partial C_A}{\partial x}$. (2.52)

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$$\therefore \quad J_B' = -\tilde{D}\frac{\partial C_B}{\partial x} = \tilde{D}\frac{\partial C_A}{\partial x} \qquad \text{i.e} \quad J_B' = -J_A'$$

$$(2.53) \quad (2.48) + (2.52) \rightarrow \quad \frac{\partial C_A}{\partial t} = \frac{\partial}{\partial x}\left(\tilde{D}\frac{\partial C_A}{\partial x}\right) \Rightarrow$$ Fick's 2nd law for subst'l alloy.

called Darken's eq

① !! Solve this eq. w/ proper b.c. to have $C_A(x,t)$, $C_B(x,t)$!!

Also.

② Sec. 2-2-5. sol'ns for 1.D. applicable to Subs. alloy as long as the range of comp. is small that $\tilde{D} \neq f(\text{comp})$. strongly.

$X_A D_B + X_B D_A$ if X_B very small

— Measurement of $\tilde{D}$ ① by determining X_A, X_B after annealing. w/ known D_A, D_B

② if $\tilde{D}$ is assumed const, a comparison of eq. 2.26 & conc. profile (measured) would give $\tilde{D}$.

(2.26) $$\boxed{C = \left(\frac{C_1+C_2}{2}\right) - \left(\frac{C_1-C_2}{2}\right)\mathrm{erf}\left(\frac{x}{2\sqrt{\tilde{D}t}}\right)}$$

C_1 C_2

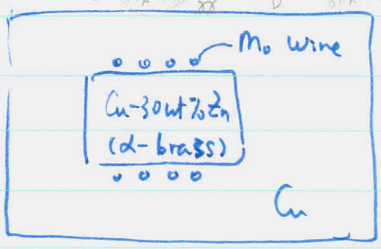
③ when $\tilde{D}$ is not const. solve it graphically. to meet Fick's 2nd law. (2.53)

— • To know D_A, D_B. need to know the the velocity of the lattice, v, by inserting insoluble wires at the interface before welding the two blocks together. ↳ fixed to the lattice place ∴ moves to give v

• w/ known v & $\tilde{D}$, use (2.47), (2.51) to have D_A & D_B.

$v = (D_A - D_B)\frac{\partial X_A}{\partial x}$ $\quad \tilde{D} = X_B D_A + X_A D_B$

— Kirkendall Effect

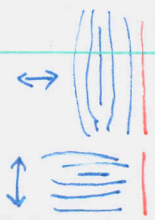


∴ $D_{Zn} > D_{Cu}$ atom radius 1.55Å, 1.57Å

low Tm → bonding force 약함

In general, in a given couple like above, atoms w/ low Tm. have higher D.

— Growth or Shrink of Voids



expansion parallel to or perpendicular to the weld interface

" restricted by surrounding : compressive or tensile force.

→ induce plastic def → hot deformation

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(2.55) $\tilde{D} = \tilde{D}_0 \exp\left(\frac{-Q}{RT}\right)$

(2.56) $D_A = D_{A0} \exp\left(-\frac{Q_A}{RT}\right)$ likewise. $D_B =$

- Rough estimate of $\tilde{D}$ based on the exp. results.

1. For a given xtal struc. $\tilde{D}(T_{m1}) = \text{const.}$
   ∴ if B was added to A, resulting in lower T_{m2}
   then $\tilde{D}(T_{m2})$ after B addition will be higher than before.
2. - Interstitial C in Fe. at 910°C $D_C^{\alpha} / D_C^{\gamma} \sim 100$
   - Sub - self Diffusion at 850°C $D_{Fe}^{\alpha} / D_{Fe}^{\gamma} \sim 100$
   → open struc allows higher Diffusivities.

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4. Diffusion in Dilute Subst'l Alloys.

- when $X_B \sim 0$, $X_A \sim 1$.

(2.58) (2.51) → $\tilde{D} = X_A D_B + X_B D_A = D_B$. ← controlling factor for homogenization only by B migration.

impurity diff. coeff.

- Normally $D_B > D_A$ in a dilute soln ∵ the solute atoms attract vac.
  → more than a random prob. of finding vac.
  ⇒ It is true when the solute atom is larger than the solvent atom in size. and when it has a higher valency
  ⇒ It could form the solute - vac. pair if the binding E is very large.
  ⇒ The pair diffuse together.

for neutrality it requires vacancy

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2.4 Atomic Mobility.

- Diffusion near defects is usually influenced by both the conc. grad. & the grad. of interaction E. ∴ Fick's law is insufficient.

  ex. a large solute atom want to find a large space like g.b or near disl. to reduce the strain.

  ex. segregation of atoms to g.b. interfaces and disl.
  - C, N in mild steel to disl. → responsible for strain ageing & blue brittleness (?)
  - Sb. Sn. P. As to g.b. in low-alloy steel → temper embrittlement
    ~~causes~~ affects the g.b. mobility. recrystallizing texture, g. growth

- Soln to this ; thermo condition for equil^m (i.e. const. chem. potential. the same in the sys.) diffusion continues until this cond. is satisfied.

  ∴ $J \propto \frac{\partial \mu}{\partial x}$ (= Chem. pot. grad.)

- An other way to describe a flux of atom. = net drift ~~of~~ vel ($\bar{v}$) (by defects (vac. disl. g.b)) + random jumping.

(2.59) J_B [= atoms/cm²·sec] = v_B [cm/sec] C_B [atoms/cm³]

(2.60) drift vel. = $v_B = -M_B \frac{\partial \mu_B}{\partial x}$; atoms move to remove the diff. in chem pot.

force = $(D_A - D_B)\frac{\partial x_A}{\partial x} = v$ ↳ prop'lty const. = atomic mobility.

(2.61) (2.59) + (2.60) $J_B = -M_B C_B \frac{\partial \mu_B}{\partial x}$

- The relationship bet'n mobility and D_o. (→ related) by relating $\frac{\partial \mu}{\partial x}$ to $\frac{\partial c}{\partial x}$

$-x_A d\mu_A = x_B d\mu_B = RT\left\{1 + \frac{d\ln\gamma_B}{d\ln x_B}\right\} dx_B$

(2.62) (1.70) + $C_B = x_B / V_m$ → $J_B = -M_B \frac{x_B}{V_m} \cdot \frac{RT}{x_B}\left\{1 + \frac{d\ln\gamma_B}{d\ln x_B}\right\}\frac{\partial x_B}{\partial x}$

(2.63) $\boxed{J_B = -M_B RT\left\{1 + \frac{d\ln\gamma_B}{d\ln x_B}\right\}\frac{\partial C_B}{\partial x}}$

No strain field.

(2.64) ∴ $\boxed{D_B = M_B RT \left\{ 1 + \frac{d \ln \gamma_B}{d \ln X_B} \right\}}$, $D_A = M_A RT \left\{ 1 + \frac{d \ln \gamma_A}{d \ln X_A} \right\}$

(2.65)

for ideal, dilute soln s ($X_B \to 0$) $\gamma_B = \text{const}$.

(2.66) ∴ $D_B = M_B RT$.

related the curvature of the molar free E.-comp. curve.

– w/ a strain E. grad.

(2.67) $\mu_B = G_B^\circ + RT \ln \gamma_B X_B + E$

(2.63)

(2.68) $J_B = -D_B \frac{\partial C_B}{\partial x} - \frac{D_B C_B}{RT} \cdot \frac{\partial E}{\partial x}$

$J_B = -M_B C_B \frac{\partial \mu_B}{\partial x}$ $\mu_B : G_B^\circ + RT \cdots$

$= -M_B C_B \left[RT \left(\frac{\partial \ln \gamma_B}{\partial x} + \frac{\partial \ln X_B}{\partial x} \right) + \frac{\partial E}{\partial x} \right]$

$= -M_B C_B \left[RT \frac{\partial X_B}{\partial x} \frac{1}{X_B} \left\{ \frac{d \gamma_B}{d X_B} \frac{1}{\gamma_B} + 1 \right\} + \frac{\partial E}{\partial x} \right]$

$= -M_B \frac{X_B}{V_m} \left[RT \frac{\partial}{\partial x} \frac{V_m dC_B}{X_B} \left\{ 1 + \frac{d \ln \gamma_B}{d \ln X_B} \right\} \right]$

$- M_B C_B \frac{\partial E}{\partial x} = -D_B \frac{\partial C_B}{\partial x} - \frac{D_B C_B}{RT} \frac{\partial E}{\partial x}$

$D_B \cong M_B RT$

$\left(C_B = \frac{X_B}{V_m} \right)$

– Uphill diffusion w/ an elec. field or temp. grad.
(= electromigration, thermo migration).

2.5. Tracer Diffusion in Binary Alloys.

- Use radioactive tracers to determine the intrinsic diff. coeff.
  - B^* (small quantity) on A/B soln bar. (homogeneous bar)
  - use eq. (2.35) → $C = \frac{M}{2\sqrt{\pi D t}} \exp\left(-\frac{X^2}{4Dt}\right)$ → the tracer diff. coeff. D_B^*

D_B^*

- for Au–Ni alloy w/ miscibility gap at low temps. ⟺ Au, Ni dislike each other
  no conc. grad, chem. homo. — conc. grad
  D_{Au}^*, D_{Ni}^* (tracer) ≠ D_{Au}, D_{Ni} (marker)의 의미는 없는 것
  D_{Au}^* : Au^* diffuse in a chemically homo. alloy
  D_{Au} : when a conc. grad. is present.

D_{Au}^* : the jumps by Au in a chem. homo. equally probable in all dir.
D_{Au} : in a conc. grad. the jump will be biased away from Ni-rich region.

∴ $D_{Au}^* > D_{Au}$, $D_{Ni}^* > D_{Ni}$

– If assuming $M \neq f(\text{conc.})$ atomic mobility

$D_B = M_B RT \left\{ 1 + \frac{d \ln \gamma_B}{d \ln X_B} \right\}$

$J = -D_B \frac{\partial C}{\partial x}$

특정 (A/B comp.가 정해진) sys.에서의 D^* (ratio)

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$D = f(\text{conc})$

Intrinsic w/ conc. grad. homo

* Relationship bet'n chemical diffusivity & tracer diffusivity.
  - the tracer forms a dilute sol'n

tracer B 극희 dilute sol'n A/B homo. only

assuming $M \neq f(\text{conc.})$ $\therefore M_B^* = M_B$

(2.69) (2.66) → $D_B^* = M_B^* RT = M_B RT$ $\therefore D_B = M_B RT\left\{1 + \frac{d\ln\gamma_B}{d\ln x_B}\right\} = D_B^* \cdot F$

$\because M \neq f(\text{conc})$

$\tilde{D} = X_B D_A + X_A D_B$ $D_B = M_B RT[1 + \frac{d\ln\gamma_B}{d\ln x_B}] = D_B^* [1 + \frac{d\ln\gamma_B}{d\ln x_B}]$ F

(2.70) (2.51) + (2.64) $D_A = F D_A^*$, $D_B = F D_B^*$

(2.71)

F = thermodymic factor.

(2.72) $\therefore \tilde{D} = F(X_B D_A^* + X_A D_B^*)$ when $F = \left\{1 + \frac{d\ln\gamma_A}{d\ln X_A}\right\}$

$= \left\{1 + \frac{d\ln\gamma_B}{d\ln X_B}\right\}$

$= \frac{X_A X_B}{RT} \frac{d^2G}{dX^2}$ (1.71)

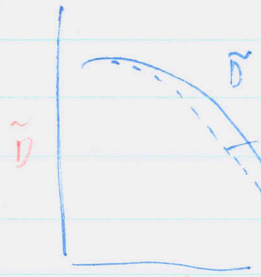


Fig 2.22d

$\tilde{D}$ (calculated assuming (2.69) holds using Fig 2.22b & c)

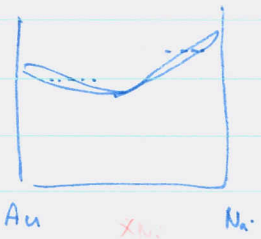
observed. (Diffusion couple)

!! in good agreement !!

Thus, we can say $\therefore$ $D_B^* = M_B^* RT = M_B RT$. in a "dilute" sol'n.

yes $M \neq f(\text{com})$!

- Relationship bet'n D & Conc. : 10^3 (3 orders of mag. over the range).
  (Au-rich & Ni-rich)



lower liquidus at Au-rich. $D = f(\text{conc})$

$\boxed{D(T_m) = \text{cmst.}}$ -- the same D_m.

$\therefore$ Au diffuses faster than Ni at all comp.
low T_m

2.6. Diffusion in Ternary Alloys

- 3rd component addition → cause math. complexity.

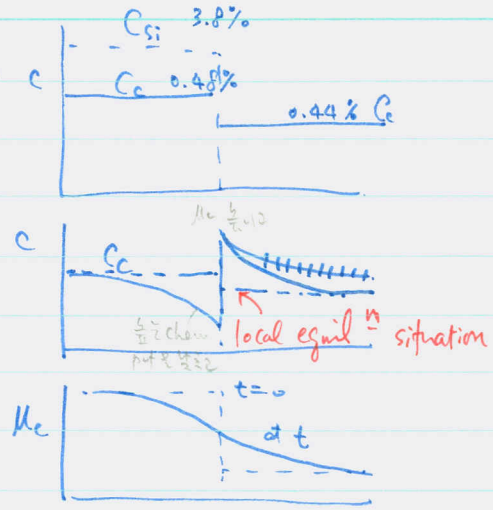
$\mu_C = \mu_C^\circ + RT\ln a_C$

a_C

* Consider !
- Fe-Si-C alloy instructive $\because$ ① Si raises the μ_C in sol'n. (C diffuses to lower μ_C)
  ② $M_{Si} \ll M_C$
  Sub. interstitial

No C diffusion from high C

Fig 2.23



two pieces welded, austenized at 1050°c.

(atoms on Si-side → to Si-free side until equal μ_c (= a_c) at the interface. the interfacial comp. remain const as long as the Si atoms do not migrate.

- If 3 component diffuse substitutionally (← D_0's not equal): less obvious effect.

2.7. High-Diffusivity Path.

- All defects are ass'ted w/ more open struc, resulting in a higher jump freq. - could be dominant diff. path.

1. Diffusion along G.B and FreeSurface (sheets)

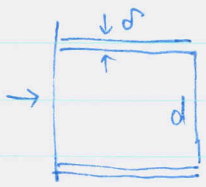
(2.73) diff. along g.b. $D_b = D_{b0} \exp\left(-\frac{Q_b}{RT}\right)$

(2.74) diff. along free surface $D_s = D_{s0} \exp\left(-\frac{Q_s}{RT}\right)$

(2.75) $D_s > D_b > D_{lattice}$. but areal fraction. (∵)
(lattice > g.b area > surface)
↑ became important

- when the diffusion along g.b. is important?

(2.76)



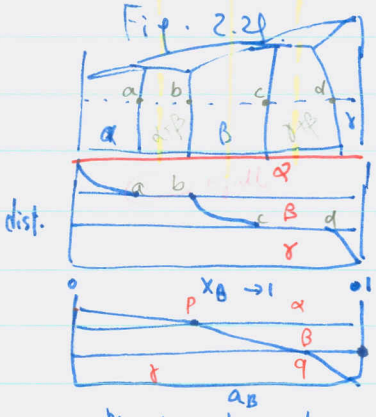
$J_l = -D_l \frac{dC}{dx}$ $J_b = -D_b \frac{dC}{dx}$

contribution ∝ cross-sectional area.

(2.77) $J = (J_b\delta + J_l d)/d = -\left(\frac{D_b\delta + D_l d}{d}\right)\frac{dC}{dx}$.

- welding two metals that are not completely miscible.

Fig. 2.29

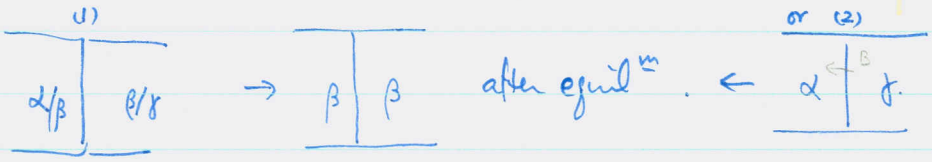


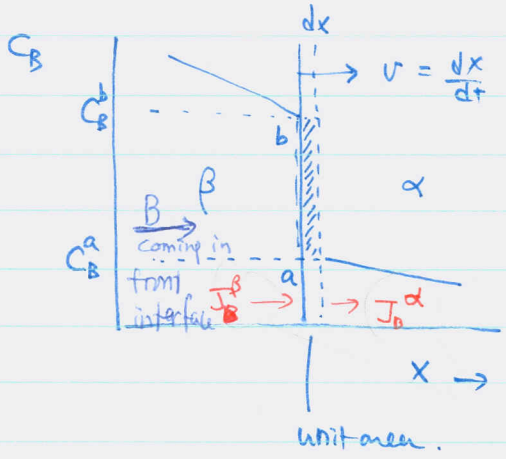
welding pure A & B. → gives a layered struc containing α, β, γ.

The comp. a, b are seen to be the equil^m comp. of α, β in the α+β field – α, β are in local equil^m across α/β interface. & β, γ " across β/γ interface

- Moving boundary.

· If the overall comp. of the diff. couple lies bet'n b & c, the final equil^n ⇒ β.





If $(dx \cdot 1)$ vol. converted from α (C_B^b) to β (C_B^b)

then $(C_B^b - C_B^a)(dx \cdot 1)$ must accumulate at the α/β interface. (unit area.)

$$\therefore J_B^\beta = -\tilde{D}(\beta)\frac{\partial C_B^b}{\partial x}$$ flux toward the interf

$$J_B^\alpha = -\tilde{D}(\alpha)\frac{\partial C_B^a}{\partial x}.$$ flux away from "

∴ for a given time dt.

the accumulation will be

$$(J_B^\beta - J_B^\alpha)\,dt = \int\left\{\left(-\tilde{D}(\beta)\frac{\partial C_B^b}{\partial x}\right) - \left(-\tilde{D}(\alpha)\frac{\partial C_B^a}{\partial x}\right)\right\}dt = (C_B^b - C_B^a)\,dx$$

Date . .

(2.18) $\therefore D_{apparent} = D_l + D_b \delta/d$ or $\frac{D_{app}}{D_l} = 1 + \frac{D_b}{D_l} \cdot \frac{\delta}{d}$

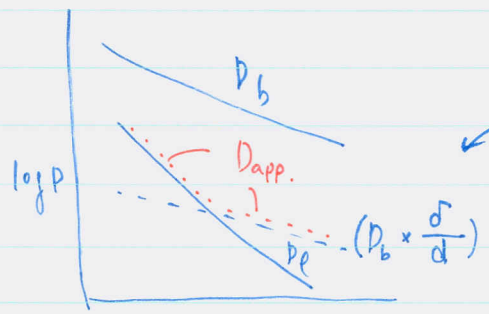
when $D_b \delta \gg D_l \cdot d$, lattice diffusion → small. & g.b diffusion → $\frac{2}{3}$ la.

$D_b > D_l$ always

width of a g.b ~0.5 nm.

g. sizes 1 ~ 1000 μm.

→ Also, sensitive to temp., ∵ $Q_b \sim 0.5 Q_l$ in fcc. metals.



D g.b. important below 0.75 ~ 0.8 Tm.

2. Diffusion along ⊥s (pipe).

Cross-sectional area of ⊥ / unit area (δ/d).

(2.21) $\frac{D_{app}}{D_l} = 1 + g \cdot \frac{D_p}{D_l}$

Ex. Roughly 10^5 dislocations / mm^2 (annealed mat'l), a ⊥ accomodates 10 atms in the cross-section

matrix contains 10^{15} atoms / mm^2. ∴ $g \simeq 10^{-7}$

at low temp. $g \cdot \frac{D_p}{D_l}$ become important.

at h. temp negligible.

2.8. Diffusion in Multiphase Binary Systems.

practical case.

diff.'l transformation

Date . .

vel. of moving boundary

$$(2.82) \quad \therefore \frac{dx}{dt} = \underline{\underline{v}} = \frac{1}{C_B^b - C_B^a} \left\{ \left(-\tilde{D}(\beta)\frac{\partial C_B^b}{\partial x}\right) - \left(-\tilde{D}(\alpha)\frac{\partial C_B^a}{\partial x}\right) \right\}$$

Δ in the diffusive flux → determines the vel.

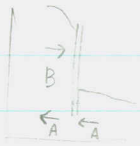
~~fit~~ → true when two phases are in local equil^{m}. → $\mu_B(\sim a_B)$ continuous across the boundary

usually true ~~when~~ in diffusion-couple experiment.

not true for all moving interfaces.

this is not true. not continuous.

If the interface has a low mobility, $(C_B^b - C_B^a)$ will increase → creating a discontinuity of μ_B across the boundary. → evaluate vel. Complex issue.



~~※~~ the flux of atoms to the interface balance ① the rate of accumulating & ② the rate of diffusion away into the other phase.

② balance w/ the rate of transfer across the interface.

① due to boundary moving

Interface controlled - if rxn so slow, no conc. grad (!?) why?

∵ diffusion limited by conc. grad

→ diffusion is not its limit → no conc. grad.

reaction takes place so slow.