HOMEWORK
From Dieter
4-2, 4-7

Module #9

Slip by Dislocation Motion and Dislocation Theory

READING LIST

DIETER: Ch. 4, p. 114-132; Ch. 5, p. 145-183; Ch. 8, p. 310-314

Ch. 2 in Hertzberg

Ch. 3, Pages 85-139 in Courtney

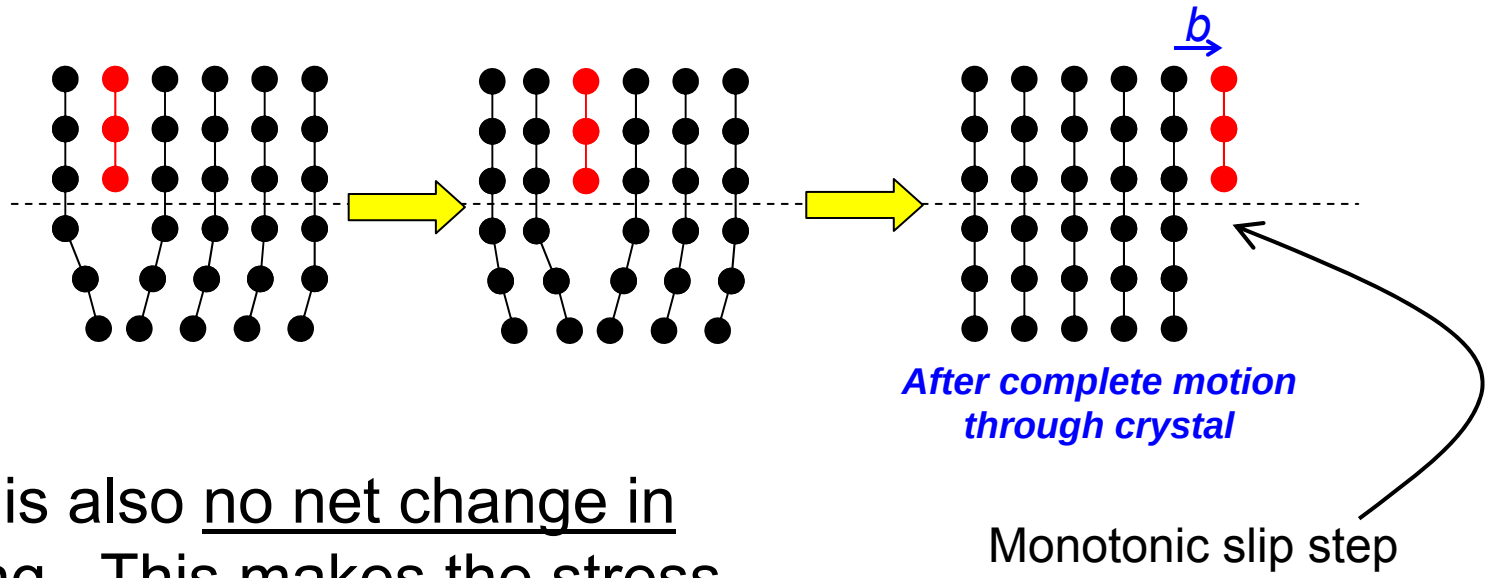
Ch. 3, Pages 42-61 in Hull & Bacon

Types of Dislocation Motion

- Glide (conservative motion):
  - ⊥ moves on a plane that contains both its line and Burgers vector.
  - A ⊥ that moves glides is called glissile.
  - A ⊥ that can't move is called sessile.
  - ⊥ glide plane and direction depend upon crystal structure.
- Climb (non-conservative motion)
  - ⊥ moves out of the glide plane perpendicular to the Burgers vector.
- Glide of many dislocations leads to slip which is the most common manifestation of plastic deformation.

Slip by Dislocation Glide

Glide requires relatively little atomic motion compared with the process for slip that we outlined for perfect (i.e., defect free) crystals.



There is also no net change in bonding. This makes the stress to move a dislocation smaller than the theoretical stress to shear a perfect dislocation free crystal.

Force is required to move individual dislocations

- We call it the Peierls-Nabarro force.
- The concept was originally developed in 1940 by Peierls[1], but later refined by others[2-4].
- It is the frictional force that must be overcome to move an individual dislocation.
- It is a consequence of the distortion caused by the presence of a dislocation in a crystal lattice.

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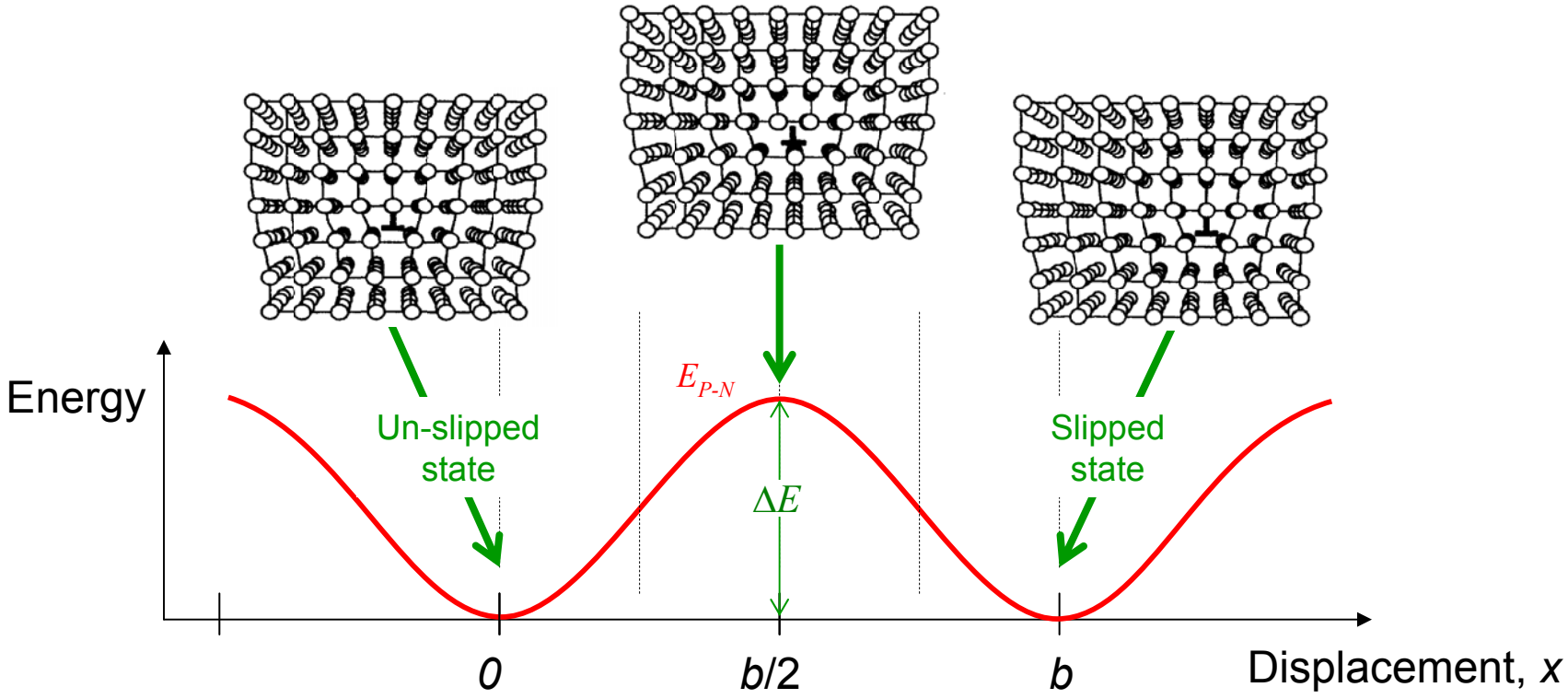
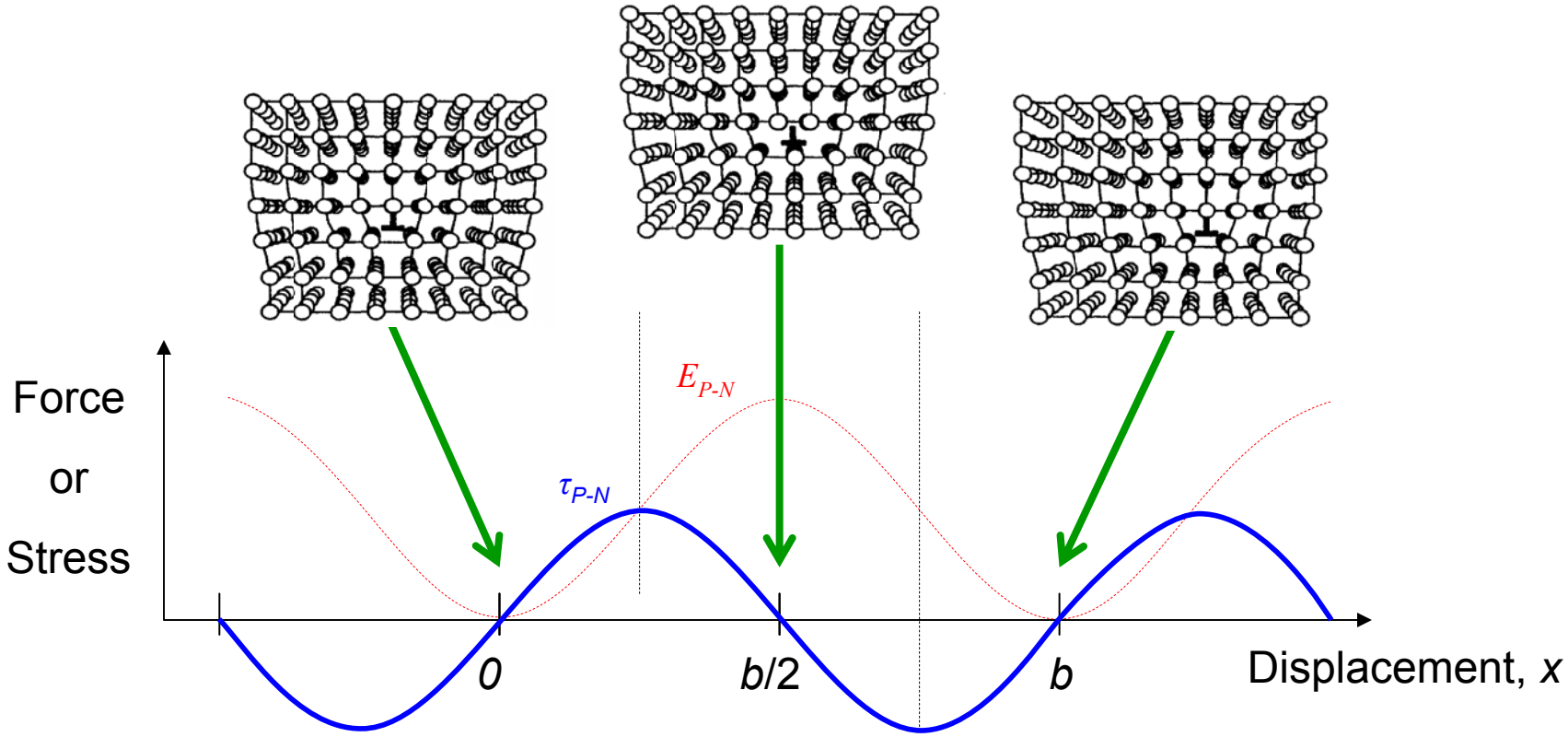
[1]R. Peierls, *Proc. Phys. Soc.*, v. 52, p. 34 (1940).
[2]F.R.N. Nabarro, *Proc. Phys. Soc.*, v. 59, p. 256 (1947).
[3]G. Leibfried and K. Lücke, *Z. Phys.*, v. 126, p. 450 (1949).
[4]A.J. Foreman, M.A. Jawson, and J.K. Wood, *Proc. Phys. Soc.*, v. 64, p. 156 (1951).

- A dislocation must pass through a higher energy configuration to move.

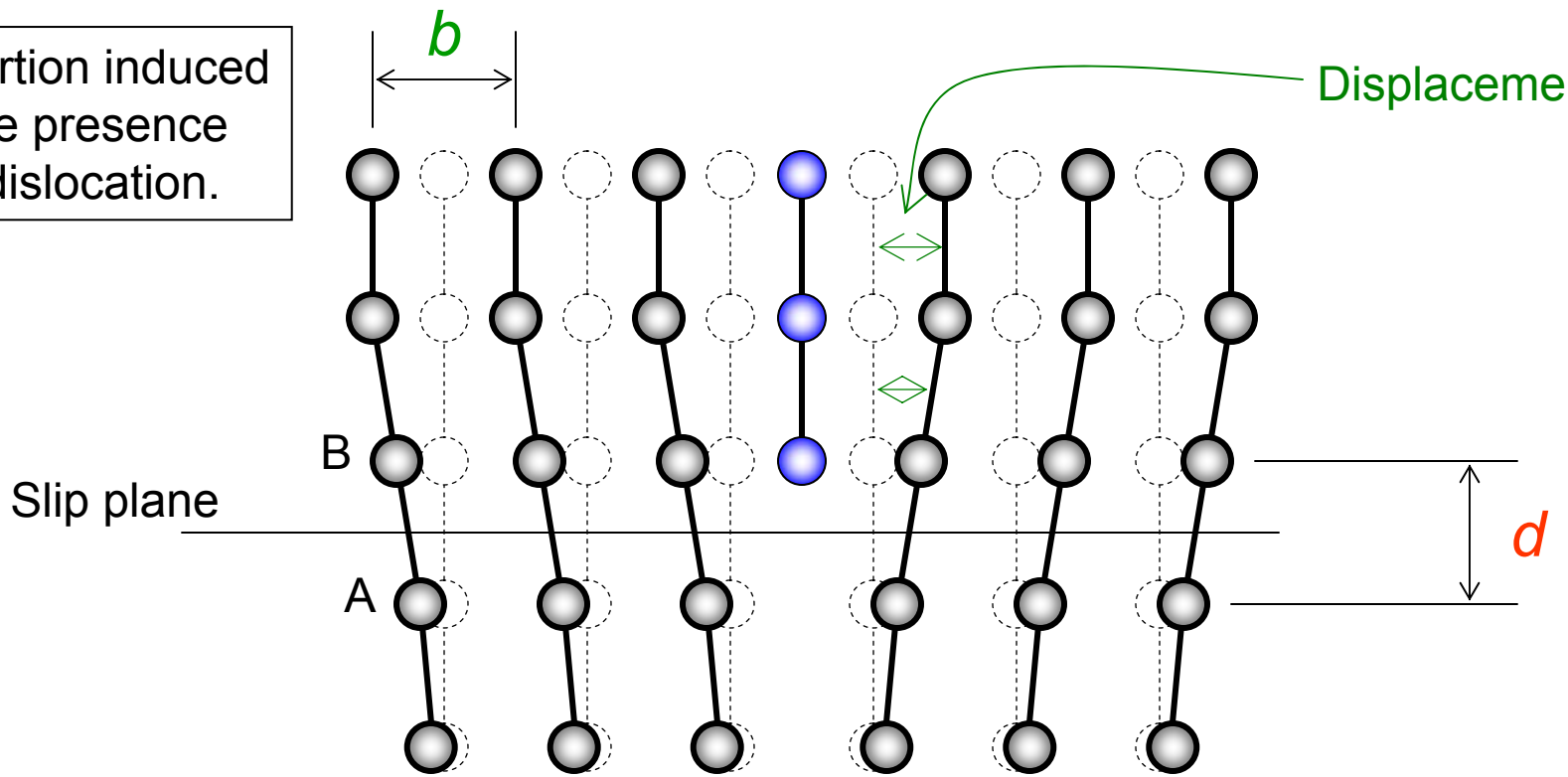


- The force required to move the dislocation (i.e., to overcome stress field caused by lattice distortion) $\propto$ shear stress on a slip plane.
- The Peierls-Nabarro force depends on the form of the force distance relation between atoms.



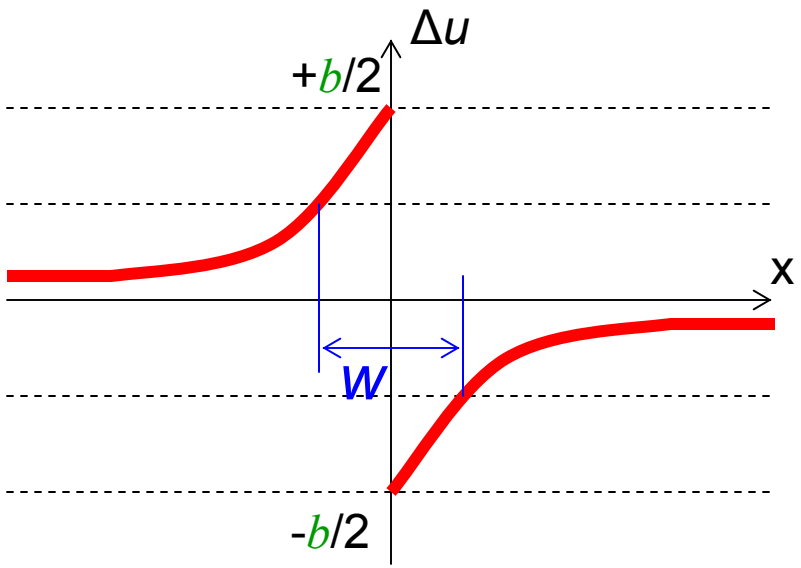
- The Peierls-Nabarro stress is the shear stress required to move an individual dislocation on its slip plane.
- Its value depends upon the amount the lattice is distorted by the dislocation.
- Amount of distortion described by dislocation width (*w*).

Distortion induced by the presence of a dislocation.



- Above and below the slip plane, atoms are displaced from their equilibrium positions (u); this represents the distortion in the lattice caused by the dislocation.
- To accommodate the dislocation, there is differential displacement across the slip plane ($\Delta u = u_B - u_A$); this produces shear.
- The maximum value of Δu is $\pm b/4$.

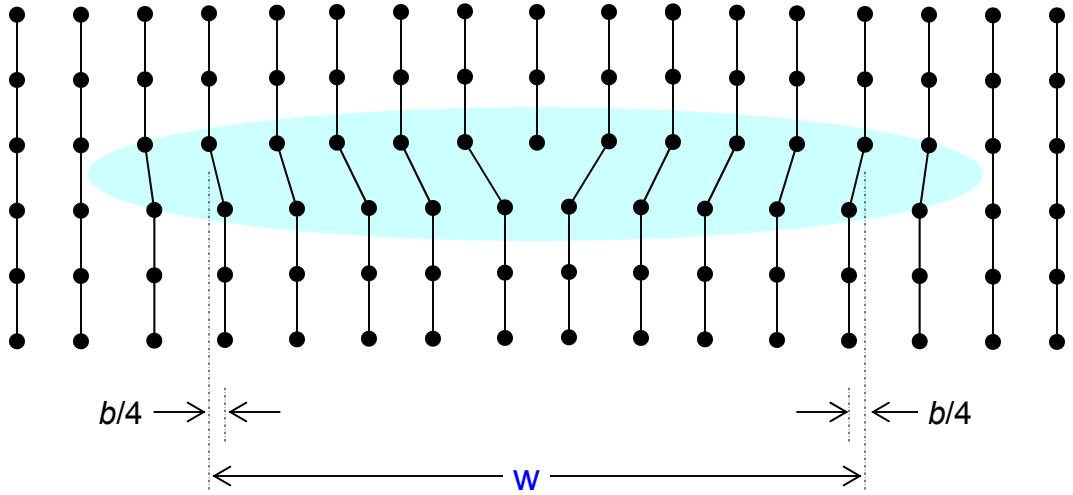
Distortion described in terms of dislocation width, *w*



- The width, *w*, is the distance over which the magnitude of Δu > ½ of its maximum value (i.e., $\Delta u/b$ > ½ or $-b/4 \le \Delta u \le b/4$).
- The width also provides a measure of the size of the core.
- Core widths vary between b and $5b$ and depend upon:
  - Interatomic potential,
  - Crystal structure.

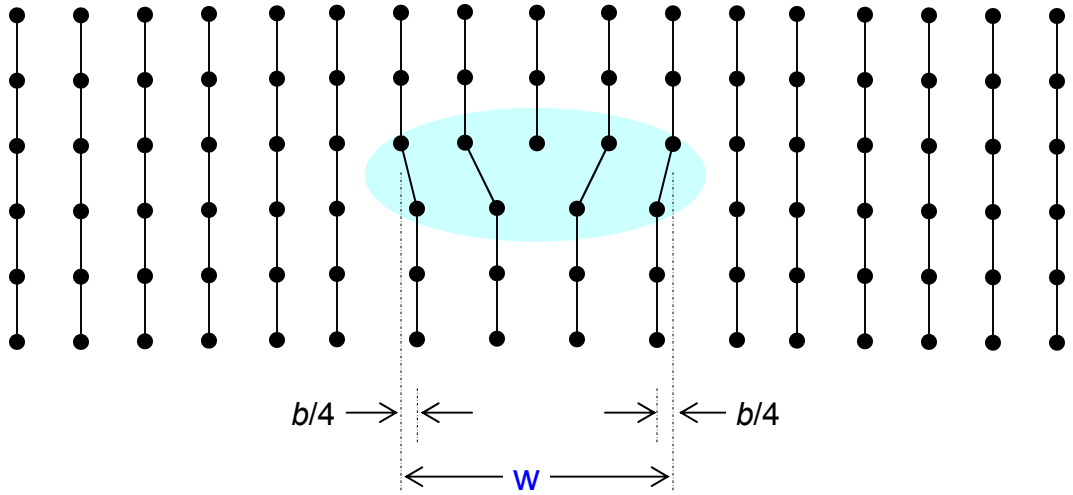
The width, *w*, is directly related to crystal structure

Fig. A.
Schematic illustration of a wide dislocation; more typical of close-packed metals.



Close packed structures: *w* is larger. Distortion distributed through crystal lattice.

Fig. B.
Schematic illustration of a narrow dislocation; more typical of ceramics, intermetallics, and non-close-packed metals.



Non-close packed structures: *w* is smaller. Distortion concentrated into a smaller area.

Peierls and Nabarro estimated the energy of the dislocation per unit length as a function of dislocation position as:

$$E_{P-N} = \frac{Gb^2}{\pi(1-\nu)} \exp\left(\frac{-2\pi w}{b}\right)$$

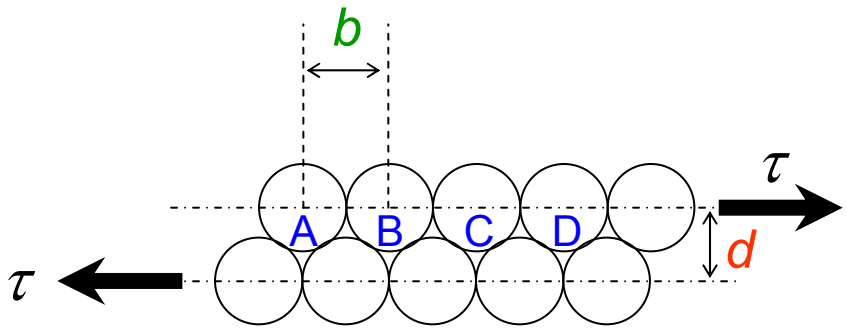
from which the shear stress required to move a dislocation (i.e., Peierls stress) can be determined as:

$$\tau_{P-N} = \frac{2\pi}{b^2} E_{P-N} = \frac{2G}{(1-\nu)} \exp\left(\frac{-2\pi w}{b}\right) \approx \frac{2G}{(1-\nu)} \exp\left(\frac{-2\pi d}{(1-\nu)b}\right)$$

Values of $\tau_{P\text{-}N}$ vary with crystal structure. In general,

$$\tau_{P\text{-}N} << \tau_{theo}.$$

Structure	$\tau_{P\text{-}N}$
FCC & HCP	$\leq 10^{-5}$ to 10^{-6} G
Covalent crystals	~10^{-2} G

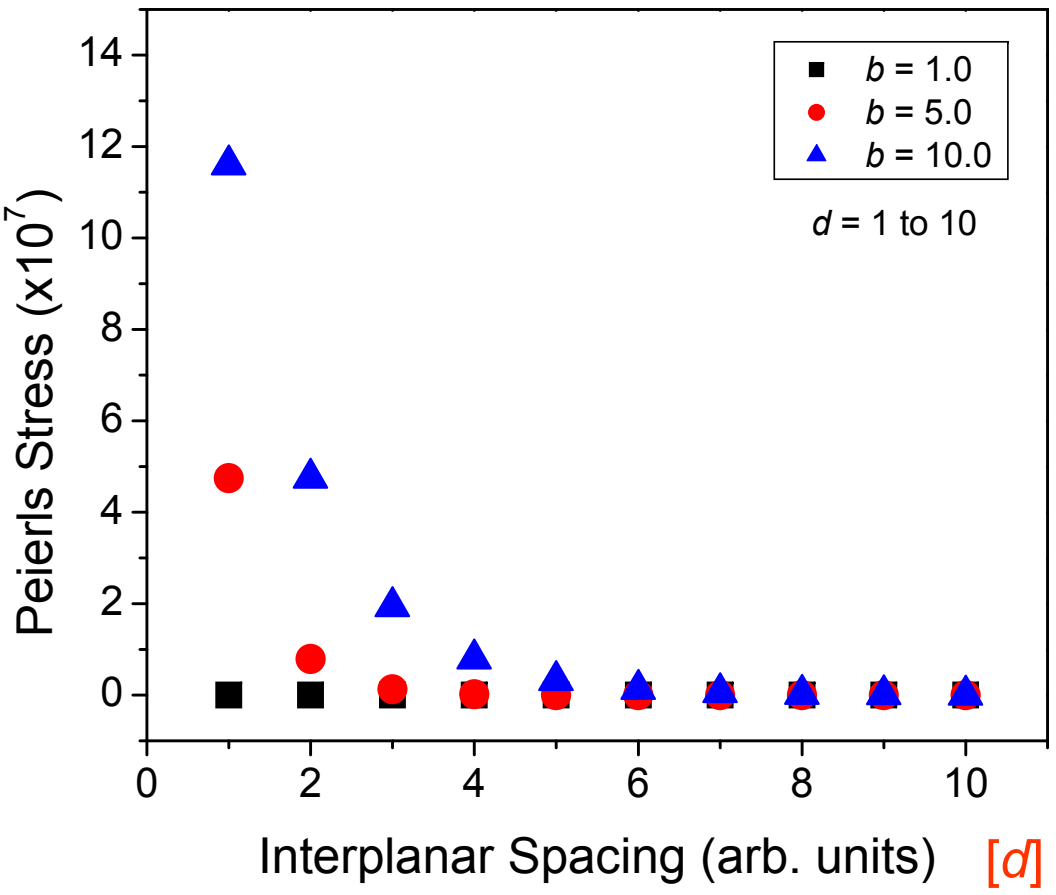


$$\tau_{P-N} = \frac{2\pi}{b^2} E_{P-N} = \frac{2G}{(1-\nu)} \exp\left(\frac{-2\pi w}{b}\right) \approx \frac{2G}{(1-\nu)} \exp\left(\frac{-2\pi d}{(1-\nu)b}\right)$$

Let G = 100 GPa and ν = 0.3.

What is the impact of crystal structure on $\tau_{P\text{-}N}$?

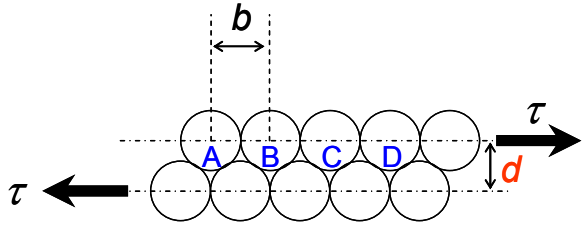
Variation of Peierls Stress w/ *d* and *b*



$$\tau_{P-N} \approx \frac{2G}{(1-\nu)} \exp\left(\frac{-2\pi d}{(1-\nu)b}\right)$$

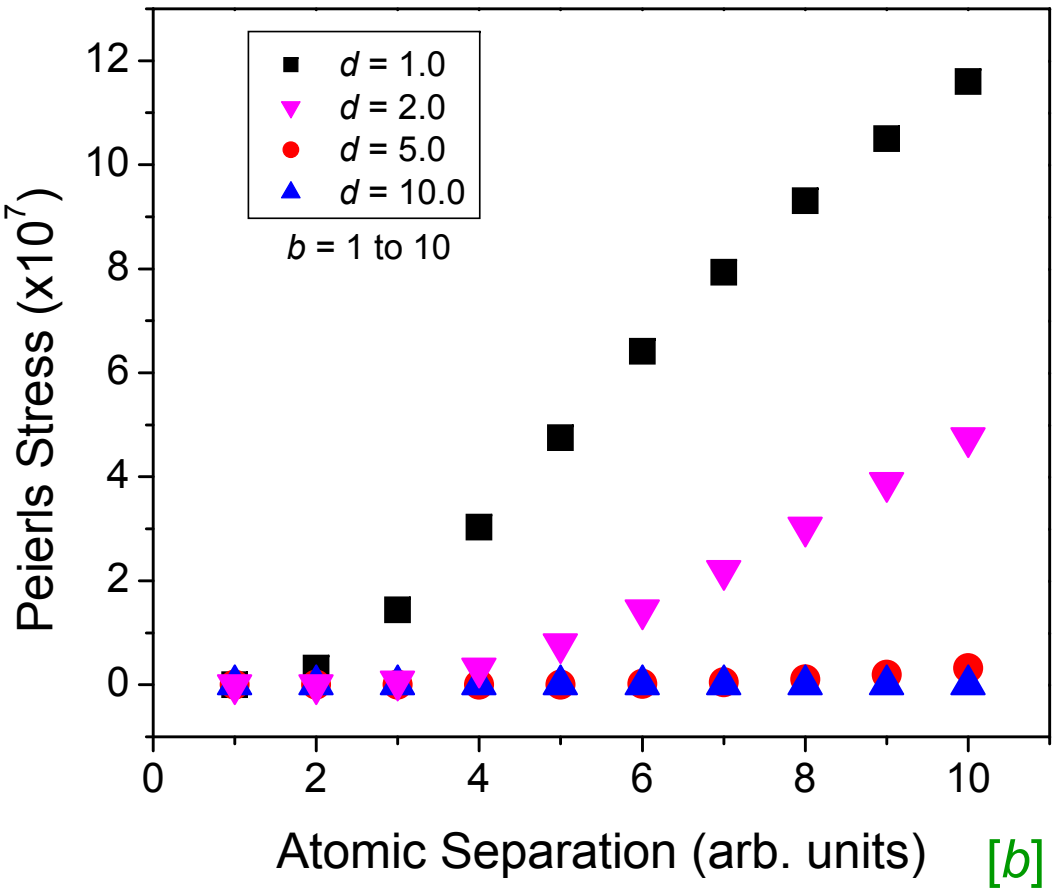
- For fixed *b*, $\tau_{P\text{-}N}$ decreases as *d* increases.
- An increase in *b* results in a larger $\tau_{P\text{-}N}$.
- "Slip via dislocation motion occurs more readily in close packed directions (lowest *b*) and on widely spaced planes (highest *d*)." This is because $\tau_{P\text{-}N}$ values are lowest on these planes.

B*



The width, *w*, is directly related to crystal structure

Variation of Peierls Stress w/ *d* and *b*

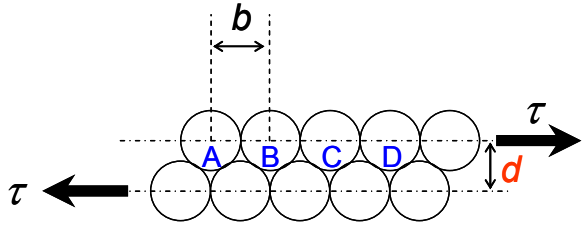


[*b*]

$$\tau_{P-N} \approx \frac{2G}{(1-\nu)} \exp\left(\frac{-2\pi d}{(1-\nu)b}\right)$$

- For fixed *d*, $\tau_{P\text{-}N}$ increases as *b* increases.
- An increase in *d* results in a reduced $\tau_{P\text{-}N}$.
- "Slip via dislocation is more likely to occur more readily in close packed directions (lowest *b*) and on widely spaced planes (highest *d*)." This is because $\tau_{P\text{-}N}$ values are lowest on these planes.

B*



The width, *w*, is directly related to crystal structure

Close Packed Planes

- Recall from your introductory materials courses:

- Close packed planes (i.e., those with the smallest interatomic separation, *d*) are the ones that are spaced farthest apart (i.e., those with the largest *b*).

- We can relate properties to atomic/ionic packing factors (APF/IPF) or planar density.

				Ionic				Covalent
	FCC/HCP	BCC	SC	KCl	NaCl	CsCl	MgO	Diamond cubic (Si)
APF IPF	0.74	0.68	0.52	0.725	0.67	0.68	0.627	0.34
☆ τ_{P-N}								

☆ Rank 1 – 8 where 1 is lowest and 8 is highest.

A*

Close Packed Planes

- Recall from your introductory materials courses:

- Close packed planes (i.e., those with the smallest interatomic separation, *a*) are the ones that are spaced farthest apart (i.e., those with the largest *b*).

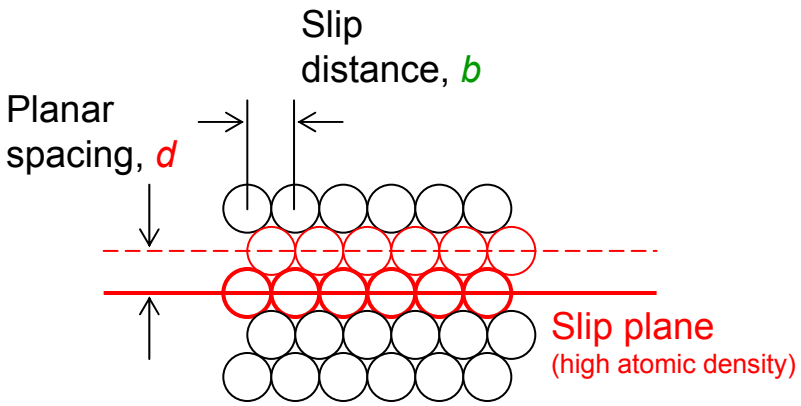
- We can relate properties to atomic/ionic packing factors (APF/IPF) or planar density.

				Ionic				Covalent
	FCC/HCP	BCC	SC	KCl	NaCl	CsCl	MgO	Diamond cubic (Si)
APF IPF	0.74	0.68	0.52	0.725	0.67	0.68	0.627	0.34
☆ τ_{P-N}	1	3	7	2	5	3	6	8

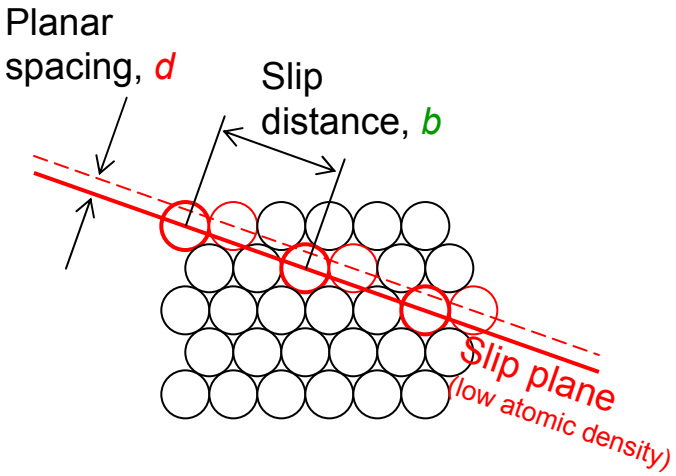
☆ Rank 1 – 8 where 1 is lowest and 8 is highest.

B*

Slip vs. atomic density



- Close-packed planes/structures
  - Smaller *b*
  - Larger *d*
  - *Smaller* τ_{P-N}



- Non close-packed planes/structures
  - Larger *b*
  - Smaller *d*
  - *Larger* τ_{P-N}

The Peierls-Nabarro stress is smaller than the critical resolved shear stress or the yield stress.

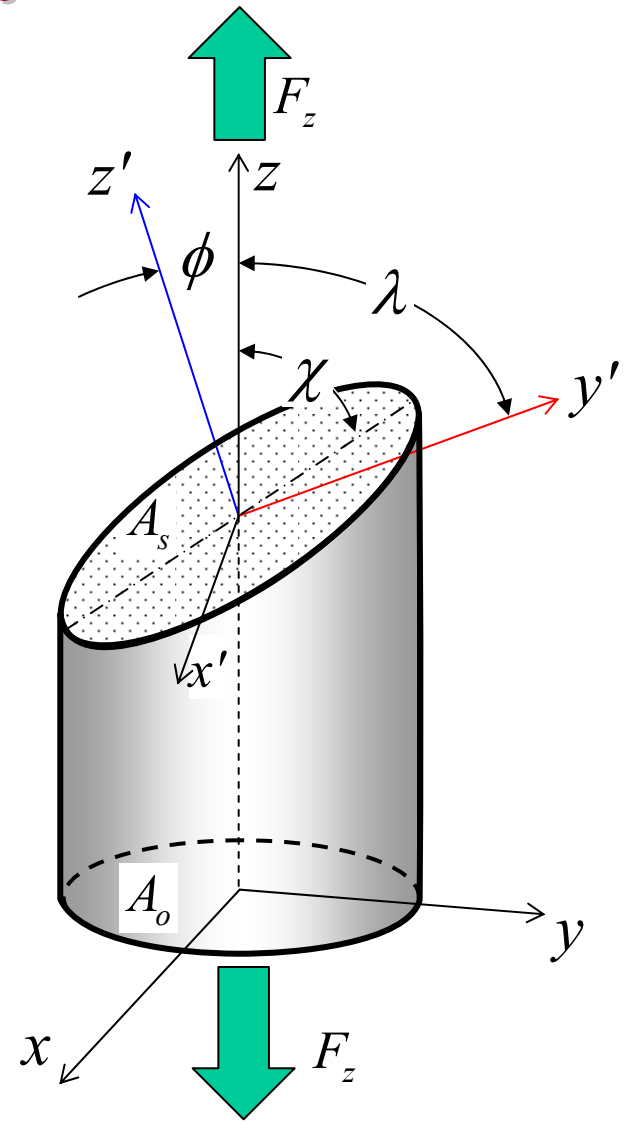
The CRSS and YS represent the conditions to move lots of dislocations

Slip Systems in Crystals

- A specific shear stress is required to induce dislocations to move.
- Dislocations slip on specific slip systems (i.e., specific crystal plane + specific crystal direction on that plane).
- The resolved shear stress on plane A_s in direction y' is:

$$\tau_{z'y'} = \frac{F_{y'}}{A_{z'}} = \frac{F_z \cos\lambda}{A_o / \cos\phi} = \sigma_{zz} \cos\lambda \cos\phi$$

- Applies for single crystals and individual grains in polycrystals.



$\phi + \chi = 90°$

$\phi + \lambda$ is not necessarily 90°

Resolved Shear Stress

- Consider an arbitrary plane oriented at angle ϕ with respect to the applied load *F*. Let y' = slip direction and z' = slip plane normal.
- Consider SLIP on the ϕ-plane. → **Refer to illustration on previous viewgraph**
- **NORMAL Force**: $F_N = F_{z'} = F_z \cos\phi$
- **SHEAR Force** in the slip direction (y'): $F_s = F_z \cos\lambda$.
- **Area of slip-plane:** $A_s = A/\cos\phi$ (check: A_s must have larger area than A_o.)

Resolved NORMAL Stress on the slip plane:

$$\sigma_N = F_N / A_s = \left(F\cos\phi\right)/\left(A_o/\cos\phi\right) = \sigma\cos^2\phi$$

Resolved SHEAR Stress on the slip plane in the slip direction:

$$\tau_s = \tau_{RSS} = F_s / A_s = \left(F\cos\lambda\right)/\left(A_o/\cos\phi\right) = \sigma\cos\phi\cos\lambda$$

- **The slip direction is not necessarily in same direction as tilt of the slip plane!**

Critical Resolved Shear Stress

$$\tau_{RSS} = \frac{F}{A_o}\underbrace{\cos\phi\cos\lambda}_{\text{Schmid Factor}} = \sigma_{\text{flow}}\cos\phi\cos\lambda = \sigma_{\text{flow}} m$$

- The active slip system will have the largest Schmid factor.

- If we relate the resolved shear stress to the macroscopic tensile yield stress as opposed to the flow stress, we get:

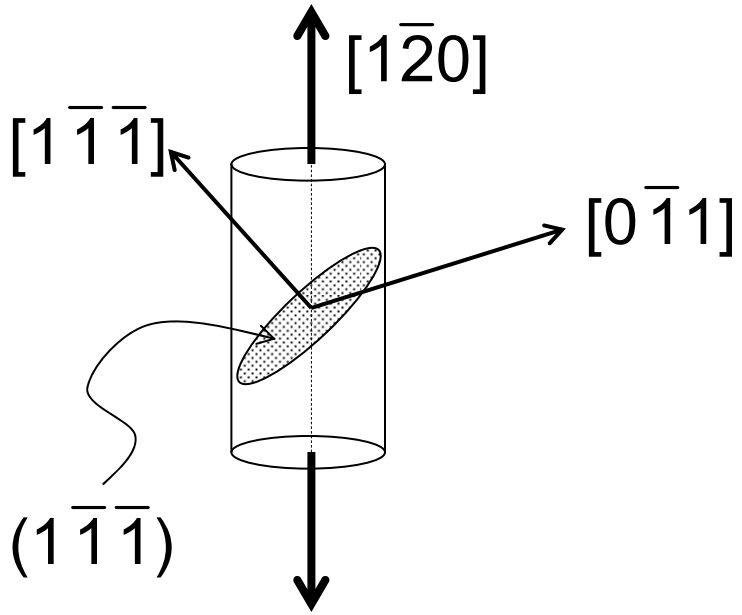
$$\sigma_y = \frac{\tau_{CRSS}}{\cos\phi\cos\lambda} \text{ or } \tau_{CRSS} = \sigma_y \cos\phi\cos\lambda = \sigma_y m$$

- τ_{CRSS} is the *resolved shear stress* required *to cause plastic deformation via slip*.

Skip example problems

Example Problem 1

Calculate the tensile stress that is applied along the $[1\bar{2}0]$ axis of a gold crystal to cause slip on the $(1\bar{1}\bar{1})[0\bar{1}1]$ slip system. The critical resolved shear stress is 10 MPa.



Solution to Example Problem 1

The angle between the tensile axis $[1\bar{2}0]$ and the normal to the slip plane $(1\bar{1}\bar{1})$ is:

Angle btw. Tensile axis & slip plane normal

$$\cos\phi =$$

The angle between the tensile axis $[1\bar{2}0]$ and the slip direction $[0\bar{1}1]$ is:

Angle btw. Tensile axis & slip direction

$$\cos\lambda =$$

Since CRSS = 10 MPa,

$$\sigma = \frac{P}{A} = \frac{\tau_{CRSS}}{\cos\phi\cos\lambda} =$$

Solution to Example Problem 1

The angle between the tensile axis $[1\bar{2}0]$ and the normal to the slip plane $(1\bar{1}\bar{1})$ is:

$$\cos\phi = \frac{(1)(1)+(-2)(-1)+(0)(-1)}{\sqrt{(1)^2+(-2)^2+(0)^2}\sqrt{(1)^2+(-1)^2+(-1)^2}} = \frac{3}{\sqrt{5}\sqrt{3}} = \frac{3}{\sqrt{15}}$$

The angle between the tensile axis $[1\bar{2}0]$ and the slip direction $[0\bar{1}1]$ is:

$$\cos\lambda = \frac{(1)(0)+(-2)(-1)+(0)(1)}{\sqrt{(1)^2+(-2)^2+(0)^2}\sqrt{(0)^2+(-1)^2+(-1)^2}} = \frac{2}{\sqrt{5}\sqrt{2}} = \frac{2}{\sqrt{10}}$$

Since CRSS = 10 MPa,

B*

$$\sigma = \frac{P}{A} = \frac{\tau_{CRSS}}{\cos\phi\cos\lambda} = \frac{10}{\left(3/\sqrt{15}\right)\left(2/\sqrt{10}\right)} = \boxed{20.41 \text{ MPa}}$$

Room temperature slip systems and critical resolved shear stress for metal single crystals (from Dieter, 3rd Edition, p. 126).

Metal	Crystal Structure	Purity	Slip Plane	Slip Direction	Critical Shear stress (MPa)	Reference
Zn	*HCP*	99.999	(0001)	$[11\bar{2}0]$	0.18	[1]
Mg	*HCP*	99.996	(0001)	$[11\bar{2}0]$	0.77	[2]
Cd	*HCP*	99.996	(0001)	$[11\bar{2}0]$	0.58	[3]
Ti	*HCP*	99.99 99.9	$(10\bar{1}0)$ (1010)	$[11\bar{2}0]$ $[11\bar{2}0]$	13.7 90.1	[4]
Ag	*FCC*	99.999 99.97 99.93	(111) (111) (111) (111)	[110] [110] [110] [110]	0.48 0.73 1.3	[5]
Cu	*FCC*	99.999 99.98	(111) (111)	$[10\bar{1}]$ $[10\bar{1}]$	0.65 0.94	[5]
Ni	*FCC*	99.8	(111)	[110]	5.7	[5]
Fe	*BCC*	99.96	(110) (112) (123)	[111]	27.5	[6]
Mo	*BCC*	⋯	(110)	[111]	49.0	[7]

[1] D.C. Jillson, *Trans. AIME*, v. 188, p. 1129 (1950).
[2] E.C. Burke and W.R. Hibbard, Jr., *Trans. AIME*, v. 194, p. 295 (1952).
[3] E. Schmid, "International Conference on Physics," v. 2, Physical Society of London (1935)
[4] A.T. Churchman, *Proc. R. Soc. London Ser. A*, v. 226A, p. 216 (1954)
[5] F.D. Rosi, *Trans. AIME*, v. 200, p. 1009 (1954)
[6] J.J. Cox, R.F. Mehl, and G.T. Horne, *Trans. Am. Soc. Met.*, v. 49, p. 118 (1957)
[7] R. Maddin and N.K. Chen, *Trans. AIME*, v. 191, p. 937 (1951)

Note the differences in slip systems for different crystal structures. Slip occurs when m is maximum. This means that we must determine which particular slip system has the maximum m to obtain the CRSS.

Example Problem 2

Consider a cylindrical single crystal of silver of 5 mm diameter with its axis parallel to [321]. This crystal begins to deform plastically in compression at a load of 39 N. Determine the CRSS for this crystal.

Solution to Example Problem 2

Silver has an FCC crystal structure. Thus the slip system is $\{111\}\langle 110\rangle$. There are 12 distinct slip systems for FCC. For each we must compute and tabulate the corresponding angles ϕ and λ, as well as the Schmid factors. This is done via the cosine law.

$$\cos\angle(h_1k_1l_1)(h_2k_2l_2) = \frac{(h_1h_2)+(k_2k_1)+(l_1l_2)}{\sqrt{(h_1^2+k_1^2+l_1^2)}\sqrt{(h_2^2+k_2^2+l_2^2)}}$$

Slip System			λ	ϕ	Schmid factor cos□ cos□
(a/2)	[1 $\underline{1}$ 0]	(1 1 1)			
(a/2)	[1 0 $\underline{1}$]	(1 1 1)			
(a/2)	[0 1 $\underline{1}$]	(1 1 1)			
(a/2)	[0 1 1]	(1 1 $\underline{1}$)			
(a/2)	[1 0 1]	(1 1 $\underline{1}$)			
(a/2)	[1 $\underline{1}$ 0]	(1 1 $\underline{1}$)			
(a/2)	[1 1 0]	(1 $\underline{1}$ 1)			
(a/2)	[1 0 $\underline{1}$]	(1 $\underline{1}$ 1)			
(a/2)	[0 1 1]	(1 $\underline{1}$ 1)			
(a/2)	[0 1 $\underline{1}$]	(1 $\underline{1}$ $\underline{1}$)			
(a/2)	[1 0 1]	(1 $\underline{1}$ $\underline{1}$)			
(a/2)	[1 1 0]	(1 $\underline{1}$ $\underline{1}$)			

$$\tau_{CRSS} = \frac{P}{A}\cos\phi\cos\lambda = \frac{39\ \text{N}}{\pi(5\ \text{mm}^2)/4}\cos\phi\cos\lambda = 2\ \text{MPa}\times m_{\max}$$

$$\tau_{CRSS} = 2\ \text{MPa}\ \times\ (0.4665) = \boxed{0.93\ \text{MPa}}$$

Solution to Example Problem 2

Silver has an FCC crystal structure. Thus the slip system is $\{111\}\langle 110\rangle$. There are 12 distinct slip systems for FCC. For each we must compute and tabulate the corresponding angles ϕ and λ, as well as the Schmid factors. This is done via the cosine law.

$$\cos\angle(h_1k_1l_1)(h_2k_2l_2) = \frac{(h_1h_2)+(k_2k_1)+(l_1l_2)}{\sqrt{(h_1^2+k_1^2+l_1^2)}\sqrt{(h_2^2+k_2^2+l_2^2)}}$$

Slip System			λ	ϕ	Schmid factor cos□ cos□
(a/2)	[1 $\underline{1}$ 0]	(1 1 1)	79.11	22.21	0.1749
(a/2)	[1 0 $\underline{1}$]	(1 1 1)	67.79	22.21	0.3500
(a/2)	[0 1 $\underline{1}$]	(1 1 1)	79.11	22.21	0.1749
(a/2)	[0 1 1]	(1 1 $\underline{1}$)	55.46	51.89	0.3499
(a/2)	[1 0 1]	(1 1 $\underline{1}$)	40.89	51.89	0.4666 → m_{max}
(a/2)	[1 $\underline{1}$ 0]	(1 1 $\underline{1}$)	79.11	51.89	0.1166
(a/2)	[1 1 0]	(1 $\underline{1}$ 1)	19.11	72.02	0.2917
(a/2)	[1 0 $\underline{1}$]	(1 $\underline{1}$ 1)	67.79	72.02	0.1167
(a/2)	[0 1 1]	(1 $\underline{1}$ 1)	55.46	72.02	0.1750
(a/2)	[0 1 $\underline{1}$]	(1 $\underline{1}$ $\underline{1}$)	79.11	90.00	0.0000
(a/2)	[1 0 1]	(1 $\underline{1}$ $\underline{1}$)	40.89	90.00	0.0000
(a/2)	[1 1 0]	(1 $\underline{1}$ $\underline{1}$)	19.11	90.00	0.0000

$$\tau_{CRSS} = \frac{P}{A}\cos\phi\cos\lambda = \frac{39\ \text{N}}{\pi(5\ \text{mm}^2)/4}\cos\phi\cos\lambda = 2\ \text{MPa}\times m_{\max}$$

$$\tau_{CRSS} = 2\ \text{MPa}\ \times\ (0.4665) = \boxed{0.93\ \text{MPa}}$$

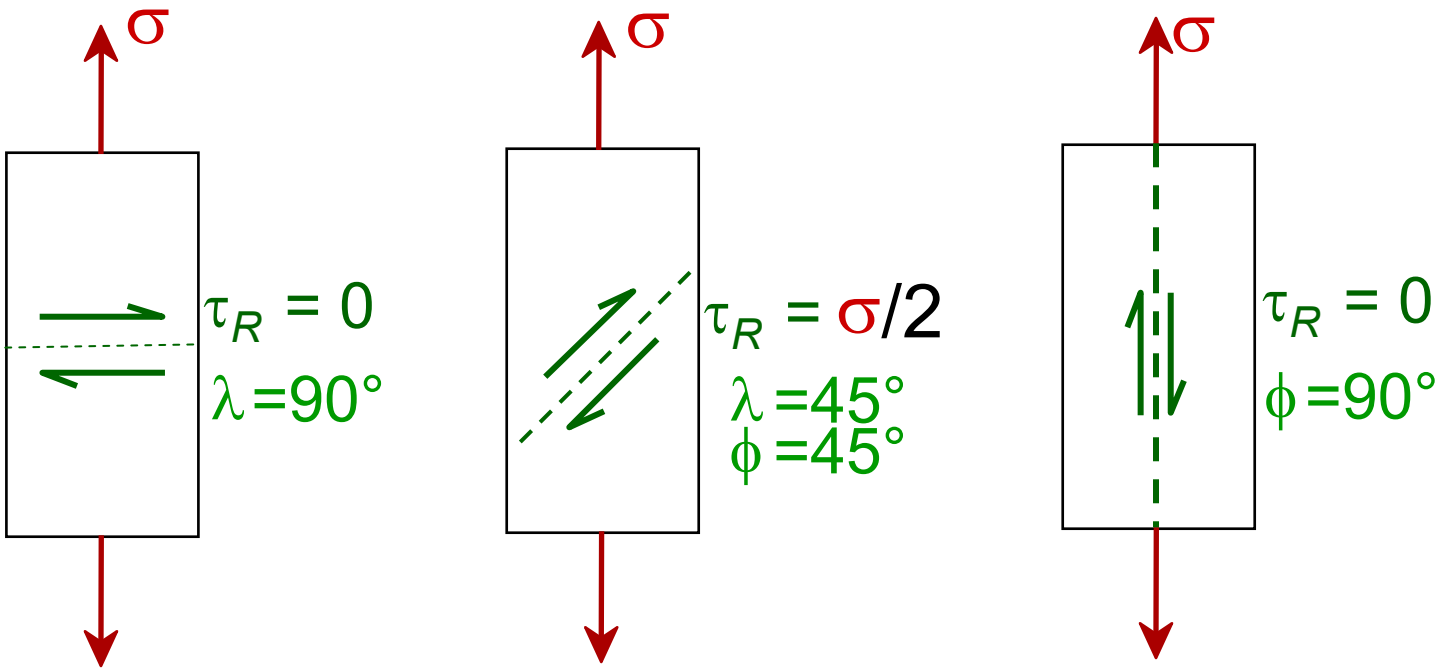
Critical Resolved Shear Stress

- Condition for dislocation motion: $\tau_R > \tau_{CRSS}$

- Crystal orientation can make it easy or hard to move dislocation

typically 10^{-4} GPa to 10^{-2} GPa

$$\tau_R = \sigma \cos\lambda \cos\phi$$



τ maximum at $\lambda = \phi = 45°$

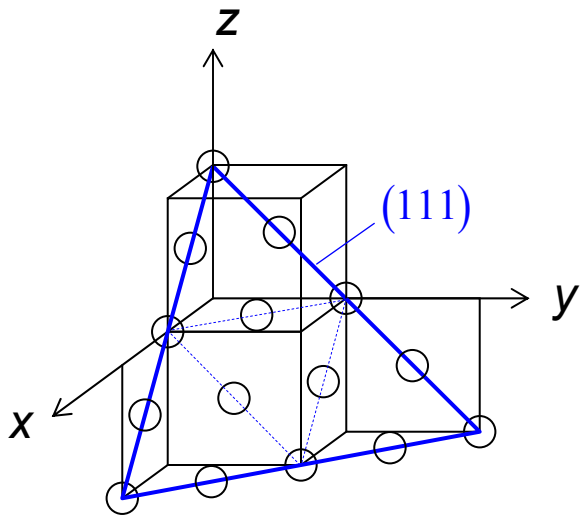
The next 14 viewgraphs provide illustrations of the operative slip systems in different crystals.

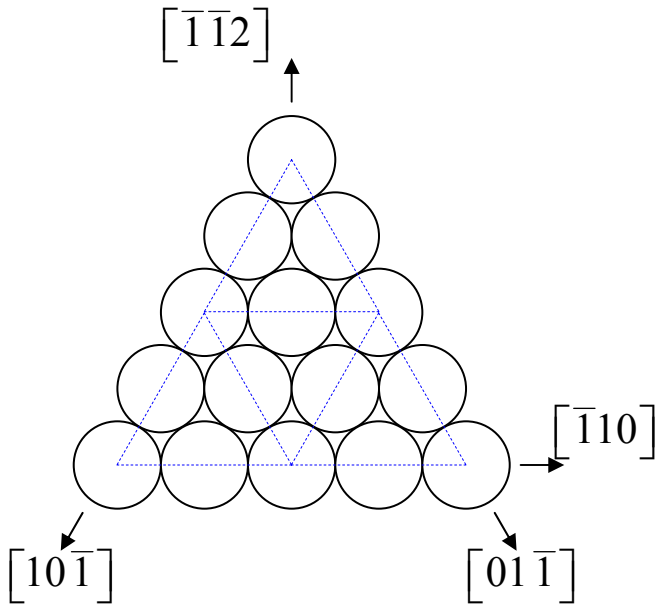
We will cover basic structures (i.e., fcc, hcp, bcc) in lecture.

You should review the material on ionic, covalent, and ordered (intermetallic) crystals on your own time.

Common Slip Planes in Metals

fcc

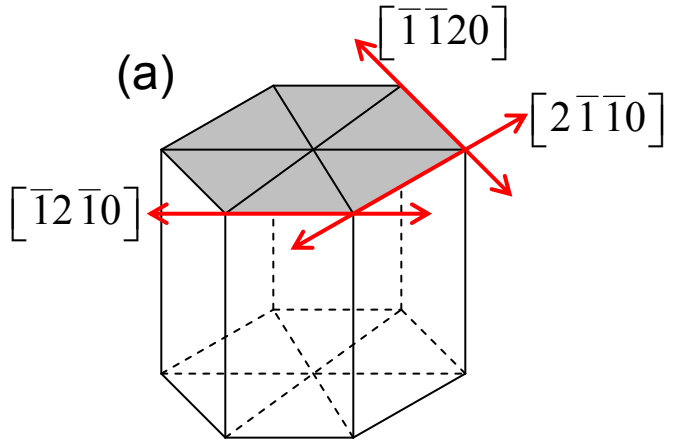


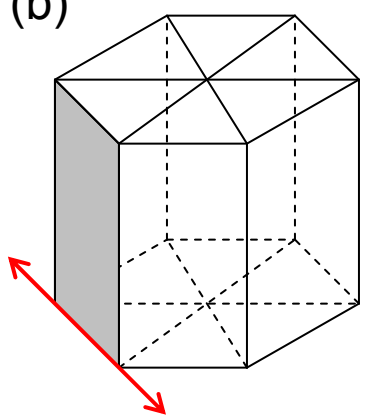


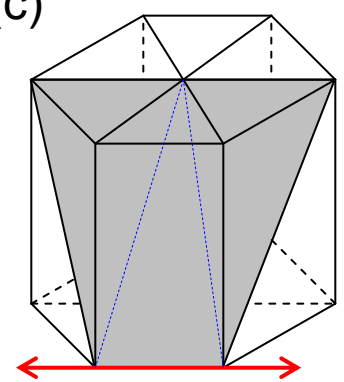
Four $\{111\}$ planes each with 3 $\langle 110 \rangle$ slip directions

12 slip systems

[Felbeck and Atkins, 2nd Ed., p. 117]

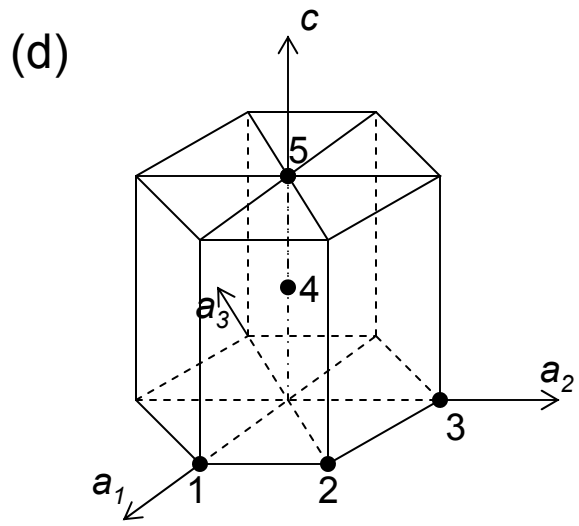






Basal Plane $\{0001\}$ Prismatic Plane $\{10\bar{1}0\}$ Pyramidal Plane $\{10\bar{1}1\}$

hcp



Pyramidal Plane 124 $\{10\bar{1}2\}$

Pyramidal Plane 135 $\{11\bar{2}1\}$

Pyramidal Plane 134 $\{11\bar{2}2\}$

Figure. Primary slip planes and directions for HCP crystals. (a) Basal slip plane; (b) prismatic slip plane; (c) pyramidal plane; and (d) other possible slip planes.

Common Slip Planes in Metals

hcp

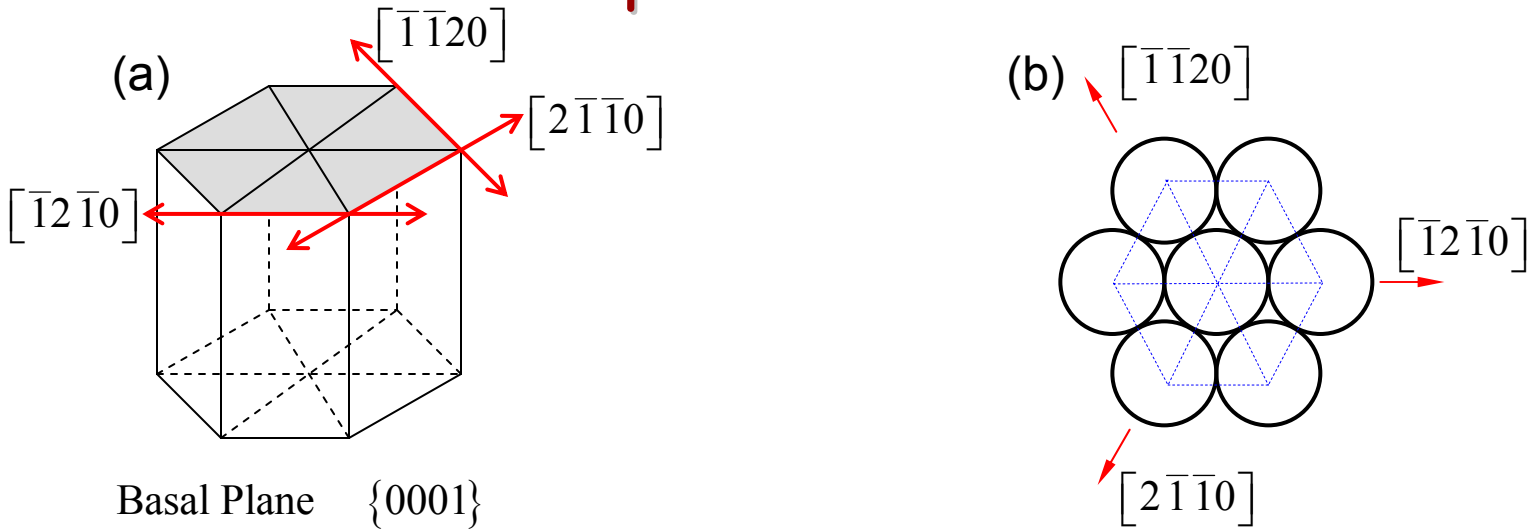


Figure. (a) Basal slip plane; (b) atomic arrangement on basal plane with possible slip directions indicated.

One $\{0001\}$ plane with three $\langle 11\bar{2}0\rangle$ slip directions

Three $\{10\bar{1}0\}$ planes with one $\langle 11\bar{2}0\rangle$ slip direction on each

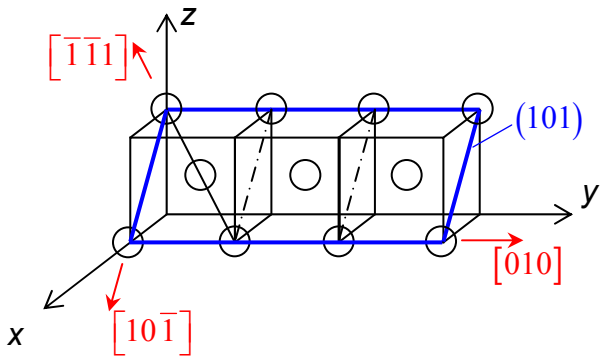
Six $\{10\bar{1}1\}$ planes with one $\langle 11\bar{2}0\rangle$ slip direction on each

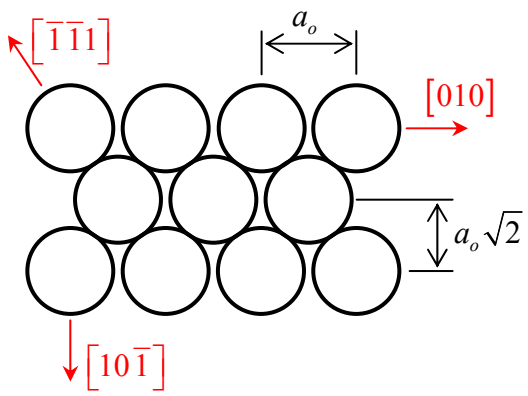
12 slip systems possible

active depends on *c/a* ratio

Common Slip Planes in Metals

bcc





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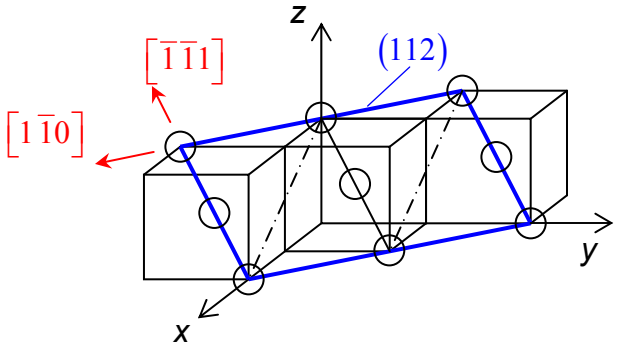

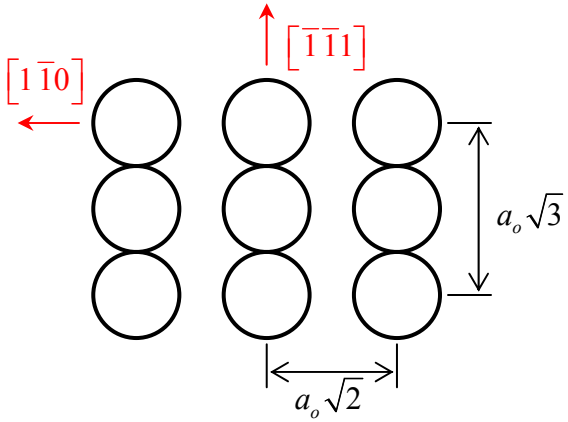

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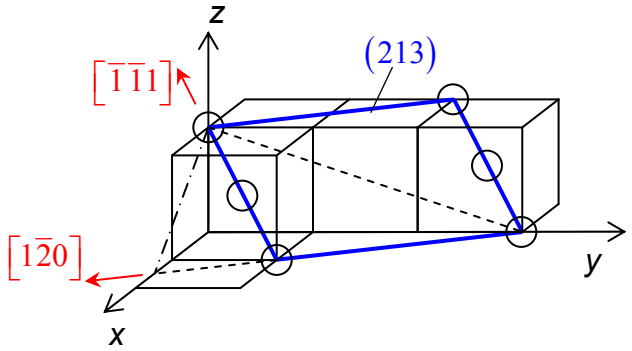
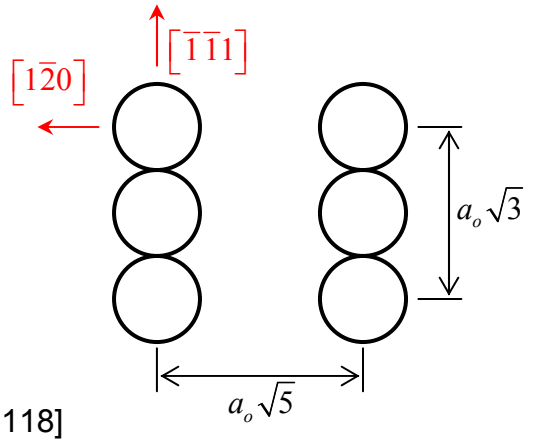




24

48 slip systems

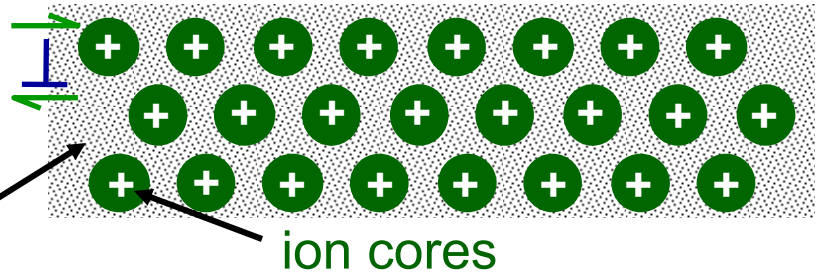
[After Felbeck and Atkins, 2nd Ed., p. 118]

Slip systems for the most common lattice types.

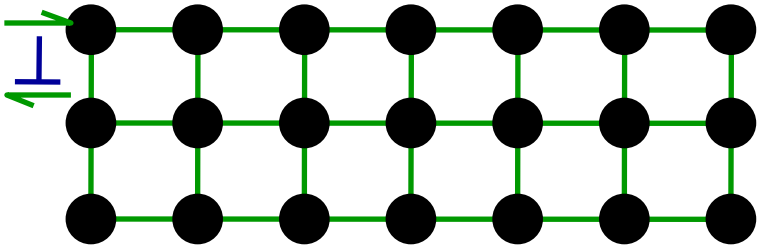
Crystal structure	Slip plane	Slip direction	Number of non-parallel planes	Slip directions per plane	Number of slip systems
fcc	$\{111\}$	$\langle 1\bar{1}0\rangle$	4	3	$(4\times 3)=12$
bcc	$\{110\}$	$\langle \bar{1}11\rangle$	6	2	$(6\times 2)=12$
	$\{112\}$	$\langle 11\bar{1}\rangle$	12	1	$(12\times 1)=12$
	$\{123\}$	$\langle 111\rangle$	24	1	$(24\times 1)=24$
hcp	$\{0001\}$	$\langle 11\bar{2}0\rangle$	1	3	$(1\times 3)=3$
	$\{10\bar{1}0\}$	$\langle 11\bar{2}0\rangle$	3	1	$(3\times 1)=3$
	$\{10\bar{1}1\}$	$\langle 11\bar{2}0\rangle$	6	1	$(6\times 1)=6$

Dislocations and Materials Classes

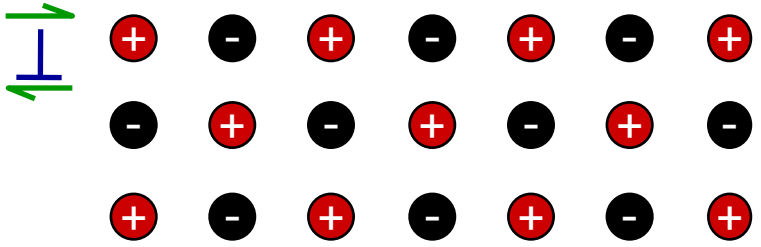
- Metals: Disl. motion easier.
  - -non-directional bonding
  - -close-packed directions for slip.



► Covalent Ceramics (Si, diamond): Motion hard.
- -directional (angular) bonding



► Ionic Ceramics (NaCl): Motion hard.
- -need to avoid ++ and - - neighbors.



Ionic Solids

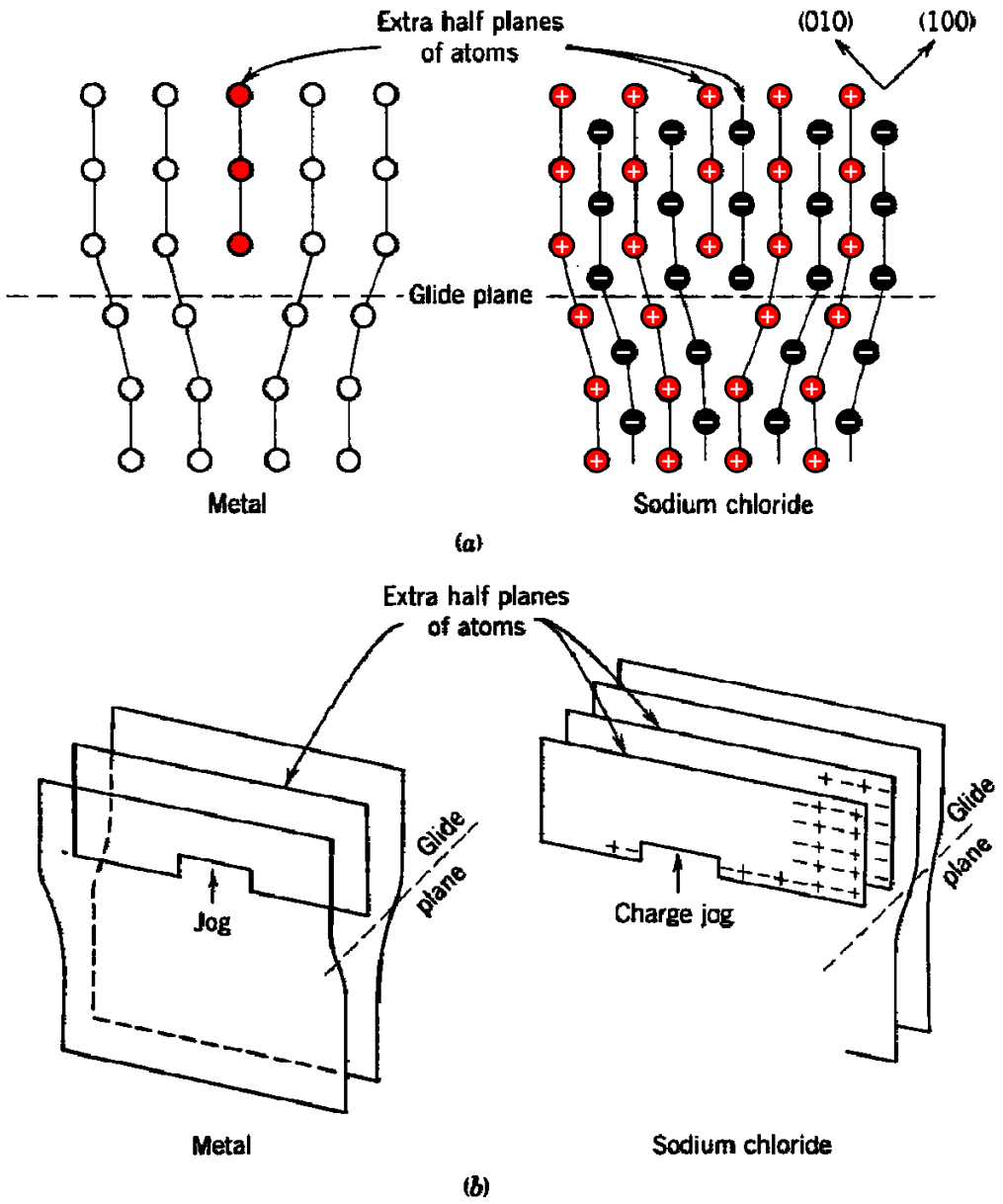
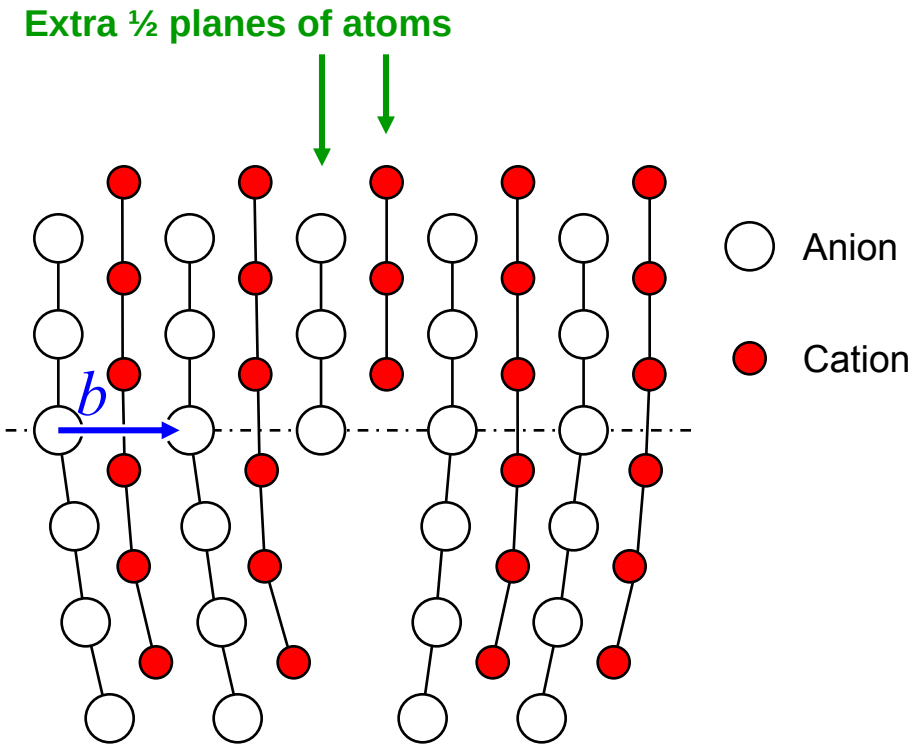


Fig. 4.21. (a) Schematic representation of an edge dislocation in NaCl; (b) demonstration of how dislocation jogs in ionic crystals can have effective charges. [Figure adapted from Kingery et al, p. 172].

The edge dislocation in the ionic crystal consists of two extra half planes of ions to maintain charge neutrality.

Ionic Solids – cont'd



Schematic representation of an edge dislocation in a solid with a NaCl structure. You are looking at the (100) plane. In this image the Burgers vector is [011].

- It's a bit easier to see on this diagram.
- To preserve electrical neutrality, an extra half plane of atoms must consist of:
  - a half plane of cations
  - a half plane of anions.

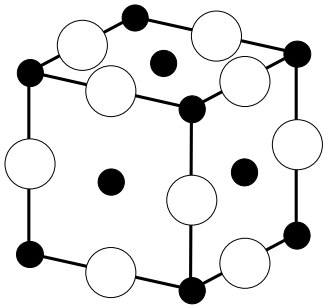
NOTE

- To maintain structural regularity (and charge neutrality), two extra half planes of atoms are required.

- This makes Burgers vectors more complicated in ionic crystals compared to metallic counterparts with the same crystal structures (see next slide).

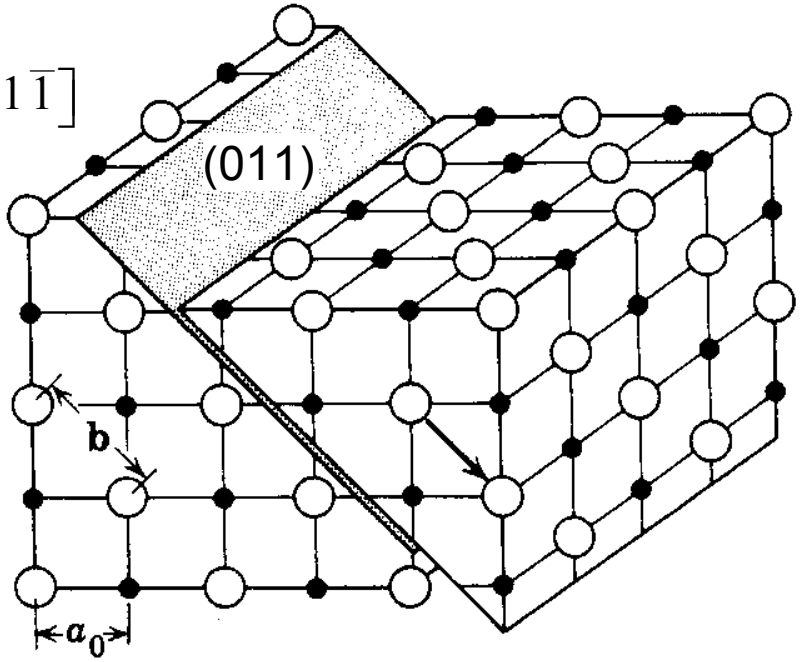
- Dislocations in ionic crystals can have an effective charge which can influence mobility.

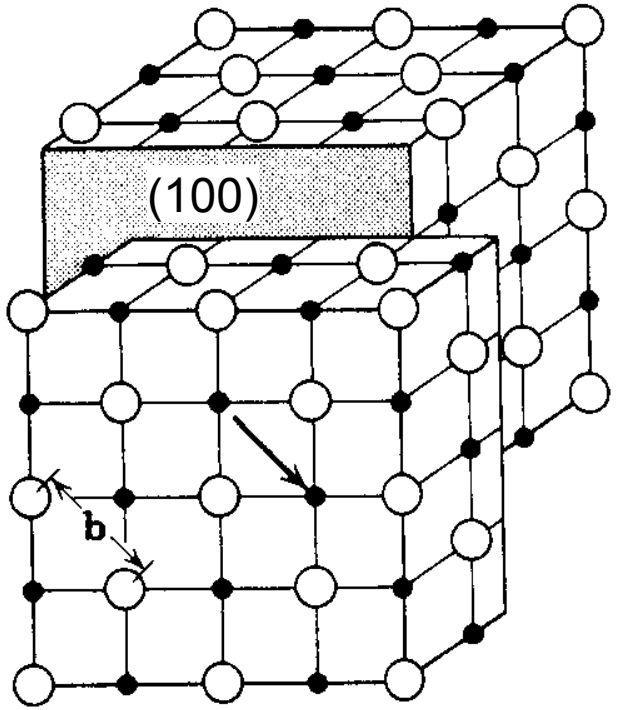
Ionic Solids



NaCl has an FCC Bravais lattice

$b = [01\bar{1}]$





Must maintain charge balance and same nearest neighbors

Fig. 14.10. Translation gliding in the <110> direction on (a) the {110} plane and (b) the {100} plane for crystals with the rock salt (i.e., NaCl) structure. {110}<110> glide is preferred. [Figure adapted from Kingery et al, p. 713].

Limited # independent slip systems

Table 17.4 Independent slip systems for some ceramics.

Lattice type	*Crystal*	*Slip system*	*Number of independent systems*
Rocksalt	MgO, NaCl, LiF, NaF	$\{110\}\langle 1\bar{1}0\rangle$	2
Rocksalt	MgO, NaCl, LiF, NaF	$\{110\}\langle 1\bar{1}0\rangle$ $\{001\}\langle 1\bar{1}0\rangle$ $\{111\}\langle 1\bar{1}0\rangle$	5 at high temperature
Fluorite	UO_2 and CaF_2	$\{001\}\langle 1\bar{1}0\rangle$	3
	TiC and UC	$\{111\}\langle 1\bar{1}0\rangle$	5
Spinel	$MgAl_2O_4$	$\{111\}\langle 1\bar{1}0\rangle$ $\{110\}\langle 1\bar{1}0\rangle$	5
Fluorite	UO_2 and CaF_2	$\{001\}\langle 1\bar{1}0\rangle$ $\{110\}\langle 1\bar{1}0\rangle$ $\{111\}\langle 1\bar{1}0\rangle$	5 at high temperatures
Hexagonal	Al_2O_3, C (graphite), BeO	$\{0001\}\langle 11\bar{2}0\rangle$	2
Hexagonal	Al_2O_3, C (graphite), BeO	$\{0001\}\langle 11\bar{2}0\rangle$ $\{1\bar{2}10\}\langle 10\bar{1}0\rangle$ $\{1\bar{2}10\}\langle 10\bar{1}1\rangle$ $\{1\bar{1}02\}\langle 01\bar{1}1\rangle$ $\{10\bar{1}1\}\langle 01\bar{1}1\rangle$	5 at high temperatures
Sphalerite	ZnS, β-SiC	$(111\}\langle 1\bar{1}0\rangle$	5

[from Carter and Norton, Ceramic Materials Science and Engineering, p. 314]

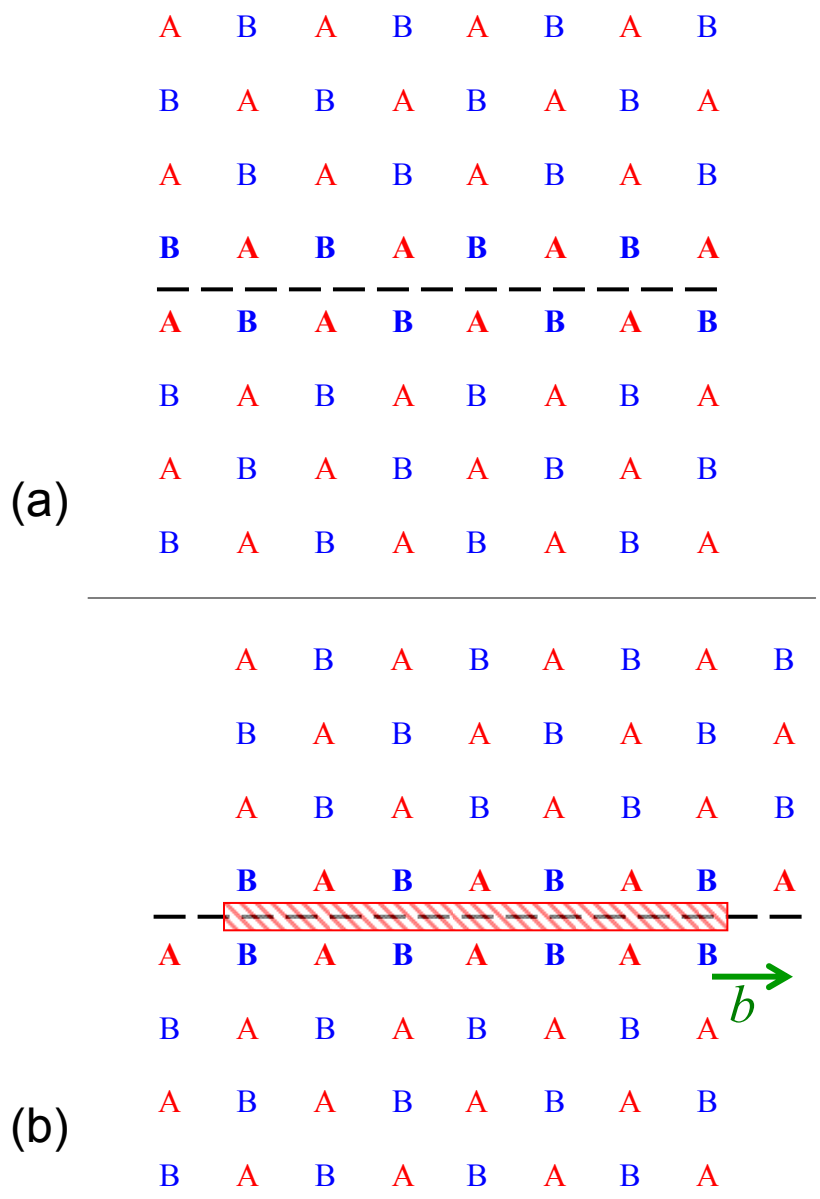
Ordered Structures

a) Schematic of an ordered AB crystal.

A atoms are next to B atoms.

b) Slip by passage of a single edge dislocation (Burgers vector = b) produces like bonds (A-A, B-B).

Undesirable. Antiphase boundary (APB) forms.



[Figures adapted from Courtney, p. 122]

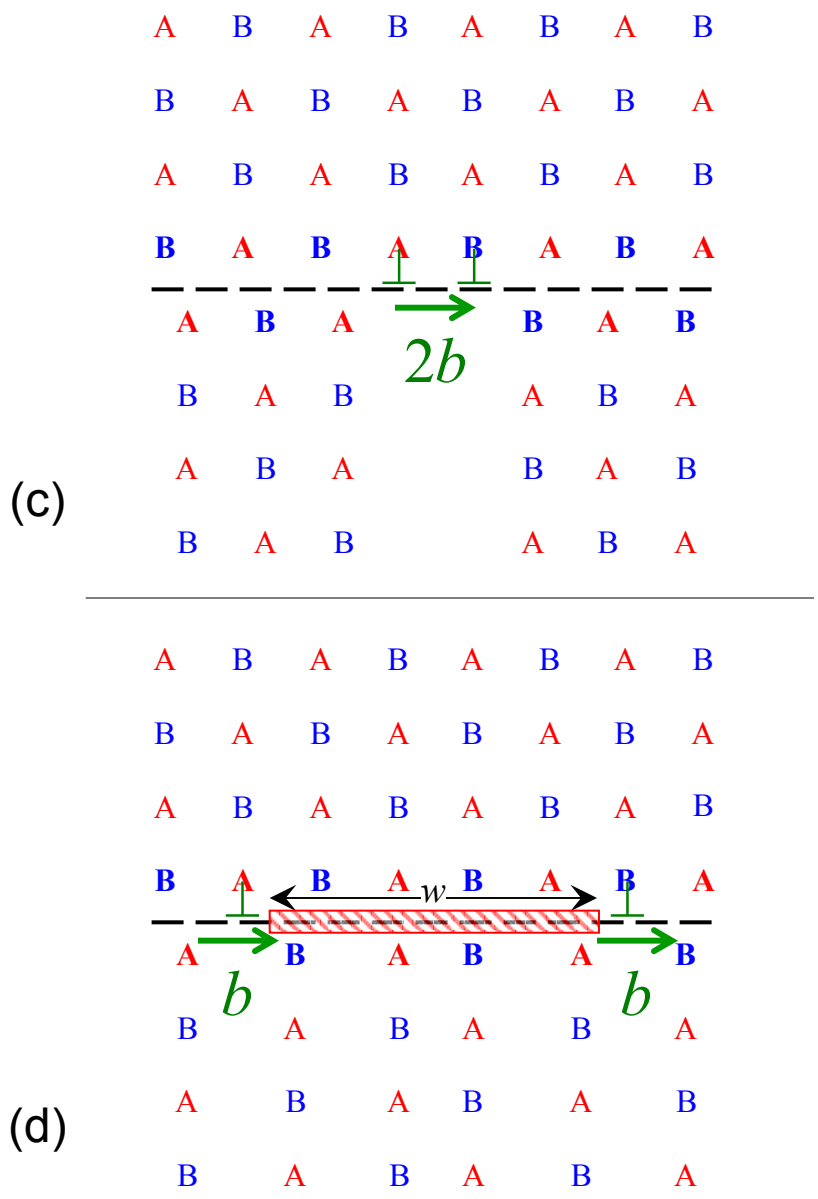
Ordered Structures

c) Slip by movement of a "superlattice" dislocation (i.e., 2 partial dislocations, Burgers vector = 2b) retains desired bonding.

A atoms next to B atoms.

d) Due to elastic repulsion, superlattice dislocations split into partials dislocations separation and an APB.

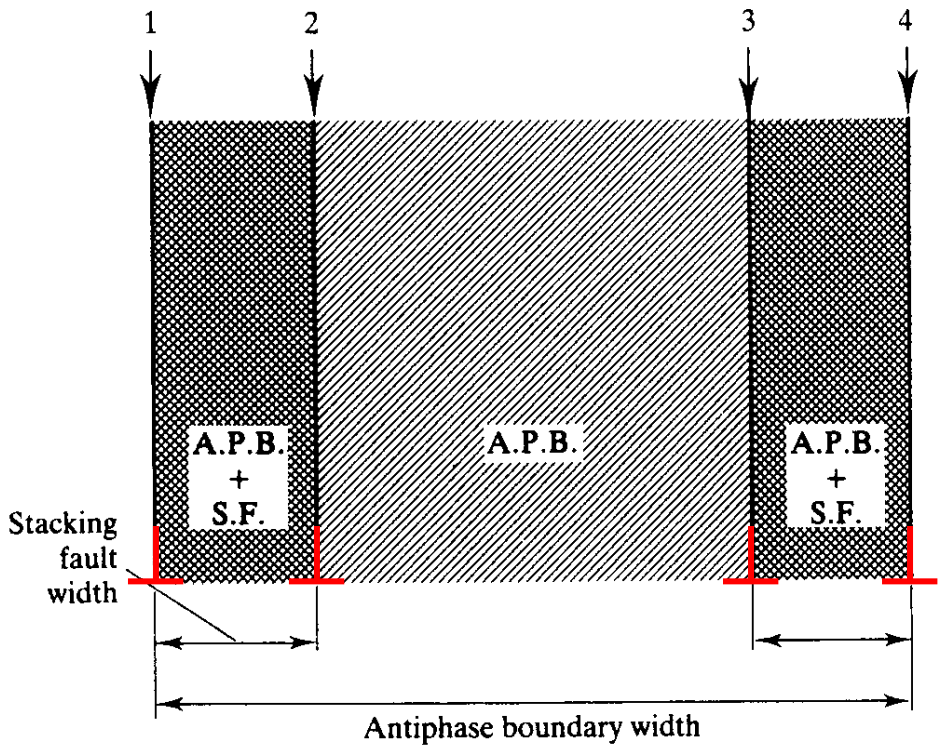
APB width (w) depends on APB energy.



[Figures adapted from Courtney, p. 122]

Ordered Structures

- In some crystals such as FCC Cu_3Au or Ni_3Al, the superdislocations are composed of two unit (a/2)[110] dislocations.
- In this structure the superdislocations can dissociate producing an additional APB bound on each side by individual stacking faults.
- Really complicates dislocation motion.



Schematic looking down on a slip plane in an ordered A_3B crystal. [Figure adapted from Courtney, p. 123; Originally from Marcinkowski, et al., *Acta Metall.*, **9** (1961) 129]

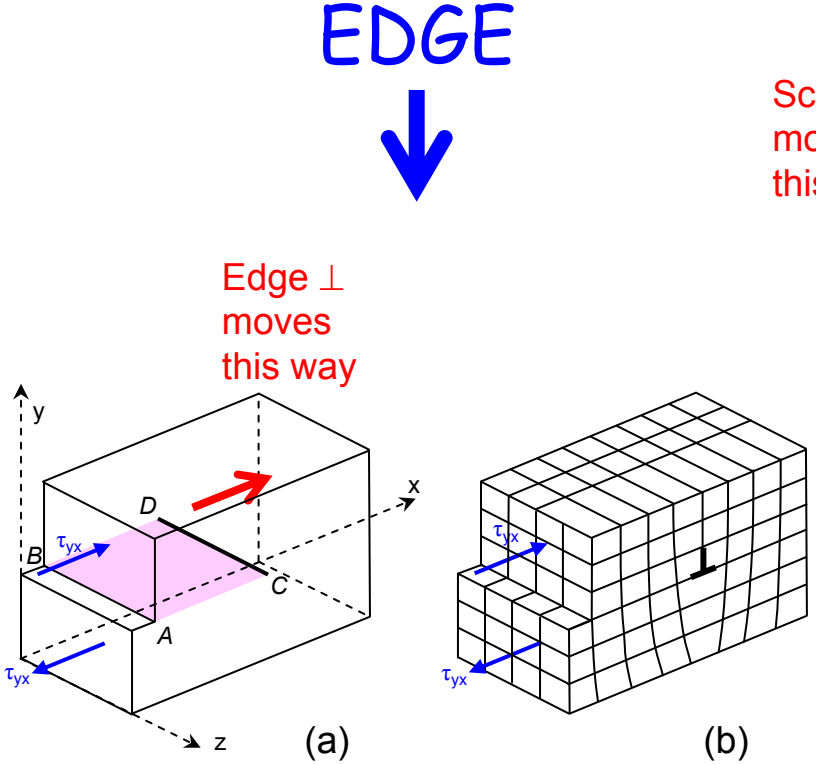
RECALL
dislocations move via

- Glide (conservative motion):
  - ⊥ moves on a surface that contains both its line and Burgers vector.
  - A ⊥ that moves this way is glissile.
  - A ⊥ that can't move is sessile.
  - ⊥ glide surface and direction depend upon crystal structure.

- Climb (non-conservative motion)
  - ⊥ moves out of the glide surface, perpendicular to the Burgers vector.

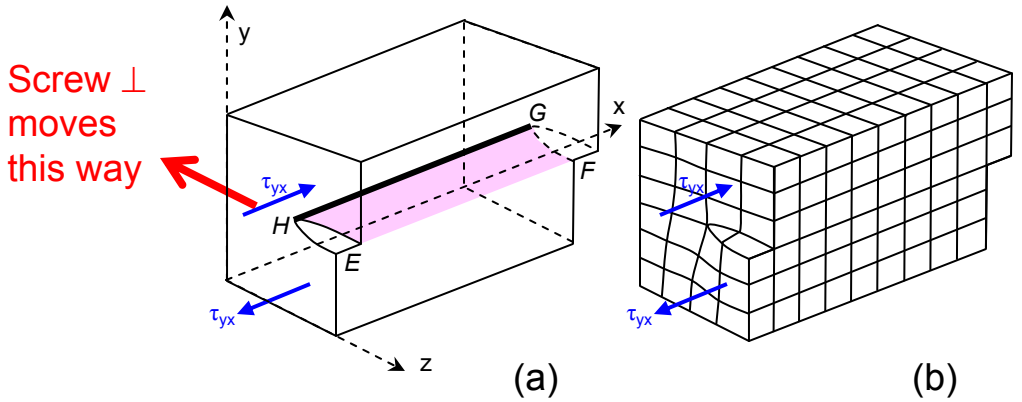
Dislocation motion via “glide” / “slip”

EDGE



Schematic illustration of single crystal deformation by motion of an edge dislocation. (a) Application of shear stress τ_{yx} can introduce an edge dislocation into a crystal along AB and cause it to move to position *DC*. (b) Lattice representation clearly showing that the dislocation has edge character.

Adapted from S.M. Allen and E.L. Thomas; The Structure of Metals; Wiley, New York, (1998) p. 284

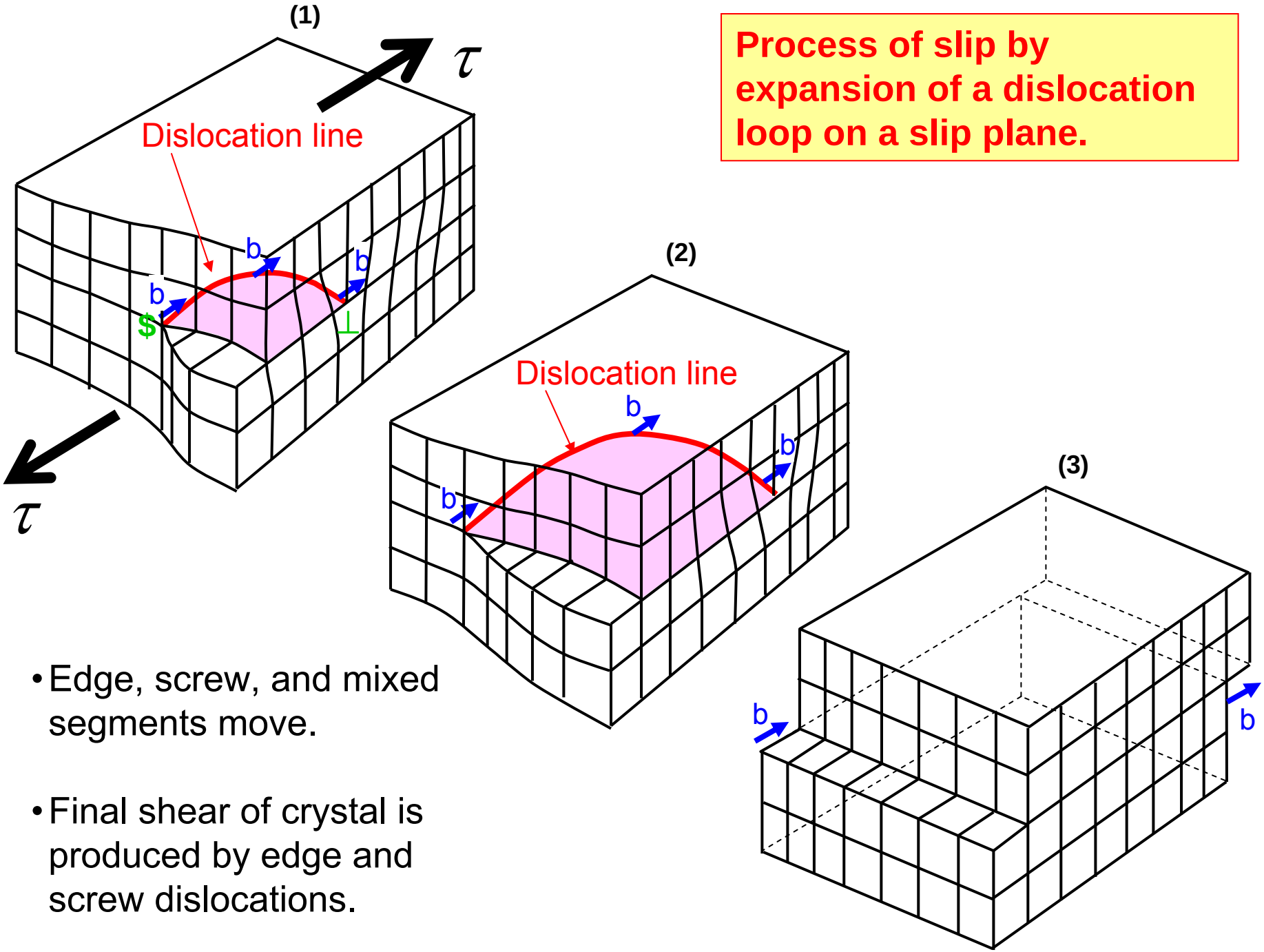


Schematic illustration of single crystal deformation via motion of a screw dislocation. (a) Application of a shear stress τ_{yx} can introduce a screw dislocation into a crystal along line *EF* and cause it to move to position *HG*. (b) Lattice presentation clearly showing that the dislocation has right-hand screw character.

Adapted from S.M. Allen and E.L. Thomas; The Structure of Metals; Wiley, New York, (1998) p. 285

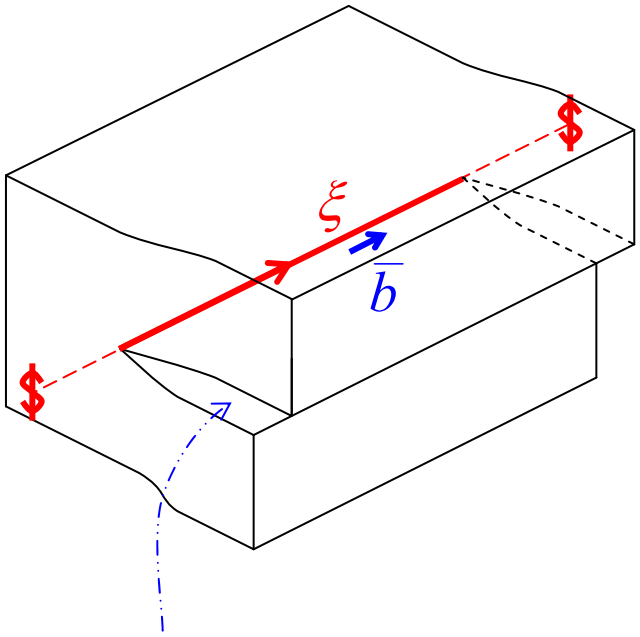
SCREW

Process of slip by expansion of a dislocation loop on a slip plane.



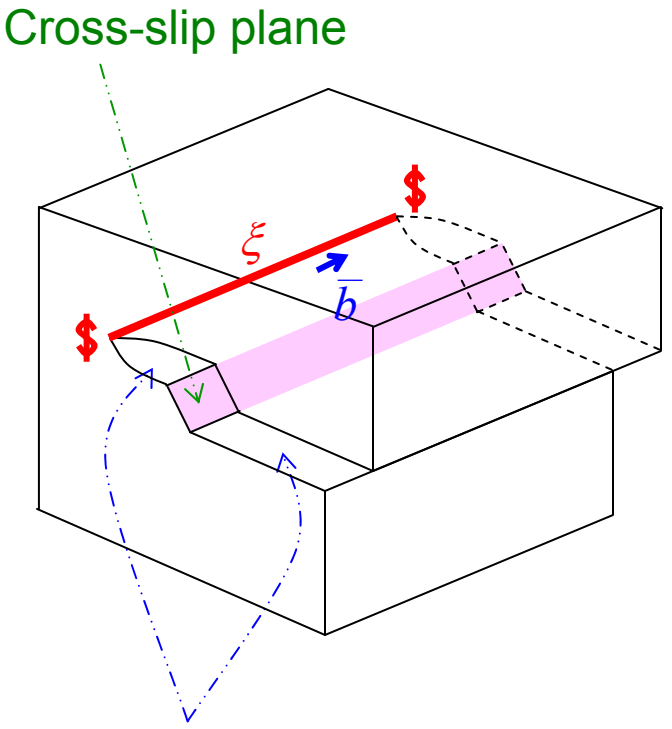
- Edge, screw, and mixed segments move.
- Final shear of crystal is produced by edge and screw dislocations.

Slip



Slip plane

Cross-Slip



Primary slip plane

Screw dislocations are not restricted to a single plane. They can cross-slip

Edge dislocations cannot

Dislocation Glide

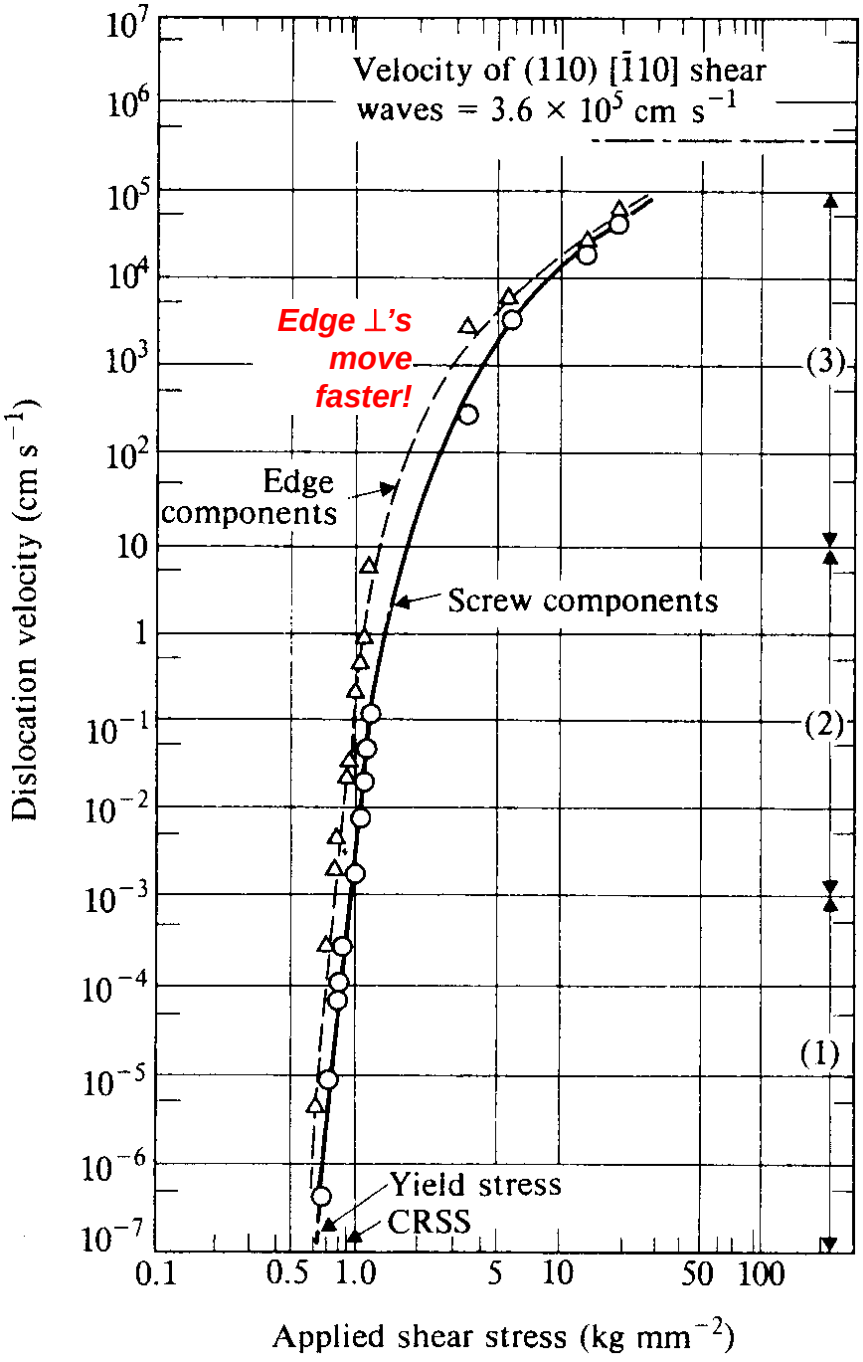
- Dislocations glide at glide velocities that depend on:
  - Applied stress;
  - Purity of the crystal;
  - Temperature;
  - Type of dislocation.

- Johnston and Gilman who showed that the *dislocation velocity* for a number of ionic crystals and metals *is a strong function of the shear stress in the slip plane* as follows:

$$v = A\left(\frac{\tau}{\tau_o}\right)^m$$

This equation is empirical in nature and applies for a specific velocity range: 10^{-9} to 10^{-3} m/s

- Where v is the dislocation velocity, τ is the applied shear stress in the slip plane, τ_o is the shear stress for v = 1 m/s, and m is a constant.

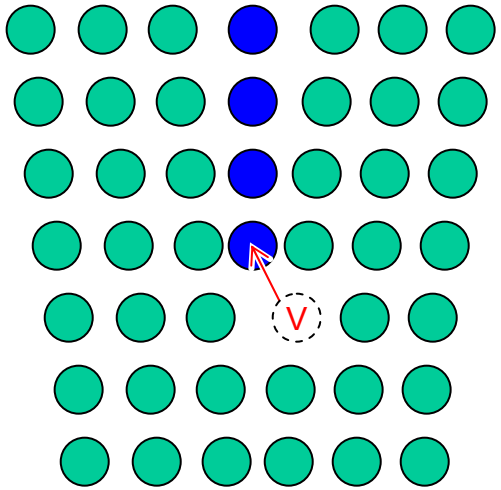


- Dislocation velocity increases rapidly at the *critical resolved shear stress* (τ_{crss} or CRSS).
- This is where plastic deformation actually begins.
- We address the details later.

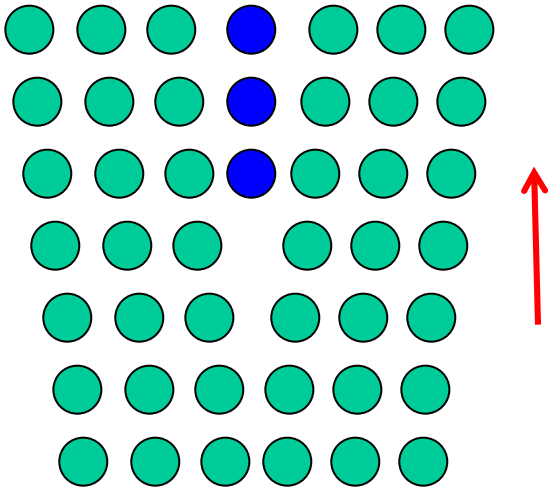
Figure

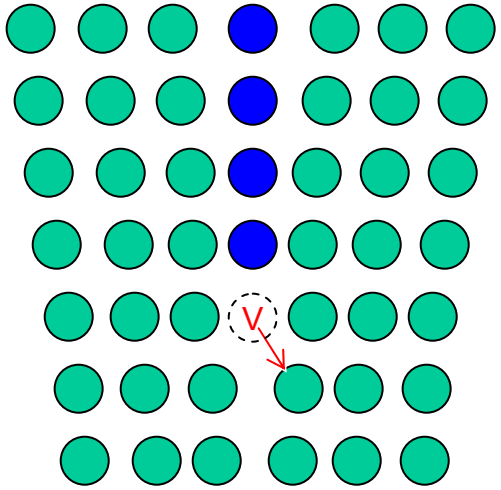
Stress dependence of the velocity of edge and screw dislocations in LiF (after Johnston and Gilman, *J. Appl. Phys.* **30**, 129, 1959). Scanned from E.W. Billington and A. Tate, The Physics of Deformation and Flow, McGraw-Hill, New York, 1981, pages 418 and 420.

Climb

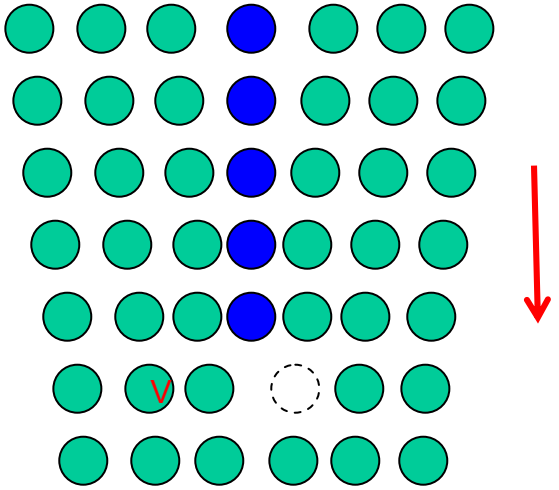


Positive climb of a dislocation due to vacancy annihilation





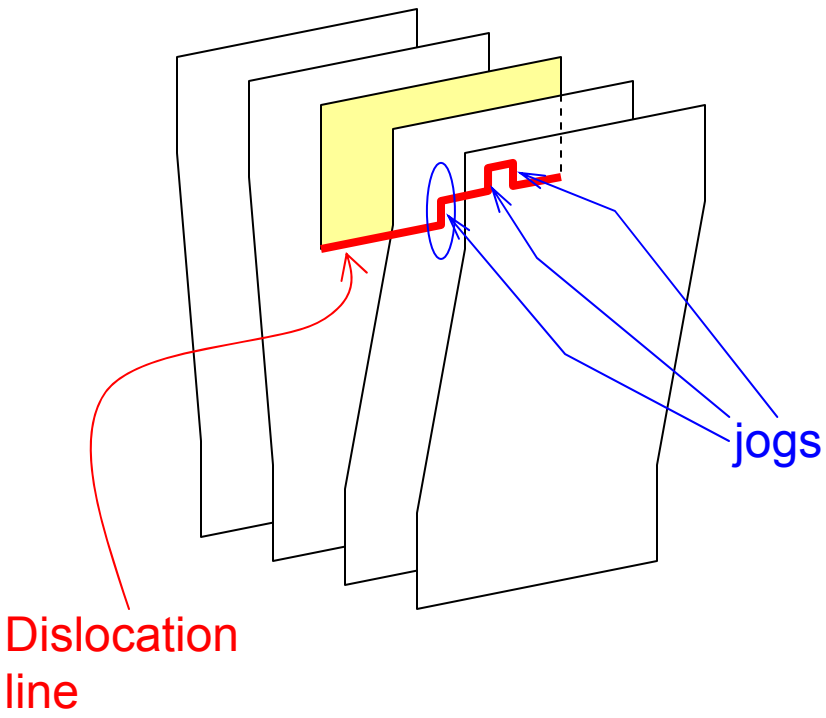
Negative climb of a dislocation due to vacancy generation



Climb

- Climb is a diffusion dominated process.
- It will be minimal at low temperatures where diffusion is difficult.
- It can be significant at high temperatures where diffusion is easier.

Implications of Climb



- Climb of short sections of dislocation lines result in the formation of steps called *jogs*.
- Dislocation climb proceeds by the nucleation and motion of jogs.

► *Jogs* are steps on a dislocation that move it from one atomic plane to another.

•There is also another type of dislocation step called a kink.

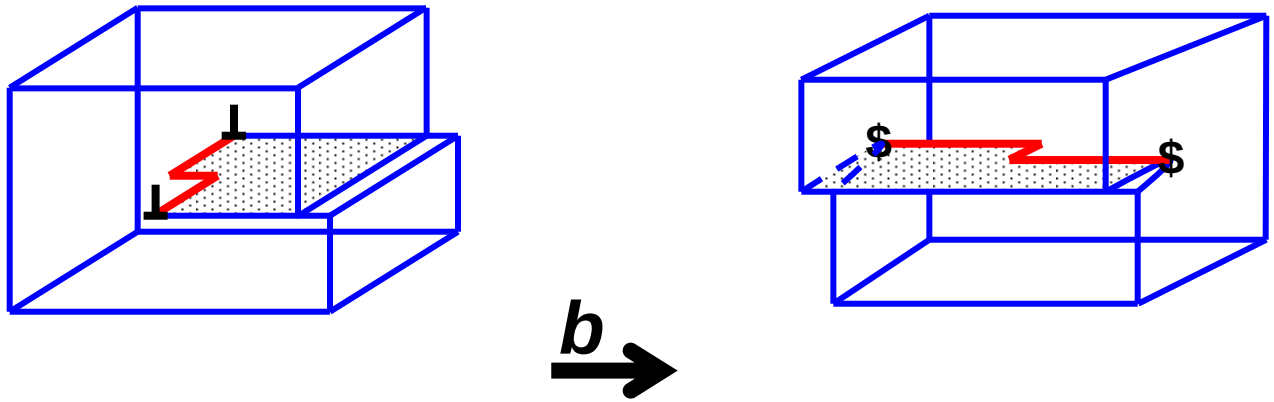
► *Kinks* are steps that displace the dislocation within the slip plane.

Jogs and kinks

- Jogs and kinks are short segments of a dislocation.
- They have the same Burgers vector as the line on which they lie.
- The same rules apply for conservative and non-conservative motion of jogs and kinks as regular dislocations.

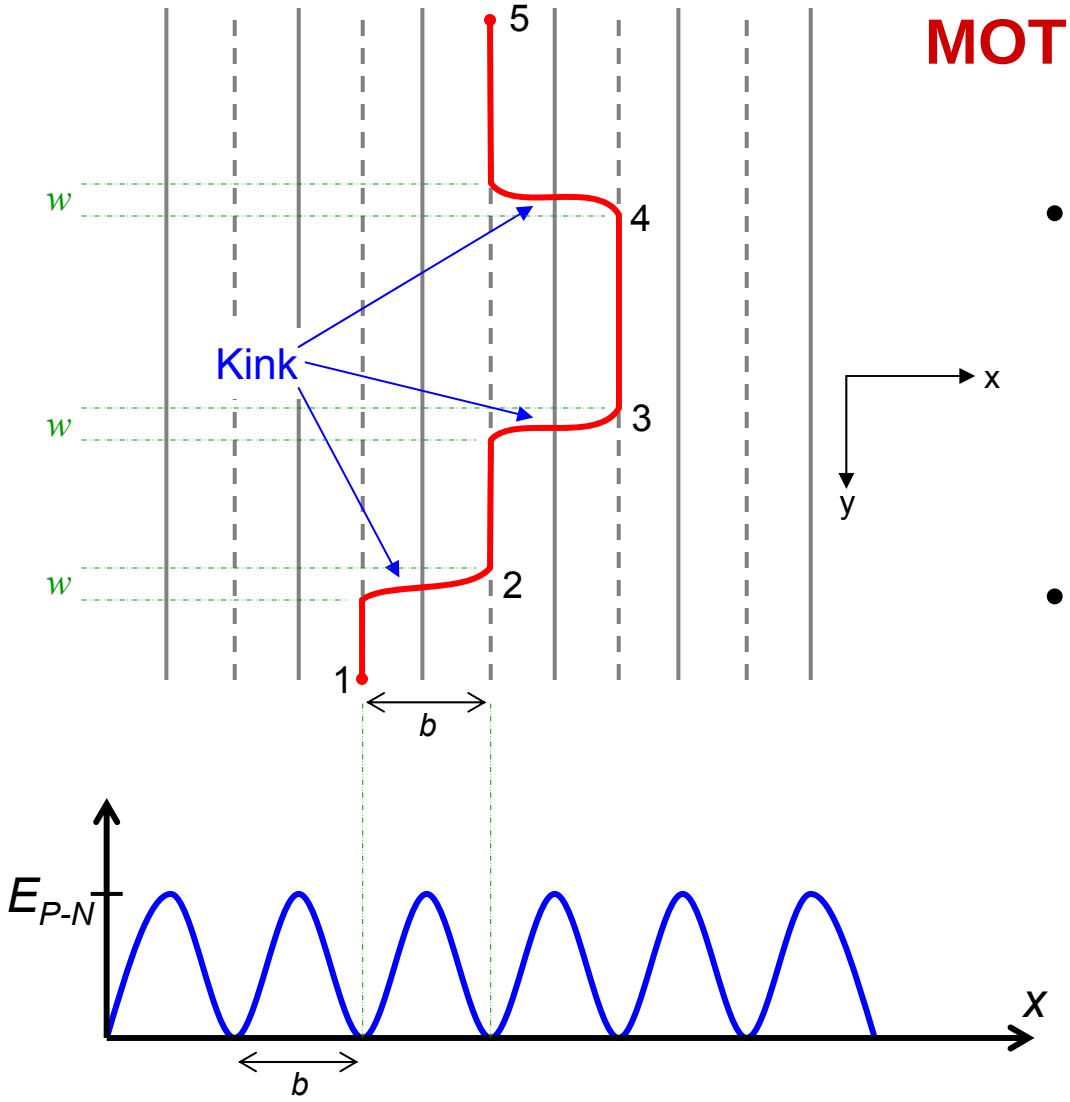
Kinks in edge and screw dislocations

These figures were adapted from Hull and Bacon's text (Fig. 3.16).



- Kinks have the same slip plane as the dislocation line.
- Kinks do not impede glide of a dislocation line.
- Kinks can actually assist glide.

MOTION OF DISLOCATIONS



- To glide, dislocations must overcome the Peierls-Nabarro barrier.
- Dislocation lines do this in a step-like fashion where a small segment proceeds beyond the Peierls barrier first producing a *kink*.

- Kinks spread laterally along the length of the dislocation line resulting in forward motion of the dislocation line.

Kink Propagation

- Dislocation velocity, $v_\perp$, is a function of kink velocity, v_k:

$$v_\perp = v_k \frac{b}{L} \qquad (v_k \gg v_\perp)$$

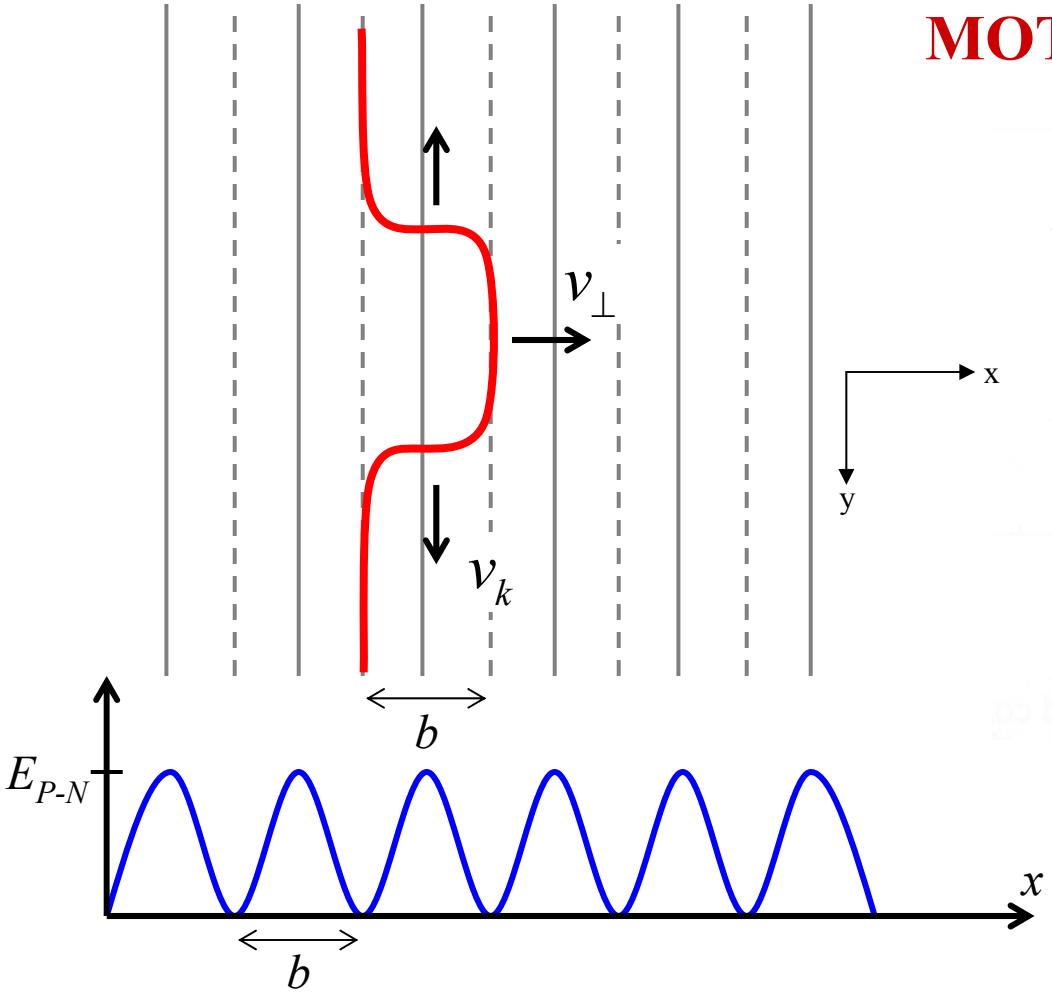
where L is the length of the dislocation segment and b is the Burgers vector.

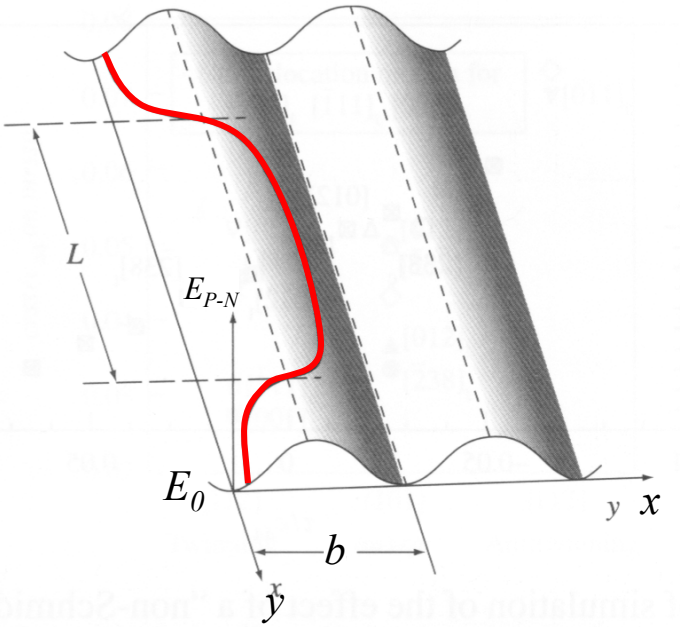
- Kink velocity is a function of kink formation energy (W_k):

$$v_k = \frac{2\sigma b^2 D_k}{kT} \exp\left(\frac{-Q}{kT}\right)$$

where Q = W_k for a single kink (#2) and $2W_k$ for a double kink (#3 and #4).

MOTION OF DISLOCATIONS





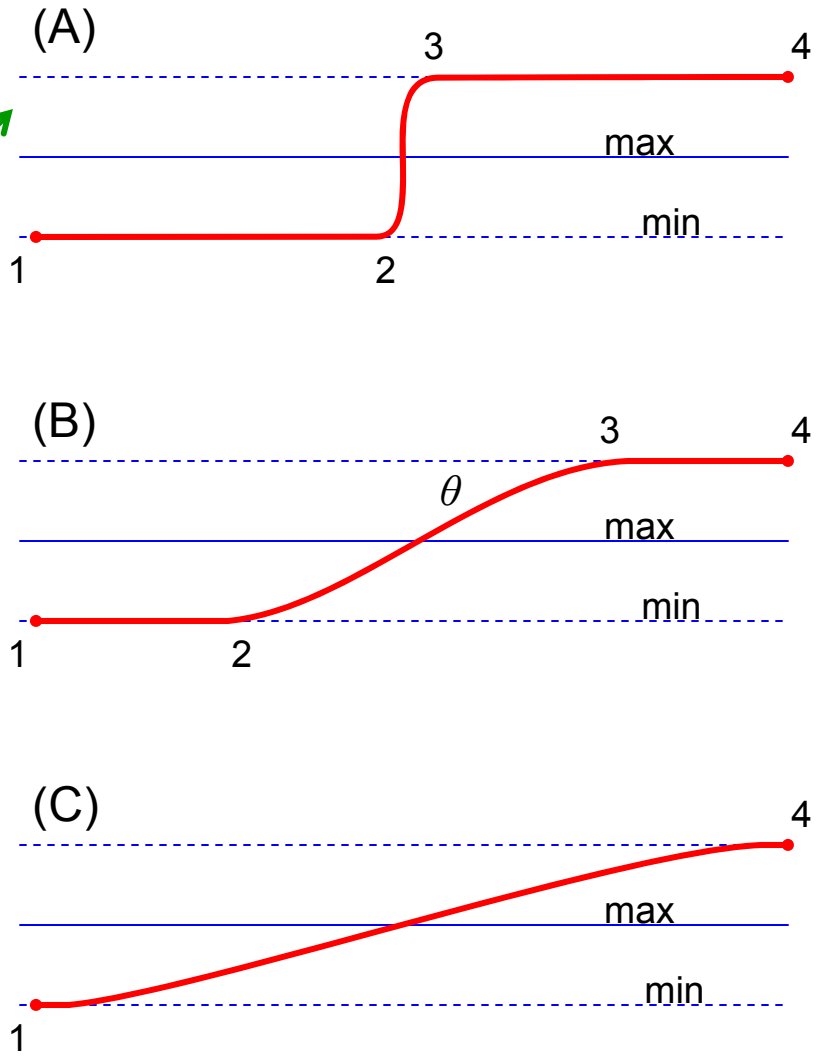
[Argon, p.96]

There is a driving force for the kink to move from a high energy to a low energy state.

- This process occurs because lateral propagation of kinks occurs more readily than forward motion of an entire line over the Peierls-Nabarro barrier.
- This is because $v_k \gg v_\perp$.

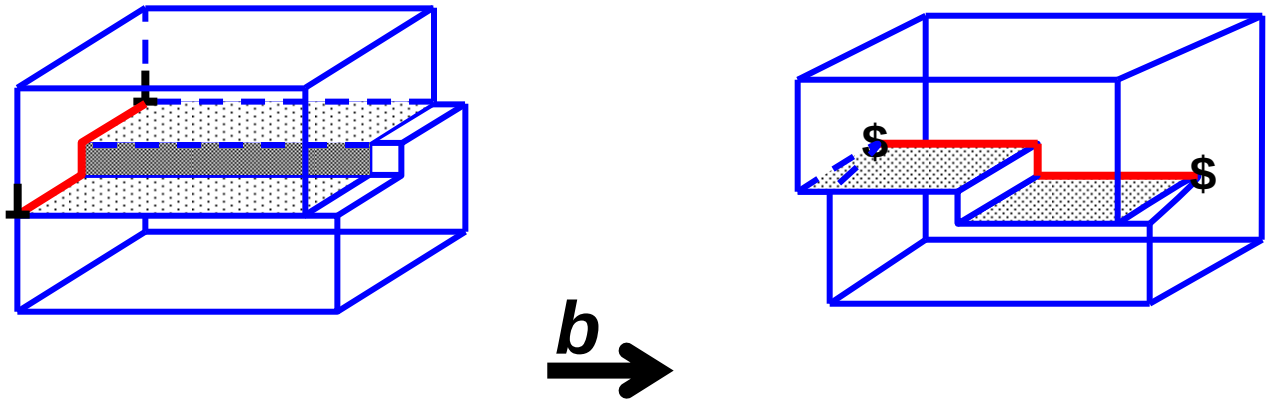
Kink Geometry

- Kink widths are determined by a balance between line tension forces and the Peierls barrier.
- (A) Kinks tend to be narrower and more difficult to move in less close packed structures.
- (B, C) Kinks tend to be more diffuse and easier to move in close-packed structures.



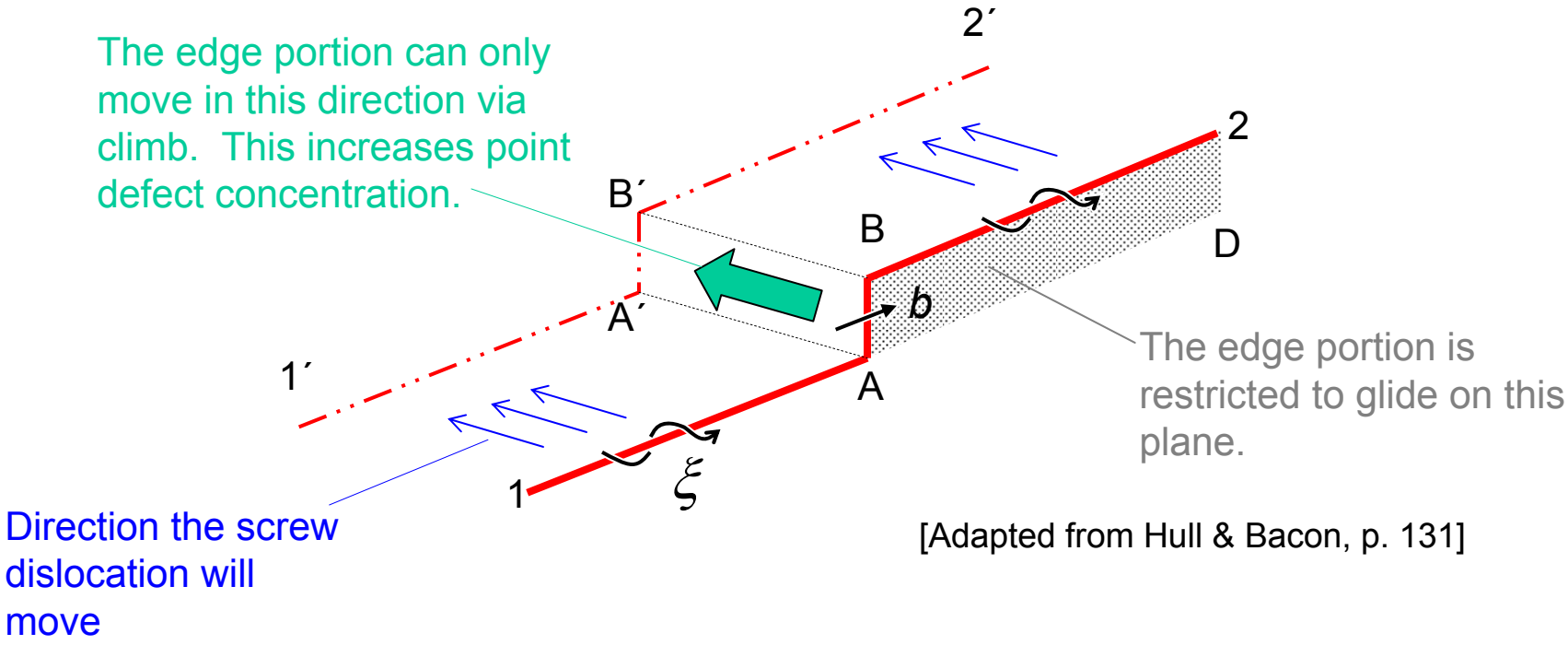
Jogs in edge and screw dislocations

These figures were adapted from Hull and Bacon's text (Fig. 3.16).



- Edge dislocations:
  - Jogs do not impede the glide of edge dislocations.
- Screw dislocations:
  - Jogs have "edge character" and are be restricted to glide along the dislocation line (normal to *b*). *This requires climb*. Thus, they do impede motion of screw dislocations. (illustrated on the next viewgraph)

Movement of a jog on a screw ⊥

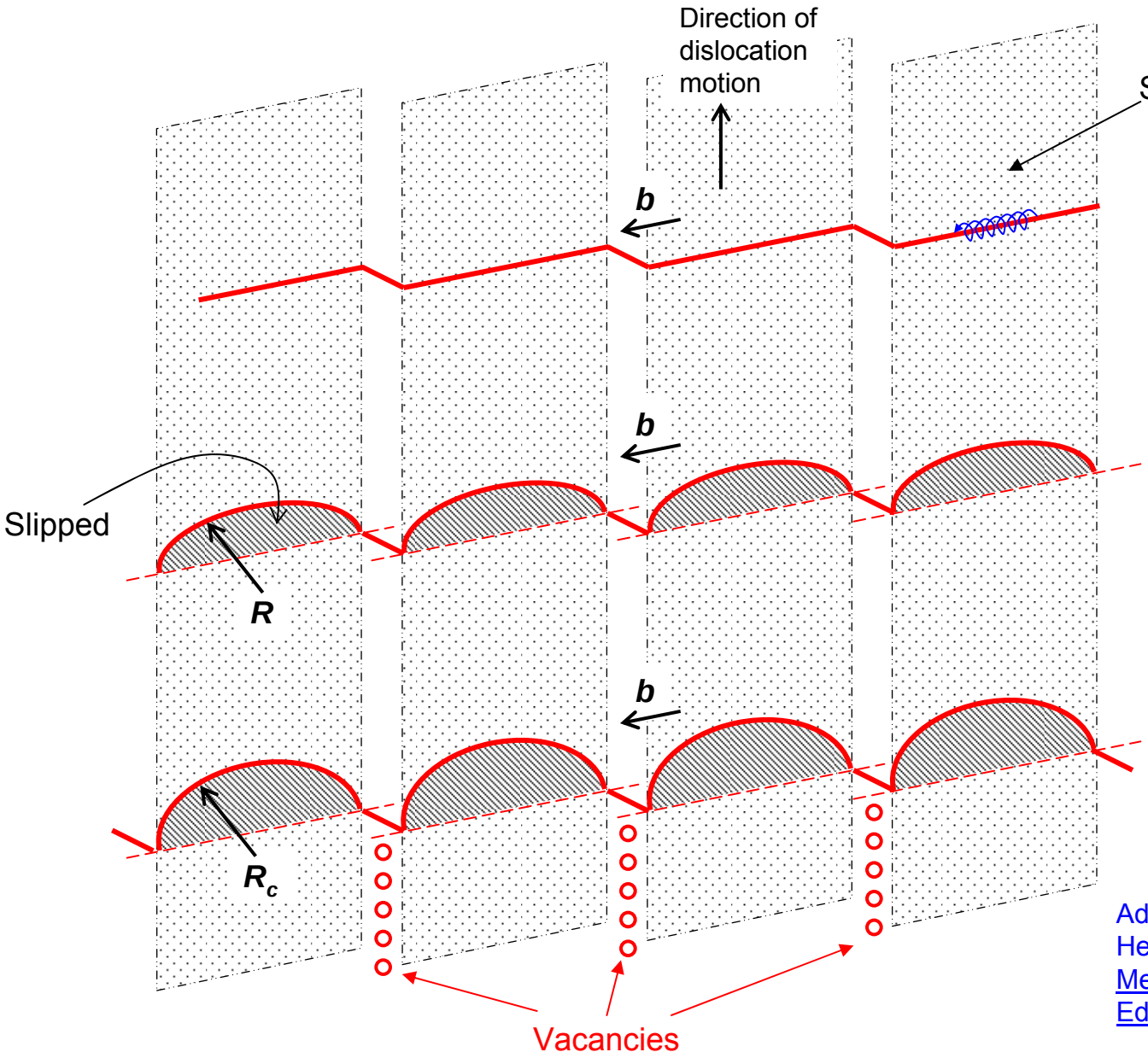


[Adapted from Hull & Bacon, p. 131]

Movement of a jog on a screw dislocation. The jog AB has a Burgers vector normal to AB. It is therefore a short length of edge dislocation. The plane defined by AB and it's Burgers vector is AB2D. It is the plane upon which AB can glide. Movement of the screw dislocation 1AB2 to 1´A´B´2´ requires climb of jog AB to A´B´

Motion of a jogged screw dislocation

implications



A: No stress

B: Under stress

The jog, which has edge character can't move on the available slip plane. The screw segments continue moving forming a dislocation dipole.

C: At critical stress

At this stress, the dislocation dipole pinches off leaving behind interstitials or dislocation loops.

Adapted from Fig. 2.26 in R.W. Hertzberg, Deformation and Facture Mechanics of Engineering Materials, 4th Ed., (Wiley, New York, 1996) p. 84.

Origin of dislocation debris and dipoles

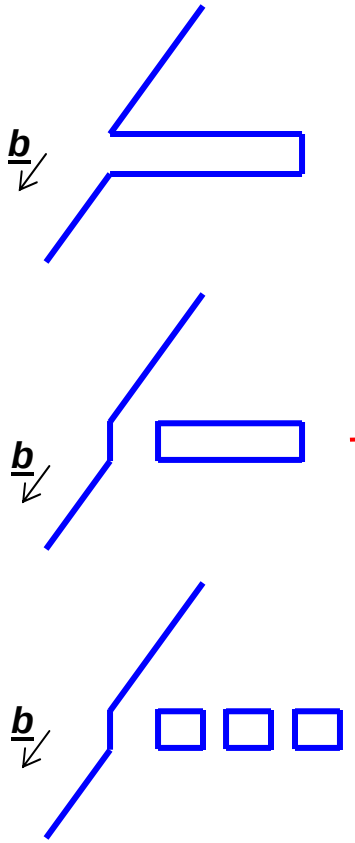


Figure Formation of dislocation loops from a dislocation dipole. (a) Dislocation dipole; (b) Elongated dislocation loop; and (c) row of small loops (i.e., debris). [Adapted from Fig. 7.11 in Hull & Bacon].

The trails of defects are often produced during plastic deformation usually appear as dislocation loops.

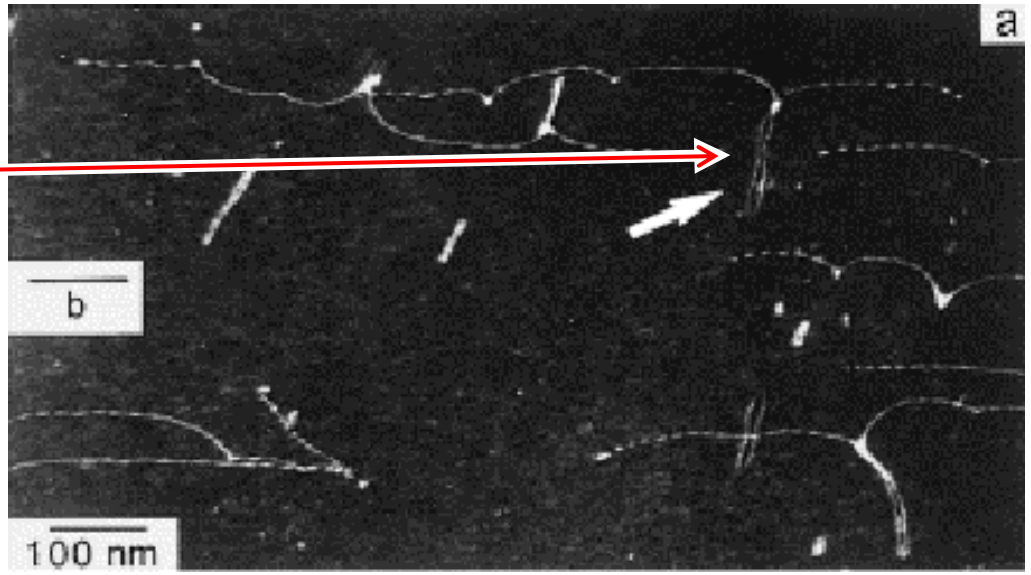
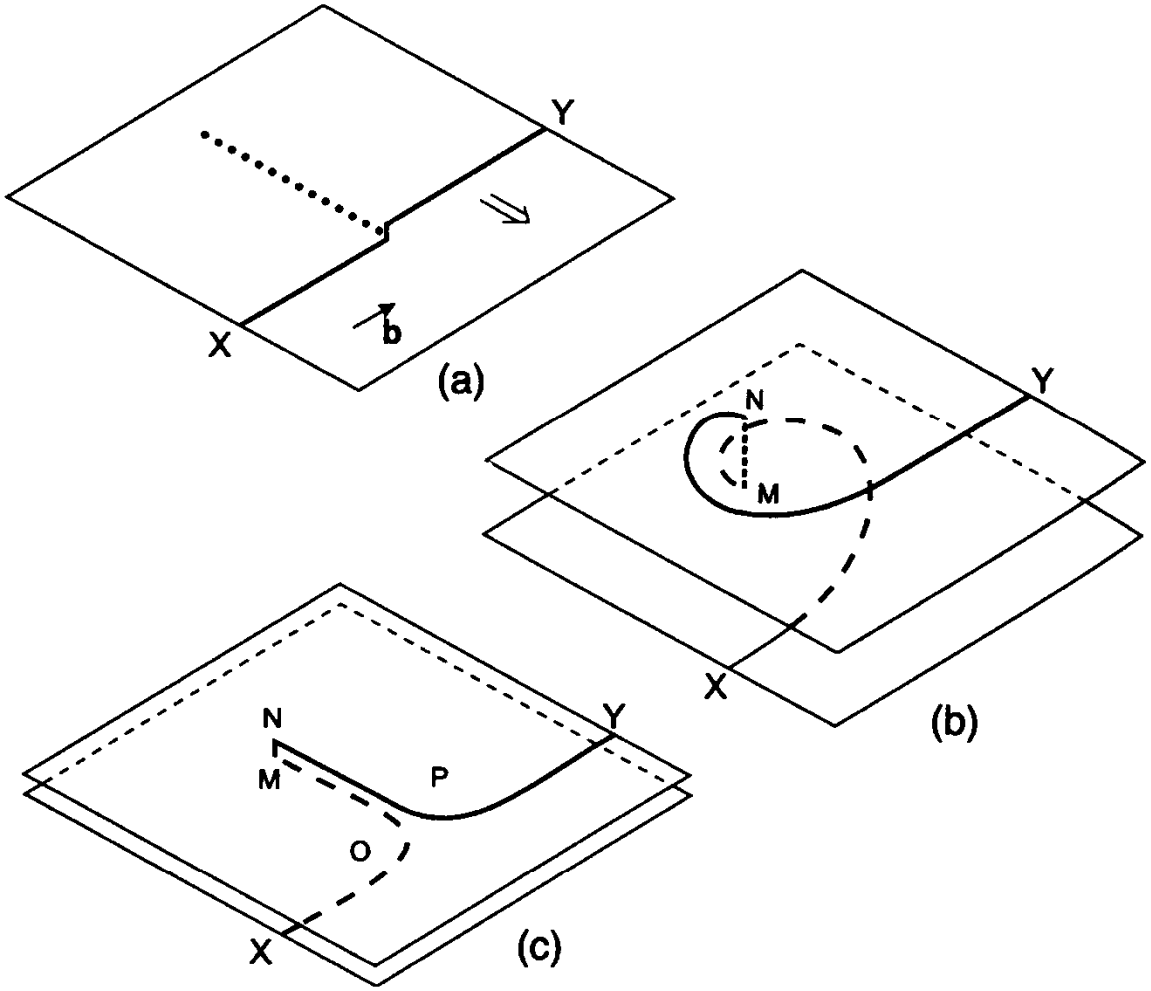


Fig. 7. Dislocation dipoles and debris in two-phase titanium aluminides. Alloy 1, compression at *T*=295 K to strain =3%. (a) Dislocation dipoles and debris (arrowed) are trailed and terminated at jogs in screw dislocations. From F. Appel, U. Sparka and R. Wagner, *Intermetallics* **v.7**, n. 3-4 (1999) pp. 325-334.



Synopsis:

- Small jogs are dragged behind
- Large jogs, dislocations move independently
- Intermediate jogs, dislocations interact and cannot pass each other except at very high stress.

From D. Hull and D.J. Bacon, Introduction to Dislocations, 4th Ed., (Butterworth-Heinemann, Oxford, 2001).

Figure 7.8 Behaviour of jogs with different heights on a screw dislocation moving in the direction shown by the double arrow. (a) Small jog is dragged along, creating point defects as it moves. (b) Very large jog – the dislocations *NY* and *XM* move independently. (c) Intermediate jog – the dislocations *NP* and *MO* interact and cannot pass by one another except at a high stress. (After Gilman and Johnston, *Solid State Physics*, **13**, 147, 1962.)

Jogs and kinks can also result from the intersection of dislocations

Intersection of Dislocations

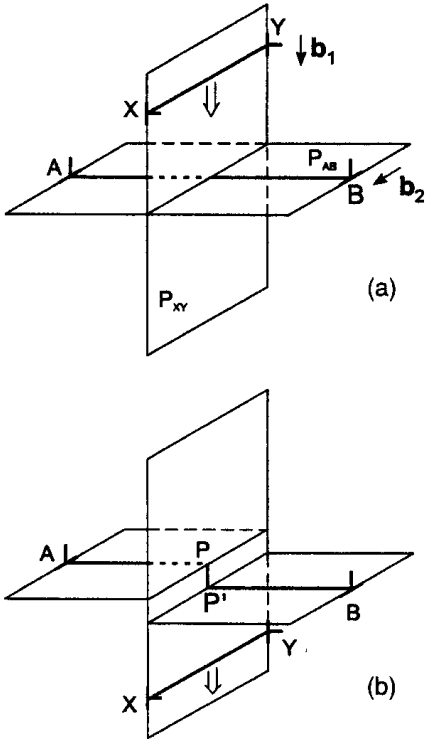


Figure 7.1 Intersection of edge dislocations with Burgers vectors at right angles to each other. (a) before intersection and (b) after intersection producing jog *PP′* in *AB*. [Adapted from Hull & Bacon].

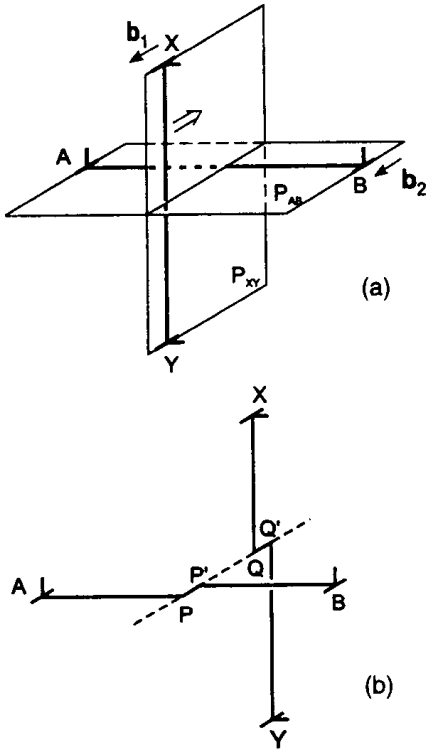


Figure 7.2 Intersection of edge dislocations with parallel Burgers vectors. (a) before intersection and (b) after intersection producing kink *PP′* in *AB* and kink *QQ′* in *XY* [Adapted from Hull & Bacon].

Dislocation Intersections

Creation of 1 edge jog which impedes further screw dislocation motion

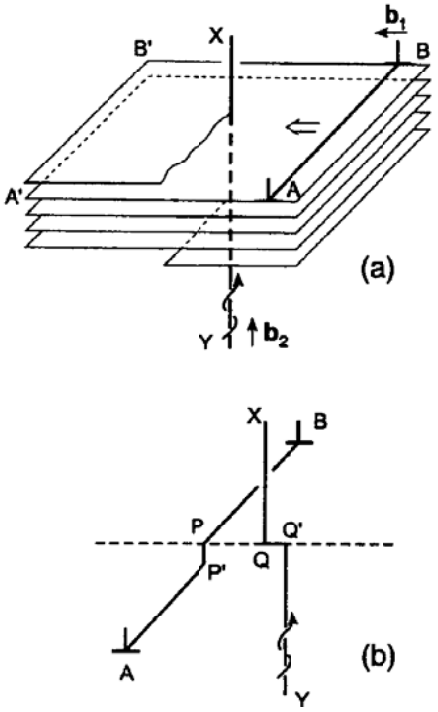
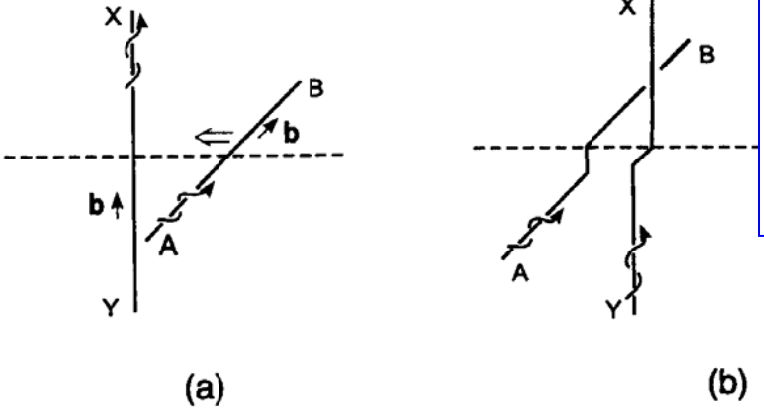


Figure 7.3 Intersection of edge dislocation *AB* with right-handed screw dislocation *XY*. (a) Before intersection. (b) After intersection jog *PP′* is produced on *AB* and jog *QQ′* is produced on *XY* [Adapted from Hull & Bacon].



Creation of 2 edge jogs which impedes further dislocation motion

Figure 7.4 Intersection of screw dislocation *AB* and *XY*. (a) Before intersection. (b) After intersection jogs are produced on both screw dislocations. [Adapted from Hull & Bacon].

Dislocation intersections can lead to dislocation multiplication and work hardening.

What happens when dislocations collide (5)?

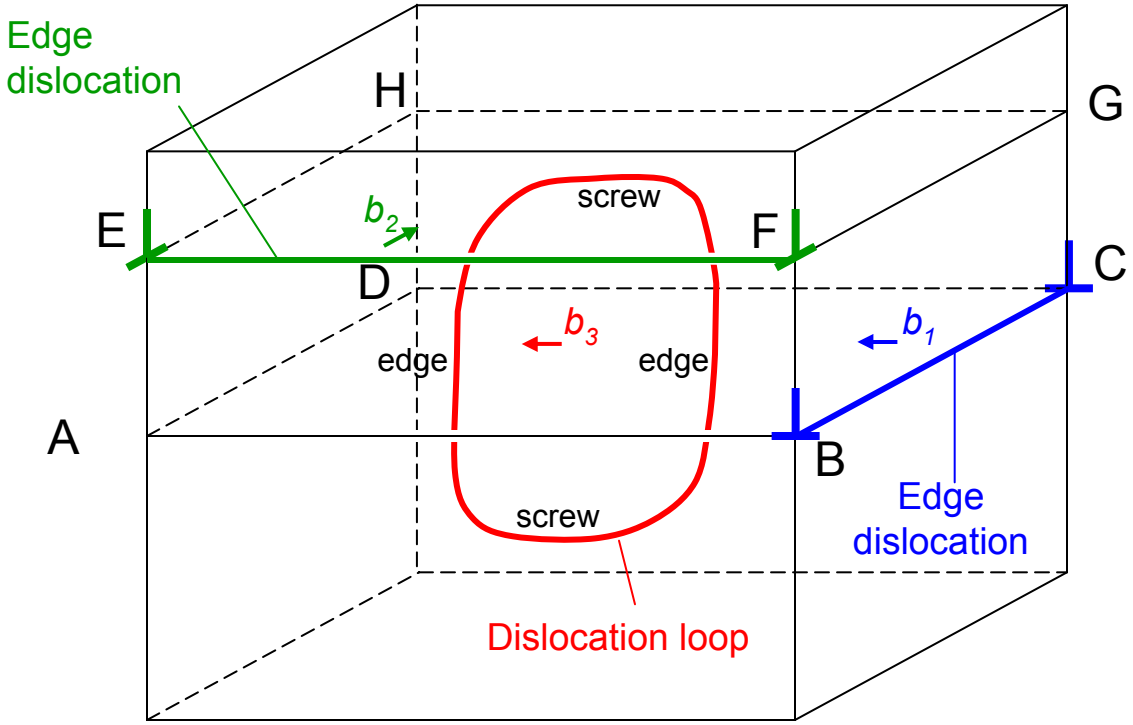


Figure A. What happens when two edge dislocations with different Burgers vectors move across the slip planes ABCD and EFGH and to cut a dislocation loop.

What happens when dislocations collide (6)?

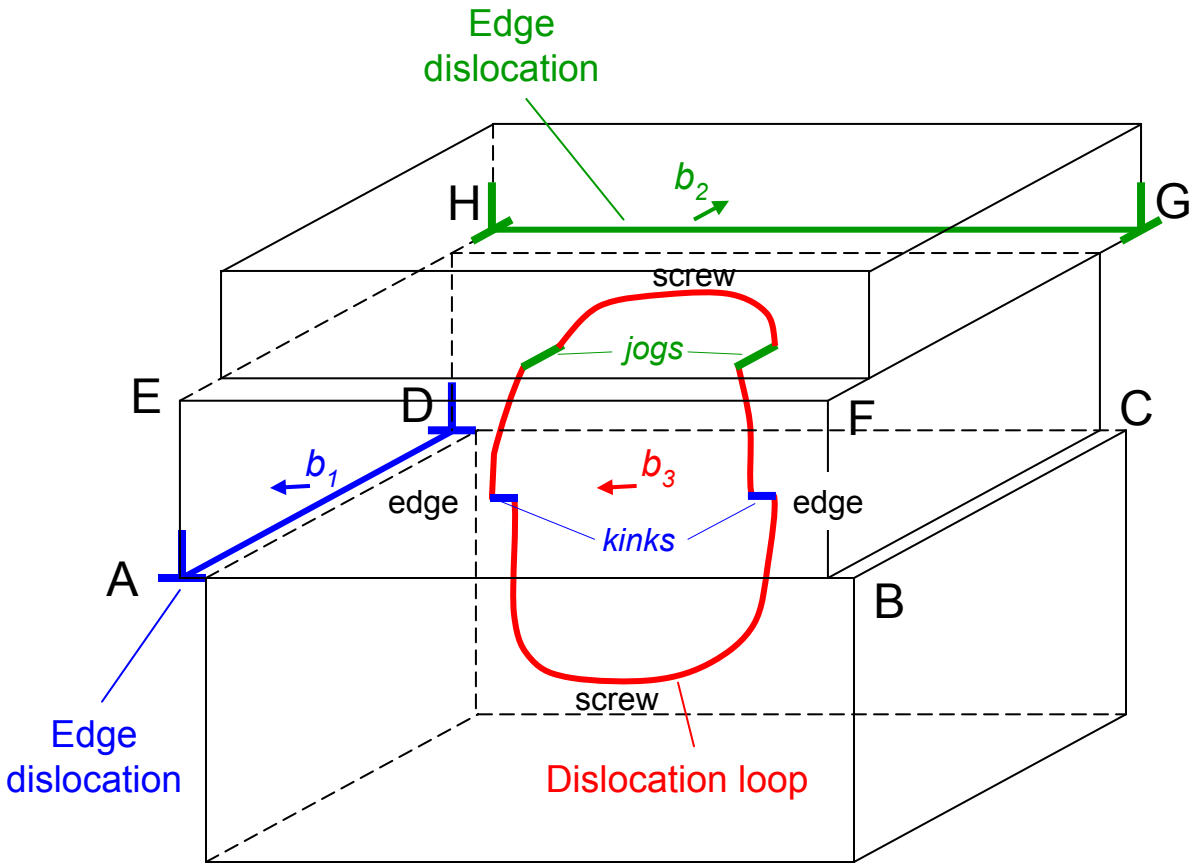


Figure B. Two edge dislocations are assumed to have moved across the slip planes ABCD and EFGH and to have cut a dislocation loop. The result of this intersection is the formation of a pair of kinks and jogs on the dislocation loop. In this drawing the kinks will be of magnitude b_1 while the jogs will be of magnitude b_2.