HOMEWORK
From Dieter
4-12

Module #10

Plastic Strain via Dislocation Motion and Dislocation Multiplication

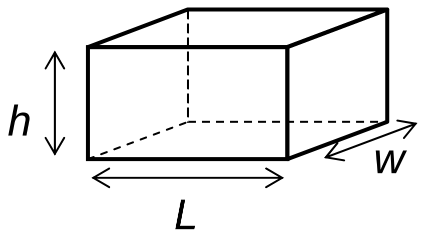
READING LIST

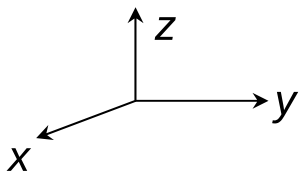
DIETER: Ch. 4, Pages 119-123

What are the implications of dislocation motion on the properties of crystals?

How much strain is caused by dislocation motion?

Implications of dislocation motion





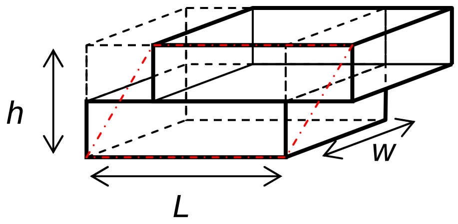
- Consider the crystal illustrated to the left.

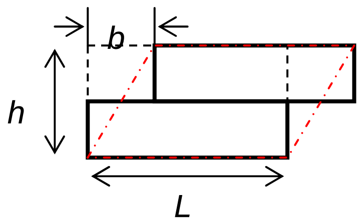
- If a single dislocation passes through the crystal, what will be the resulting strain?

3-D

2-D

Cubic crystal after passage of a single dislocation





- First we recognize that the shear strain is simply defined by the equation:

$$\text{shear strain} = \gamma = \frac{b}{h}$$

- Let's assume that the original crystal dimension are 1 cm × 1 cm × 1 cm and that *b* = 1 Å (this is roughly of the correct order of magnitude).

- The shear strain can be calculated as:

$$\gamma = \frac{b}{h} = \frac{1\times10^{-10}\ \text{m}}{1\times10^{-2}\ \text{m}} = 1\times10^{-8}\ \text{or}\ 1\times10^{-6}\%$$

This amount is not really perceptible so how can dislocations cause strain?

How can dislocations be responsible for plastic strain?

- The solution is simple. There must be multiple dislocations to cause perceptible amounts of strain.

- We can calculate this too.

- Assuming a random number of dislocations, *N,* the shear strain becomes:

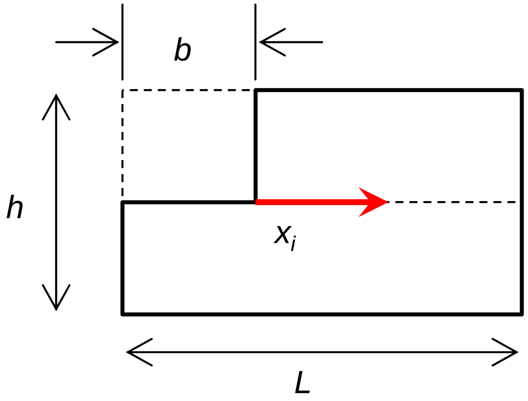
$$\gamma = \frac{Nb}{h}$$

- This expression yields better results, however, it assumes that all dislocations pass all of the way through the crystal.

  In reality, we can't make this assumption. Sometimes dislocations can form nodes within a crystal or just annihilate out.

  Thus we need to revise our estimate.

When dislocations pass partially through a crystal

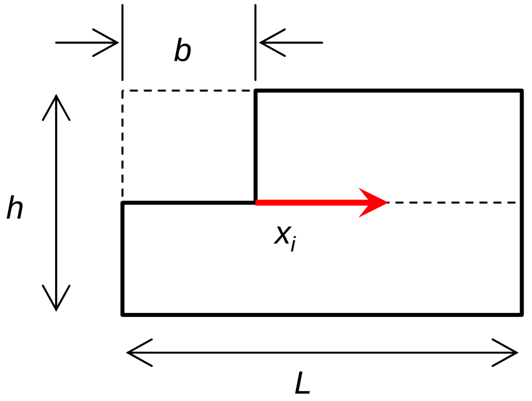


- Assuming that all of the dislocations move a distance x_i along the crystal length, the strain equations for single and multiple dislocations can be expressed as follows:
  - Single dislocation:

$$\gamma = \frac{x_i}{L}\frac{b}{h}$$

  - Multiple dislocations:

$$\gamma = \frac{1}{L}\frac{b}{h}\sum_{1}^{N} x_i$$



If N dislocations move an average distance $\overline{x}$ then,

$$\gamma = \frac{N\overline{x}}{L}\frac{b}{h} \text{ for multiple dislocations}$$

$L \times h =$ area of the end of the crystal

$$\text{dislocation density} = \rho_\perp = \frac{N}{Lh} \equiv \frac{\# \perp \text{ lines}}{\text{area}}$$

$$\therefore$$

$$\boxed{\gamma = \rho_\perp b\overline{x}}$$

Actually $\rho_\perp$ is defined as the total dislocation line length per unit volume OR more simply as the number of dislocation lines that cut a unit cross-sectional area.

The shear strain rate associated with this type of motion is:

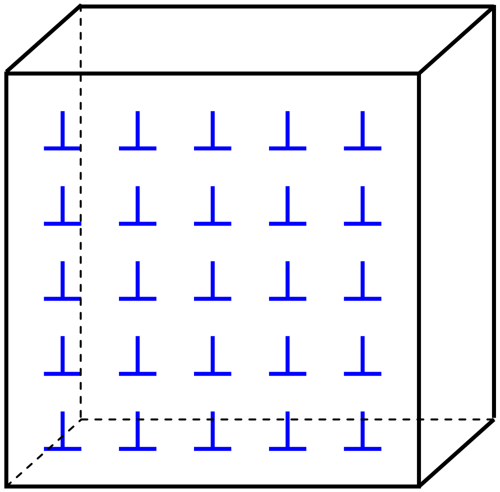
$$\dot{\gamma} = \frac{d\gamma}{dt} = \rho_\perp b\frac{dx}{dt} = \rho_\perp b\dot{x} \text{ or } \boxed{\dot{\gamma} = \rho_\perp bv}$$

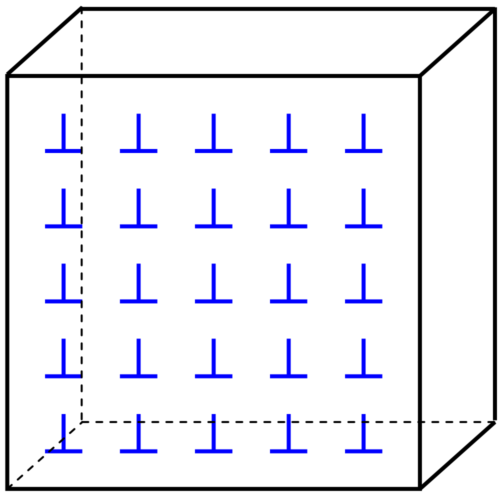
where v is the dislocation velocity.

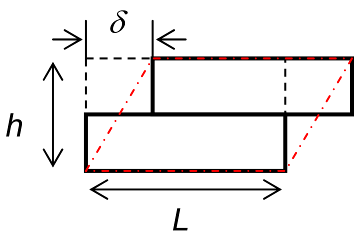
This is the Taylor-Orowan relation, which relates dislocation motion to strain rate. We'll use it again a bit later.

Example Problem

- Consider a 1 cm × 1 cm × 1 cm cubic crystal containing a uniform distribution of edge dislocations with an average spacing of *x* as schematically illustrated below. Keep in mind that we do not know exactly how many dislocations there are. How many dislocations per unit area are required to cause 1% strain (i.e., $\gamma = 0.01$) in this crystal? How many dislocations per unit area are required to cause 10% strain (i.e., $\gamma = 0.10$) in this crystal? Once again, we will let b = 1 Å.







$$\gamma = 10^{-2} = \rho_{\perp}(10^{-8}\ \text{cm})x$$

We find x by invoking the definition of shear strain:

$$\gamma = \frac{\delta}{h} = \frac{Nb}{h} = \frac{x}{h} = \frac{x}{1\ \text{cm}}; \quad x = 10^{-2}$$

1% strain: $\rho_{\perp} = \frac{\gamma}{bx} = \frac{10^{-2}}{(10^{-8})(10^{-2})}$, $\boxed{10^{8}\ \text{dislocation lines per cm}^{2}}$

10% strain: $\rho_{\perp} = \frac{\gamma}{bx} = \frac{10^{-1}}{(10^{-8})(10^{-2})}$, $\boxed{10^{9}\ \text{dislocation lines per cm}^{2}}$

How does the calculation compare with reality?

Typical dislocation densities encountered in a parallepiped sample (3 x 3 x 8 mm)

Sample history	dislocation density ρ (m^{-2})	Length of dislocation line per sample (km)	Mean distance *D* between dislocations (μm)
As grown	10^{10}	100	10
As grown and annealed	10^{8}	1	100
→ Deformed	$10^{13} - 10^{15}$	$10^{5} - 10^{8}$	0.1

[from Veyssiere, in Mechanics of Materials Fundamentals and Linkages, 1999, p.272]

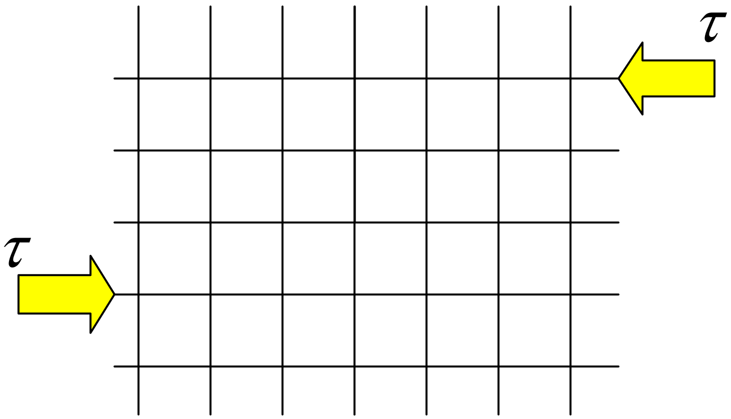
- There generally many dislocations to begin with.
- Many more are required for deformation.
- Dislocation density (ρ) increases significantly during deformation.

Dislocation Sources

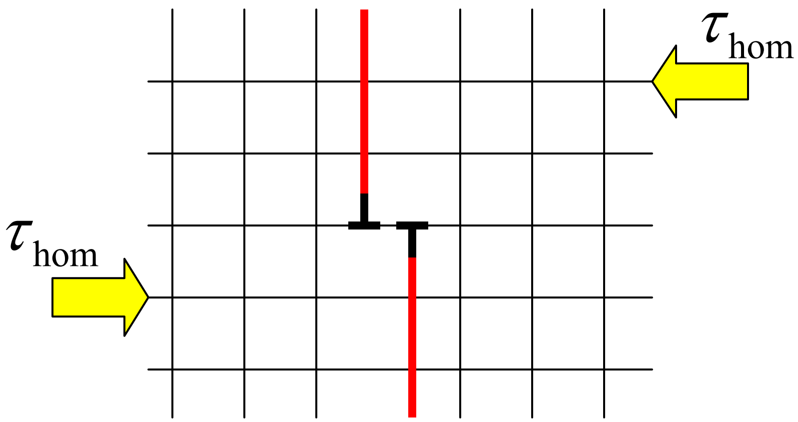
- Homogeneous nucleation
- ► Interfaces
  - Grain or phase boundaries
  - Surfaces or surface films
- ► Frank-Read sources
- Form during crystal for film growth
- To accommodate of strains at interfaces
- ► Condensation of point defects
- Etc…

Dislocation sources

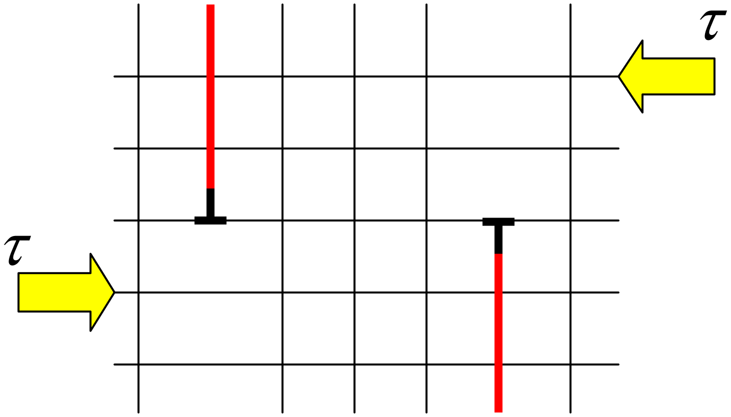
- *Homogeneous nucleation* of dislocations is unlikely under normal circumstances.



(a)



(b)



(c)

Requires very high stresses

$$\tau_{\text{hom}} \cong \frac{G}{14} \text{ to } \frac{G}{30}$$

Possible under "shock loading" conditions

Meyers Model for Homogeneous Nucleation During Shock Loading

- Dislocations are homogeneously nucleated at (or near) a shock front by deviatoric stresses set up by the state of uniaxial strain; generation of these dislocations relaxes deviatoric stresses.

- These dislocations move short distances at subsonic speeds.

- New dislocation interfaces are generated as the shock wave propagates through the material.

- M.A. Meyers, Dynamic Behavior of Materials, (Wiley, New York, 1994) P. 405.

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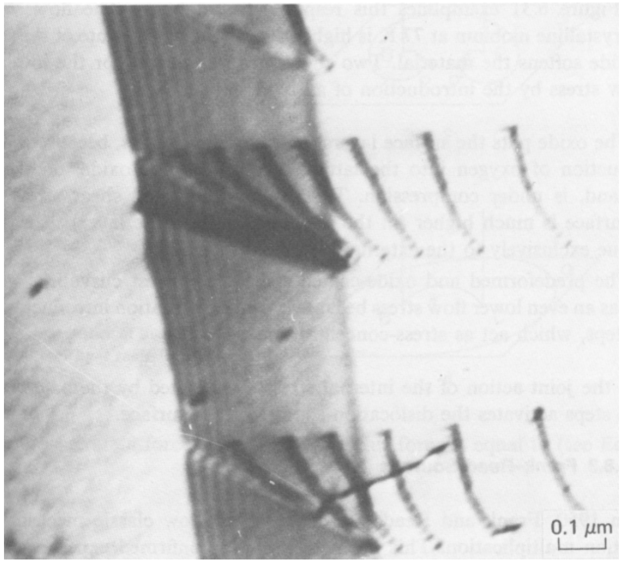
Dislocation sources

M.A. Meyers and K.K. Chawla, Mechanical Metallurgy: Principles and Applications, Prentice-Hall (1984) p. 261

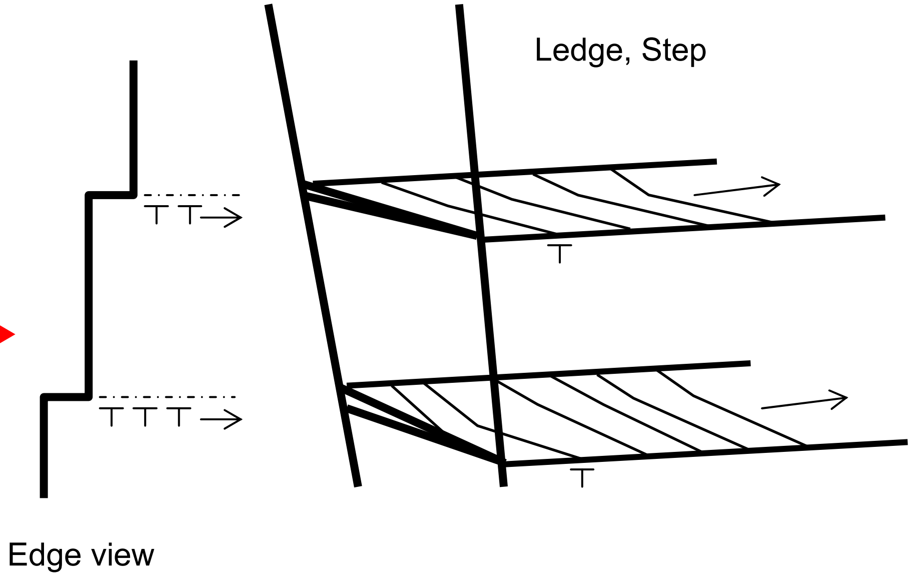
- Steps and ledges on grain boundaries are potent dislocation sources during the early stages of deformation just as small surface steps are in single crystals.

- They act as stress concentrators.

- Second phase particles and inclusions can also have the same effect (see next page).

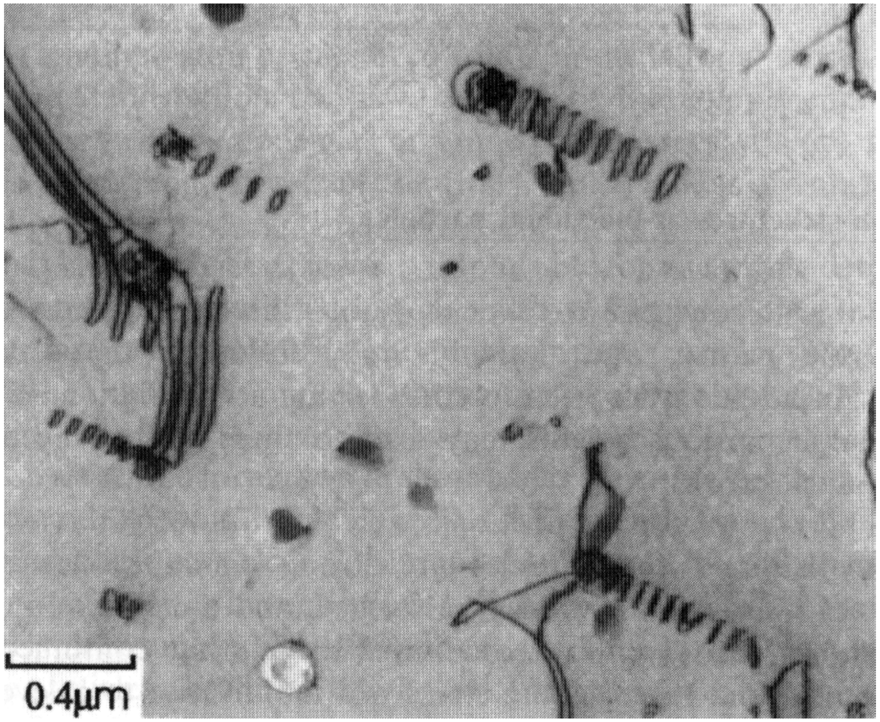


Schematic of dislocation emission from grain boundary sources.



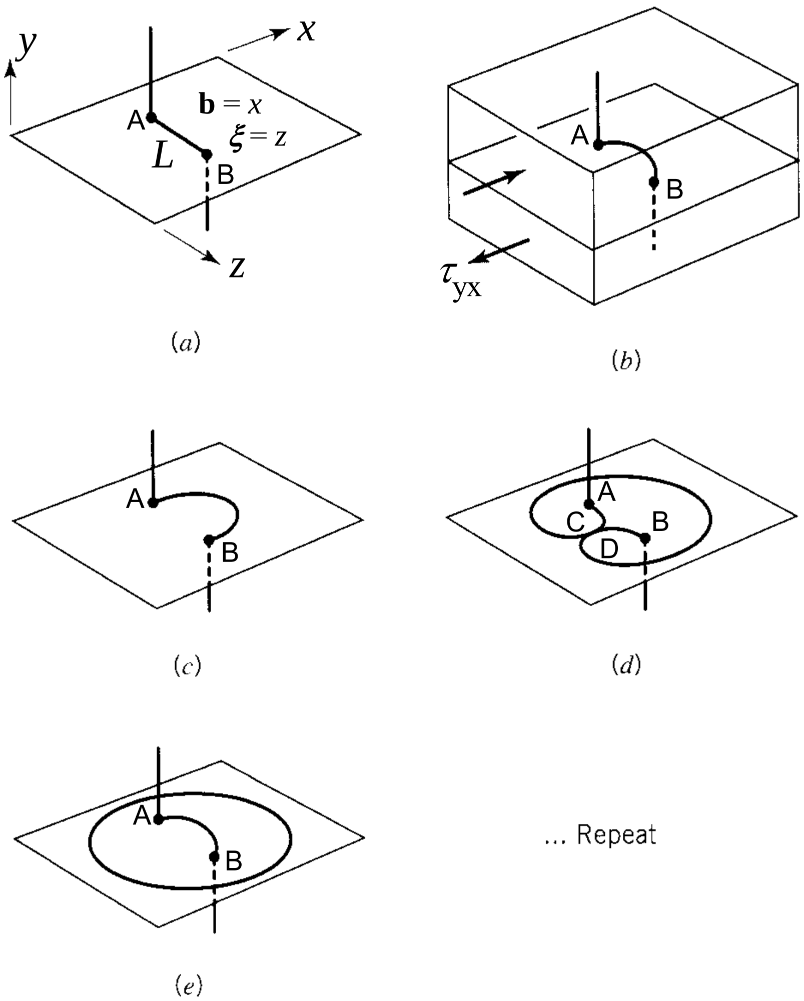
Dislocation sources

- High local stresses at second phase particles and those near second phase particles (or other defects) make nucleation easier.



◀ Primary prismatic loops at Al_2O_3 particles in an α-brass crystal

F.J. Humphreys & M. Hatherly, Recrystallization and Related Annealing Phenomena, 2nd Edition, Elsevier (2004) p. 58



Formation of a dislocation loop via the Frank-Read mechanism. Figure adapted from Allen and Thomas, Structure of Materials, (Wiley, New York, 1999) p. 306.

Frank-Read Source

a) Moving dislocation is pinned.

b) Applied shear stress causes the segment to bow out to a radius of curvature (R). Recall:

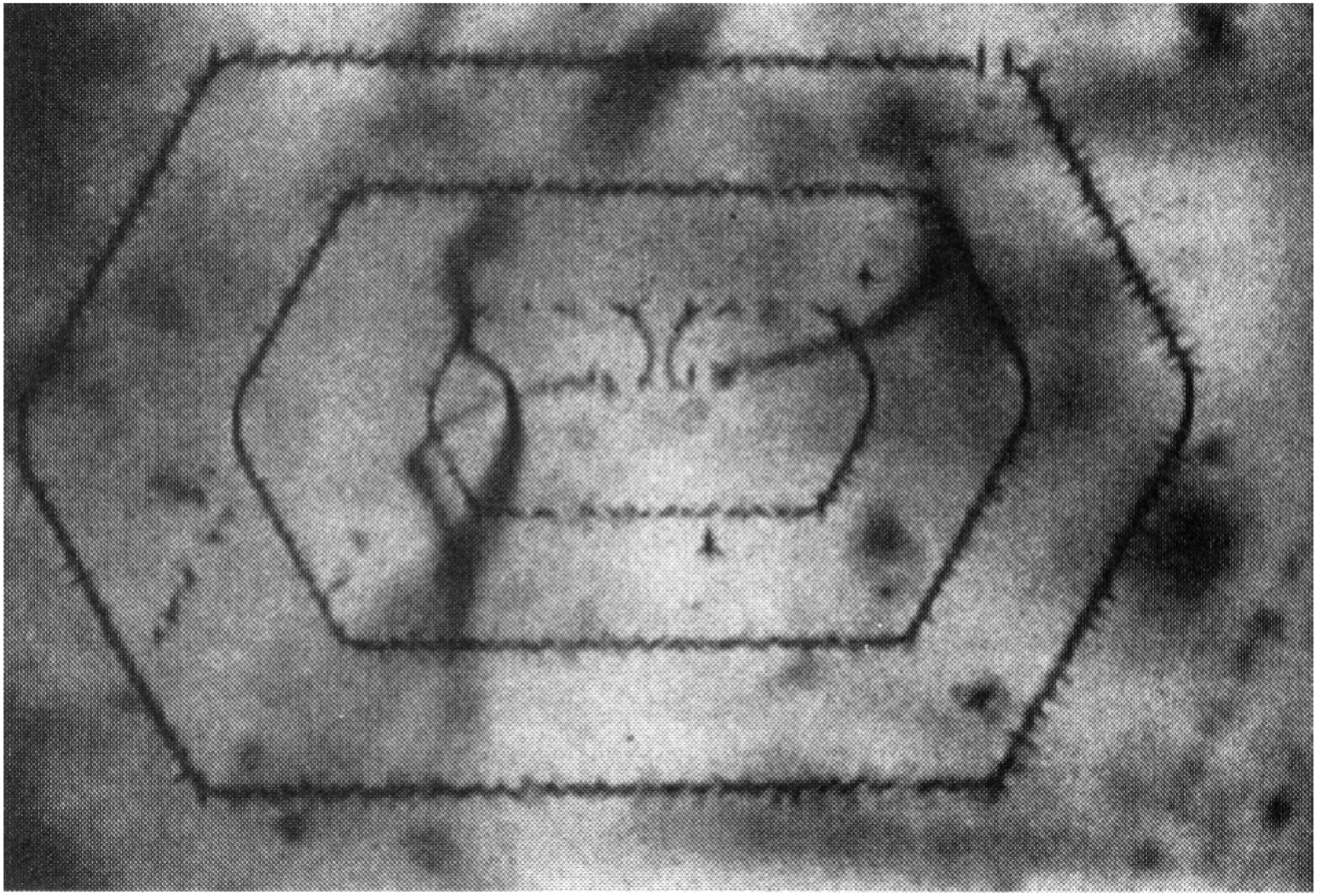
$$\tau = (\alpha G b)/R$$

c) Bowing continues until $R = L/2$. Shear stress is maximum. Loop begins to bend around upon itself. Here:

$$\tau = (2\alpha G b)/L \approx (Gb)/L$$

d) Loop expands spontaneously. This continues until points C and D which are dislocations of opposite sign annihilate each other.

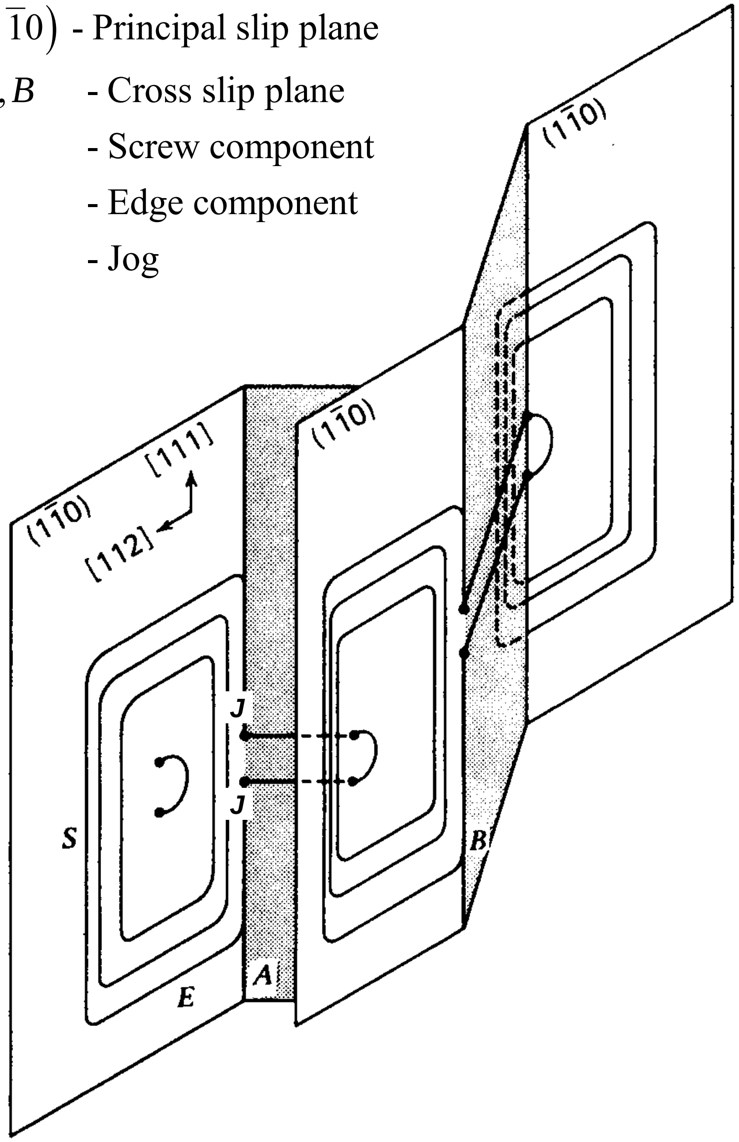
e) The loop grows while segment AB repeats the process.



Frank-Read source in a Si crystal

[Fig. 8.6 from Hull & Bacon]

$(1\bar{1}0)$ - Principal slip plane
A, B - Cross slip plane
S - Screw component
E - Edge component
J - Jog



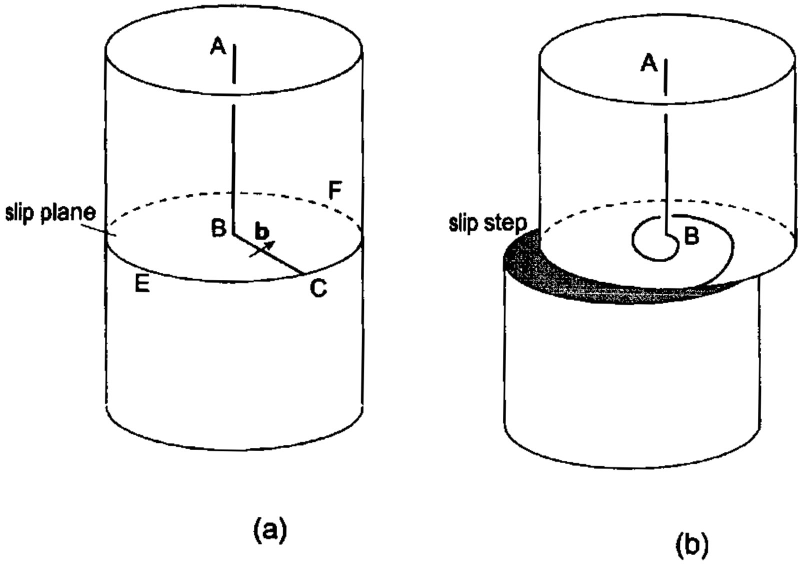
[adapted from J.R. Low and R.W. Guard, *Acta Metall.*, v. 7 (1959) pp. 171-179.]

Figure Schematic representation of dislocation multiplication via double cross-slip.

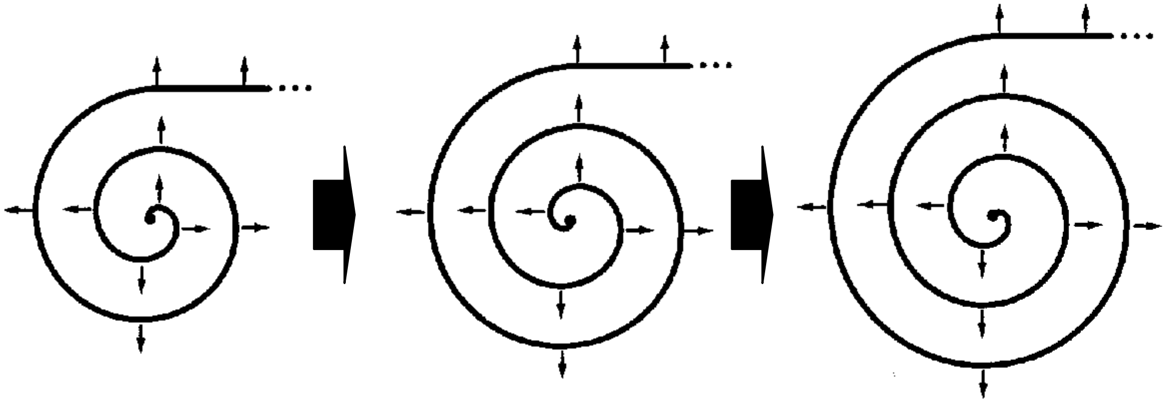
- Loops given off by a single source move through the lattice.
- Edge segments are restrained to a single slip plane.
- When a screw oriented segment of the loop encounters an unfavorable local stress, it can move off onto another plane until it reaches a position where the local stress allows it to move on a plane parallel to the original one.
- The segment lying on the new slip plane can now operate as a Frank-Read source and generate new loops.
- The process repeats.

Other types of Frank-Read sources

Figure 8.4 Single ended Frank-Read source. (a) Dislocation lying partly in a slip plane CEF. (b) Formation of a slip step and spiral dislocation by rotation of BC about B.



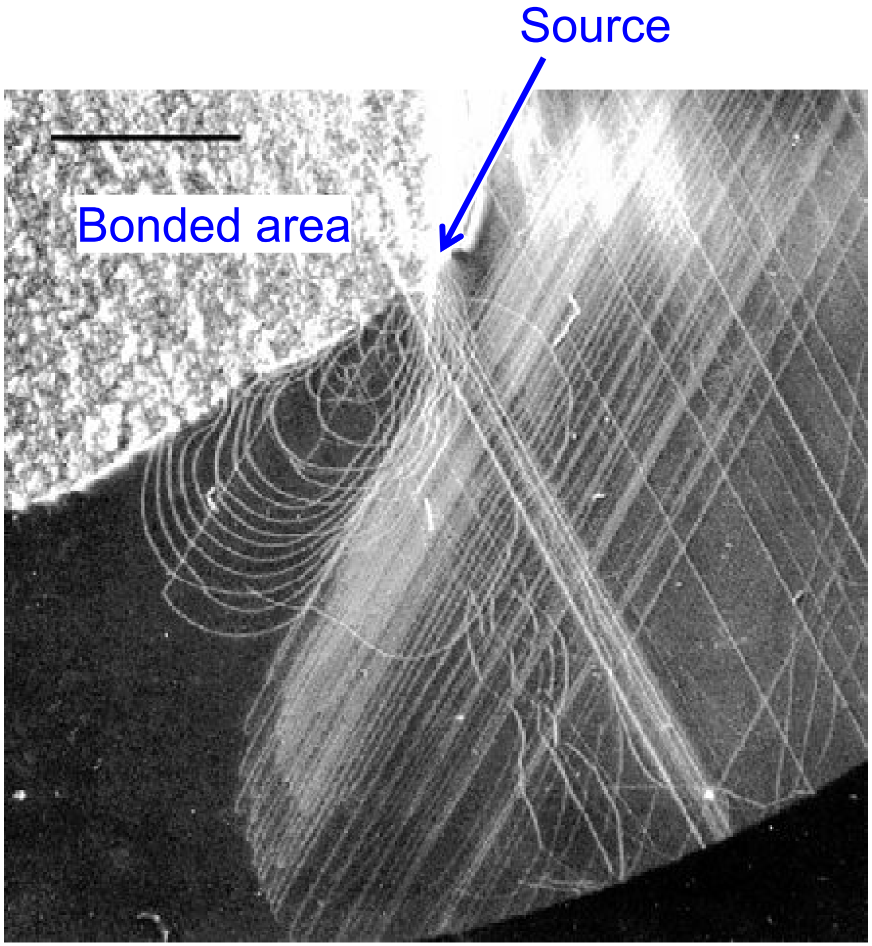
[Hull & Bacon]



Spiral propagation of a dislocation

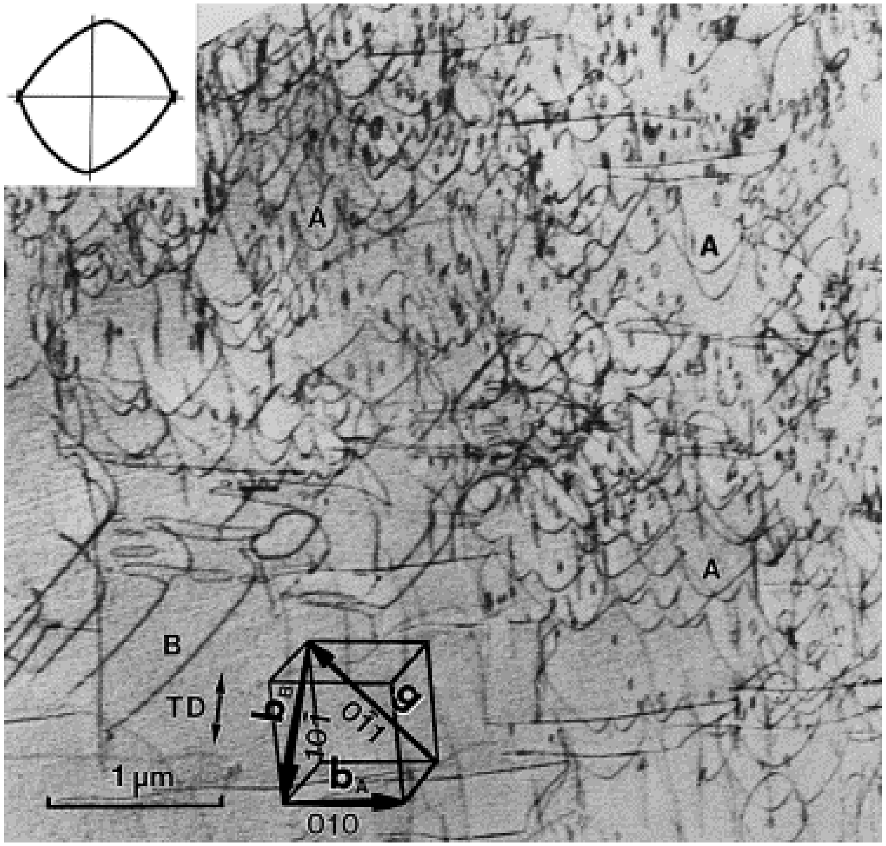
[adapted from Fig. 6.20, Roesler, Harders, & Baeker]

Images of Actual Dislocation Sources



Dislocation emission at an interface between bonded Si wafers

http://www.tf.uni-kiel.de/matwis/amat/def_en/kap_5/illustr/x_topo_constantin_mittel.jpg

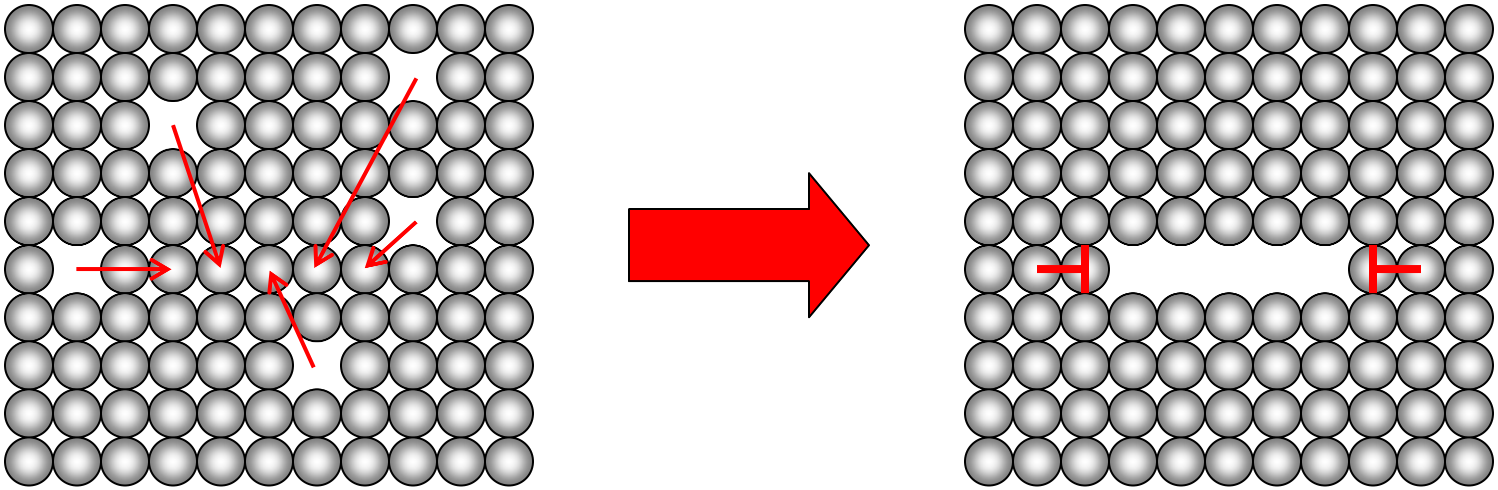


Numerous F-R sources in NiAl

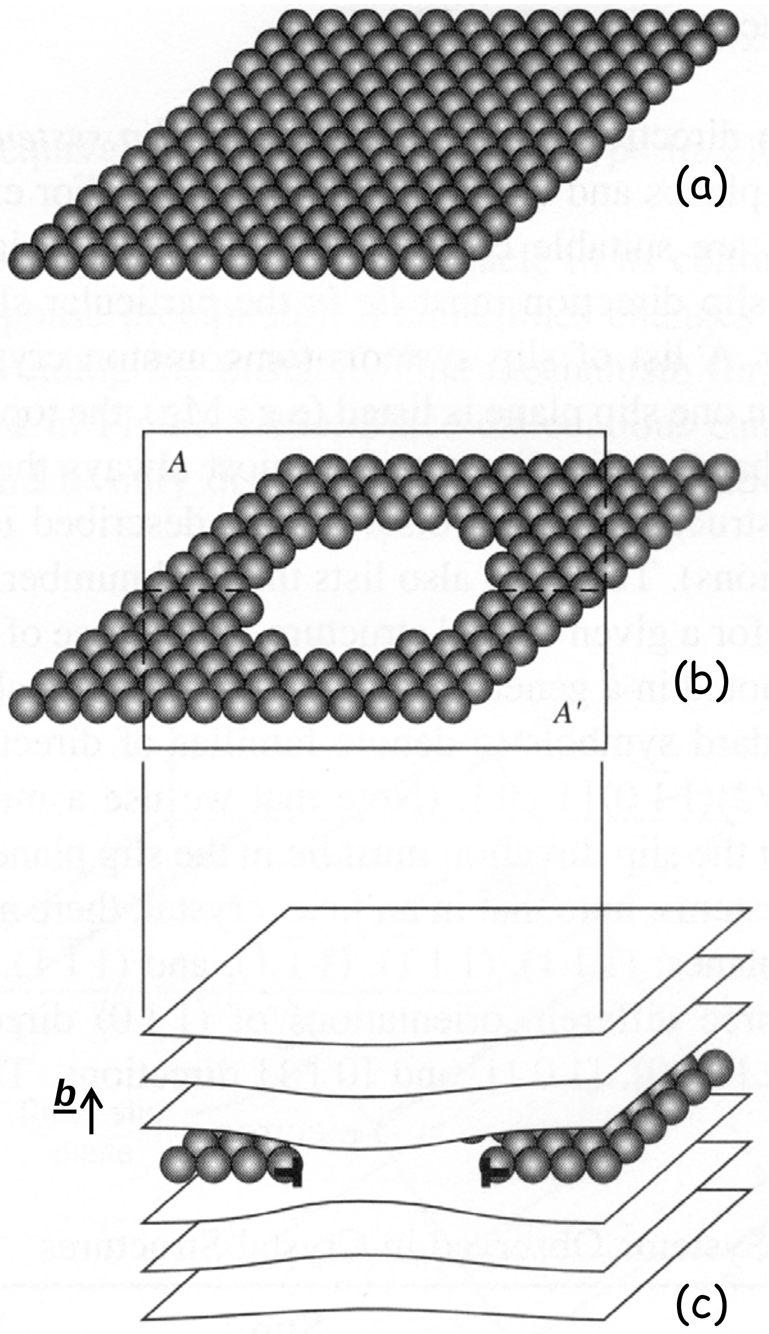
[Intermetallics, v.7 (1999) p.455-466]

Dislocation sources

- Dislocations can also form by the collapse or aggregation of vacancies (or interstitials) into discs or prismatic loops.



A 3-D perspective drawing is provided on the next viewgraph



Plane of atoms

Plane with disk of
vacancies

Stack of planes
showing half of
Frank loop

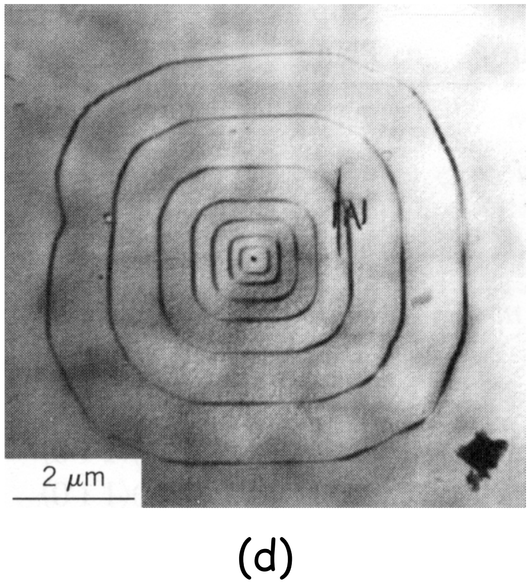
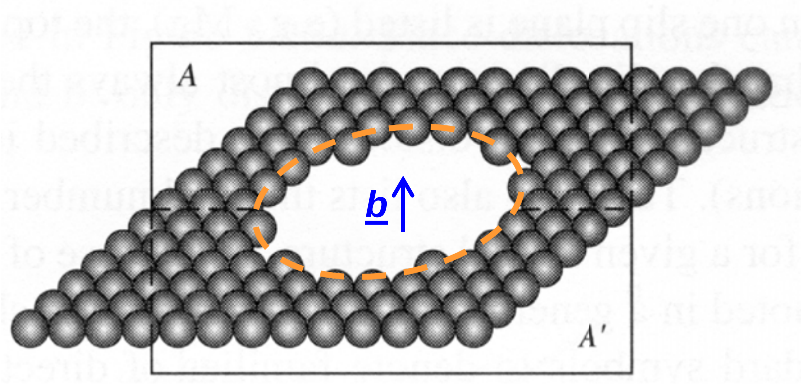


Figure 5.25 Schematic illustration of a Frank dislocation loop in single a crystal. (a) Single plane of atoms. (b) A small number of atoms has been removed from the center of the plane, and its relation to the normal section plane AA´ is indicated. (c) Stack of planes with defective plane from (b) inserted, as viewed looking onto section plane AA´. Except for the plane of the imperfection, planes are illustrated schematically. Half of the dislocation loop is seen in (c). (d) TEM image of annealed an NiAl single crystal showing unusually large concentric Frank dislocation loops formed by vacancy condensation.

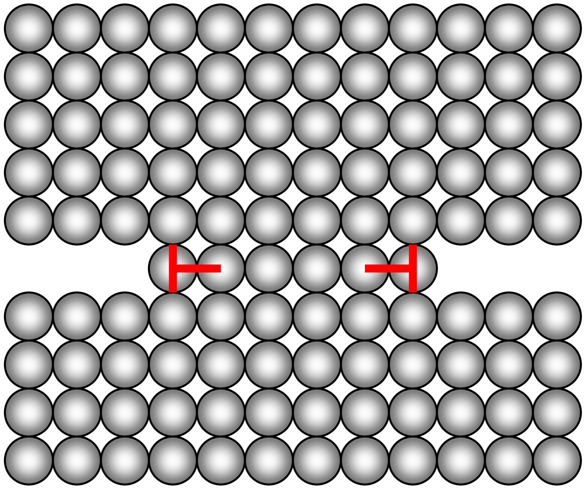
[From S.M. Allen and E.L. Thomas, The Structure of Materials, Wiley (1999) p. 291.]

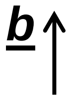
From S.M. Allen and E.L. Thomas, The Structure of Materials, Wiley (1999) p. 291.



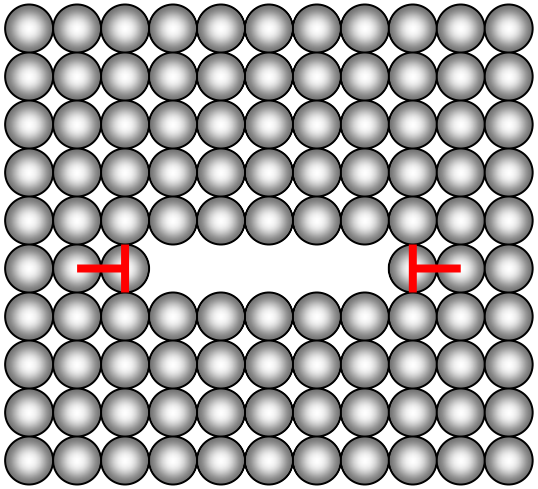
A prismatic (vacancy) loop. The Burgers vector is orthogonal to the line direction around the entire loop. The loop consists of the vacant circle of atoms within the crystal.

Prismatic loop (Interstitial)





Prismatic loop (Vacancy)

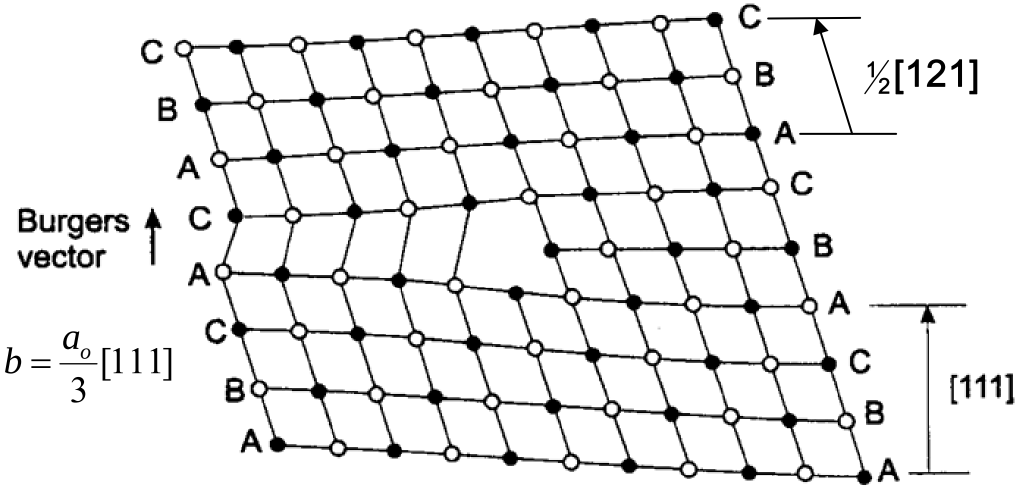


These dislocation loops are restricted to glide parallel to their Burgers vectors.
Loop expansion or contraction requires climb.

Frank Partial dislocations in FCC crystals

- Formed by inserting or removing one close-packed {111} layer of atoms. This results in either an intrinsic or an extrinsic stacking fault.

- This results in an edge dislocation with a Burgers vector is normal to the {111} plane of the fault. This dislocation is sessile.



$$b = \frac{a_o}{3}[111]$$

Formation of a 1/3[111] Frank partial dislocation by removal of a close-packed layer of atoms. [Figure adapted from Hull & Bacon, Introduction to Dislocations, 4th Edition, (Butterworth-Heinemann, Oxford, 2001) p. 92].

Dislocations generation during crystal growth

- Near the melting point, small stresses are required to cause plastic deformation. Such stresses arise for a variety of reasons. A few are listed below.
  - Thermal stresses,
  - Constitutional stresses,
  - Supersaturation of vacancies.
- These mechanisms, and others, are detailed in P. Haasen, Physical Metallurgy, 3rd Edition, (Cambridge University Press, Cambridge, 1996) p. 63.

B*

Dislocations generation during crystal growth

- Gradients in composition and/or temperature can result in dendrite misalignment during solidification and growth. This can result in dislocation networks and grain boundaries.*

- Similar misalignments can occur between growing islands in thin films resulting in dislocations and/or grain boundaries.

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* P. Haasen, Physical Metallurgy, 3rd Edition, (Cambridge University Press, Cambridge, 1996) p. 63.

Dislocations at interfaces between coherent/epitaxial phases

- Slight variances in lattice parameter can lead dislocations during epitaxial growth.

- There is a critical thickness that must be reached to form misfit dislocations.

- Misfit dislocations reduce the total strain energy in a system.

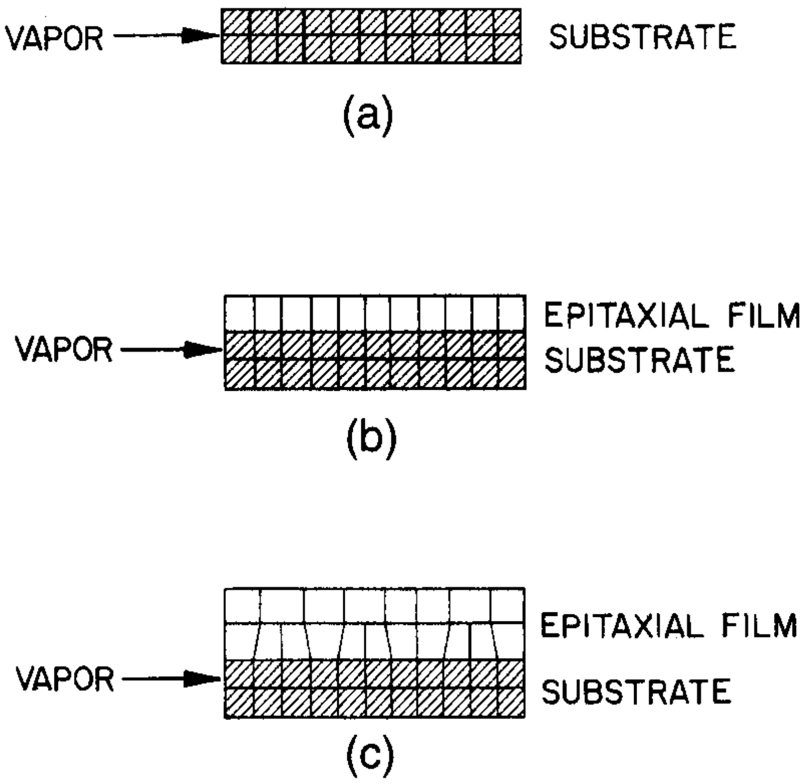


Figure 4.51 Epitaxial growth of a thin film. (a) Substrate. (b) Start of epitaxial growth. (c) Formation of "misfit" dislocations at substrate/film interface. [Figure copied from Meyers and Chawla, 1st ed.].

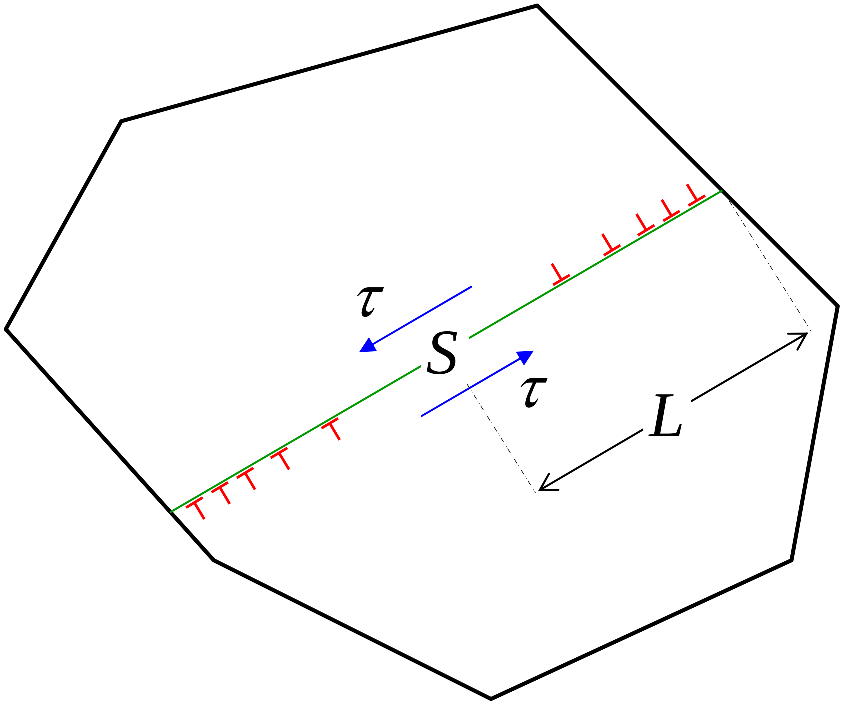
B*

Dislocation Pileups

- Dislocations moving on slip planes often pile up at barriers:
  - Grain boundaries
  - Second phases
  - Sessile dislocations
  - Etc…

- Lead dislocation is acted on by applied shear stress and interaction forces (i.e., back stress) from other dislocations.

- # dislocations in pileup is:



$$\tau(\text{lead dislocation}) \cong n\tau$$

$$n = \frac{k\pi\tau L}{Gb} \quad \text{or} \quad n = \frac{k\pi\tau D}{4Gb}$$

[For $\perp$ source in center of grain]

where $k = 1$ for screw dislocations and $(1\text{-}\nu)$ for edge dislocations.

Dislocation pileups – cont'd

- Pileup at a barrier causes a stress concentration at the lead dislocation.

- As dislocation density increases in the pileup, the stress on the dislocation can become high enough to cause yielding on the other side of the barrier or to nucleate a crack at the barrier.

- Pileups produce a back stress that acts to oppose movement of additional dislocations along the slip plane in the slip direction.

- The number of dislocations than an obstacle can support depends on:
  - Type of barrier
  - Orientation relationship between slip plane and barrier
  - Material
  - Temperature

Dislocation pileups – cont'd

- Pileup can be overcome by:
  - Cross-slip (screw dislocations)
  - Climb (illustrated below)
  - Generation of cracks.

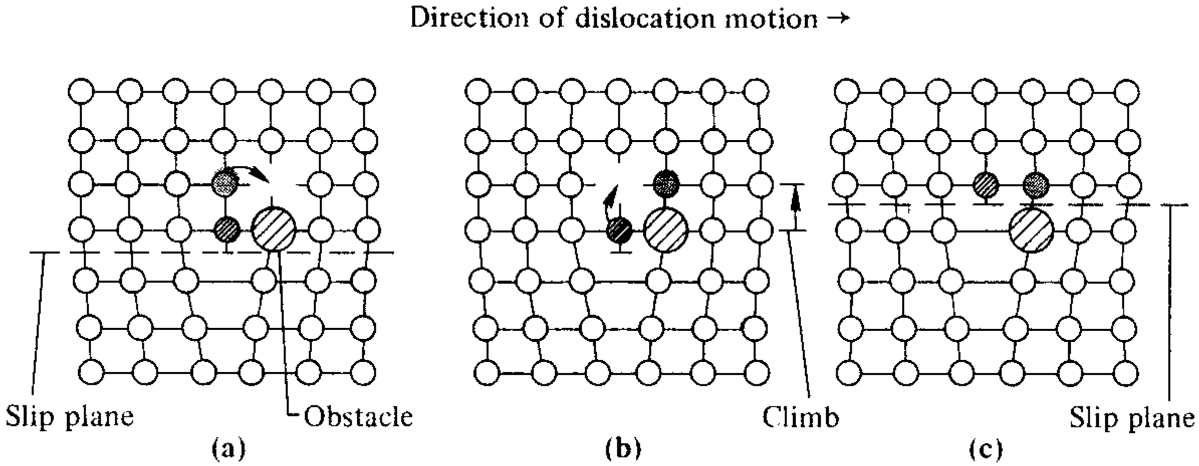


Figure from Eisenstadt, Introduction to Mechanical Properties of Materials, (Macmillan, New York, 1971) p.250

Dislocation-Point Defect Interaction

- Vacancies and isolated solute atoms distort the crystal lattice and <u>can</u> interact with dislocations.
- Strain fields surrounding point defects are spherically symmetric. They usually do not influence the motion of screw dislocations.
- Exceptions: interstitial carbon or nitrogen in BCC iron.
- Combined operation of a number of dislocation sources and dislocation-defect interactions are the basis for work/strain hardening.

Dislocation-Point Defect Interaction – cont’d

- Differences in elastic modulus between solutes and the lattice can lead to interactions with dislocations.
- Point defects that are elastically softer than the matrix are attracted to the dislocation line and visa versa.
- An increase in point defect content around a dislocation is called an impurity cloud/atmosphere.
- Condensation of impurity atmospheres on dislocation lines is one cause for upper yield points, strain aging and solid solution strengthening.