ME 477/677

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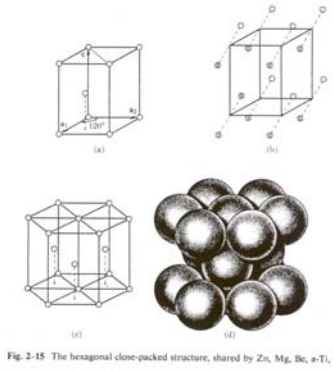
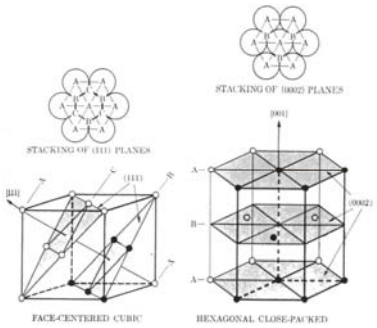


Fig. 2-15 The hexagonal close-packed structure, shared by Zn, Mg, Be, α-Ti, etc.

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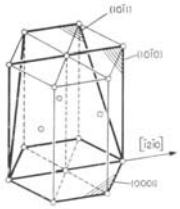
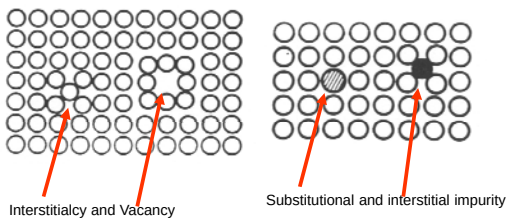
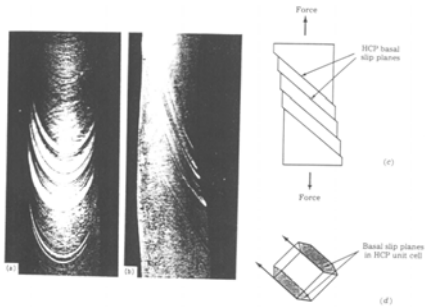


FIG. 6.3. Planes in an hexagonal lattice with a common [1210] direction.

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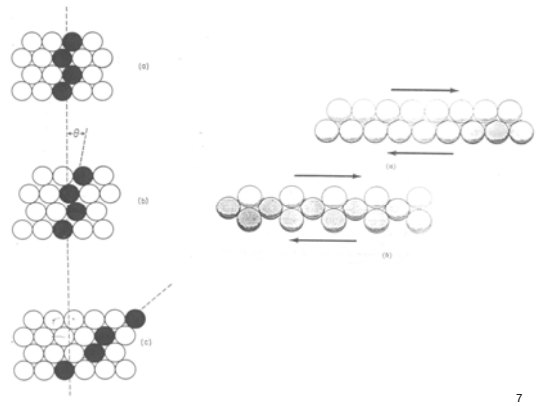


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Deformation of a Zinc single crystal

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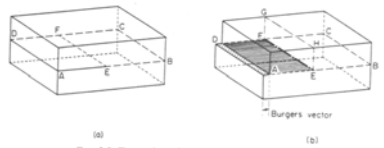


FIG. 3.3. Formation of a pure edge dislocation *FE*.

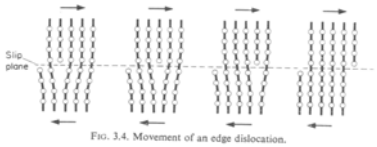


FIG. 3.4. Movement of an edge dislocation.

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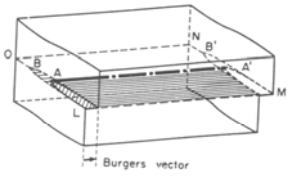


FIG. 3.5. Formation and movement of a pure screw dislocation AA' to BB' by slip.

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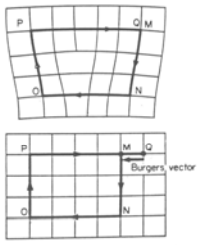


FIG. 1.16. (a) Burgers circuit round an edge dislocation, (b) the same circuit in a perfect crystal; the closure failure is the Burgers vector.

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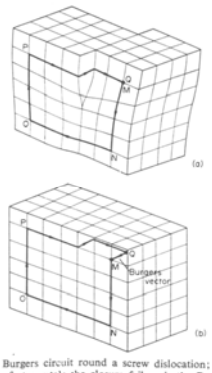


FIG. 1.17. (a) Burgers circuit round a screw dislocation; (b) the same circuit in a perfect crystal; the closure failure is the Burgers vector.

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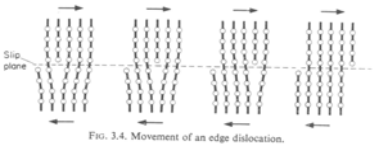


FIG. 3.4. Movement of an edge dislocation.

The *"Peierls-Nabarro force"*

Shear stress required to move a dislocation

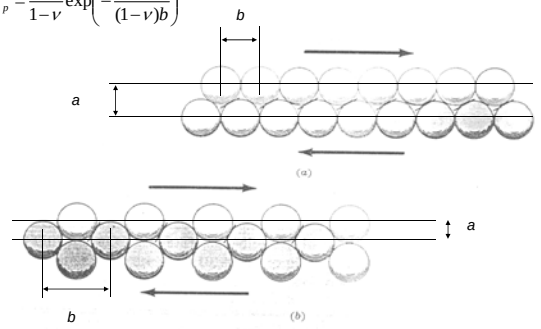
For an edge dislocation

$$\tau_p = \frac{2G}{1-\nu}\exp\left(-\frac{2\pi a}{(1-\nu)b}\right)$$

$$= \frac{2G}{1-\nu}\exp\left(-\frac{2\pi w}{b}\right)$$

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$$\tau_p = \frac{2G}{1-\nu}\exp\left(-\frac{2\pi a}{(1-\nu)b}\right)$$



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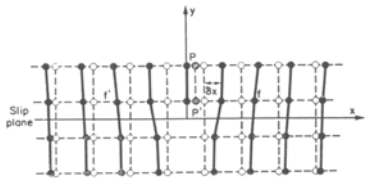


FIG. 10.2. Displacement of atoms at an edge dislocation. Open and full circles represent the atom positions before and after the extra half plane is inserted.

$$\tau_p = \frac{2G}{1-\nu}\exp\left(-\frac{2\pi w}{b}\right)$$

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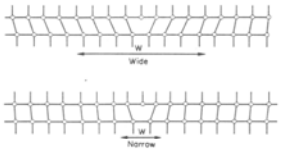


FIG. 10.3. Wide and narrow edge dislocations. (From COTTRELL (1957), *The Properties of Materials at High Rates of Strain*, Inst. Mech. Eng., London.)

$$\tau_p = \frac{2G}{1-\nu}\exp\left(-\frac{2\pi w}{b}\right)$$

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Slip Systems

- Since the stress required to move dislocations is lowest on a close-packed plane and in the close packed direction, the combination of a closed packed plane and a close packed direction on that plane is called a slip system.
- The fcc, hcp, and bcc crystal structures have different combinations of close-packed planes and directions.
- fcc:
  - Atoms touch on {111}
  - Atoms touch on <110>
- hcp:
  - Atoms touch on basal plane (0001) when (c/a) ≥1.633
    - The prism planes and the first order pyramidal planes may also become slip planes if (c/a) < ideal
  - Atoms touch along $\langle 11\bar{2}0 \rangle$
- bcc:
  - Atoms touch along <111> direction, but there is no plane of dense packing as in the other two crystal structures
  - Three families of planes {110}, {112} and {123} have similar atomic densities, and can all act as slip planes

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Slip Systems for Face-Centered Cubic, Body-Centered Cubic, and Hexagonal Close-Packed Metals

Metals	*Slip Plane*	*Slip Direction*	*Number of Slip Systems*
	Face-Centered Cubic		
Cu, Al, Ni, Ag, Au	{111}	$\langle 1\bar{1}0 \rangle$	12
	Body-Centered Cubic		
α-Fe, W, Mo	{110}	$\langle \bar{1}11 \rangle$	12
α-Fe, W	{211}	$\langle \bar{1}11 \rangle$	12
α-Fe, K	{321}	$\langle \bar{1}11 \rangle$	24
	Hexagonal Close-Packed		
Cd, Zn, Mg, Ti, Be	{0001}	$\langle 11\bar{2}0 \rangle$	3
Ti, Mg, Zr	$\{10\bar{1}0\}$	$\langle 11\bar{2}0 \rangle$	3
Ti, Mg	$\{10\bar{1}1\}$	$\langle 11\bar{2}0 \rangle$	6

For Ti, Mg, Zr, (c/a) < ideal (1.633). Therefore, there are additional planes with a high density of atoms
Prism plane: $\{10\bar{1}0\}$ and Pyramidal plane: $\{10\bar{1}1\}$

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Table 4-4 Room-temperature slip systems and critical resolved shear stress for metal single crystals

Metal	Crystal structure	Purity, %	Slip plane	Slip direction	Critical shear stress, MPa
Zn	hcp	99.999	(0001)	$[11\bar{2}0]$	0.18
Mg	hcp	99.996	(0001)	$[11\bar{2}0]$	0.77
Cd	hcp	99.996	(0001)	$[11\bar{2}0]$	0.58
Ti	hcp	99.99	(1010)	$[11\bar{2}0]$	13.7
		99.9	(1010)	$[11\bar{2}0]$	90.1
Ag	fcc	99.99	(111)	[110]	0.48
		99.97	(111)	[110]	0.73
		99.93	(111)	[110]	1.3
Cu	fcc	99.999	(111)	[110]	0.65
		99.98	(111)	[110]	0.94
Ni	fcc	99.8	(111)	[110]	5.7
Fe	bcc	99.96	(110)	[111]	27.5
			(112)		
			(123)		
Mo	bcc	...	(110)	[111]	49.0

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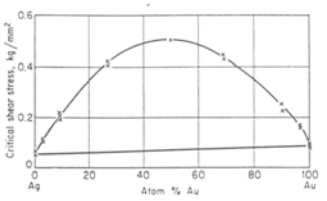


Figure 4-19 Variation of critical resolved shear stress with composition in silver-gold-alloy single crystals. (From G. Sachs and J. Weerts, Z. Phys., vol. 62, p.473, 1930.)

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- Single dislocation causes a step of b on each side
- Shear strain in a crystal of height h is

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

- If there are multiple dislocations that pass part way through the crystal, then the associated strain depends on how many dislocations are present and how far each dislocation has moved
- If *L* is the length of the crystal, and a typical dislocation has moved a distance x_i through the crystal then
- Displacement of the top due to one dislocation is

$$\delta_i = \frac{x_i}{L} b$$

- Displacement of the top due to all dislocations is

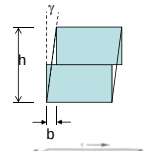
$$\Delta = \sum \delta_i$$

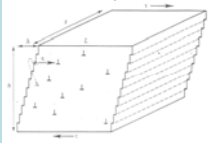
- The corresponding strain is

$$\gamma = \frac{\Delta}{h} = \frac{b}{hL} \sum x_i = \frac{bN\bar{x}}{hL}$$

or

$$\gamma = \rho b \bar{x}$$



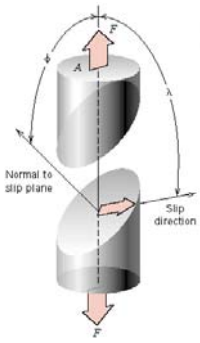


N = Total number of dislocations crossing front plane

Average displacement of each dislocation

Density of dislocations

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Geometrical relationships between the tensile axis, slip plane, and slip direction used in calculating the resolved shear stress for a single crystal.

When you have a large number of slip systems, the one with the highest resolved shear stress is the one on which plastic deformation (movement of dislocations) occurs.

Resolved shear force $= F \cos\lambda$

Area of plane $= \dfrac{A}{\cos\phi}$

Resolved shear stress $= \tau_R = \dfrac{F}{A}\cos\phi\cos\lambda$

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

$= m.\sigma$

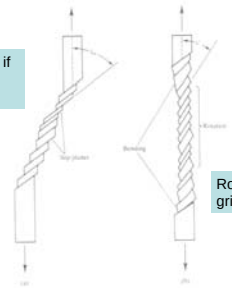
Schmid Factor $m = \cos\phi\cos\lambda$

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- The Schmid factor $\boldsymbol{m} = \cos\boldsymbol{\phi}\cos\boldsymbol{\lambda}$ is used to determine the slip system with the highest resolved shear stress
- Each slip system of a particular crystal structure, whether fcc, bcc or hcp, may have a different $\boldsymbol{\phi}$ (angle between loading direction and slip plane normal) and $\boldsymbol{\lambda}$ (angle between the loading direction and the slip direction)
- The values of $\cos\boldsymbol{\phi}$ and $\cos\boldsymbol{\lambda}$ can be determined from the vector dot product rules discussed earlier.
- The largest value of $\boldsymbol{m}$ is 0.5 when both angles are 45º
- If the ***tensile*** loading direction is perpendicular to the slip plane, all directions on the slip plane are perpendicular to the loading direction. Therefore $\boldsymbol{\lambda}$ = 90º, and $\boldsymbol{m}$ = *0.*
- If the ***tensile*** loading direction is parallel to the slip plane, $\boldsymbol{\phi}$ = *0*, and $\boldsymbol{m}$ = *0.*
- The fcc and bcc crystal structures have numerous slip system, such that for any tensile loading direction, there will be at least one slip system with a non-zero Schmid factor
- For the hcp, especially when (c/a) ≥1.633, there are only three slip systems, which all have the same slip plane. This has a very significant impact on the ductility of polycrystalline hcp metal such as Zn or Cd.

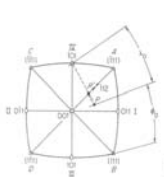
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Grips need to move if slip system is to be preserved



Rotation of the crystal if the grips are fixed

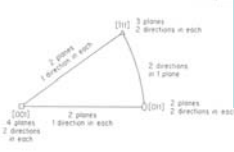
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PLASTIC DEFORMATION OF SINGLE CRYSTALS 131

Figure 4-23 Standard (001) stereographic projection for a cubic crystal.

- The slip system with the highest Schmid factor becomes active first.
- m can be determined using vector multiplication rules



- Alternatively, one can use the stereographic projection.
- Locate the loading direction within a triangle
- Locate the vertices on the far side of the adjacent triangles to identify the primary slip system

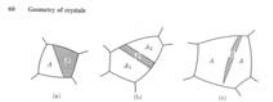
Figure 4-24 Operative slip systems along boundaries of stereographic triangle.

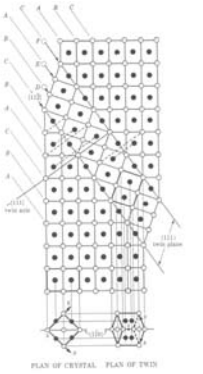
Primary slip system is B-IV or $[\bar{1}01](111)$
Secondary or conjugate slip system is C-I $[011](\bar{1}11)$

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Twinning

- A deformation mode observed when slip or movement of dislocations is not possible
- A part of the crystal re-orients to a mirror image about the "twinning" plane
- Figure shows the twinned region between regions of the original orientation
- Twins can form during annealing also



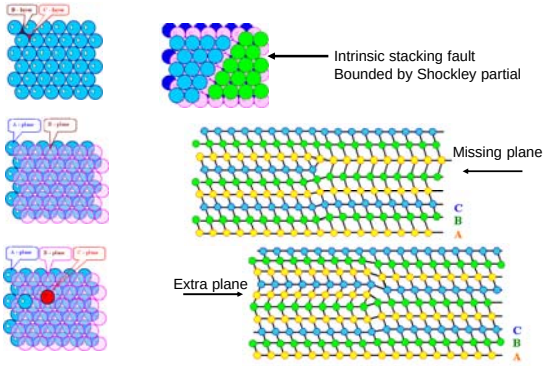


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Planar or 2-D Defects

- Grain boundaries
  - Where crystals of two different orientations meet
- Stacking faults
  - Errors in the stacking of planes within a crystal
  - Example: fcc crystal
    - Consists of stacking of close packed {111} planes in an ABCABCABC… sequence
    - If this sequence is disrupted, then there will be a region of hcp stacking (for example ABAB …)
    - This can happen:
      - During deformation if a portion of a plane (say the B plane) instead of sliding to the next B position, slides part way to a C position – this is called an intrinsic stacking fault
      - By the introduction of piece of a plane in the "wrong" stacking sequence, or removal of a piece of a plane – this is called an extrinsic stacking fault or twin fault
  - Stacking faults are bounded by "partial dislocations" (discussed later)

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Extrinsic stacking faults bounded by Frank partials

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Microstrain behavior

- Tension test – off set yield stress 0.2% plastic deformation
- Limit of elastic deformation - microscopic level
  - No movement of dislocations
  - Reversible movement of dislocations
  - Permanent plastic deformation
  - Permanent plastic deformation of a fixed amount

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Deformation of a Single crystal

- Annealed crystals contain 10^5 to 10^6 cm/cm^3 of dislocations due to errors in crystal growth
  - These dislocations may be glissile or sessile
    - Glissile – capable of moving under applied stress – close-packed plane and close-packed direction of the correct orientation
    - Sessile – does not move under applied stress – wrong orientation or the slip plane and burgers vector are not close-packed
- Deformation results from the movement of glissile dislocations which may run into other dislocations
  - Higher stress to move dislocations (strain hardening)
  - Dislocation multiplication
- Dislocations that are prevented from moving may multiply by
  - Frank-Reed source
  - Single-ended dislocation source
- Dislocation density may increase by
  - Grain boundaries emitting dislocations
  - Condensation of vacancies

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Deformation of a single crystal

- Increased dislocation density leads to further dislocation-dislocation interaction leading to a further multiplication of dislocations
- Cold worked material may have 10^{10} to 10^{12} cm/cm^3 of dislocations
- Cold working results in strain hardening, i.e., plastic deformation or movement of dislocations becomes more difficult.
  - The (111) plane of an fcc crystal like Cu has 1.5×10^{15} atoms per cm^2
  - Even with 10^{10} to 10^{12} dislocations per cm^2 only about 1 in 1,000 or 100,000 atoms is out of place – ie the crystal is 99.9% perfect or better!

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Bauschinger Effect

- When the resolved shear stress on a plane and direction reach the critical value, dislocation on that system will start moving and will continue to move until
  - They leave the crystal – leaving behind a step on the surface
  - Run into obstacles and stop
    - When this happens, the dislocations that follow will feel a back stress from the stopped dislocation
    - With continued deformation, this back pressure builds up and it becomes more difficult to cause plastic deformation – i.e. move dislocations.
- If the direction of loading is reversed after some plastic deformation, the back stress from the pile up will aid in the "reverse" plastic deformation
- Plastic deformation can now occur at a lower value of externally applied stress

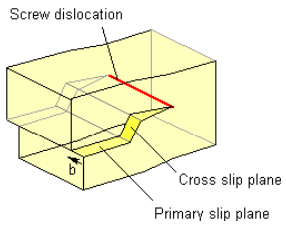
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Deformation of an fcc single crystal

- Three stages
  - Easy glide
    - Relatively few dislocations which move under low stress
    - Some increase in stress due to reorientation
    - Laminar flow
  - Linear work hardening
    - Multiple slip systems
    - Dislocation multiplication due to dislocation-dislocation interaction
    - Dislocation pile up
    - Slope independent of temperature (after correction for changes in G with temperature), purity of crystal and initial orientation.
  - Dynamical recovery
    - Stress level is high enough for dislocations to overcome barriers
      - Cross-slip of screw dislocation
      - Climb of edge dislocation
  - Orientation determines overall shape of τ-γ curve

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Cross Slip



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