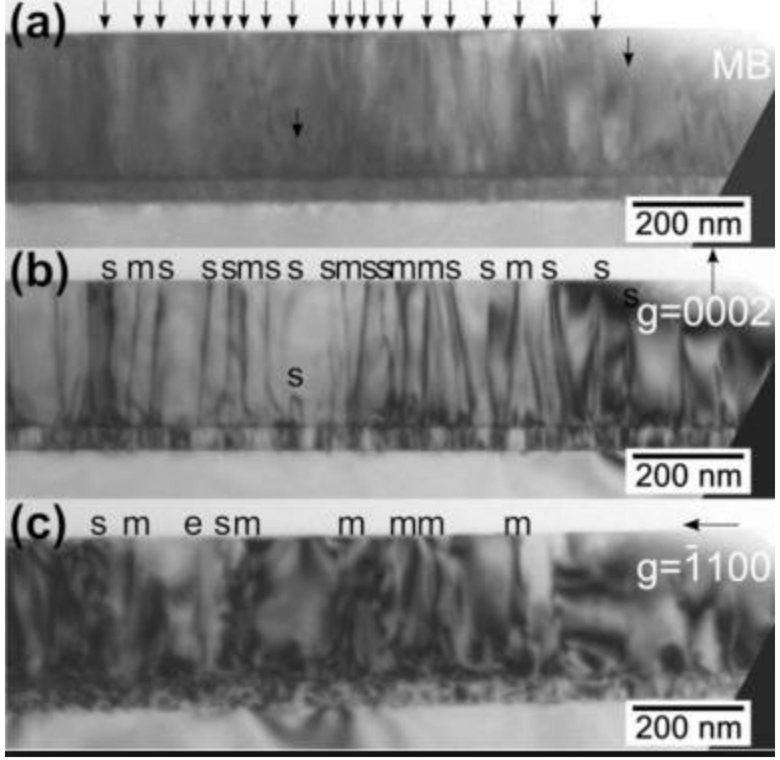
Scienza dei Materiali

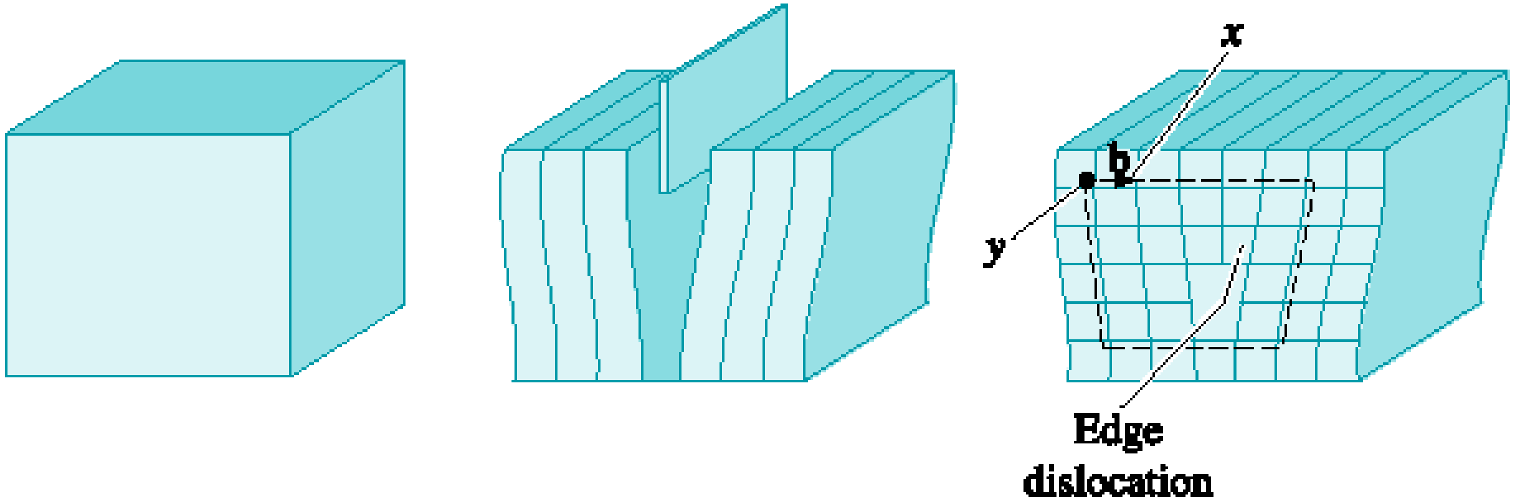
Dislocazioni

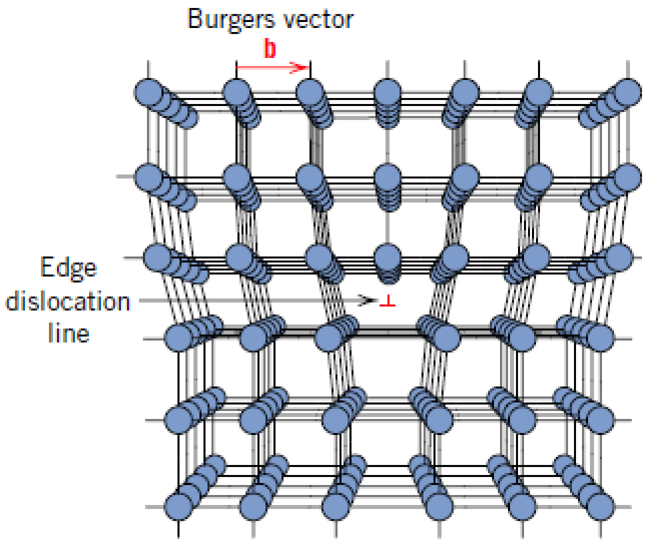
Difetti lineari (1-dimensionali): Dislocazioni

- Termodinamicamente instabili
- Imperfezioni del reticolo localizzate lungo una linea
- Termodinamicamente non stabili
- Perturbano localmente la simmetria del reticolo
- Ruolo chiave nella determinazione delle proprietà dei materiali (deformazione plastica)

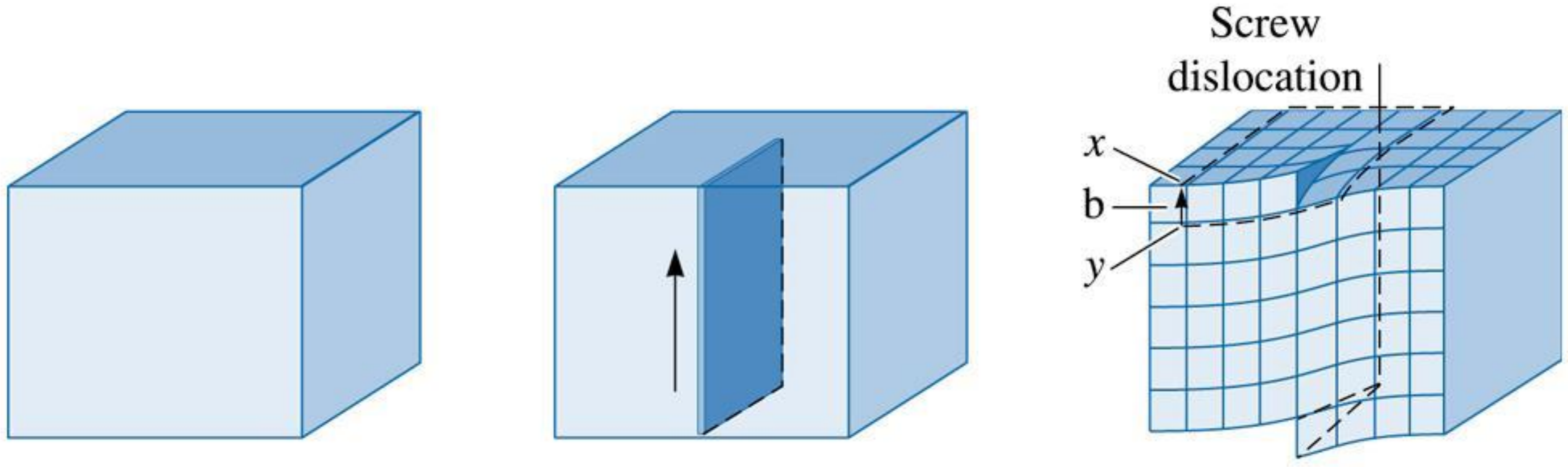


Dislocazione a spigolo (edge dislocation)

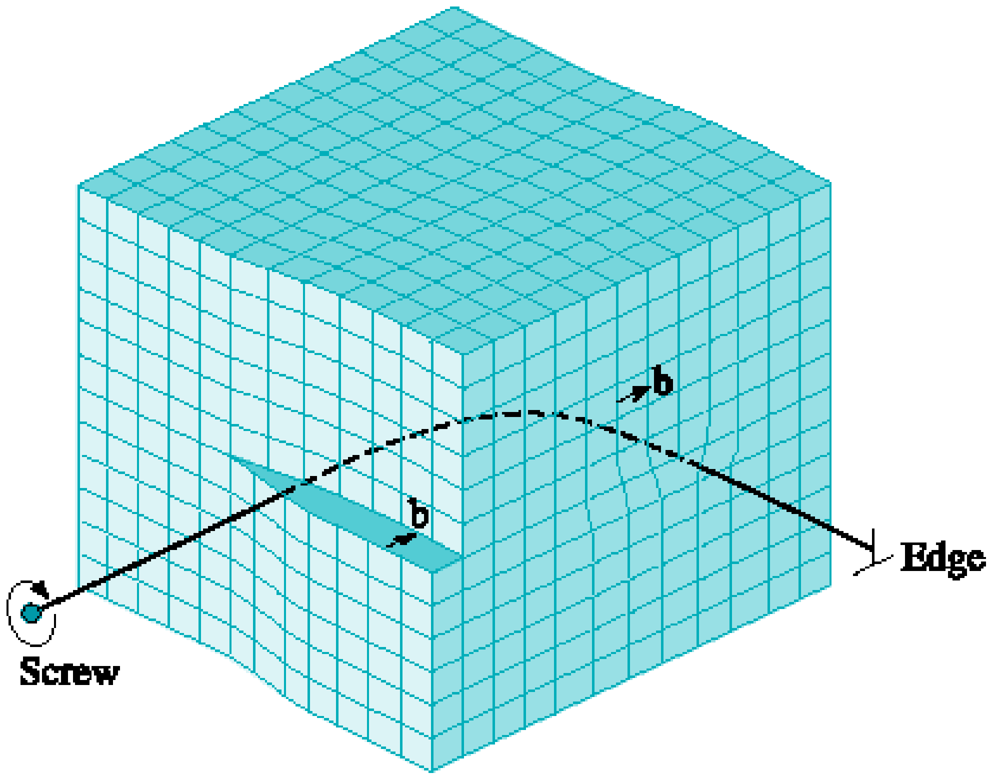




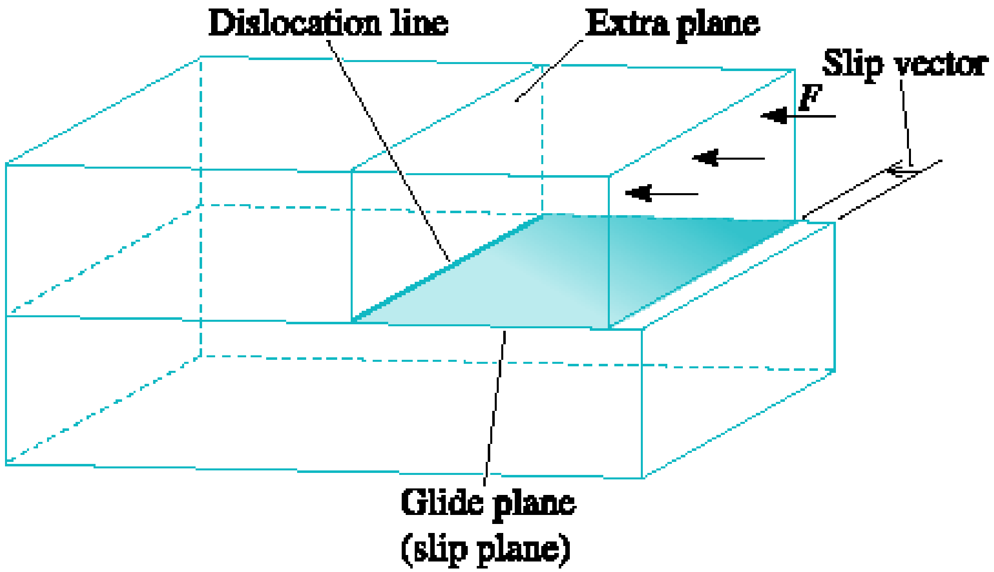
Dislocazione a vite (screw dislocation)

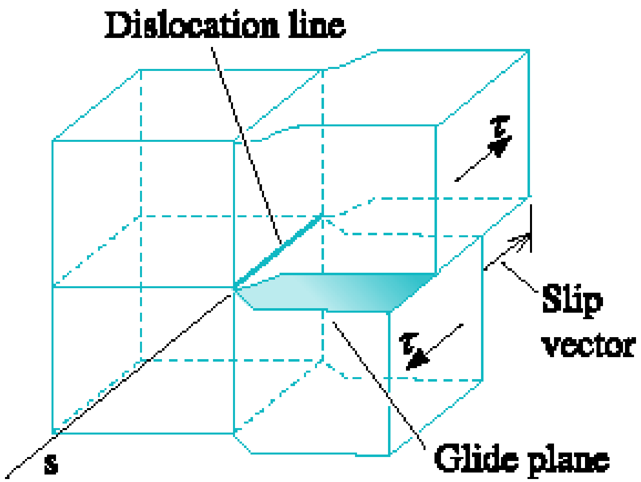


Dislocazione mista



Piano di scorrimento e vettore di Burgers



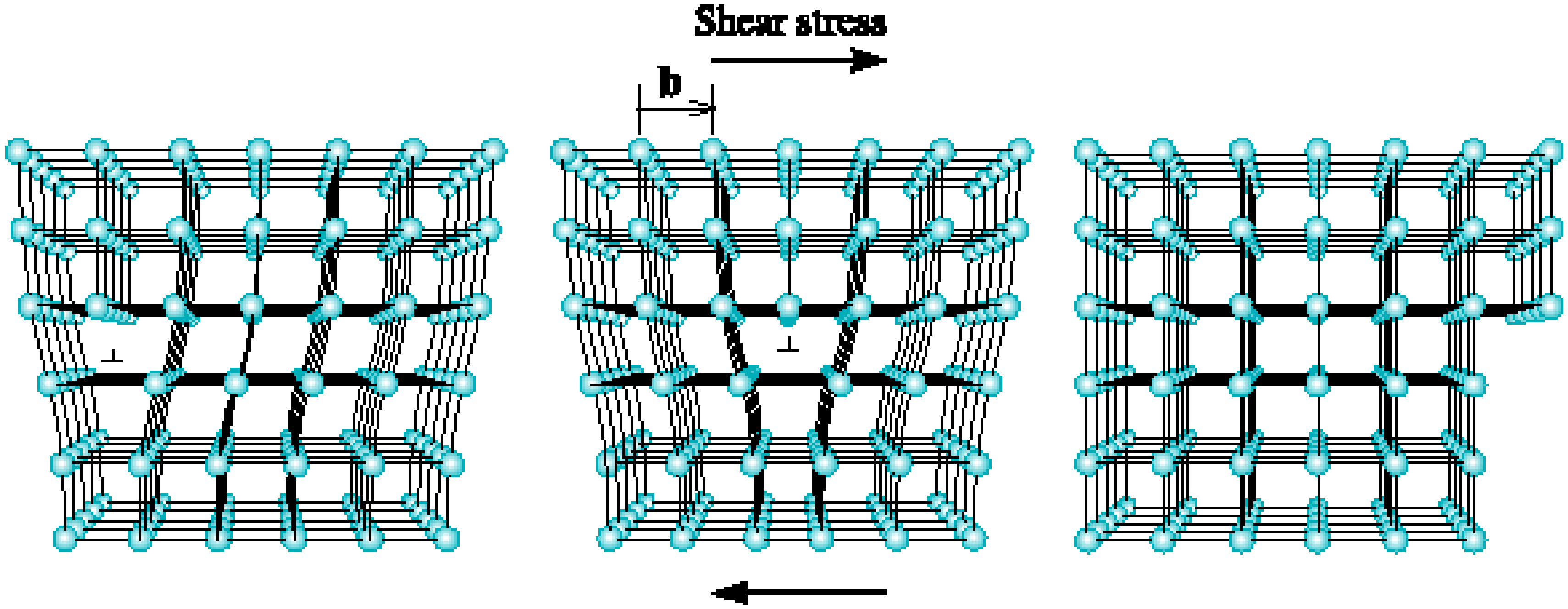


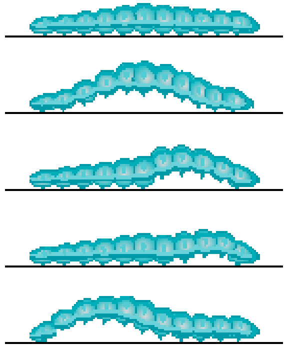
Piano di scorrimento

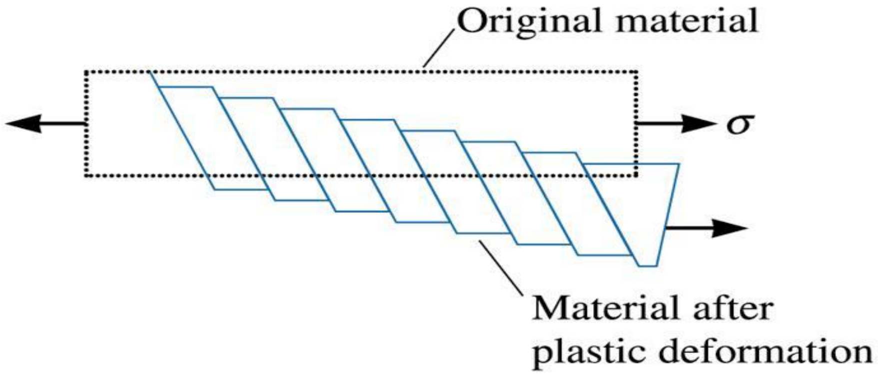
Vettore di Burgers <u>ortogonale</u> alla dislocazione

Vettore di Burgers <u>parallelo</u> alla dislocazione

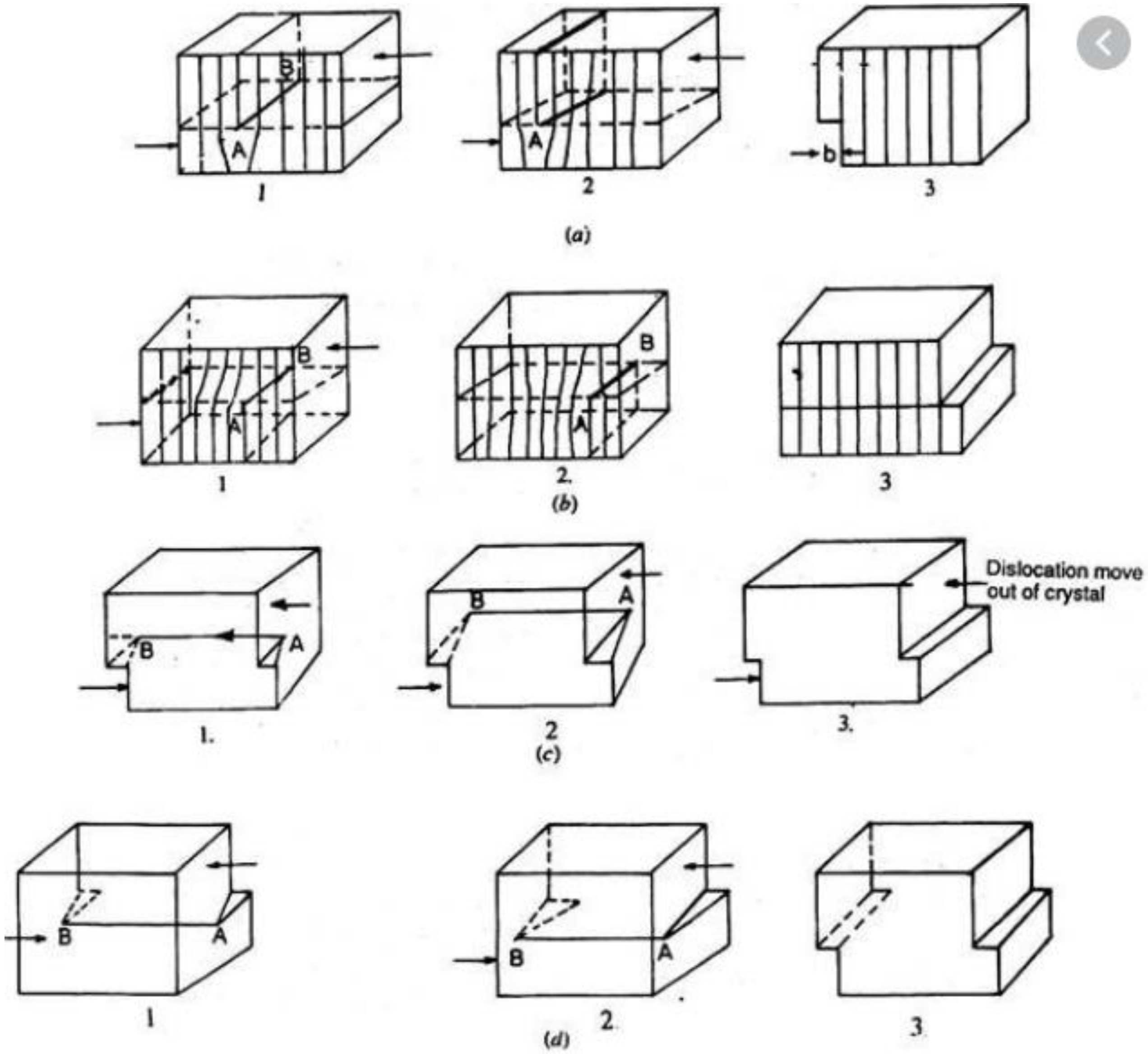
Scorrimento: Meccanismo Deformazione Plastica







Scorrimento: Meccanismo Deformazione Plastica

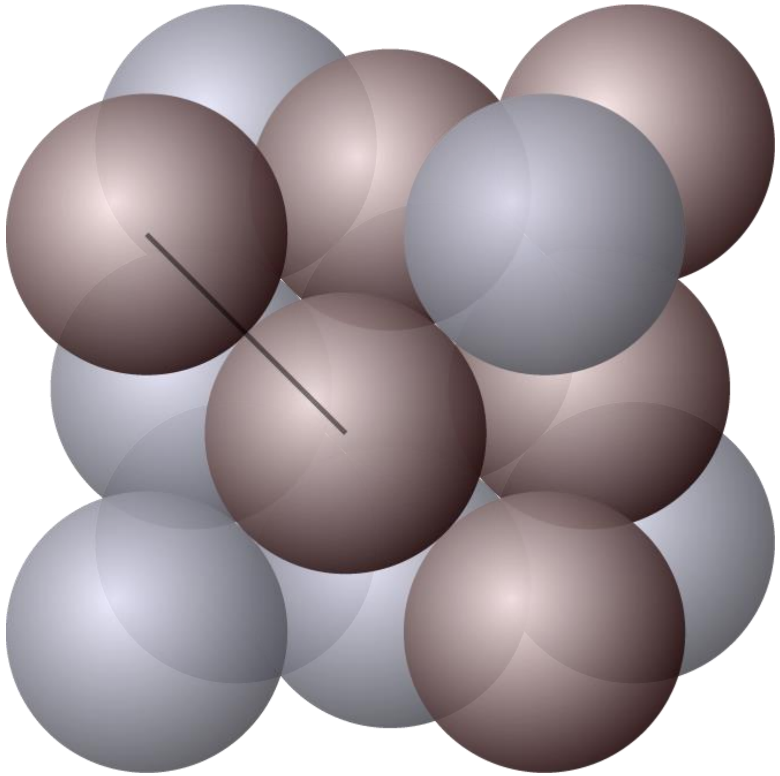


Su quali piani e in quali direzioni è più facile che avvenga lo scorrimento?

(«sistemi di scorrimento»)

- Piani ad alta densità atomica superficiale
- Direzioni ad alta densità atomica lineare
- Elevata distanza interplanare
- Piccolo vettore di scorrimento

Identificare sistemi di scorrimento: FCC



Identificare sistemi di scorrimento : BCC

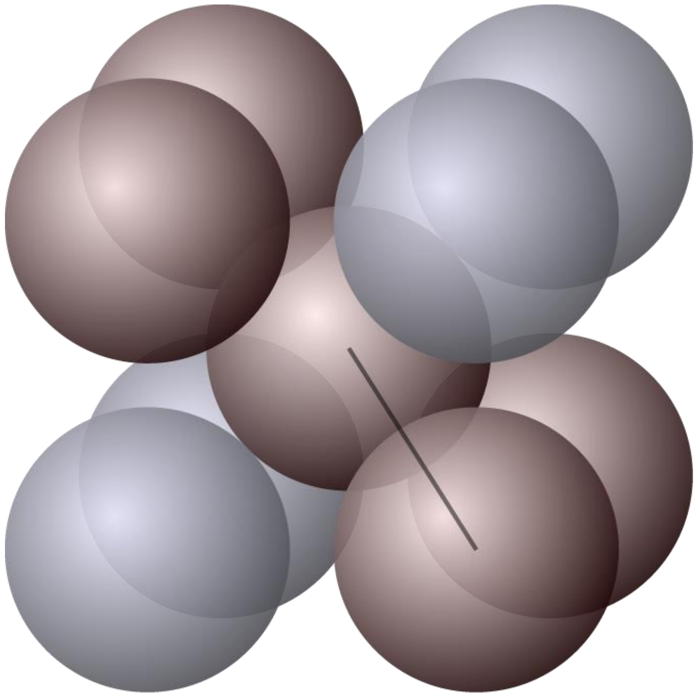


TABLE 4-1 ■ ***Slip planes and directions in metallic structures***

Crystal Structure	Slip Plane	Slip Direction
BCC metals	{110}	⟨111⟩
	{112}	
	{123}	
FCC metals	{111}	⟨110⟩
HCP metals	{0001}	⟨100⟩
	$\{11\bar{2}0\}$ See Note	⟨110⟩
	$\{10\bar{1}0\}$ See Note	or $\langle 11\bar{2}0 \rangle$
	$\{10\bar{1}1\}$ See Note	
MgO, NaCl (ionic)	{110}	⟨110⟩
Silicon (covalent)	{111}	⟨110⟩

Note: These planes are active in some metals and alloys or at elevated temperatures.

- Dislocation - A line imperfection in a crystalline material.
- Screw dislocation - A dislocation produced by skewing a crystal so that one atomic plane produces a spiral ramp about the dislocation.
- Edge dislocation - A dislocation introduced into the crystal by adding an ‘‘extra half plane’’ of atoms.
- Mixed dislocation - A dislocation that contains partly edge components and partly screw components.
- Slip - Deformation of a metallic material by the movement of dislocations through the crystal.

Example: Burgers Vector Calculation

Calculate the length of the Burgers vector in copper.

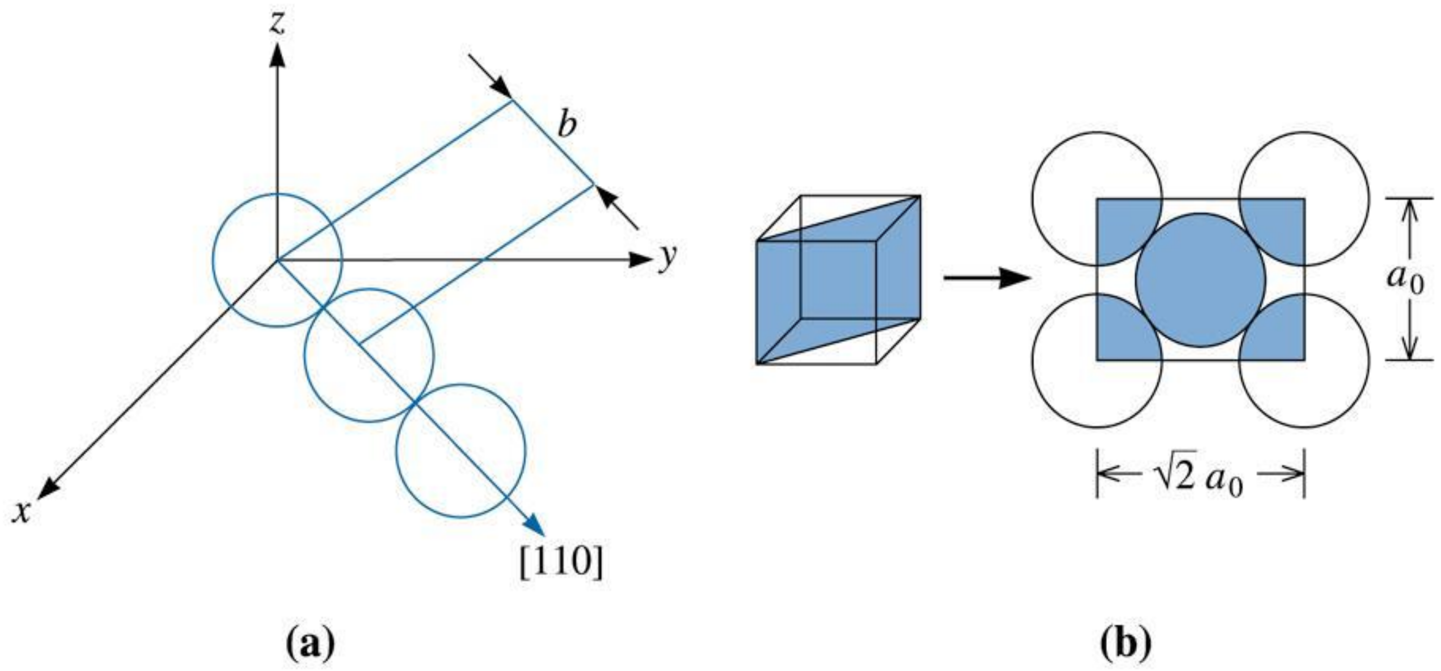


Figure 4.10 (a) Burgers vector for FCC copper. (b) The atom locations on a (110) plane in a BCC unit cell (for example 4.8 and 4.9, respectively)

Example SOLUTION

Copper has an FCC crystal structure. The lattice parameter of copper (Cu) is 0.36151 nm. The close-packed directions, or the directions of the Burgers vector, are of the form $\langle 110 \rangle$. The repeat distance along the $\langle 110 \rangle$ directions is one-half the face diagonal, since lattice points are located at corners and centers of faces [Figure 4.10(a)].

$$\text{Face diagonal} = \sqrt{2}a_0 = (\sqrt{2})(0.36151) = 0.51125 \text{ nm}$$

The length of the Burgers vector, or the repeat distance, is:

b = 1/2(0.51125 nm) = 0.25563 nm

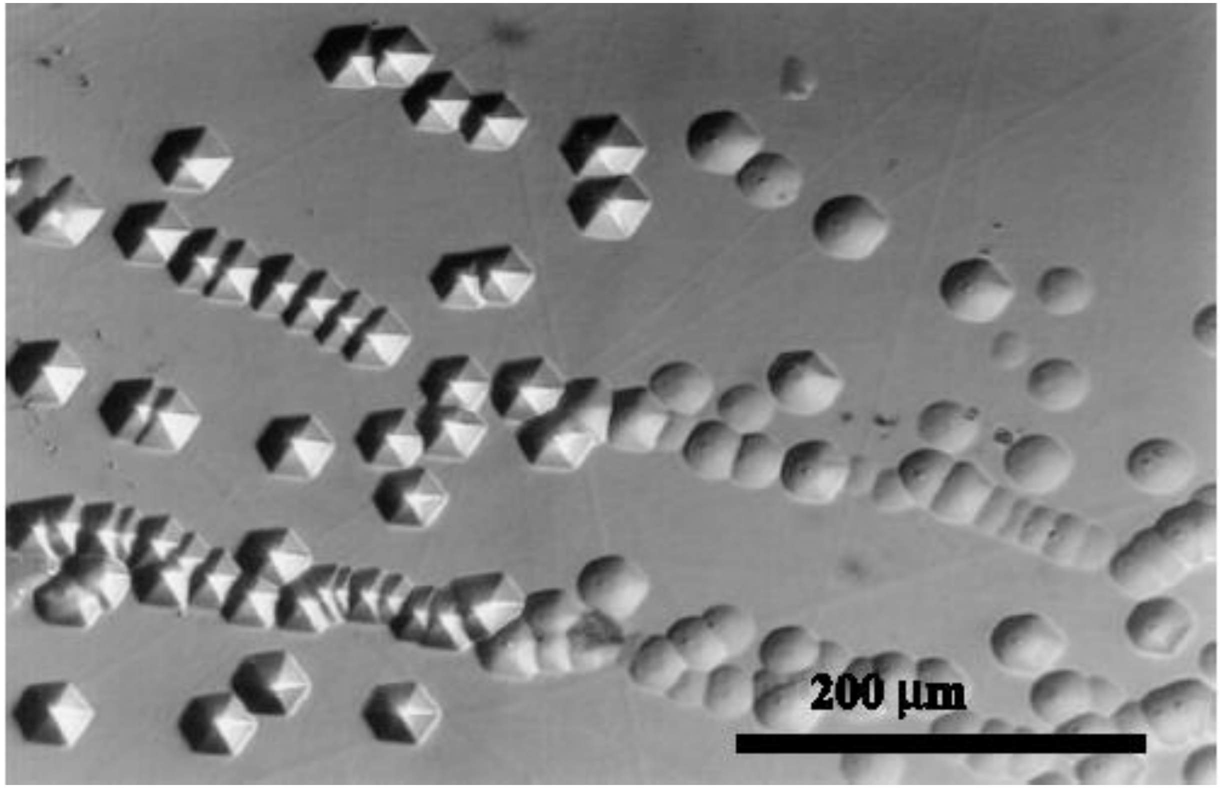
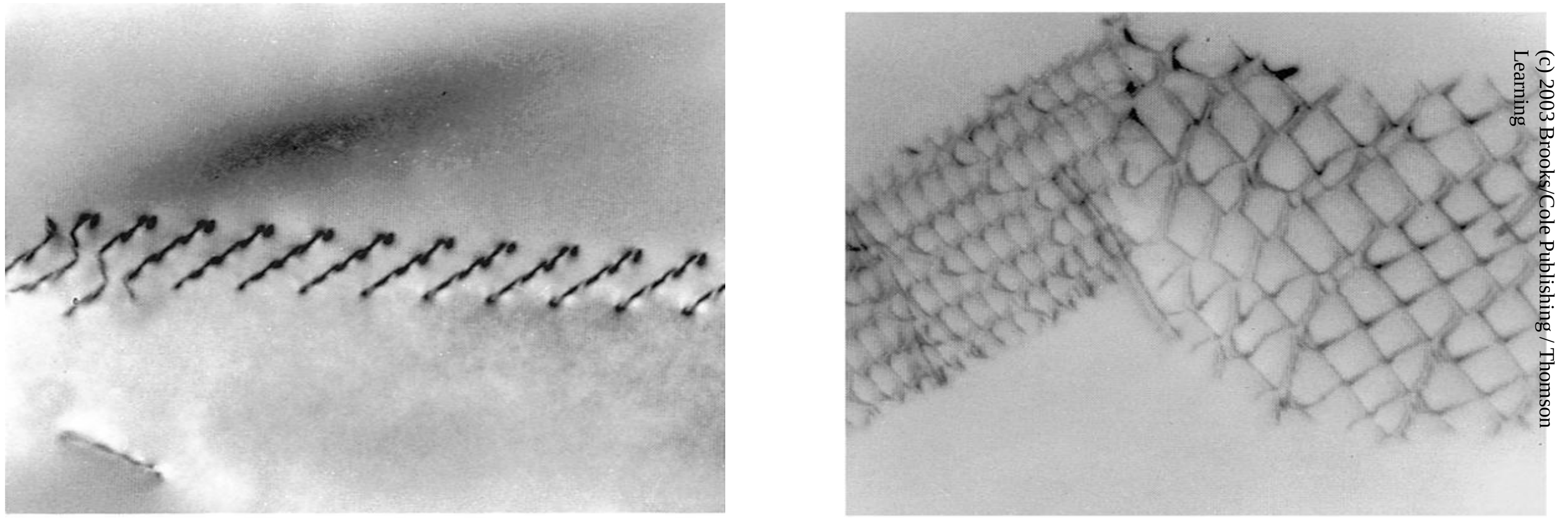


Figure 4.12 Optical image of etch pits in silicon carbide (SiC). The etch pits correspond to intersection points of pure edge dislocations with Burgers vector a/3 $\langle 1\bar{1}20\rangle$and the dislocation line direction along [0001] (perpendicular to the etched surface). Lines of etch pits represent low angle grain boundaries (*Courtesy of Dr. Marek Skowronski, Carnegie Mellon University*.)



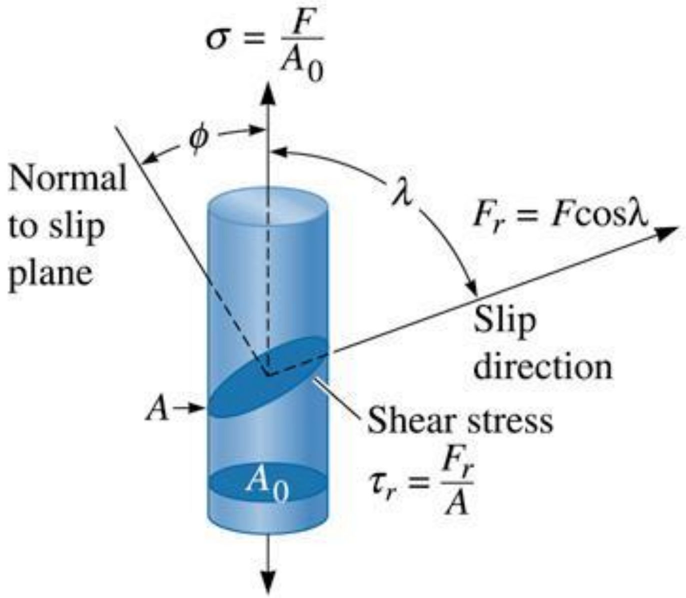
(a) (b)

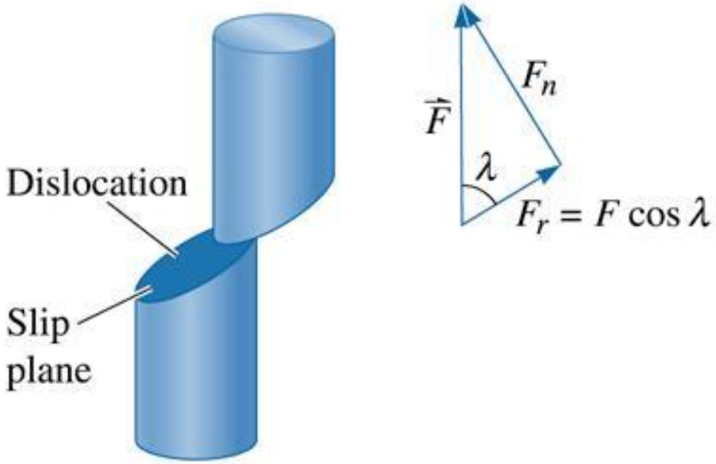
Figure 4.13 Electron photomicrographs of dislocations in *Ti_3Al*: (a) Dislocation pileups (x26,500). (b) Micrograph at x 100 showing slip lines and grain boundaries in Al.

Resolved Shear Stress

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

Legge di Schmid





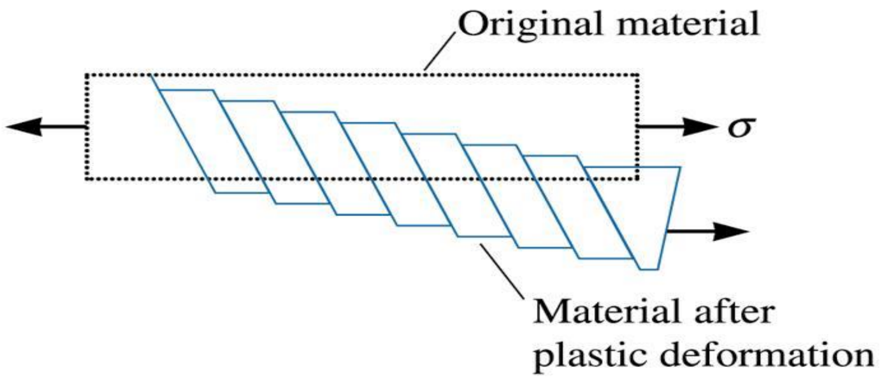


TABLE 4-2 ■ ***Summary of factors affecting slip in metallic structures***

Factor	FCC	BCC	HCP $\left(\frac{c}{a} > 1.633\right)$
Critical resolved shear stress (psi)	50–100	5,000–10,000	50–100[a]
Number of slip systems	12	48	3[b]
Cross-slip	Can occur	Can occur	Cannot occur[b]
Summary of properties	Ductile	Strong	Relatively brittle

[a] *For slip on basal planes.*

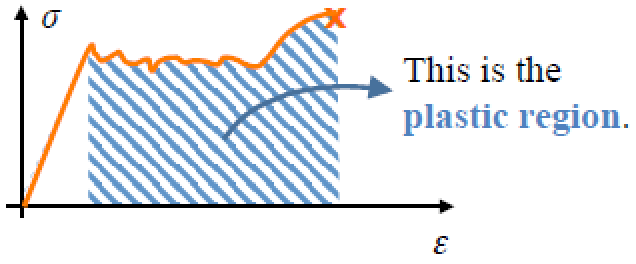
[b] *By alloying or heating to elevated temperatures, additional slip systems are active in HCP metals, permitting cross-slip to occur and thereby improving ductility.*

DISLOCATIONS AND PLASTICITY

Defects in solids are above the simple 0 dimensionality:

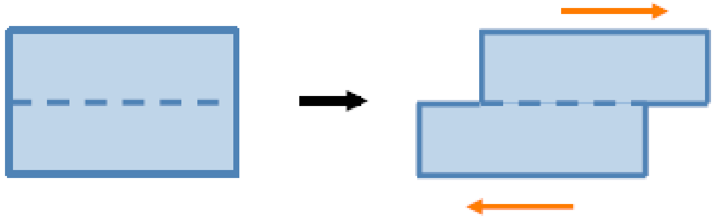
- 1D defects = dislocations
- 2D defects = surfaces, grain boundaries

They are all accompanied by energy and configuration changes. Here, we look at **plasticity**.



Q: What happens in the solid?

A: Internal flow, something like:



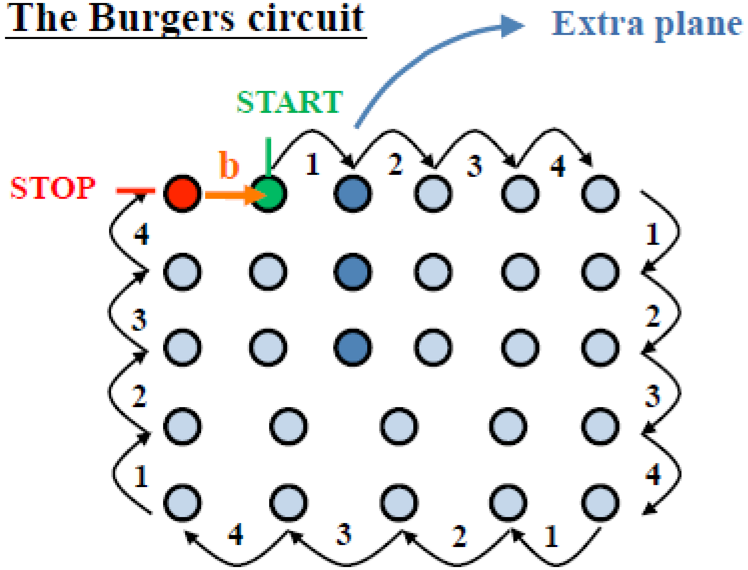
Q: Surely breaking bonds over an entire surface involves an enormous barrier?

$$1\,\text{bond} \cong 1\ eV;\ 1m^2 \cong 10^{20}\ \text{bonds} \rightarrow 10^{20}eV \cong 10J \rightarrow e^{-\frac{10J}{k_bT}} \equiv 0!!$$

A: in reality, only **one bond at the time** is broken, this is along a *dislocation* line-defect.

DISLOCATION AND PLASTICITY

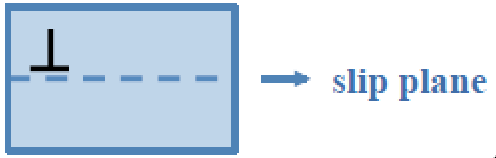
The Burgers circuit



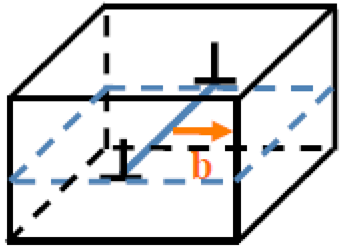
A crystal with an "**edge**" dislocation ≃ an **extra plane** (but there is no unique such plane for the same dislocation!).

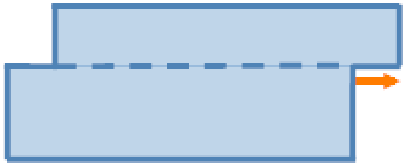
The Burgers circuit, clockwise, must be completed by $\vec{b}$, the **Burgers Vector**.

Note: an incomplete Burgers circuit is sufficient, but not necessary, for dislocations to exist inside the loop – the sum of their **b** vectors could be zero!



In edge dislocations the **Burgers vector** is perpendicular to the **dislocation line**.





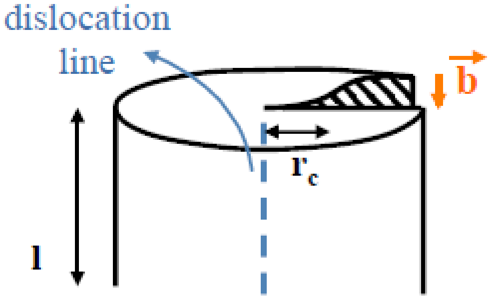
Note: the top moved only by $\vec{b}$, while the dislocation moved across the entire length of the solid

ENERGY OF A DISLOCATION

To summarise, in a dislocation loop:

•the handedness (or chirality) varies along the line (it's a *local, physical* property).

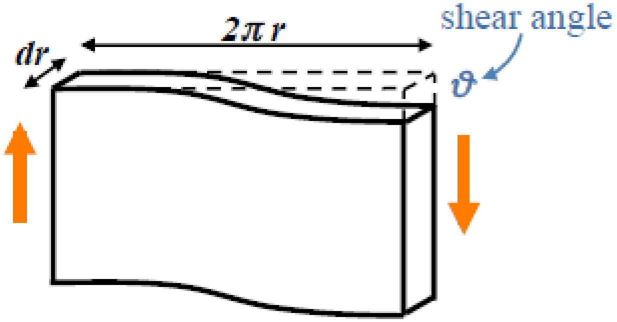
•the b vector is invariant it's a global, topological property (but its sign has no physical meaning, since choosing the opposite orientation for the dislocation will reverse b).



a screw dislocation can be modelled as a cylinder: after a "chemical zone" of radius r_c, the lattice is displaced by $\vec{b}$, parallel to dislocation line.

at all distances r from the dislocation line, the lattice is slipped by $\vec{b}$. The "cylindrical crust" of radius $\boldsymbol{r}$ and thickness $\boldsymbol{dr}$ is slipped by $|\vec{b}|$: it's **shear**!!

To see that shearing in going on, we "unroll" the cylinder:



The elastic energy density is:

$$E = \frac{1}{2}\sigma\varepsilon = \frac{1}{2}C_{44}\varepsilon^2 = \frac{1}{2}G\vartheta^2$$

$$C_{44} = G$$
$$\varepsilon = \vartheta$$
$$\vartheta \cong b/2\pi r$$

ENERGY OF A DISLOCATION

Total energy of the cylindrical crust:

$$l(2\pi r dr)\cdot\frac{1}{2}G\left(\frac{b}{2\pi r}\right)^2; \qquad \left(\frac{E}{l}\right)_{TOT} = \left(\frac{E}{l}\right)_{CHEM} + \left(\frac{E}{l}\right)_{ELAST}$$

but $$\left(\frac{E}{l}\right)_{ELAST} = L = \int_{R_C}^{R_{MAX}} \frac{1}{2}G\frac{b^2}{2\pi r}dr = \frac{1}{4\pi}Gb^2 \ln\left(\frac{R_{MAX}}{R_C}\right)$$

where the (double!) infinity of the integral is avoided because:

1) $\boldsymbol{R_M < \infty}$ as every sample must have a finite size
2) $\boldsymbol{R_C > 0}$ where the atomistic structure kinks in, in the chemical zone, at a distance from the line $r \cong |\vec{b}|$, the elastic approximation holds no more

If we set $\ln\left(\frac{R_{MAX}}{R_C}\right) \cong 4\pi$ we get $$\boxed{\left(\frac{E}{l}\right)_{EL} \cong Gb^2}$$ **elastic energy per unit length**

this has the dimension of a *tension*: a **force** trying to keep the dislocation line straight to minimise its overall energy cost

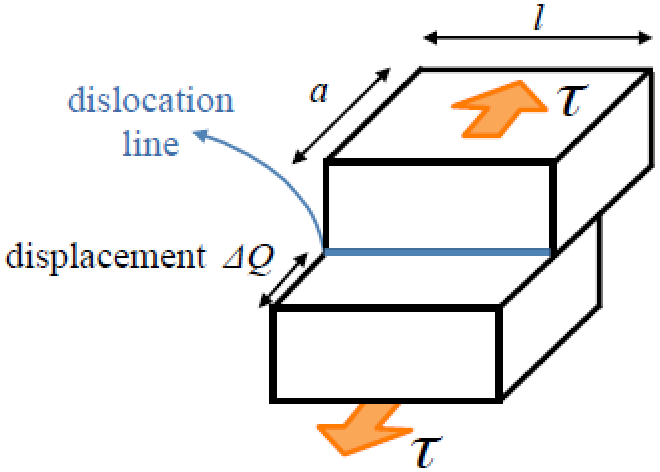
Suppose $G = 100\,GPa;\ |\vec{b}| = 1\,\mathring{A}$ ⇨ $\left(\frac{E}{l}\right)_{EL} = 10^{11}\cdot 10^{-20} = 10^{-9}\,J/m$

Note that this is still $\boldsymbol{\sim 10^{-19}\ J/\mathring{A}}$, so for every atom along the dislocation line we have $\cong \boldsymbol{1eV}$ of elastic energy → this is bigger than the chemical energy, usually. This rough estimate suggests that the dislocation energy is mostly elastic

FORCE ON THE DISLOCATION

Q: what "pushes" the dislocations?

A: the applied external stresses, of course. In fact "dislocation" *means* plasticity (plastic deformation) typically occurring as a response to applied stress

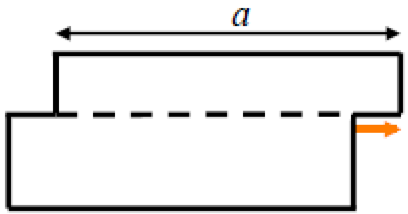


Sliding caused by a shear stress τ (it's the common name for $\sigma_{4,5,6}$).

$\tau \cdot (a \cdot l) =$ Shear Force

$\tau \cdot (a \cdot l) \cdot \Delta Q = \Delta L$ Work made by the external stress

$\frac{\Delta L}{l} = \tau \cdot a \cdot \Delta Q =$ Work per unit of dislocation length



Note: for $\Delta Q = \vec{b}$ = the dislocation has moved across the entire solid! so:

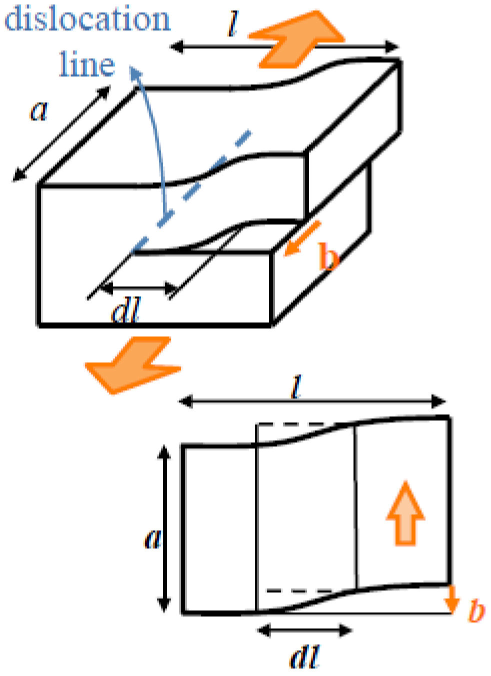
$\frac{\Delta L}{l} = \vec{f} \cdot \vec{a}$ with $\vec{f} =$ force per unit length on the dislocation

$\therefore \ |\vec{f}| = \tau \cdot b$ $\boxed{f = \tau b}$ **force exerted on the dislocation line (orthogonal to it, per unit length)**

FORCE ON THE DISLOCATION

Q: what happens in the case of screw dislocations?

A:



The displacement to the left of a surface $(a \cdot dl)$ corresponds to an external work

$$[a \cdot dl \cdot \tau] \cdot b = \Delta L$$

$$\frac{\Delta L}{a} = \text{ Work per unit of dislocation length}$$

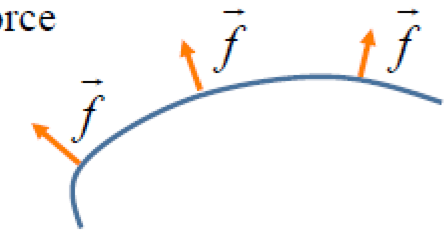
so in this case $\dfrac{\Delta L}{a} = f \cdot dl$

$$\Rightarrow f = \tau \cdot b$$

The same result as before!

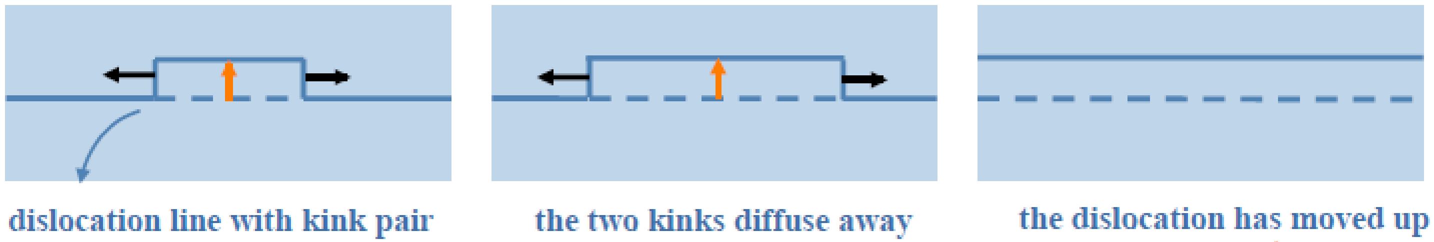
This relation is valid in every case: in general, the force per unit length, orthogonal to the dislocation line is

$$\boxed{f = \tau b}$$



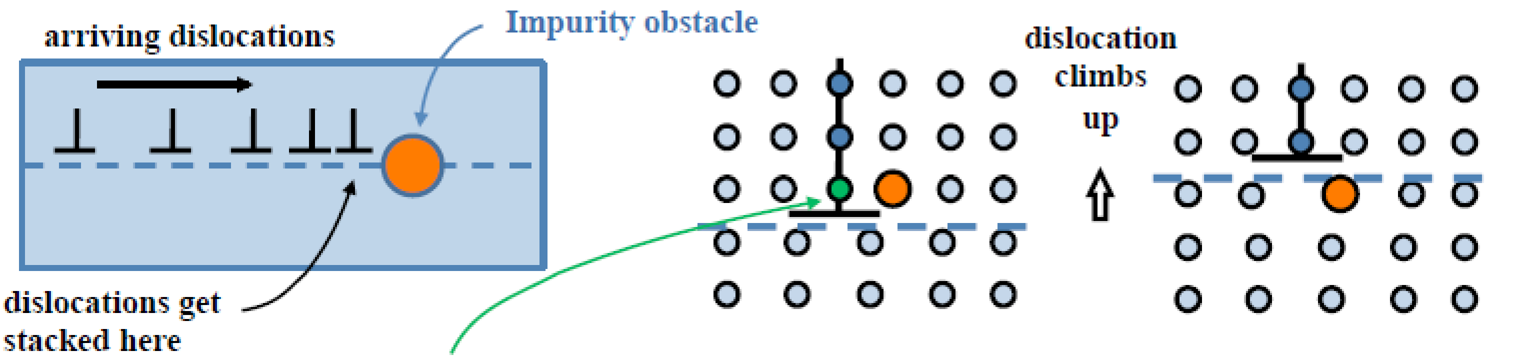
THE DISLOCATION MOTION

The motion normally happens via the “kink” mechanism:



Note that an edge dislocation has a fixed slip plane, identified by the line and $\vec{b}$.
However the screw dislocation does not!

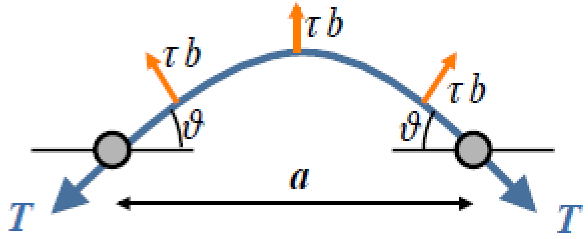
So an edge dislocation cannot normally change slip plane: in the presence of impurities the dislocation could be blocked. However they can “climb”!



This atom can move away (if a “vacancy” arrives, e.g.) enabling dislocation climb. It’s slow, however, as it has to wait for the vacancy diffusion

DISLOCATION PINNING

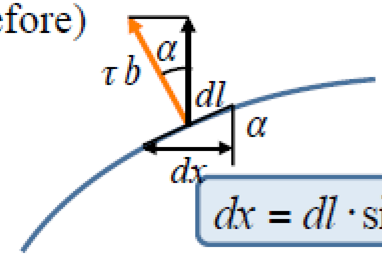
This may happen when a dislocation hits an impurity or defect located his path:



we compare two effects:

1) $2T\sin\vartheta =$ Pull back: sum of the two tension components opposing the motion of the dislocation

$T \cong Gb^2$ (as seen before)



$dx = dl \cdot \sin\alpha(x)$

2) PUSH FORWARD: total effect of the external forces:

$$\int_0^a dx \cdot \tau b \cdot \sin\alpha(x) = \int_0^l dl \cdot \tau b = \tau\, b\, l$$

$\rightarrow \quad 2 \cdot Gb^2 \cdot \sin\vartheta = \tau \cdot b \cdot l$ ⇒ $\tau = \dfrac{2 \cdot Gb \cdot \sin\vartheta}{l}$ **Equilibrium tension**

the maximum occurs for $\sin\vartheta = 1;\ \vartheta = \pi/2$

$$\tau_c = \frac{2 \cdot Gb}{l}$$

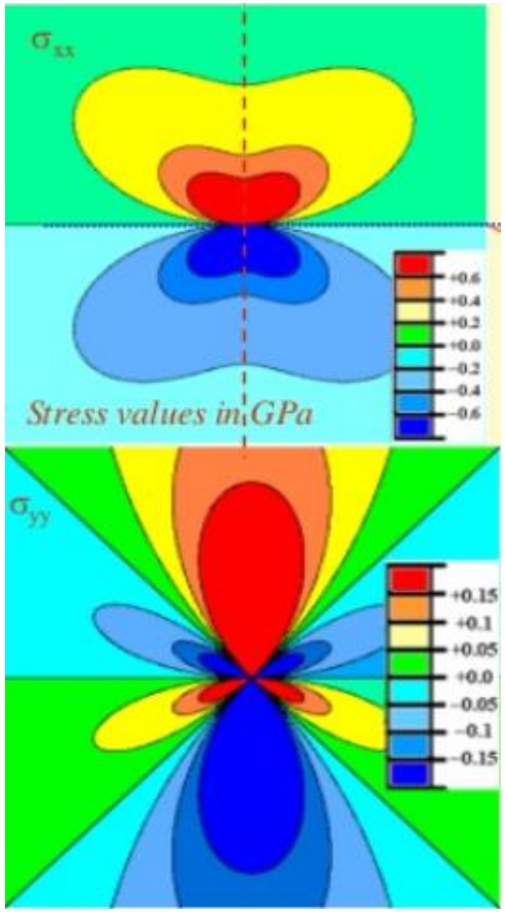
Critical Tension (before the yielding)

clearly bigger for closer defects!

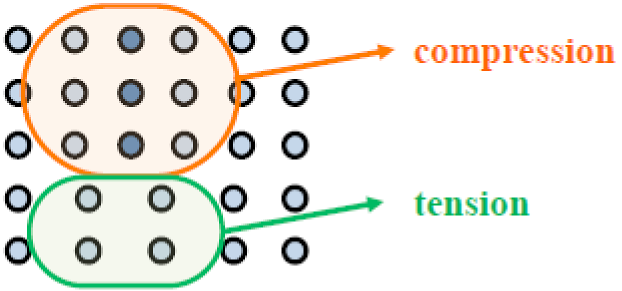
Example: $\begin{cases} l \cong 1\mu m = 10^{-6}m \\ b \cong 10^{-10}m \\ G = 100GPa \end{cases}$ $\tau_c = \dfrac{2 \cdot 10^{11} \cdot 10^{-10}}{10^{-6}} = 2 \cdot 10^7\,Pa = 20\,MPa$

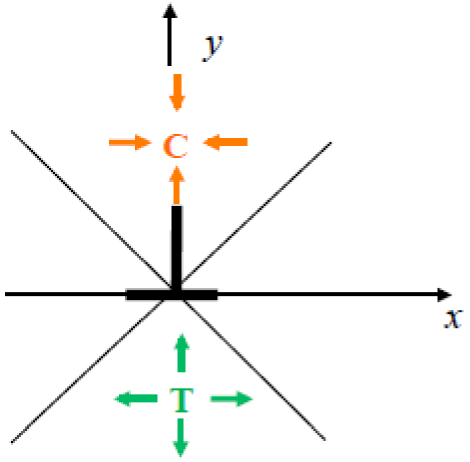
Note: the (b/l) ratio lowers the τ_c for large l. This is crucial in “heat treatments” designed to control the distribution of l.

ELASTIC STRESS FIELD



Emitted by a dislocation. The shape of this elastic stress field helps to understand the interaction between dislocation. For an edge Dislocation we have:





$$\sigma_1 = \sigma_{xx} = \frac{-Gby}{2\pi(1-\nu)} \frac{3x^2 + y^2}{(x^2 + y^2)^2}$$

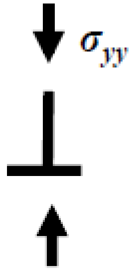
$$\sigma_2 = \sigma_{yy} = \frac{+Gby}{2\pi(1-\nu)} \frac{x^2 - y^2}{(x^2 + y^2)^2}$$

$$\sigma_6 = \sigma_{xy} = \frac{+Gbx}{2\pi(1-\nu)} \frac{x^2 - y^2}{(x^2 + y^2)^2}$$

Different from the cylindric case, it's not only a function of $r = \sqrt{x^2 + y^2}$ but the "long range" part is still $\sigma \propto 1/r$ in simple cases.

ELASTIC STRESS FIELD

e.g.: y axis ($x=0, y$)



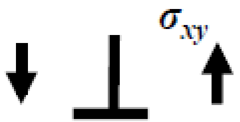
$$\begin{cases} \sigma_{yy} \approx -y^3/y^4 \approx 1/y \approx 1/r \\ \sigma_{xx} = \sigma_{yy} \\ \sigma_{xy} = 0 \end{cases}$$

note: $\varepsilon \propto 1/r$ as well $\Rightarrow \frac{1}{2}\varepsilon\sigma \propto \frac{1}{r^2}$

$\int \frac{1}{2}\varepsilon\sigma \cdot 2\pi r \cdot dr \rightarrow \propto 1/r$

The elastic energy associated with distance r is $\propto 1/r$, causing a logarithmic divergence

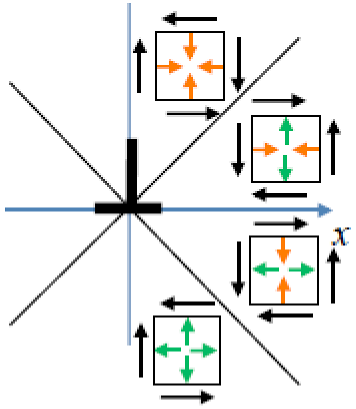
<u>**Shear**</u>



$$\begin{cases} \sigma_{yy} = \sigma_{xx} = 0 \\ \sigma_{xy} \approx 1/x \approx 1/r \end{cases}$$

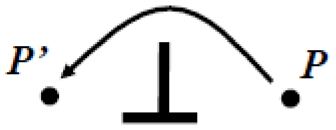
→ by symmetry

it is possible to divide the scheme in octants and plot as arrows the sign of all components.



(x = y), (x = - y) : The shear components change sign.

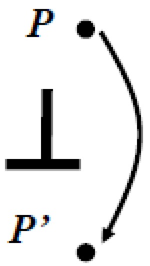
(x = 0), (y = 0) : The compressive components change sign.



$(x, y) \rightarrow (-x, y)$

σ_{yy}, σ_{xx} unvaried

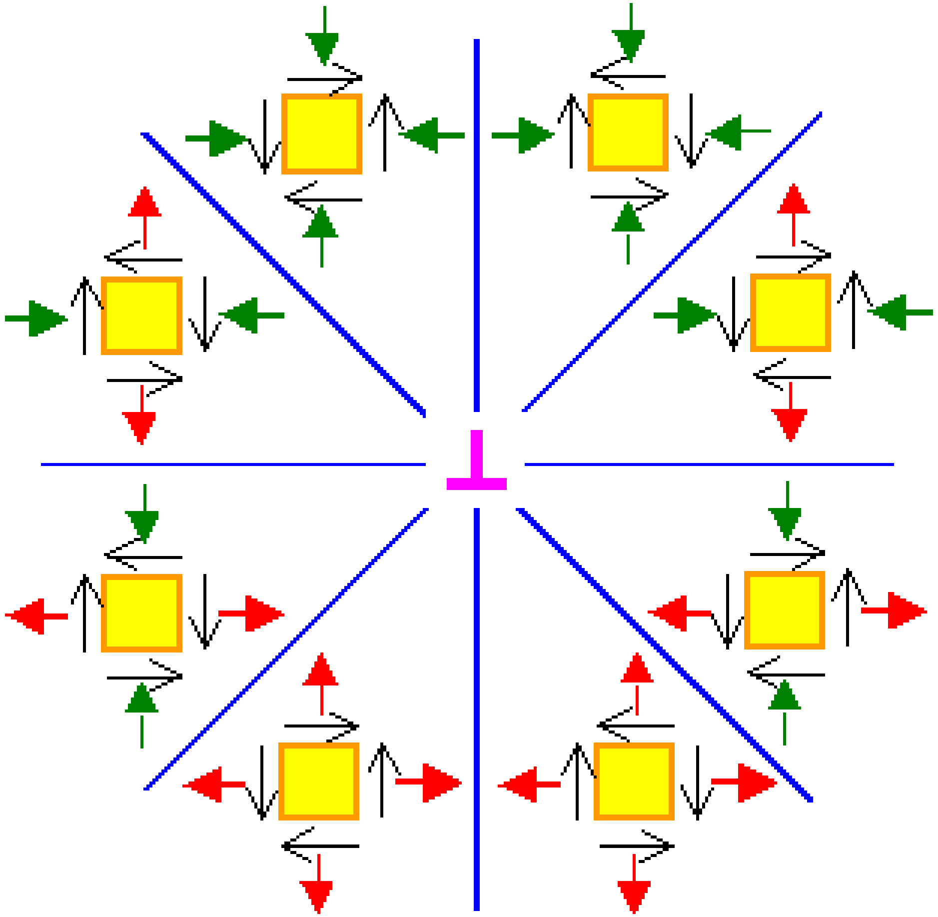
σ_{xy} reversed

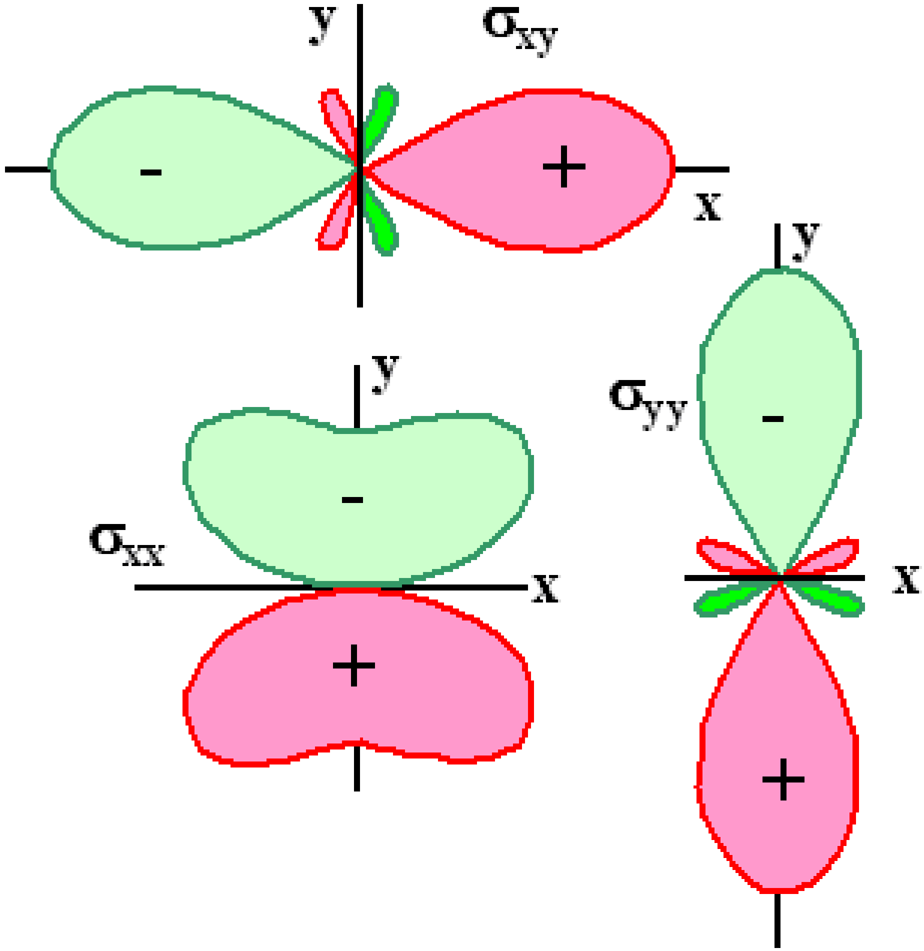


$(x, y) \rightarrow (x, -y)$

σ_{yy}, σ_{xx} reversed

σ_{xy} unvaried




y
σxy
-
+
x
y
σyy
-
x
y
σxx
-
x
+
+

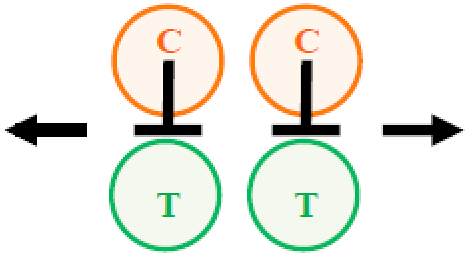
INTERACTION BETWEEN DISLOCATIONS

Parallel dislocations

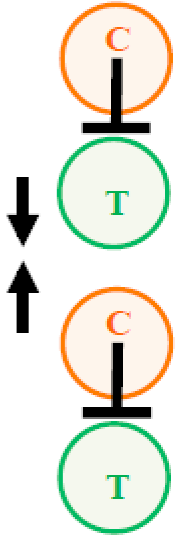
Overlap principle: $E = \frac{1}{2}\underline{\sigma}\cdot\underline{\varepsilon}$ with $\sigma \propto \varepsilon$ (remember: $\underline{\sigma} = \underline{\underline{C}}\ \underline{\varepsilon}$)

1) overlap (sum, at first order) of the two elastic strain fields ⇨ **but the energy varies quadratically!**

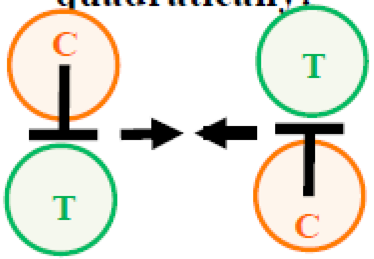
2) the stress is therefore the sum of the two stresses



Dislocations of the *same sign* (e.g., same slip plane) *repel* each other



With this geometry they attract each other!

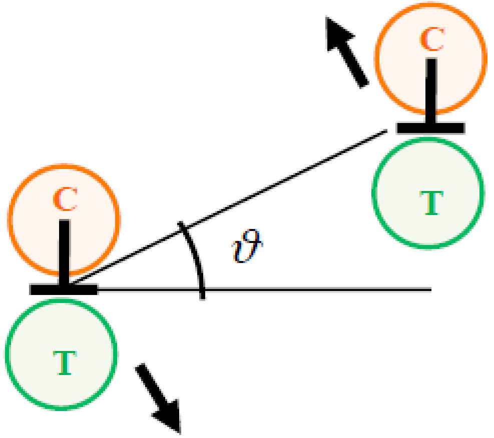


Dislocations of opposite sign attract each other.

notice the similarity with the interaction of electric dipoles. However here the position can change, but not the orientation (as the orientation of b and the dislocation line are fixed)

INTERACTION BETWEEN DISLOCATIONS

If the two edge dislocations don't have the same slip plane, shear forces will try to minimize the elastic energy cost



The force has to be $F_\vartheta = 0$ for $\begin{cases} \vartheta = 0 \\ \vartheta = \pi/2 \end{cases}$

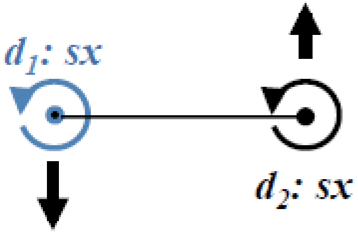
The correct expression turns out to be:

$$\left|\vec{F}_\vartheta\right| = C\,\frac{\vec{b}_1 \cdot \vec{b}_2}{r} \cdot \sin(2\vartheta)$$

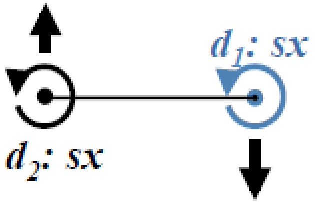
where $\vec{b}_1 \cdot \vec{b}_2$ reproduces the correct sign

for parallel dislocations $F_r = C_{11}\,\dfrac{\vec{b}_1 \cdot \vec{b}_2}{r}$

Note: parallel *screw* dislocations *cannot* involve $F_\vartheta \neq 0$



← supposing a couple of forces as shown, an observer looking t the system from the Opposite side would see still two ***sx*** dislocations, but an opposite (clockwise) force couple.



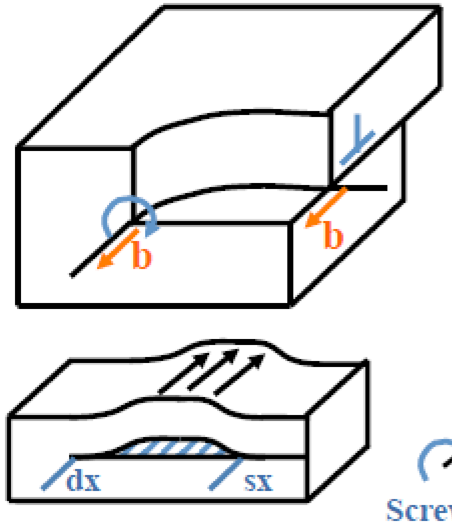
⇨ $F_\vartheta \neq 0$ it's not possible. (this is because the chirality of screw dislocations is invariant under rotations by 180°)

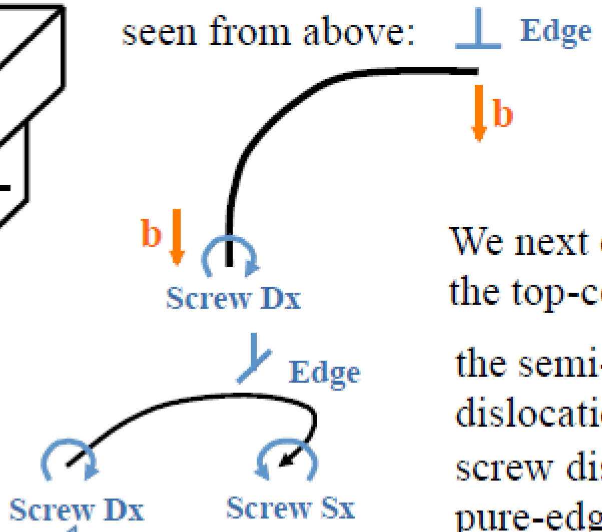
So we have only $F_r = C\,\dfrac{\vec{b}_1 \cdot \vec{b}_2}{r}$!! (radial part not derived here)

DISLOCATION GENERATION

THE FRANK–READ SOURCE

we start from the example already seen.

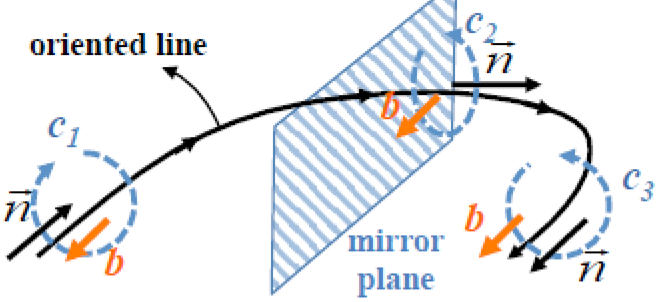




Clearly, $\vec{b}$ is a topological invariant for closed Burgers circuit enclosing the dislocation.

We next consider a symmetric system, with the top-central part slipped forward

the semi-loop start with a ***dx*** screw dislocation on the left, and ends with a ***sx*** screw dislocation on the right, with a pure-edge section in the front

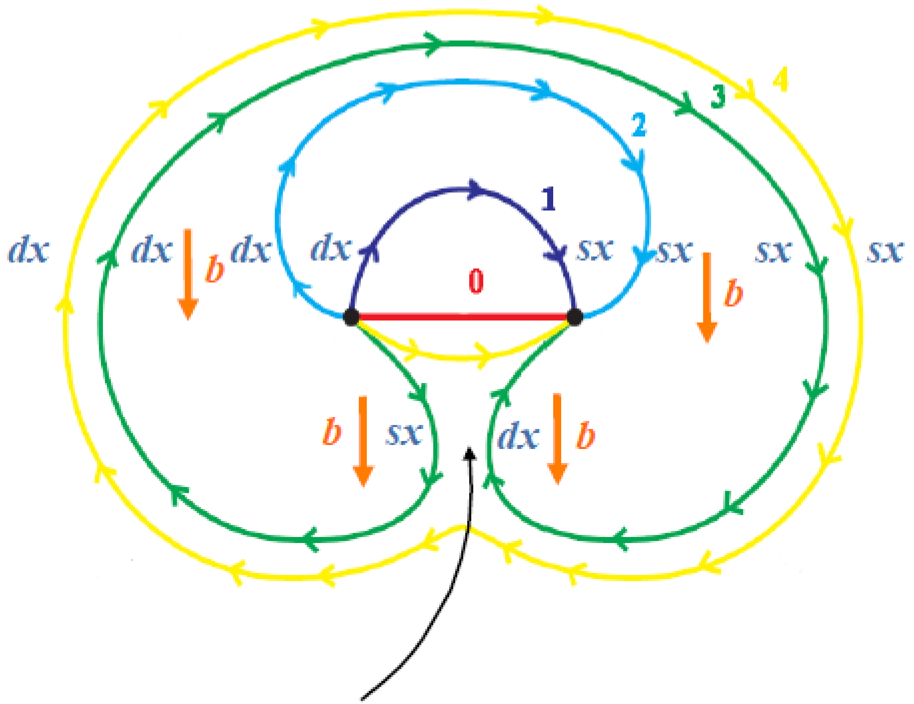


Note: All the c_n circuits are clockwise respect the orientation of the line.

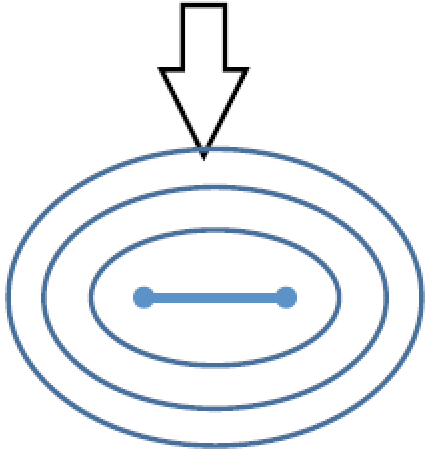
The invariance of b implies that on left the dislocation is ***dx*** ($\vec{b}\cdot\vec{n} < 0$), while on the right is ***sx*** ($\vec{b}\cdot\vec{n} > 0$). Note once more that the ***sx*** or ***dx*** chirality are physical, local, properties of the dislocation. b it's a global property.In the middle section ($\vec{b}\cdot\vec{n} = 0$ there is no chirality and the dislocation is pure edge.

⇨ The systems has a **"mirror plane symmetry"**

DISLOCATION GENERATION
THE FRANK – READ SOURCE



These are the 4 evolution steps of a pinned dislocation. At the end of each 4th step, all external loop widen, and the mechanism keeps iterating, emitting concentric dislocation loops which enable plastic behaviour



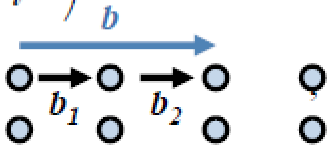
Frank – Read source

Here there are two mirror symmetric screw dislocation sections with the same *b*: one is ***sx***, the other is ***dx***. The opposite chirality implies that they attract each other. The local reaction decays into tow new separate branches: a large loop and a new (“0”) source section.

PARTIAL DISLOCATIONS

Partial dislocations occur in different materials (e.g., covalent and metallic).
Up to now $\vec{b} \in$ to the direct lattice (the Burgers circuit is composed only by lattice sites).

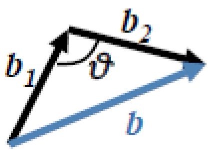
but $\left(\frac{E_{EL}}{l}\right) \propto Gb^2$ so, if we could write $\vec{b} = \vec{b}_1 + \vec{b}_2$, and create two "partial" dislocations, able to separate, this will be energetically favourable.

e.g., if 

$\vec{b}, \vec{b}_1, \vec{b}_2$ are all direct lattice vectors; then clearly:

all the dislocations $\vec{b} = 2\vec{b}_1 \in R$ decay in two dislocations with vector $\vec{b}_1$.

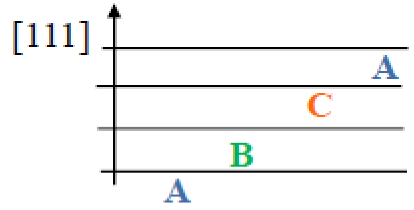
Once the minimum $\vec{b} \in R$ is reached, further decomposing becomes harder.

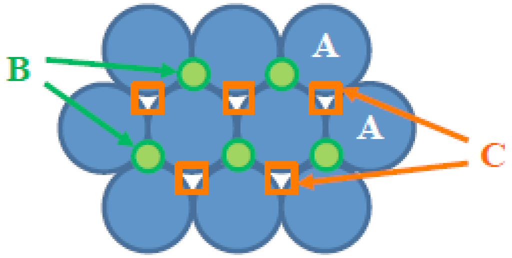
but, if $\frac{\vec{b}_1 \cdot \vec{b}_2}{b_1 b_2} = \cos\vartheta < 1$ 

i.e.: for angles $> \frac{\pi}{2}$ $b^2 > b_1^2 + b_2^2$

$$(\vec{b}_1 + \vec{b}_2)^2 = b_1^2 + b_2^2 + \underbrace{2\vec{b}_1 \cdot \vec{b}_2}_{>0}$$

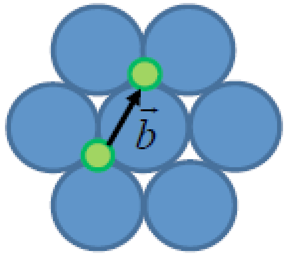
This situation arises in the FCC lattices on the (111) plane:

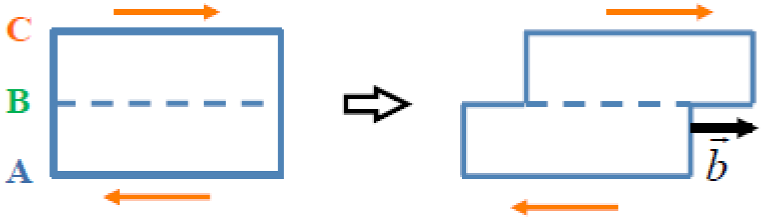




PARTIAL DISLOCATIONS

Now, the vector $\vec{b}$ that connects two **B** lattice sites is $\vec{b} \in R$, a good Burgers vector

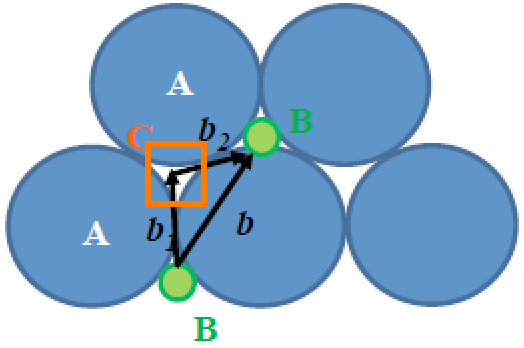




However there is another way:

$$\vec{b} = \vec{b}_1 + \vec{b}_2$$

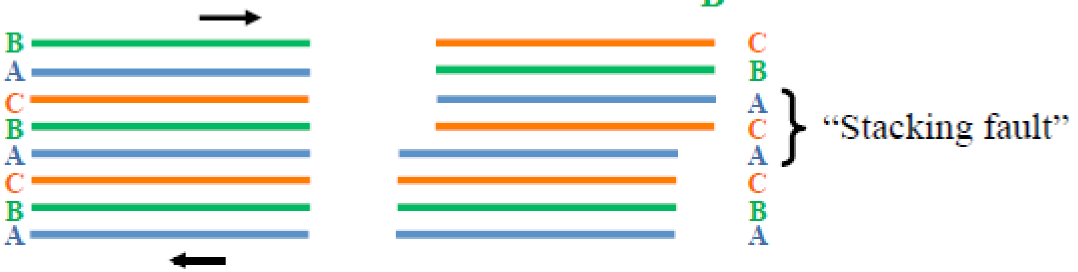
where $\vec{b}_1$ goes from a **B** site to a C site and $\vec{b}_2$ returns from a C site back to a **B** site.



e.g.: $\vec{b}_1$: **A** → **B**

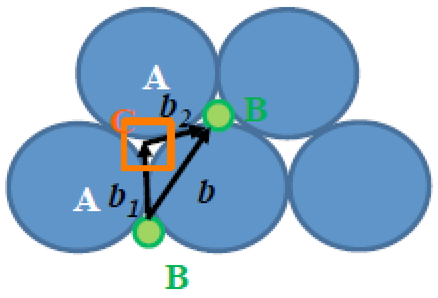
B → **C**

C → **A**

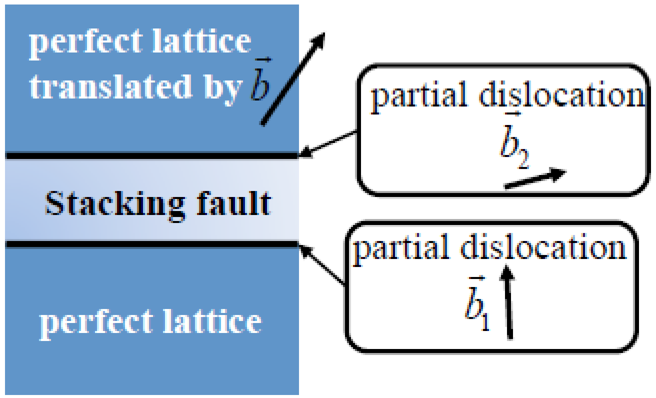


this way, two **A** planes see each other across the **C** plane in the stacking fault region

PARTIAL DISLOCATIONS



seen from far away:



The two partial dislocations repeal each other with a force $\propto \frac{1}{r}$ (per unit length).

The stacking fault costs $\gamma \cdot r \cdot l = E$ where γ [Joule/m^2] is its energy cost per m^2

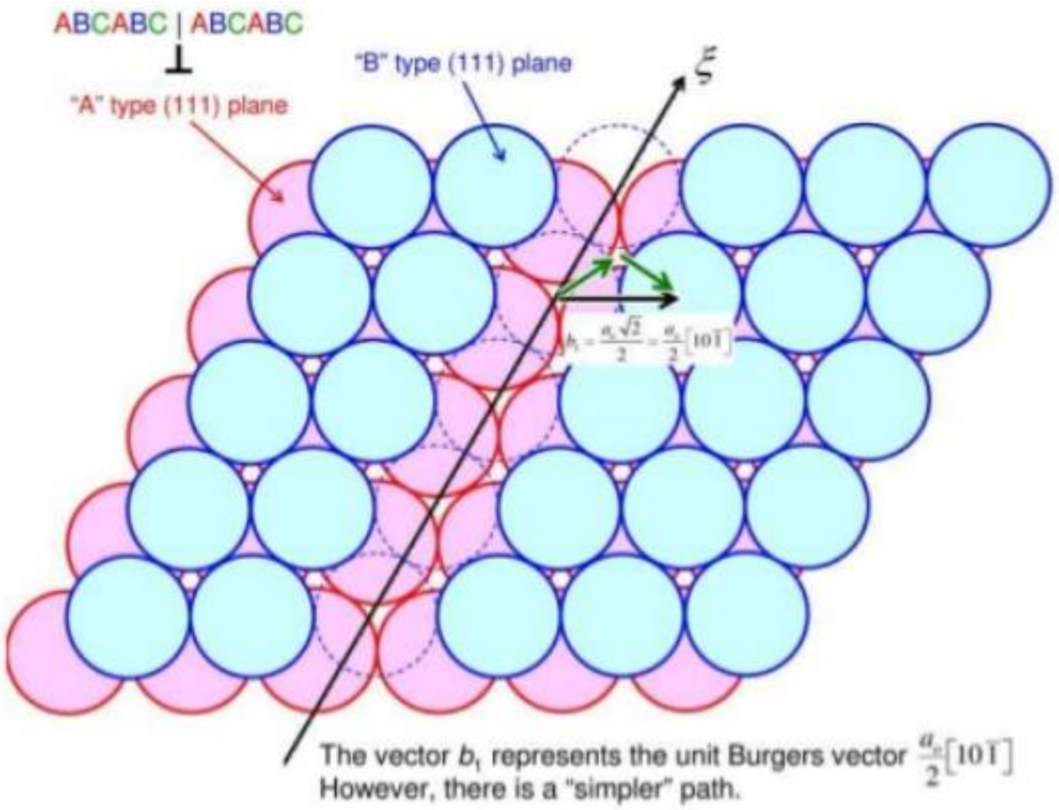
$$\frac{\partial}{\partial r}\frac{E}{l} = \gamma = \text{restoring “force”} \Rightarrow C\frac{\vec{b}_1 \cdot \vec{b}_2}{r} = \gamma$$

gives an equilibrium distance r_{eq}, which determines the typical lateral dimension of the stacking fault.

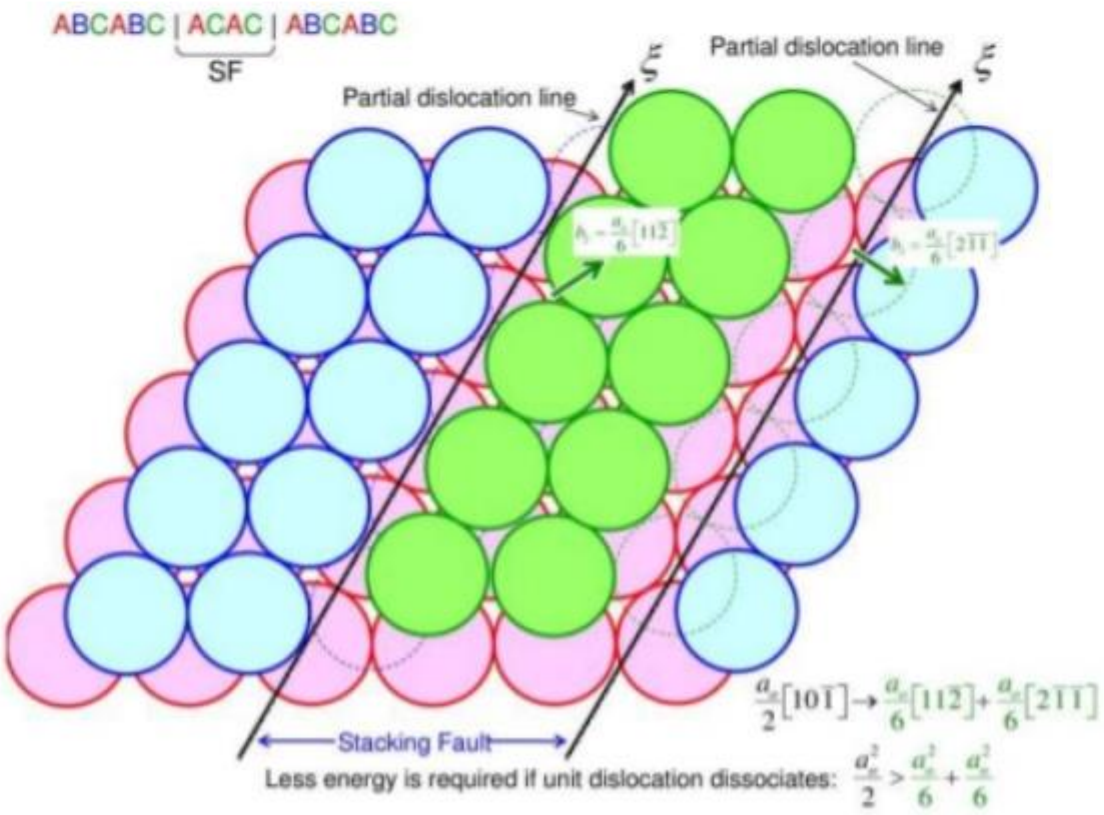
Note: A screw dislocation doesn’t have preferential slip plane, but, if it decays into two partial dislocations + stacking fault, it becomes “committed” to a slip plane.
Thus, cross-slipping to a new plane becomes problematic!

Partial Dislocations in HCP

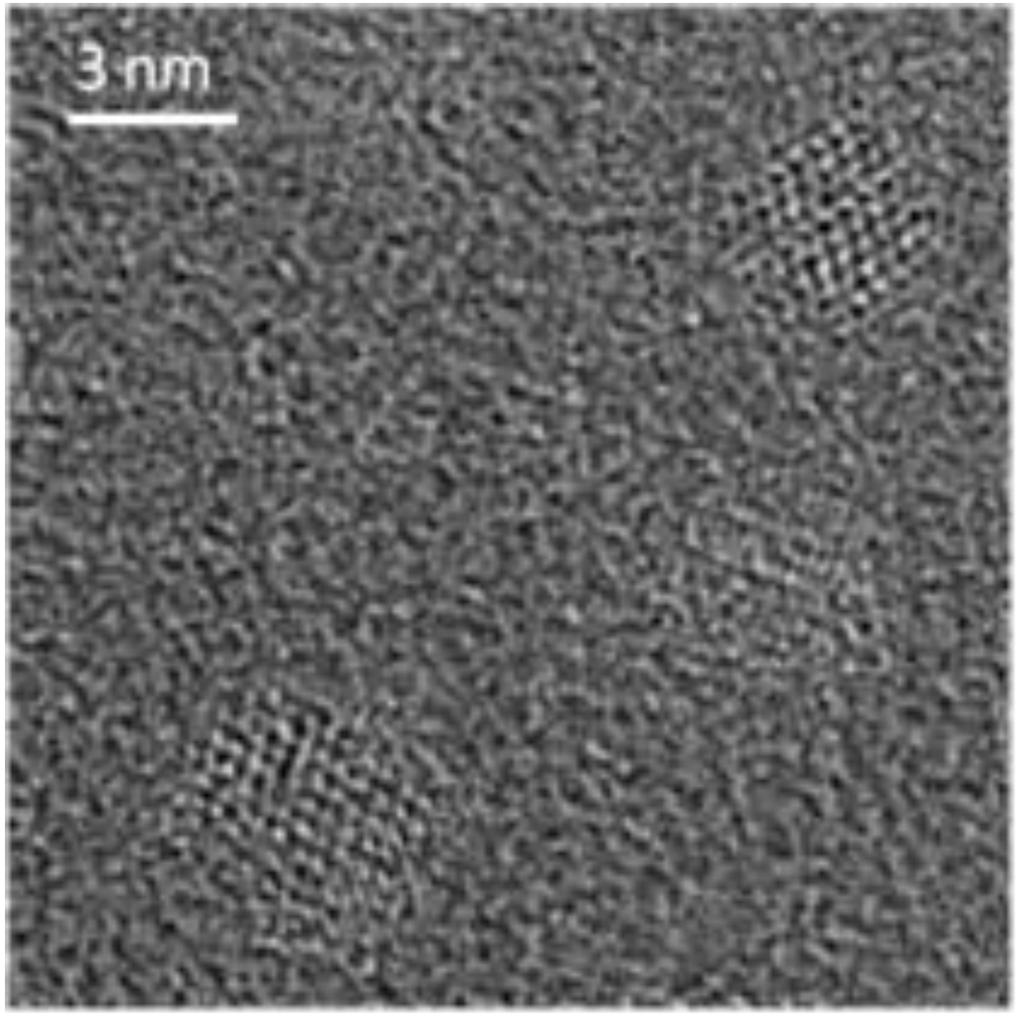
Burgers vector belongs to direct lattice



Decomposition of Burgers vector – less energy required

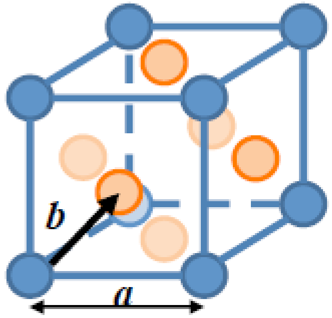


Stacking Faults



PARTIAL DISLOCATIONS

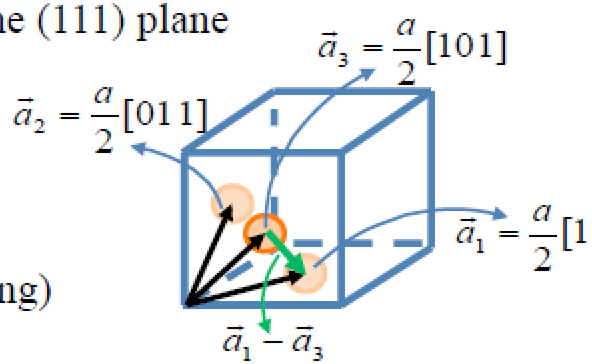
It's possible to develop a "Burgers vectors algebra":



Take $\boldsymbol{a}$ as a FCC lattice constant, with nearest neighbours connected by $\frac{a}{2}[101] = \vec{b}$ (cf. picture on left)

This is true everywhere, e.g., on the (111) plane

the vectors lying on the (111) plane can be obtained as differences:



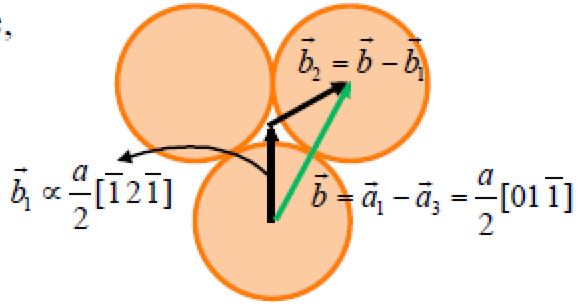
for example: $\vec{a}_1 - \vec{a}_3 = \frac{a}{2}[01\bar{1}]$

$\vec{a}_2 - \vec{a}_3 = \frac{a}{2}[\bar{1}10]$ (both $\frac{a}{\sqrt{2}}$ long)

the bisector between two atomic positions of the (111) plane, e.g. $[01\bar{1}]$ and $[\bar{1}10]$ is $[\bar{1}2\bar{1}]$

⇨ so $\vec{b}_1$ is $\propto [\bar{1}2\bar{1}]$ vector and is $\frac{b}{\sqrt{3}}$ long

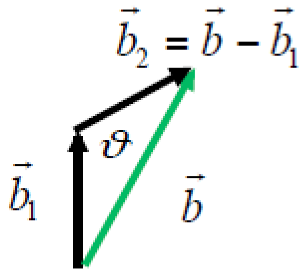
⇨ $\vec{b}_1 = \frac{a}{\sqrt{6}} \cdot \frac{[\bar{1}2\bar{1}]}{\sqrt{6}} = \frac{a}{6}[\bar{1}2\bar{1}]$



Now: $\vec{b}_2 = \vec{b} - \vec{b}_1 = a\left[0, \frac{1}{2}, -\frac{1}{2}\right] - a\left[-\frac{1}{6}, \frac{2}{6}, -\frac{1}{6}\right] = a\left[\frac{1}{6}, \frac{1}{6}, -\frac{2}{6}\right] = \frac{a}{6}[11\bar{2}]$

PARTIAL DISLOCATIONS

So the reaction of formation of two partial dislocation, in this case, is:



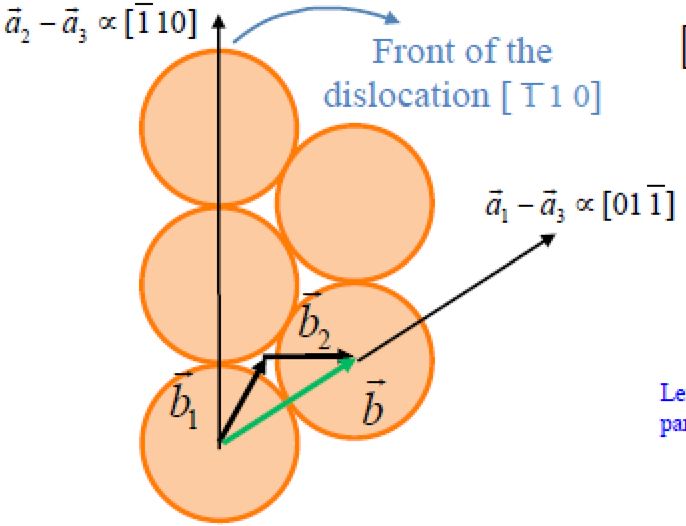
$$\vec{b} \rightarrow \vec{b}_1 + \vec{b}_2 \qquad \frac{a}{2}[01\bar{1}] \rightarrow \frac{a}{6}[\bar{1}2\bar{1}] + \frac{a}{6}[11\bar{2}]$$

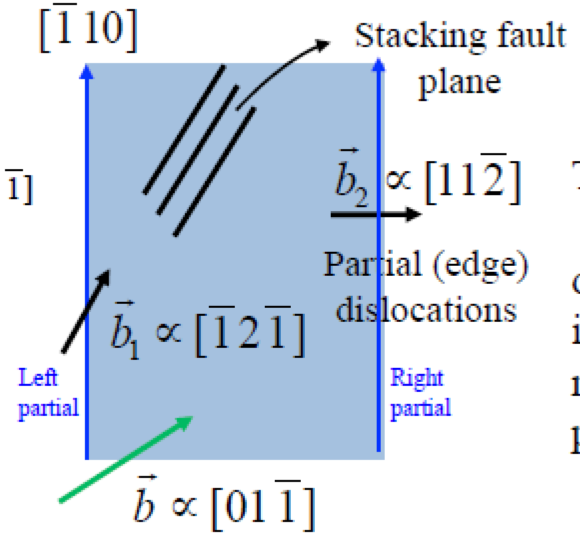
moreover:

$$b^2 = \frac{a^2}{4} \cdot 2 = \frac{a^2}{2}$$

$$b_1^{\,2} = b_2^{\,2} = \frac{a^2}{36} \cdot (1+4+1) = \frac{a^2}{6} \Rightarrow b_1^{\,2} + b_2^{\,2} = \frac{a^2}{3} < \frac{a^2}{2} = b^2$$

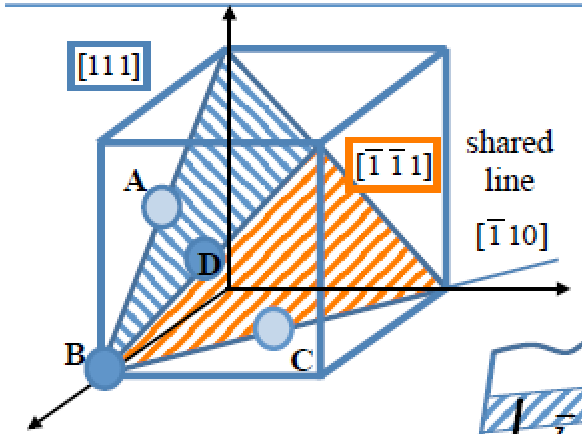
obvious as $\vartheta = 120°$





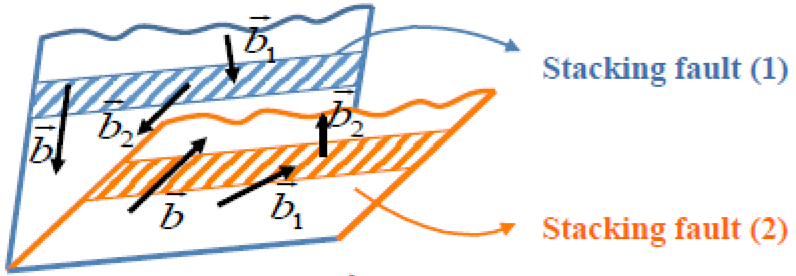
This is the situation with $\vec{b}$ tilted by 60° from the dislocation line (so this is a mixed dislocation, neither pure screw nor pure edge)

LOMER – COTTRELL LOCK



we next need two (111) glide planes, e.g., [1 1 1] and [$\bar{1}\,\bar{1}\,1$], hosting two systems of dislocations.

we consider the elementary tetrahedron ABCD and observe that in both planes it is possible to build a system with the dislocation line $[\bar{1}10] \propto \overline{BC}$



$$\begin{cases} b = a/2\cdot[10\bar{1}] & \overrightarrow{AB} \\ b_1 = a/6\cdot[11\bar{2}] \\ b_2 = a/6\cdot[2\bar{1}\bar{1}] \end{cases}$$

$$\begin{cases} b = a/2\cdot[011] & \overrightarrow{BD} \\ b_1 = a/6\cdot[\bar{1}21] \\ b_2 = a/6\cdot[112] \end{cases}$$

If we write the reaction between the two $\vec{b}$ dislocations, we obtain:

$$\overrightarrow{AB} + \overrightarrow{BD} = \overrightarrow{AD} \iff \frac{a}{2}[10\bar{1}] + \frac{a}{2}[011] = \frac{a}{2}[110]$$ a new **b** vector $\perp \overline{BC}$ (a good edge dislocation)

However, the reaction occurs between the two front partial dislocations in the two planes:

$\frac{a}{6}[2\bar{1}\bar{1}] + \frac{a}{6}[\bar{1}21] = \frac{a}{6}[110]$ this is of the same kind, but it is **not** a lattice vector!

So, the total reaction is: $\frac{a}{2}[10\bar{1}] + \frac{a}{2}[011] = \frac{a}{6}[11\bar{2}] + \frac{a}{6}[110] + \frac{a}{6}[112]$

$\frac{a}{6}[110]$ is not a good Burgers vector: neither a lattice, nor a good partial vector.
The system is locked: **sessile dislocation** ⇨ **work hardening**