1.2.3 Kinematics of dislocations

In general the movement of dislocation can be conservative or non conservative. In Fig.1.16 two extreme cases for the movement were drawn: a vector $\vec{r_1}$ within the glide plane and $\vec{r_2}$ perpendicular to it.

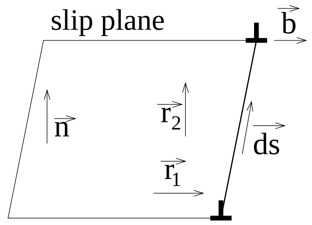

Figure 1.16: Slip plane with Burger's vector $\vec{b}$, dislocation line vector $\vec{ds}$ and two possible (extreme) vectors of movement $\vec{r_1}$ and $\vec{r_2}$.

Two possibilities:
a.) vector of movement: $\vec{r_1} \parallel \vec{b} \perp \vec{ds}$
b.) vector of movement: $\vec{r_2} \perp \vec{b} \perp \vec{ds}$
The **Volume of transport** i. e. the amount of atoms to be moved is

$$dV = \vec{b} \cdot (\vec{ds} \times \vec{r})$$

In order to discriminize whether the movement of dislocations is conservative or not we distinguish between edge- and screw dislocations:

- for screw dislocations
  $\vec{b} \parallel \vec{ds} \Rightarrow dV = 0$ (because: $\vec{ds} \times \vec{r} \perp \vec{b}$)
- for edge dislocations
  $\vec{b} \perp \vec{ds} = \vec{n}$ (normal vector of the glide plane)
  - dislocation glide $\vec{r_1} \times \vec{ds} \parallel \vec{n}$ and hence:
    $\vec{r_1} \times \vec{ds} \perp \vec{b}$ (see Fig.1.16) $\Rightarrow dV = 0$

    This is called **conservative movement** of dislocations, because there is no volume change connected with it.
  - dislocation climb
    $\vec{r_2} \times \vec{ds} \parallel \vec{b} \Rightarrow dV \neq 0$

    This is called **non conservative** movement of dislocations, because there is a volume change connected with it.

Some conclusions:

- any screw dislocation moving through a crystal in any direction does not lead to a change of volume
- a dislocation glide mechanism as well does not remove or replace atoms from their original position while gliding
- but a dislocation climb process needs the replacement of atoms from their original locations and thus changes volume

To visualize this the two different processes have been drawn in Fig.1.17

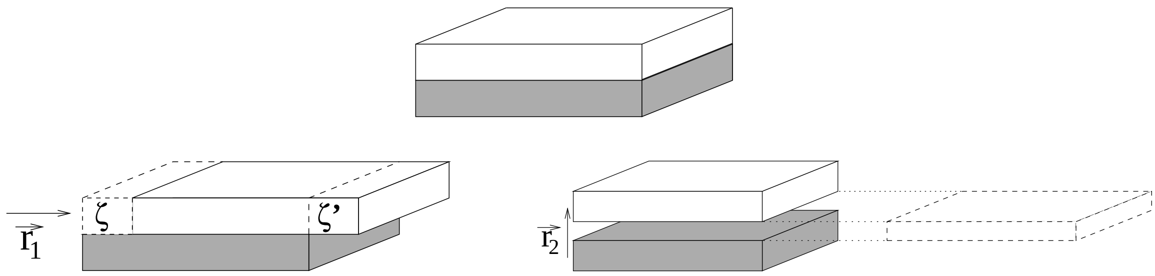

Figure 1.17: A glide process with $\vec{r_1}$ in the glide plane and a climb process with $\vec{r_2}$ perpendicular to it.

As the volume ζ is as large as ζ' no volume is produced due to gliding with $\vec{r} \parallel \vec{b}$ (left part in Fig.1.17). In contrast in the case of $\vec{r} \perp \vec{b}$ (right part) a volume is produced as sketched.

Conservative movement

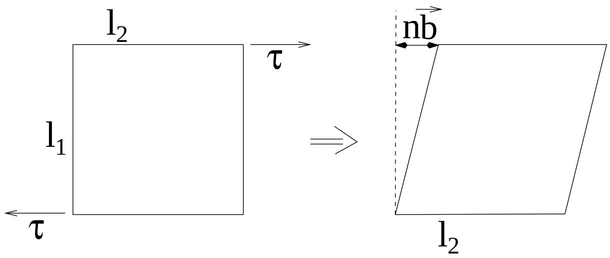

Figure 1.18: shear strain in a crystal

The shear strain in a crystal can be calculated from the average slip length of dislocations L and the number of dislocations $\#_{dl}$ as:

$$\gamma = \#_{dl} \frac{b}{l_1} \frac{L}{l_2}$$

making use of the definition of the disloaction density $\rho = \frac{\#_{dl}}{l_1 l_2}$ and of b as the absolute value of the Burger's vector $\vec{b}$ we obtain.

Orowan equation of plastic deformation

$\Rightarrow$

$$\gamma = \rho \frac{l_1 l_2 b L}{l_1 l_2} = \rho b L$$

time derivative $\Rightarrow$

$$\frac{d\gamma}{dt} = \dot{\gamma} = \rho b \frac{dL}{dt} = \rho b v$$

The shear strain rate is linked to the mean velocity of the dislocations, their density and the Burger's vector.

Non conservative movement

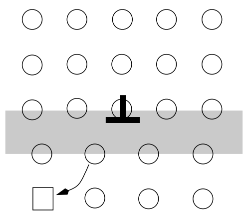

Figure 1.19: Process of dislocation climb; ⊥ represents the dislocation, ◯ atoms and □ a vacancy.

The scenario of a climbing edge dislocation is depicted two-dimensionally in Fig.1.19. One atom jumps on the place of a vacancy. This process reocurs until the lowest atom of the additional half plane jumps down und thus moves the dislocation upwards → dislocation climb.
However, only one atom has climbed and the whole row of atoms in the dislocation line has to follow. In Fig.1.20 this step-by-step climbing is shown. Hereby the third dimension of the dislocation comes into play.

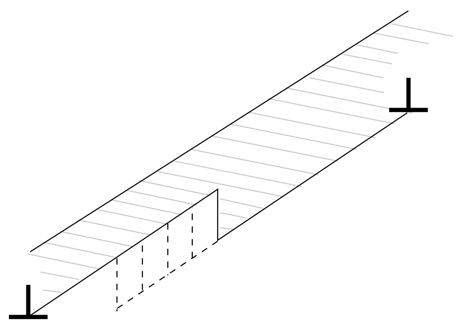

Figure 1.20: Dislocation climb of a whole row (dislocation line) as a step-by-step process of a jog moving.

There is an initial jog of atomic height which causes the dislocation climb.
Conclusions:

- climb is connected with vacancies and their movement
- diffusion is involved
- high temperatures are needed ($> 0.5T_{\mathrm{melt}}$)
- <u>but</u>: a whole column of atoms cannot be removed at once
  → do it sequentially
  → formation of jog for which the interaction of 1 vacancy and 1 atom is needed
  → movement of the jog parallel to the dislocation line $\vec{ds}$

vacancy concentration and movement is connected to Arrhenius-type equation i.e. $0.5T_{\mathrm{melt}}$ yields sufficient concentration of vacancies and their mobility.

How can a jog be created?

⇒ by intersection of dislocations on different glide planes!

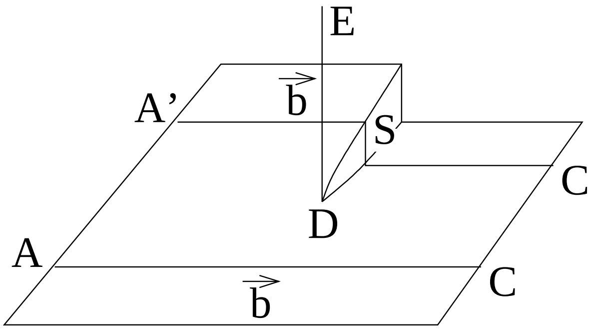

Figure 1.21: Two screw dislocations crossing each other.

In Fig.1.21 we have two screw dislocations with $\vec{b_1} \perp \vec{b_2}$. If the dislocation line $\overline{\mathrm{AC}}$ passes point D a jog is created by dislocation $\overline{\mathrm{DE}}$ (line S in $\overline{'\mathrm{AC}'}$. As a dislocation line cannot end in a crystal the segments left and right to the jog <u>have</u> to be linked by jog itself.
The resulting jog S in Fig.1.21 has edge configuration as the vector of the dislocation line is vertical and $\vec{b}$ is horizontal. Thus $\vec{ds} \perp \vec{b} \Rightarrow$ edge type configuration.

Conclusions:

- configuration $\overline{\mathrm{A'C'}}$ can no longer glide in the plane displayed as it has now an edge component S
- segment S has to climb with screw dislocation $\overline{\mathrm{A'C'}}$ (needs diffusion, high temperatures, ...)
- strain hardening if not mobile