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Abstract

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ON THE ROLE OF DIFFUSION OF POINT DEFECTS ALONG DISLOCATIONS IN THE PROCESS OF PLASTIC DEFORMATION

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Diffusion processes along the boundaries of blocks and grains, in comparison with volume diffusion, proceed considerably faster; this is closely connected with the preferential motion of point defects along dislocation lines ([1–3]). Most conclusions concerning the possibility of diffusion along dislocations are qualitative in character ([4–13]). Some estimates of diffusion along dislocations have been obtained for germanium ([14,15]). Investigation of the influence of molten metallic coatings on the strength of metals also leads to the conclusion that there is rapid diffusion of coating atoms into the body volume along dislocations ([16]).

The motion of point defects along a dislocation line can lead to a special diffusional interaction of dislocations. We have investigated this interaction on single-crystal zinc foil, which was obtained by electrolytic polishing of thin flakes split off from a large single-crystal specimen cooled in liquid nitrogen. In this process the (0001) plane proved to be oriented strictly parallel to the surface of the foil.

In such specimens, in the electron microscope, dislocations of four different types are visible. In the (0001) plane there are dislocations responsible for basal slip, with Burgers vectors $\frac{1}{3}\langle 11\bar{2}0\rangle$, and loops of prismatic dislocations formed during the collapse of planar accumulations of vacancies. Since in zinc crystals collapse occurs with a shear, the Burgers vectors of these dislocations prove to be equal to $\frac{1}{6}\langle\bar{2}203\rangle$, and a stacking fault is formed inside the loop ([13])*. For pyramidal slip, the responsible dislocations are those with Burgers vectors $\frac{1}{3}\langle 11\bar{2}0\rangle$, lying in the planes $\{10\bar{1}1\}$, and dislocations with Burgers vectors $\frac{1}{3}\langle 11\bar{2}3\rangle$. The last two types of dislocations intersect the foil at angles close to 90°, and are found predominantly in edge and screw orientation, respectively.

Figure 1 shows a characteristic pattern of the arrangement of dislocations in zinc foil, in which one can distinguish dislocations lying in pyramidal planes (designated A, B, and C) and prismatic loops. In some cases, under the action of the electron beam, dislocations in pyramidal slip planes begin to move. On encountering vacancy loops, they often seem to cut them (dislocation B in Fig. 1). Figure 2 shows successive stages of such a process.

The following features attract attention:

Figure 3 diagrams

Figure 1: Figure 3 diagrams

1. During cutting, the total area of the loop decreases. The width of the cutting zone fluctuates within small limits and is on average 750 Å.
2. Cutting is observed if the dislocation moves in the direction $\langle 11\bar{2}0\rangle$, characteristic of the motion of edge dislocations moving in the planes $\{10\bar{1}1\}$. Dislocations moving in other directions (apparently screw dislocations with Burgers vectors

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* The stability of such loops is apparently connected with frictional forces that hinder the diffusion of vacancies into the bulk of the crystal.

$1/3\langle 11\bar{2}3\rangle$), while crossing the prismatic ring, do not cut it. Figure 3a shows the scheme, inferred from the observations, of the interaction of a moving screw dislocation with a prismatic ring.

Usually, dislocations, upon encountering a ring, do not change the direction of their motion. In some cases, an edge dislocation, when passing through a vacancy ring, moves along a curvilinear trajec–

Fig. 3. Schemes of the interaction of a moving screw (a) and edge (b) dislocation with prismatic vacancy rings

tory, or along a rectilinear trajectory with a slight kink at the point of encounter with the ring.

4. A moving edge dislocation experiences resistance only at the moment of contact with the vacancy ring; at this point it sometimes comes to a halt. The subsequent cutting process proceeds freely, without noticeable resistance.
5. When the cutting dislocation stops inside the ring, no change in the configuration of the ring is observed. The only exception is rounding of the sharp points at the beginning of the cutting zone.
6. The entire process of cutting a ring of diameter $\sim 1\ \mu$ takes from one tenth of a second to several seconds. In this interval, the width of the cutting zone does not depend on the velocity of motion of the cutting dislocation.

The decrease in the area of the vacancy ring during cutting apparently occurs as a consequence of the flow of vacancies along the cutting dislocation. It should be noted that in the present case there is ascending diffusion [17] along dislocation lines, caused by the inhomogeneity of internal stresses and leading to relaxation of these stresses. This special interaction of dislocations may be called a diffusion interaction.

Figure 1

Figure 2: Figure 1

Figure 2

Figure 3: Figure 2

In Fig. 3b the successive stages of the diffusion interaction of gliding edge dislocations and vacancy dislocation rings are shown schematically. If the cutting dislocation has approached the ring and has entered inside the ring by a distance x_0, then the shape of the dislocation line $f(x)$ bounding the ring near the interaction site can be determined from the variational equation

$$\delta \int_0^{x_0} \left[\gamma f(x) - W\sqrt{1 + [f'(x)]^2}\right] dx = 0,$$

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Fig. 1. General view of the arrangement of dislocations in zinc foil, observed in a transmission electron microscope

Fig. 2. Successive stages in the cutting of a prismatic ring by an edge dislocation. 20 000×

where γ is the stacking-fault energy, W is the energy per unit length of dislocation. Integration of the first term takes into account the decrease in energy due to reduction of the stacking-fault area, and of the second, the increase in energy due to the growth of the length of the dislocation line. The solution of this equation gives the expression for $f(x)$

$$f(x) = \frac{W}{\gamma}\sqrt{1 - \frac{\gamma^2}{W^2}\left(x - x_0 + \frac{W}{\gamma}\right)^2},$$

which describes a circle of radius W/γ.

After the cutting dislocation has advanced into the ring by an amount equal to the radius $r = W/\gamma$, further cutting proceeds without expenditure of energy, since the increase in energy due to the lengthening of the dislocation bounding the ring is completely compensated by the gain in energy due to the decrease in the area of the stacking fault. The width of the cutting zone $l = 2r$ is determined only by the stacking-fault energy and the dislocation energy and must be constant for a given type of crystal, as is also observed experimentally. If one takes $W = 16.3 \cdot 10^{-5}$ erg/cm*, then for a channel width $l = 750$ Å the stacking-fault energy for zinc proves to be $\gamma = 43$ erg/cm^2, which is in satisfactory agreement with the data of other authors ([13,20]). Thus, diffusive cutting of vacancy rings can be used to determine the stacking-fault energy. This method is especially convenient in those cases where the stacking-fault energy is large and the stretching of the nodes is small.

If it is assumed that all vacancies absorbed during cutting are consumed in shortening the extra half-plane of the cutting edge dislocation, then, for a foil thickness of ~ 800 Å, this should lead to a change in the direction of motion of the dislocation by $\sim 45^\circ$. But since in most cases the direction of motion practically does not change, it may be concluded that the majority of the vacancies flow to the surface of the foil. The sometimes observed curvilinear motion of the cutting edge dislocation is associated with the fact that in individual cases part of the vacancies may be consumed in shortening the extra half-plane, which leads to nonconservative displacement of the dislocation. From the observations one can estimate the magnitude of the vacancy flux along edge dislocations. The shortest time for cutting a ring of diameter $\sim 1\,\mu$ is ~ 0.1 sec. For a channel width of 750 Å, the vacancy flux along the cutting dislocation is estimated to be $\sim 10^7$ sec^{-1}. However, there is no certainty that in reality a larger flux cannot occur, since the cutting time is limited by the possibilities of observation, and not by a decrease in the channel width with increasing velocity of the cutting dislocation. It is also possible that during cutting some of the vacancies diffuse into the bulk.

In those cases where, as a result of diffusive cutting of vacancy rings, protrusions of large curvature are formed, their gradual rounding is observed. Such a case can be seen in the series of photographs in Fig. 2 in the region marked by arrows. It should be especially emphasized that in this process the area of the loop is preserved. Consequently, rounding of the ring occurs by diffusion of vacancies along the dislocation line bordering the stacking fault. Diffusion along the stacking fault is apparently small, for the presence of a dislocation inside the ring (dislocation C in Fig. 1) does not lead to a decrease of its area.

One can write an equation describing the transformation of a dislocation contour, by analogy with problems of surface diffusion ([21,22]). The chemical potential of a dislocation line of curvature χ is

$$\mu(\chi) = \mu_0 + Wa^2\chi,$$

* The linear energy of the dislocation was determined taking into account both the normal and the parallel to the plane (0001) components of the Burgers vector, as well as the anisotropy of the medium ([18,19]).

where μ_0 is the chemical potential of a rectilinear dislocation, a is the atomic size. The gradient of the chemical potential along the dislocation will determine the diffusion force, and the flux

$$j = -\frac{D_l}{kTa}\frac{\partial \mu}{\partial s} = -\frac{D_l W a}{kT}\frac{\partial \chi}{\partial s},$$

where D_l is the coefficient of linear self-diffusion. The differentiation is carried out along the arc. The normal velocity of displacement of points of the dislocation contour due to the inflow of vacancies will be determined by the equation

$$V_n = \frac{D_l W a^3}{kT} \frac{\partial^2 \chi}{\partial s^2}.$$

Since in the last expression the second derivative of the curvature is difficult to estimate experimentally, for calculating the diffusion coefficient it is more convenient to use the expression for the flux. An estimate of the coefficient of linear diffusion leads to a value of 10^{-9} cm^2/sec at a temperature close to room temperature, which is 10 orders of magnitude higher than the value of the coefficient of volume diffusion.

The large value of the coefficient emphasizes the special role of dislocations as channels through which point defects can be pumped from one regions of the crystal to another. This may be the reason for the climbing of interacting dislocations, the acceleration of the nonconservative motion of jogs, the twisting of dislocations having a screw component of the Burgers vector into helicoids, etc. The high mobility of point defects along dislocation lines may be especially substantial at high dislocation density, when their mobility is limited and the formation of a cellular structure from entangled dislocations leads to a sharp inhomogeneity of internal stresses. It may be thought that the third stage of creep is determined to a considerable extent by the semiconservative displacement of dislocations and by ascending self-diffusion along dislocation lines. The directed flux of vacancies along dislocations may also lead to the formation of micropores, which are one of the causes of the fracture of metals in long-term tests ([23–25]).

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Note: Figure translations are in progress. See original paper for figures.