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Abstract

Full Text

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PHYSICS

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ON PYRAMIDAL SLIP IN ZINC CRYSTALS

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The magnitude of the mechanical stress that causes the onset of dislocation motion is the most important parameter in a microscopic description of plastic deformation by slip and represents a physical characteristic of the crystal. Determination of this stress must be carried out on dislocations that are not impeded by Cottrell atmospheres, i.e., on "fresh" dislocations. Otherwise the value of the starting stress will be overestimated. It is natural to measure the starting stress on dislocations that are generated during the measurement process, since in this case they do not have time to become surrounded by impurities.

In the present work the stress required to initiate dislocation motion in the second-kind, second-order pyramidal plane $(\bar{1}22)^*$ in the $[0\bar{1}1]$ direction in zinc crystals was measured. This slip system was discovered in work (1).

Fig. 1. Loading scheme and crystallographic orientation of zinc crystals

Measurement of the starting stress for dislocation motion was carried out on zinc single crystals of 99.98% purity, grown by a method that makes it possible to obtain crystals of specified shape and specified crystallographic orientation. After growth the crystals were annealed at a temperature of 360° for 8 hours to relieve stresses arising during growth. From the annealed crystal, at liquid-nitrogen temperature, specimens in the form of prisms with dimensions $12 \times 2.5 \times 3$ mm^3 were cleaved along the cleavage plane (001). The specimens were oriented as shown in Fig. 1, where it is seen that the lateral surfaces of the specimen are basal planes (001), and the upper and lower surfaces are first-kind prism planes (100).

Fig. 3. Distribution plot of σ_c along the [010] axis of a zinc specimen

Figure 2: Fig. 3. Distribution plot of σ_c along the [010] axis of a zinc specimen

The zinc crystals under investigation were subjected to pure bending deformation. With the chosen orientation of the specimen, the components of stress in the easy-slip planes (001) are absent, and slip in these planes will not occur; whereas in the pyramidal planes $(\bar{1}22)$ there is always a component acting in the slip direction $[0\bar{1}1]$.

Before loading, the basal plane (001) of the zinc crystal on which observation was carried out was subjected to selective etching (^{2}). The crystal was then deformed, the load was measured, and after its removal the crystal was again subjected to selective etching. Under the action of the load, dislocations arise in the crystal; their screw components are revealed on the basal plane in the form of chains of etch pits originating from the upper and lower edges of the specimen, where the stresses are maximal.

* The classification of planes and directions in the hexagonal system is given in the three-axis coordinate system, since only three indices are required for calculating angles.

Depending on the applied load, the chains of etch pits have different lengths. The greater the load, the closer the dislocations move to the neutral line. Figure 2 shows the basal plane of a zinc crystal after loading and selective etching. It is seen that the dislocation etch pits stop at almost the same distance from the neutral line. Evidently, this distance corresponds to the condition $\sigma_\mathrm{B} \leqslant \sigma_\mathrm{c}$, where σ_B is the external stress and σ_c is the starting stress for screw dislocations lying in the pyramidal plane. The chains of etch pits are located along the trace of the pyramidal plane $(1\bar{2}2)$ on the (001) plane in the direction [210].

Fig. 3. Distribution plot of σ_c along the [010] axis of a zinc specimen

The stresses applied to the crystal under study in the process of pure-bending deformation can be calculated by formula (^{3}),

$$\sigma_y = \frac{6PK}{BH^3} y, \tag{1}$$

where P is the load in grams, y is the minimum distance from the neutral axis to the level at which the dislocations stop (see Fig. 2), and the values of K, B, and H are indicated in Fig. 1.

The resolved stress σ_c, acting in the pyramidal plane in the direction of slip, i.e., the starting stress for pyramidal dislocations, is determined by the formula:

$$\sigma_\mathrm{c} = \sin\psi \cos\varphi \sigma_y, \tag{2}$$

Fig. 2

Figure 3: Fig. 2

Fig. 4

Figure 4: Fig. 4

where ψ is the angle between the slip plane $(1\bar{2}2)$ and the direction of the specimen axis [010], and φ is the angle between the slip direction [011] and the specimen axis.

The values of the starting stresses for screw dislocations are given in Table 1,

Table 1

Specimen number	1	2	3	4	5	6	7	8	9	10	11	12	13	14	15	16
σ_c, G/mm^2	76	89	169	201	72	84	82	79	74	75	117	101	136	137	75	78

from which it is seen that the magnitude of the stresses varies from experiment to experiment by a factor of $2 \div 3$. The scatter of the starting-stress values becomes understandable if one takes into account such uncontrolled factors affecting the results as: a) the presence of basal dislocations, whose density is unknown; b) the presence of various structural imperfections in the form of block boundaries, cleavage steps, low-angle boundaries, etc.; c) fluctuations in the initial density of pyramidal dislocations. The average value of the starting stress is 105, the minimum 72 and the maximum 201 G/mm^2.

As can be seen from Fig. 2, not all chains of dislocations are at the same distance from the neutral axis, which leads to a scatter of starting-stress values even within a single specimen. This scatter may be associated with differing degrees of crystal perfection in its different parts. In such a case the degree of scatter of σ_c may serve as a measure of the perfection of the crystal.

Figure 3 shows the distribution of the starting stress σ_c along the [010] axis in the middle part of a zinc specimen. It is clearly seen that the overwhelming majority of dislocation chains correspond to very close values of the starting stresses. This indicates that either the crystal has a fairly perfect and homogeneous structure throughout its volume, or that the existing imperfections do not affect the value of the starting stress. The latter is unlikely, since the starting stress must be a structure-sensitive property of the crystal.

Fig. 2. Chains of dislocation etch pits on the (001) plane of zinc crystals after deformation and selective etching. 140×.

Fig. 4. Motion of dislocations on the (001) plane in zinc crystals.
a—before loading, dislocation in position 1; *b*—after loading, dislocation made transverse slip to position 2, 300×; *c*—dislocation under the action of an alternating load: 1—initial position of the dislocation, 2—position of the dislocation after loading, 3—position of the dislocation after repeated loading of opposite sign, 540×.

To determine the starting stress, another method was also used. In this method, an individual dislocation was observed and the stress necessary for its displacement was determined. In Fig. 4a, the dislocation etch pit 1 on an "old"dislocation before application of the load is shown. After the load was applied and selective etching was carried out, the dislocation moved to a new position. This is clearly seen in Fig. 4b, where the flat-bottomed etch pit $1'$ marks the location of the dislocation before loading, corresponding to position 1 in Fig. 4a. The smaller, pointed etch pit 2 corresponds to the new location of the dislocation. The stress under the action of which this dislocation moved is equal to 496 g/mm^2. This stress is almost three times greater than the maximum value given in Table 1. Such a discrepancy can be explained by the pinning effect of impurities, which form so-called Cottrell atmospheres (4) around dislocations. Indeed, if a load is applied to a "fresh" dislocation, it turns out that this load is much smaller than the load necessary to initiate motion of an "old" dislocation. Figure 4v shows a dislocation for which, to initiate motion, a load of 4 kg was applied. Under the action of this load, the dislocation moved from position 1, which is marked by a large flat-bottomed pit, to position 2 (a flat-bottomed pit of smaller size). The distance through which the dislocation moved is equal to 11 μ. Then a load of 2 kg of the opposite sign was applied to the specimen; under its action the dislocation moved from position 2 to position 3, which is marked by a pointed etch pit. It is clearly seen that, under the action of a load two times smaller, the dislocation moved a distance of 33 μ, i.e., three times farther than under the first loading (in this experiment it was not possible to determine the magnitude of the starting stress, since the position of the neutral line was unknown).

This experiment shows the strong pinning action of Cottrell atmospheres on dislocation motion. However, it should be noted that motion of "old"dislocations is observed very rarely. Almost always the appearance of new dislocations is observed without displacement of the "old"ones. This also indicates the influence of impurities, which envelop dislocations and thereby increase the stress required for their displacement.

Along with the motion of screw pyramidal dislocations in their slip plane $(1\bar{2}2)$, as shown in Fig. 4v, transverse motion of dislocations occurs. In Fig. 4b it is seen that dislocation $1'$ moved in the direction [010], which is normal to the trace of the slip plane $(1\bar{2}2)$ on the basal plane [001].

It is of interest to compare the magnitude of the critical shear stress for the slip system $(1\bar{2}2)$, [011] of a zinc crystal from work (1), measured from the appearance of macroscopic slip lines, with the starting stress measured in the present work. The critical shear stress in work (1) for the pyramid plane is equal to 1-1.5 kg/mm^2, which is almost 14 times greater than the minimum starting stress for motion of pyramidal dislocations.

Such a discrepancy is probably due to a more precise determination of the onset of plastic deformation in the present work, in comparison with the method of the appearance of the first slip bands used in work (1).

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