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MOTION OF TWINNING DISLOCATIONS IN CALCITE

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The rearrangement of the crystal lattice during mechanical twinning of crystals, according to well-known concepts, is brought about by the motion of twinning dislocations in the twinning plane under the action of a mechanical load ([1,2]). The elementary act of this process consists in the dislocation moving in the direction of shear from one equilibrium position to a neighboring one, located from it at a distance equal to the magnitude of the Burgers vector. However, this can occur only if the mechanical stress in the direction of shear exceeds some critical value, which depends on the magnitude of the "barrier" between two equilibrium positions and on the distortion of the lattice in the dislocation core. In other words, the force acting on the dislocation must exceed the resistance force of the crystal lattice to the motion of the dislocation.

The value of the mechanical stress that causes the onset of motion of a twinning dislocation is the most important quantity in a microscopic description of the twinning process and represents a physical characteristic of the crystal. The present work describes experiments to determine this quantity for calcite.

1. To observe the motion of twinning dislocations, the method of repeated etching was used, proposed by Gilman and successfully applied by him to the study of dislocation motion in lithium fluoride crystals ([3]). In calcite, the motion of twinning dislocations along the boundary of a twin lamella was first observed in work ([4]), where the influence of brief annealing on the distribution of twinning dislocations along the boundary of a twin lamella was studied. In these experiments, the motion of twinning dislocations could be either due to the interaction of dislocations with one another and also with the surface of the crystal, or the result of the action of thermoelastic stresses arising during nonuniform heating of the crystal.

However, these experiments do not give direct confirmation of the motion of dislocations under the action of a mechanical force. It should be said that direct observation of such dislocation motion is experimentally difficult. Indeed, if one observes the motion of a twin boundary during expansion of a twin lamella under the action of an applied load, it is easy to see that the distribution of dislocations on the boundary changes sharply during the expansion process ([5]). The number and arrangement of etch pits for two positions of the boundary are completely different, and there is no correspondence between them. Such a result was to be expected, since displacement of a dislocation along the entire twinning plane leads to displacement of the twin boundary by only one lattice constant. Consequently, the rate of loading and the magnitude of the load must be such that, under its action, the twin lamella does not expand, while the dislocations on the twin boundary move only over small distances.

The experiment in which the motion of dislocations under the action of a mechanical load was observed was carried out as follows. A previously etched calcite crystal with a twin interlayer was placed on the stage of a microhardness-measuring instrument (PMT-3), and, by means of a diamond pyramid, a load was applied perpendicular to the cleavage plane of the crystal at a preselected point of the crystal. The motion of the dislocations occurred under the action of stresses produced in the crystal by the concentrated load. The results of the experiment are seen in Fig. 1, which shows a photograph of a portion of a twin boundary, near which an indentation by the diamond pyramid had been made (to the right, below the twin boundary).

The "flat-bottomed" etch pits mark the initial positions of the twinning dislocations before indentation by the diamond pyramid. After indentation by the diamond pyramid (at a load of 10 g), the dislocations moved to the left along the boundary, and the pointed etch pits mark their new positions. Let us note that the problem of the distribution of stresses in a half-space for the case of a concentrated force is solved completely in the theory of elasticity, and it is possible to calculate the external stress, acting in the direction of shear from the applied concentrated force, for the point where the dislocation line emerges at the surface. By varying the magnitude of the load acting on the pyramid, one can determine the stress that leads to the onset of motion of a twinning dislocation (the starting stress).

However, in the present experiment it is difficult to specify the magnitude of the stress that acted on each dislocation, since the dislocations cannot be regarded as isolated. In calculating the forces acting on an individual dislocation, in addition to the external force it is necessary to take into account the stress field produced by neighboring dislocations. If one calculates the stress leading to the onset of motion of a twinning dislocation under the assumption that the stress field from the other dislocations may be neglected (which is clearly incorrect), then the obtained values of the dislocation starting stresses show a large scatter. For example, for the twinning dislocations shown in Fig. 1, the values of the starting stress lie within the range $60 \div 15$ g/mm^2. The magnitude of these stresses differs in different experiments and depends on the number and nature of the arrangement of neighboring dislocations. Nevertheless, it seems to us that such measurements may be of significance, since they make it possible to determine experimentally the interaction forces between twinning dislocations.

2. There is, however, a case of mutual arrangement of dislocations in which the starting stress for a twinning dislocation can be determined much more

Figure 1

Figure 1: Figure 1

Figure 2

Figure 2: Figure 2

simply.

Let us imagine that a twinning dislocation, while moving in the twinning plane, has encountered some obstacle and has been stopped. In turn, it will stop the motion of dislocations moving after it in neighboring twinning planes. If the external load is removed, then, under the action of the forces of mutual repulsion, the dislocations will move apart and assume a quite definite arrangement, characterized by the interaction forces between them. In this case each dislocation in the pile-up will move under the action of the repulsive forces of the remaining dislocations until this force becomes equal to the force resisting the motion of the dislocation in the twinning plane of the crystal.

To calculate the resistance force, one may use the results of work ([6]). In that work, the problem of finding the equilibrium positions of a system of identical free gliding dislocations accumulated near a fixed dislocation under the action of a constant applied stress is considered in detail. The problem is solved in the plane case for an isotropic* medium, under the assumption that the dislocations are infinite parallel lines. With certain additional simplifications–

* Allowance for anisotropy cannot change the order of magnitude of the quantities obtained in the calculation.

To the article by V. Z. Bengus, S. N. Komnik, and V. I. Startsev, p. 607

Fig. 1. Displacement of twinning dislocations under the action of a mechanical load–an indentation by a diamond pyramid (on the right, beneath the twin boundary, outside the field of view). Flat-bottomed pits A_1, A_2, A_3, A_4, A_5 mark the initial position of the dislocations. Pointed pits $A_1', A_2', A_3', A_4', A_5'$ mark the position of the dislocations after displacement, 580×.

Fig. 2. Accumulation of twinning dislocations along the twin boundary at obstacles.
a–the obstacle is a pinned twinning dislocation; 450×.
b–the obstacle is a complete dislocation D, intersecting the twinning plane deep in the crystal, at a distance of $25\,\mu$ from the surface. 600×.

To the article by E. F. Polikarpova and M. V. Nevzgodina, p. 758

Figure 1

Figure 3: Figure 1

Figure 2

Figure 4: Figure 2

Fig. 1. Section of the thyroid gland of a newborn single Romanov-breed lamb. 200×. *a*—thyroid follicles, *b*—epithelium of a thyroid follicle.

Fig. 2. Section of the thyroid gland of a newborn quintuplet Romanov-breed lamb. 200×. a, b—same as in Fig. 1; *v*—undifferentiated thyroid tissue.

Under the corresponding assumptions, the authors of work ([6]) obtained formulas determining the position of individual dislocations in a pile-up.

The formula for the total length of a pile-up L, i.e., for the distance between the leading and the last dislocation in the pile-up, is

$$L = \frac{n}{\sigma}\frac{Gb}{\pi K}, \tag{1}$$

where n is the number of dislocations in the pile-up; σ is the external stress acting on the dislocations; G is the shear modulus, taken to be $3.2 \cdot 10^{11}$ dyn/cm^2 ([7]); b is the Burgers vector; $K = 1 - \nu$ (for edge dislocations); ν is Poisson' s ratio, taken to be 0.3.

We are interested, however, in the case where for a dislocation in the crystal there exists a starting stress σ_0, and in this sense they are no longer free dislocations. Then the equilibrium condition for the forces acting on each dislocation of the pile-up,

$$\sum \sigma_i = 0, \tag{2}$$

is replaced by the condition

$$\sum \sigma_i < 0, \tag{3}$$

where σ_0 is the starting stress, while σ_i takes into account the various forces acting on the dislocation (external forces, forces of interaction with other dislocations, etc.).

The analysis carried out above of the kinetics of the process by which an equilibrium distribution of dislocations in a pile-up is established makes it possible to suppose that the resistance to dislocation motion is the same for each dislocation and is equal to the starting stress. Therefore it can be taken into account, when solving the problem, as an external stress, and then the problem reduces to that already solved in ([6]). When the applied external stress is absent, its role is played by σ_0, and its value can be obtained from formula (1):

$$\sigma_0 = \frac{n}{L}\frac{Gb}{\pi(1-\nu)}. \quad (4)$$

The indicated pile-ups are observed quite frequently; moreover, both pile-ups thickening in the direction $[00\bar{1}]$ and pile-ups thickening in the direction [001] occur. Typical photographs of dislocation pile-ups at obstacles are given in Figs. 2a and 2b, which differ from one another only in the nature of the obstacles that stop the twinning dislocations. For the boundary shown in Fig. 2a, the obstacle is a pinned twinning dislocation, while in Fig. 2b the obstacle is the line of a complete dislocation that does not lie in the twinning plane but intersects it. It should be noted, however, that formula (1) was obtained under the assumption that the obstacle has sufficient length and that the dislocations are straight lines, so that the force acting on each segment of the dislocation line is the same along its entire length (in this case the "line tension" of the dislocation may be neglected). Therefore formula (4) cannot be applied to the pile-up shown in Fig. 2b.

The results of measurements made for a number of typical photographs are given in Table 1.

Table 1

n	L, cm	σ_0, g/mm^2	n	L, cm	σ_0, g/mm^2
130	0.1126	45	26	0.0111	91
96	0.0635	59	357	0.1500	94
26	0.0170	60	27	0.0112	95
23	0.0134	67	33	0.0127	102
15	0.0068	86	45	0.0128	138

The observed scatter in the values of σ_0 may be due to the influence of the curvature of the dislocations, to their not purely edge character, and also to the different distances from the crystal surface of the obstacles pinning the dislocations on which the observation is being made. In this connection we note that, in the depth of the crystal, the general pattern of the distribution of twinning dislocations along the boundaries of twin lamellae is more uniform than at the crystal surface. This is apparently due to the circumstance that, near the surface, there are far more obstacles to the motion of twinning dislocations than inside the crystal, and is consistent with the fact that near the surface of a twinned crystal one usually observes more dislocations and other defects than in its interior ([8]).

Deep etching of the dislocation accumulations under study does not change their general character, although a compaction of the dislocations in the accumulation is often observed. Near the leader this compaction may reach such a value that, on etching, the individual pits become indistinguishable, and in counting this leads to an apparent decrease in the number of dislocations in the accumulation. Such a decrease will be insignificant for an accumulation with a large number of dislocations.

The mean value of the starting stress in the experiments described is 87 G/mm^2. The minimum value of the starting stress in our experiments was 45 G/mm^2, and the maximum 138 G/mm^2. The force required to begin the expansion of a twin lamella, referred to a single dislocation, would never exceed the starting stress if there were no obstacles in the crystal hindering the motion of twinning dislocations. Then these dislocations would move freely in the twinning planes if the external mechanical forces produced in them stresses equal to or greater than the starting stress. In reality, however, obstacles of various kinds—perfect dislocations intersecting the twinning plane, sessile dislocations, inclusions, etc. —hinder and stop twinning dislocations. It is evident that, in the process of overcoming these obstacles, phenomena occur that contribute to the hardening of the crystal during twinning.

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