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Abstract

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MICROSTRUCTURE OF STRESSES IN SLIP LINES AND DISLOCATIONS*

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The dislocation structure of the boundaries of misoriented blocks, in the light of recent experimental data, gives rise to no particular doubts, but the question of the dislocation nature of slip lines still remains insufficiently clarified. In recent works by R. I. Garber, I. V. Obreimov, and L. M. Polyakov ([1]), and by A. A. Bochvar and Yu. A. Preobrazhenskaya ([2]), doubt is again expressed as to the existence of a connection between slip lines and ordinary translational shear, and the possibility is discussed that slip lines appear as a result of the successive opening and closing of microcracks. In fact, the authors mentioned assume that the discreteness of translation is characterized by displacements over many interatomic distances, whereas according to dislocation concepts the discreteness of translation corresponds to steps of one lattice parameter along the direction of slip, i.e., to such a successive reconnection of interatomic bonds as if "extra" half-planes were moving in the crystal. In the simplest dislocation model of a slip line, the boundary of sliding blocks may be represented as a sequence of equally spaced edge dislocations lying in a common slip plane (a "horizontal" row of edge dislocations). Each dislocation line corresponds to the edge of an "extra" plane ending at the slip plane. The macroscopic stresses bordering slip lines, observed by many authors and quantitatively studied in the works of I. V. Obreimov and L. V. Shubnikov ([3]) and Nye ([4]), according to this scheme are determined unambiguously by the dislocation density.

In work ([5]) we showed, as applied to a corundum crystal, that comparison of the stresses measured by the optical method with the dislocation density found by the method of selective etching gives satisfactory agreement between calculated and experimental data. Thus, measurement of the macroscopic stresses made it possible to establish that each etch figure, presumably corresponding to the end of an atomic dislocation line, is indeed accompanied, on the average, by a shear approximately equal to the lattice parameter in the direction of slip (for Al_2O_3, 8.3 Å).

For a final solution of the question of the dislocation structure of slip lines, however, it is necessary to resolve the fine structure of the stress field, to reveal the effects caused by individual dislocations, and to show that this structure corresponds to the assumed arrangement of dislocations. The distribution of stresses around a single edge dislocation with an "extra" plane parallel to the Y axis and with Burgers vector b directed along the X axis is given by the stress function

$$\psi_0 = -\frac{Gb}{2\pi(1-\nu)}\, y \ln\sqrt{x^2+y^2} \tag{1}$$

(G is the shear modulus, ν is Poisson's ratio). To calculate the stresses caused by a horizontal row of edge dislocations, it is necessary to sum the stress functions corresponding to all dislocations of the row.

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To simplify the calculations, let us pass to the coordinates $z = x+iy$, $z^* = x-iy$. The stress function of the n-th dislocation, located at the point $x = nh,\ y = 0$, is written in the form

$$\psi_n = \frac{iGb}{4\pi(1-\nu)}(z-z^*)\ln\left|\frac{z-nh}{nh}\right|. \tag{2}$$

In order to prevent divergence in the subsequent summation, an extra term, linear in the coordinates and making no contribution to the values of the stresses, which are determined by the second derivatives of the stress function, has here been added in comparison with expression (1):

$$\frac{\sigma_x+\sigma_y}{2} = 2\frac{\partial^2\psi}{\partial z\partial z^*}, \tag{3}$$

$$\frac{\sigma_x-\sigma_y}{2} - i\tau_{xy} = -2\frac{\partial^2\psi}{\partial z^2}. \tag{4}$$

Using the well-known representation [6]

$$\sin x = x\prod_{n-1}^{\infty}\left(1-\frac{x^2}{\pi^2 n^2}\right), \tag{5}$$

one can calculate the total stress function

$$\psi = \sum_{n=-\infty}^{\infty}\psi_n = \frac{iGb}{4\pi(1-\nu)}(z-z^*)\ln\left|\frac{1}{\pi}\sin\pi\frac{z}{h}\right|. \tag{6}$$

With the aid of formulas (3) and (4) we obtain from this

$$\frac{\sigma_x + \sigma_y}{2} = -\frac{Gb}{2(1-\nu)h}\frac{\mathrm{sh}(2\pi y/h)}{\mathrm{ch}(2\pi y/h) - \cos(2\pi x/h)}, \tag{7}$$

$$\frac{\sigma_x - \sigma_y}{2} = -\frac{Gb}{2(1-\nu)h}\frac{\partial}{\partial y}\left(\frac{y\,\mathrm{sh}(2\pi y/h)}{\mathrm{ch}(2\pi y/h) - \cos(2\pi x/h)}\right), \tag{8}$$

$$\tau_{xy} = \frac{Gb}{2(1-\nu)h}\frac{\partial}{\partial y}\left(\frac{y\sin(2\pi x/h)}{\mathrm{ch}(2\pi y/h) - \cos(2\pi x/h)}\right). \tag{9}$$

Expression (7) characterizes the density field, which determines, in particular, light scattering by the active slip plane. Owing to this effect, slip lines are sometimes visible in polarized transparent crystals. Expressions (8) and (9) determine the birefringence field caused by the slip line. At distances large in comparison with the distance between dislocations, σ_y tends to zero, while σ_x approaches the macroscopic value

$$\sigma_x = -\frac{Gb}{(1-\nu)h}\,\mathrm{sign}\, y.$$

Fig. 1 illustrates the calculated polarization-optical pattern that should be observed in crossed nicols when the axes of the polarizer and analyzer are arranged diagonally with respect to the slip line (Fig. 1, *a*) and parallel to the slip line and the "extra" planes (Fig. 1, *b*). In the first case, usually used in the study of slip lines, the birefringence field is described by formula (8). In the macroscopic bands of birefringence of different sign bordering the slip line, a narrow transition band and local distortions near the lines of dislocations are revealed, similar to the birefringence pattern corresponding to a single dislocation (Fig. 1, *c*([7])). At distances of the order of 20% h, the birefringence is already practically indistinguishable from the macroscopic one.

More convenient for studying the microstructure of stresses is the second case (Fig. 1, *b*), when the bands of macroscopic birefringence are extinguished and opposite each dislocation line there is located a characteristic

Fig. 2. General view of one of the corundum specimens examined.
The polarization-optical pattern (a) agrees with the pattern of the arrangement of etch figures (b). A, B, C are slip lines; D are twin boundaries. 30×

Fig. 3. Structure of slip lines. *a* –stress microstructure revealed by the optical method; *b* –arrangement of etch figures in the same regions of the specimen. The designations of Fig. 2 are retained. 100×

a six-petal rosette with opposite signs of birefringence in neighboring petals, similar to the rosette corresponding to a single dislocation (Fig. 1, *v* ([7,8])). But even in this case, for the optical resolution of individual dislocations it is, of

Figure 1

Figure 1: Figure 1

course, necessary that the dislocations not be bent or inclined with respect to the direction of observation. Otherwise the effects from individual lines may overlap, and the dislocations cannot be resolved.

It should be emphasized that in none of the numerous works devoted to the optical investigation of slip lines has the fine structure illustrated in Fig. 1 been detected. Assuming that the main reason for this could be the superposition of effects corresponding to individual dislocations, we chose for investigation thin corundum plates cut from specimens with sparse slip lines. The advantages of corundum as the most promising material for revealing atomic dislocations may be judged from Table 1, which relates the dislocation density to the intensity of the macroscopic stresses bordering slip lines. In the calculations, the anisotropy of the elastic moduli and photoelastic constants was taken into account, and the constants given in works [9] (NaCl), [10,11] (LiF), [12] (Tl(BrJ)), [5] (Al_2O_3) were used.

Fig. 1. Polarization-optical pattern corresponding to a slip line. Calculated for the condition that the cross of the axes of the polarizer and analyzer is positioned diagonally (*a*) and parallel (*b*) to the slip line. For a unit, the birefringence corresponding to the macroscopic stresses is taken. A single edge dislocation at these orientations of the polarizer and analyzer gives, respectively, the rosettes *v* and *g*.

The sensitivity of the optical method for detecting dislocations is determined by the product of the Burgers vector by the corresponding photoelastic constant and the effective stiffness of the crystal. In the isotropic case the effective stiffness is equal to $2G/(1-\nu)$.

In NaCl crystals, which are most often used for the optical investigation of slip lines, birefringence of the order of 1 μ/cm is attained only when the distance between dislocations is of the order of the resolving power of the optical microscope. Owing to their great stiffness, Al_2O_3 crystals prove to be more convenient than Tl(Br, J) crystals, which are known for their high optical activity.

Figure 2 shows the general appearance of one of the specimens investigated. The plates were cut along the basal plane from boules of synthetic leucosapphire and were ground and polished to a thickness of the order of 0.5 mm. Observation in polarized light reveals slip lines and twins (Fig. 2, *a*). With the diagonal orientation of the polarizer and analyzer, the macroscopic stresses bordering the slip lines are clearly visible. Etching of the same specimen in a melt of potassium bisulfate or in boiling orthophosphoric acid (Fig. 2, *b*) gives a pattern of slip lines and twin boundaries that agrees well with the optical picture. Upon repeated grinding and etching, only the positions of the etch figures associated with random scratches on the surface of the specimen change.

Figure 3 illustrates the pattern of microstresses in slip lines that we found after the birefringence bands corresponding to the macroscopic stresses had been extinguished. In the same figure

photographs are presented showing the distribution of etch figures on the same regions of the specimen. The most striking picture, observed in line *A*, turns out to be much more complex than the calculation scheme of Fig. 1, b. In fact, the optical picture resembles a rope twisted from three strands rather than a single row of rosettes. It is interesting to note, however, that etching of this line gives a triple row of figures, i.e., in the present case we are indeed dealing not with a line but with a slip band.

Table 1

Resolving power of the optical method for investigating dislocations

	NaCl	LiF	Tl (Br, J)	Al_2O_3
Slip plane	(110)	(110)	(110)	$(1\bar{1}20)$
Slip direction	$[\bar{1}10]$	$[\bar{1}10]$	[001]	$[\bar{1}100]$
Direction of observation	[001]	[001]	$[\bar{1}10]$	[0001]
Photoelastic constant, cm^2/kg	$1.55 \cdot 10^{-7}$	$1.31 \cdot 10^{-7}$	$28.5 \cdot 10^{-7}$	$2.1 \cdot 10^{-7}$
Effective stiffness of the crystal, kg/cm^2	$3.6 \cdot 10^5$	$14 \cdot 10^5$	$2.9 \cdot 10^5$	$37 \cdot 10^5$
Burgers vector, Å	3.97	2.85	4.15	8.3
Distance between dislocations (in μ) at which the jump in birefringence reaches 10^{-4}	0.22	0.52	3.4	6.4

A less complex optical picture is observed in line *B*, corresponding to a double row of etch figures, but line C corresponds most closely to a single slip line. Here a single row of etch figures corresponds to a single-row sequence of birefringence rosettes. Even in the latter case it is difficult to distinguish individual rosettes and compare them with the calculated picture of Fig. 1, b; however, the correspondence between the density of the etch figures and the structure of the birefringence field is striking. This allows one to hope that the microstresses discovered are the very microstresses that appear in the dislocation scheme of a slip line and correspond to the idea of the atomic discreteness of translational shear.

The microstructure of stresses was noted by us in practically all the slip lines and bands found in the investigated specimens of synthetic corundum. In some cases, an increase in the density of etch figures and in the intensity of birefringence was observed in slip lines near the boundaries of twins and mosaic blocks, which apparently corresponded to an accumulation of dislocations at obstacles.

In conclusion, it should be emphasized that the possibilities of the optical method for investigating dislocations, illustrated by Fig. 3 and Table 1, can be used not only in studying the atomic structure of slip lines, but also in solving other problems that require investigation of the mechanism of collective displacement of atoms in crystals.

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