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Abstract

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PHYSICS

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THE NATURE OF DISLOCATION LOOPS

IN STRENGTHENED ALUMINUM

(Presented by Academician Yu. N. Rabotnov, September 1, 1969)

Recently ([1]), on the basis of a comparison of the macroscopic parameters of creep in metals and the microscopic characteristics of the dislocation structure of specimens deformed under creep conditions, the decisive role was established of the activation destruction of interblock boundaries (with subsequent motion of the dislocations that emerge from them) in the process of plastic deformation. It was also found ([2,3]) that a sharp lowering of the temperature during creep leads to the formation of dislocation loops at block boundaries. These loops, on the one hand, increase the degree of pinning of dislocations in the boundaries, as a result of which the creep resistance increases, and, on the other hand, are themselves the result of activation processes of plastic deformation occurring at block boundaries. The present work was undertaken as a continuation of investigations ([1–3]) and is devoted to an electron-microscopic analysis of the nature of dislocation loops.

The work was carried out on strengthened specimens of polycrystalline aluminum of grade AD-1 used in ([2,3]), with the aid of a JEM-150 electron microscope. Foils for electron transmission were prepared by the method of ([4]).

Dislocation loops, in agreement with ([2,3]), were observed near interblock boundaries. Their distribution and the contrast arising on them are illustrated by Fig. 1. The absence within the loops of the contrast characteristic of stacking-fault defects indicates that the loops are formed by perfect dislocations. To determine the crystallographic planes in which the dislocation loops lie, the shape and orientation of the latter in photographs of their electron-microscopic image were analyzed in the same way as was done, for example, in ([5]). Processing of photographs of more than 20 different regions of the foils showed that the loops, as a rule, lie in planes close to planes of the {110} type.

This observation agrees with the results of studies of dislocation loops in crystals quenched from high temperatures ([5–7]) and irradiated with particles ([8]). However, whereas in each region of these crystals the loops were located practically

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uniformly in all possible {110} planes, in our case, near block boundaries, a clear predominance of loops of certain orientations was always observed. Often one particular loop orientation predominated, and loops of not all possible orientations were present, but only individual ones. For example, in Fig. 1 most of the dislocation loops are located in planes intersecting the plane of the figure in the direction $[00\bar{2}]$ and, judging from the ratio of the axes of the ellipses—the images of the loops—close to $(\bar{1}10)$. There is also a loop whose plane intersects the plane of the figure in the direction $\{\bar{1}\bar{1}1\}$ and is close either to $(0\bar{1}1)$ or to $(10\bar{1})$. There are no loops in the remaining {110} planes; moreover, the absence in the figure of images of such loops cannot be explained by the fact that their Burgers vectors lie in the plane of the Bragg reflection $(\bar{1}11)$. In this case, a characteristic contrast similar to that shown in Fig. 2b should have been observed.

Fig. 1. Dislocation loops in hardened aluminum

Fig. 2. On the determination of the Burgers vector of dislocation loops (explanations in the text)

Fig. 3. *a*—image of a dislocation loop inside its true position, $s > 0$; *b*—image of a dislocation loop outside its true position, $s < 0$; *c*—image of the same loop after tilting the foil through a large angle clockwise. (B and H denote the upper and lower portions of the loop)

Fig. 4. Contrast from a dislocation loop in dark field ($as < 0$) and bright field (b, $s > 0$; c, $s < 0$). B corresponds to the top and H to the bottom of the loop

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Fig. 2 presents photographs of one and the same region of the specimen, corresponding to the conditions for formation of contrast by different Bragg reflections: in one case by the reflection $(\bar{1}11)$, in the other by $(1\bar{1}1)$ (the photographs have been corrected for magnetic rotation relative to the corresponding diffraction patterns). The plane of the drawing coincides with the (110) plane. In Fig. 2a a normally contrasted image is seen of a dislocation loop lying in one of two planes: (011) or $(\bar{1}01)$. In Fig. 2b the image of this loop corresponds to minimum contrast, and, consequently (see, for example, (9)), its Burgers vector lies in the reflecting plane.

In f.c.c. crystals, perfect dislocations have Burgers vectors $\frac{1}{2}\langle 110\rangle$. Thus the Burgers vector of the loop under consideration may be either $\frac{1}{2}[011]$ or $\frac{1}{2}[101]$. It is easy to see that, whichever of the indicated planes the loop lies in, one of these Burgers vectors will be normal to the loop, while the other will be inclined to it at an angle of 30°. However, unlike the first, the latter possibility contradicts theoretical and experimental studies of dislocation loops in f.c.c. structures possessing high stacking-fault energies, to which aluminum belongs (10). Thus Fig. 2 shows an image of an edge loop. The contrast observed in Fig. 2b is apparently due to displacements radial with respect to the loop (see (9)).

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A similar treatment of photographs of many dislocation loops showed that the loops arising in the metal upon a sharp lowering of the test temperature during creep are prismatic, with a Burgers vector close to the normal to the plane in which they lie.

It seemed most interesting and important to study the character of the loops under investigation, i.e., to determine whether they are vacancy or interstitial. This can be done by using the fact that, when observed under conditions of deviation from the Bragg angle, the image of a dislocation is located on one side of its true position. In Figs. 3a, b photographs are given of a dislocation loop obtained with the same reflection vector g responsible for the contrast, but with different deviation parameters s (Fig. 3a corresponds to $s > 0$, Fig. 3b to $s < 0$; the sign of the parameter s was determined from the position of Kikuchi lines relative to the Bragg reflections on the diffraction patterns). In the first case the image is located on the inner side of the loop, in the second—on the outer side. Therefore, according to the dynamical theory of contrast (9), the quantity $s(\mathbf{g} \cdot \mathbf{b})$ is positive in the first case and negative in the second, and, consequently, $\mathbf{g} \cdot \mathbf{b} > 0$ (here $\mathbf{b}$ is the Burgers vector).

Rotation of the foil clockwise with the aid of the goniometer stage led to the image of the loop expanding and approaching a circle. Fig. 3b corresponds to rotation of the foil through a large angle*. From these data the following follows: the contrast in Figs. 3a, b is produced by a dislocation loop inclined to the plane of the drawing, and in such a way that the regions denoted by B and H correspond respectively to the upper and lower parts of the loop.

The inclination of the plane in which the loop lies in the foil can also be determined from dark-field images, using a consequence of allowing for absorption in the dynamical theory of contrast (11). In dark field the contrast in the image of defects is stronger closer to the upper surface of the foil for $s < 0$ and closer to the lower surface of the foil for $s > 0$. Thus the designations B and H in Fig. 4a, obtained at $s < 0$, refer to the dark-field

* This photograph serves as yet another confirmation of the edge character of the dislocation loops. It shows a loop whose plane of occurrence is parallel to the plane of the drawing. Symmetry of the image with respect to the line of zero contrast is visible, the latter being orthogonal to the reflection vector. Such contrast is typical of edge loops with the Burgers vector directed parallel to the electron beam.

···to the image of portions of the loop lying in the foil above or below with respect to one another.* Images of this same dislocation loop in Fig. 4b, c were obtained in a bright field, respectively at $s > 0$ and $s < 0$. In these figures the image is located inside (4b) and outside (4c) the real position of the loop. Consequently, here as well the product $g \cdot b > 0$.

In the photographs under discussion, the vectors g have been drawn taking into account the magnetic rotation in the intermediate lens and the inversion of the image in the objective lenses of the microscope. With the described geometry

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of contrast formation and the spatial arrangement of the dislocation loops, in accordance with the theory of diffraction contrast,[9] fulfillment of the condition $g \cdot b > 0$ necessarily means that the loops recorded in Figs. 3 and 4 are vacancy loops.

In these examples, images of loops were used with the same vector g and parameters s of different sign. The use of images corresponding to opposite g and like-named s led to the same result.

Thus, as a result of rapid cooling of a metal deformed under creep conditions, prismatic dislocation loops arise at the sites of activation processes; these loops correspond to collapsed vacancy disks. In accordance with [2,3], the energy barrier determining the creep rate of the specimens used in the present work numerically coincides with the activation energy of self-diffusion of vacancies in aluminum. Therefore it may be considered that the destruction of interblock boundaries is associated with the mass generation and migration of vacancies. The inadequacy of planes of the {110} type in each local region, from the standpoint of the formation of loops in them, is apparently connected with the peculiarities of the mechanism of destruction of block boundaries.

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* The dark-field studies were carried out on a JEM-120 electron microscope.

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

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