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FUNDAMENTAL STUDIES RELATED TO THE ORIGIN AND NATURE OF CREEP OF METALS

Eighth Technical Report

Some Observations On the Work Hardening of Metals

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INTRODUCTION

The nature of work hardening accompanying plastic deformation of metals has long been the subject of speculation. As early as 1925, Becker[1] suggested a thermodynamical process to explain slip based on the statistical probability for the occurrence of local glide steps. The theory that plastic glide was due to the presence of lattice defects known as dislocations was first introduced in 1934 by Taylor [2], Orowan[3] and Polanyi[4]. Taylor presented a rather complete analysis of the behavior and interaction of dislocations. He assumed a certain type of stress field to be associated with a single dislocation, and postulated that work hardening would result from plastic flow in two ways. First, dislocations following one another across a slip plane might encounter a barrier which would impede their movement, thereby gradually diminishing and eventually stopping plastic flow on that plane. The second mechanism involved the concept of dislocations traveling in opposite directions on nearby planes attracting one another to form a metastable lattice. The interaction of dislocations in such an array was shown to be large enough to require an increase in stress before plastic flow would continue.

Kochendörfer[5,6] extended the concept of work hardening involving a back stress due to an accumulation of like dislocations on a slip plane. He suggested that the formation of a new dislocation is hindered by those dislocations already present. The hardening resulting from the interaction of newly forming dislocations with bound dislocations was designated as "formative hardening", and indicated to be the only type of hardening necessarily connected with the slip process. In another detailed picture of work hardening, Mott and Nabarro[7] assumed that sufficient dislocations are primarily present in the crystal as a consequence of growth irregularities (mosaic structure) to initiate slip. Work hardening resulted from the interaction of migrating dislocations with localized internal stresses produced during deformation. More recently, however, Mott[8] has modified his treatment to include the generation of dislocations at Frank-Read sources and to account for the hardening by the formation of sessile dislocations.

It has been widely accepted that metals of hexagonal structure, usually possessing only one set of glide planes, exhibit mechanical behavior markedly different from those having cubic structures. However, the differences in behavior of the two classes may be less fundamental than suggested. Röhm and Kochendörfer[9] have shown that a single crystal of aluminum subjected to approximately simple shear gives the type of stress-strain curve associated with the hexagonal metals. Similar behavior has been demonstrated for gold and silver crystals by Andrade and Henderson[10]. Thus when glide in cubic metals is restricted to one plane, the strain hardening characteristics closely resemble those of the hexagonal metals. It would appear therefore, that an investigation of some of the strain hardening properties of single crystals sheared in simple glide might provide a more complete understanding of the phenomenon of work hardening. This report relates a number of recent observations made on single crystals of copper, zinc and cadmium tested in simple shear. In addition, pertinent observation of effects accompanying stress-induced movement of dislocation boundaries are reported.

EXPERIMENTAL PROCEDURE AND RESULTS

Details of the production and advantages of the type of single crystal shear specimen employed for the following tests have been presented in a previous publication[11]. The method of testing makes possible the application of a shear stress accurately aligned with a crystallographic slip plane and direction and results in an unusually uniform shear strain. Spherical crystals were grown in a helium atmosphere by a modified Bridgeman technique and acid machined to final specimen contour. The gage section of the crystals used in these experiments was a cylinder having a height of one-eighth inch and a cross-sectional area of approximately one-third of a square inch.

Stress-induced motion of small angle boundaries represents one of the simplest kinds of plastic deformation. The techniques previously described for forming and moving the boundaries in zinc crystals were utilized[12]. Whereas the behavior of single dislocations may never be observed directly, these small angle boundaries apparently consist of an array of edge dislocations of like sign whose movement through a crystal may be observed and controlled. The experimental observations on this localized plastic deformation so far have been consistent with the plastic behavior of crystals deformed in simple shear. For example, the shear stress necessary to cause boundary motion is the same as the yield stress in single crystals. Results of other boundary motion experiments will be compared with results obtained from simple shear deformations.

Strain hardening in simple shear is directional. Fig. 1 represents the strain hardening curve for a zinc crystal of 99.99% purity sheared in simple glide at -196°C. At a strain of 0.08, the direction of straining was reversed. Two effects on the path of the curve may be discerned. First, plastic flow begins at a stress much lower than that accompanying the onset of slip in the original direction. Second, although the crystal was held continuously at a temperature of -196°C, the level of the stress-strain curve was appreciably lowered by the strain reversal ("strain softening"). The identical behavior demonstrated by a high purity cadmium crystal under the same experimental conditions is shown in Fig. 2. A similar situation was encountered when the direction of the stress-induced movement of a dislocation boundary was reversed.

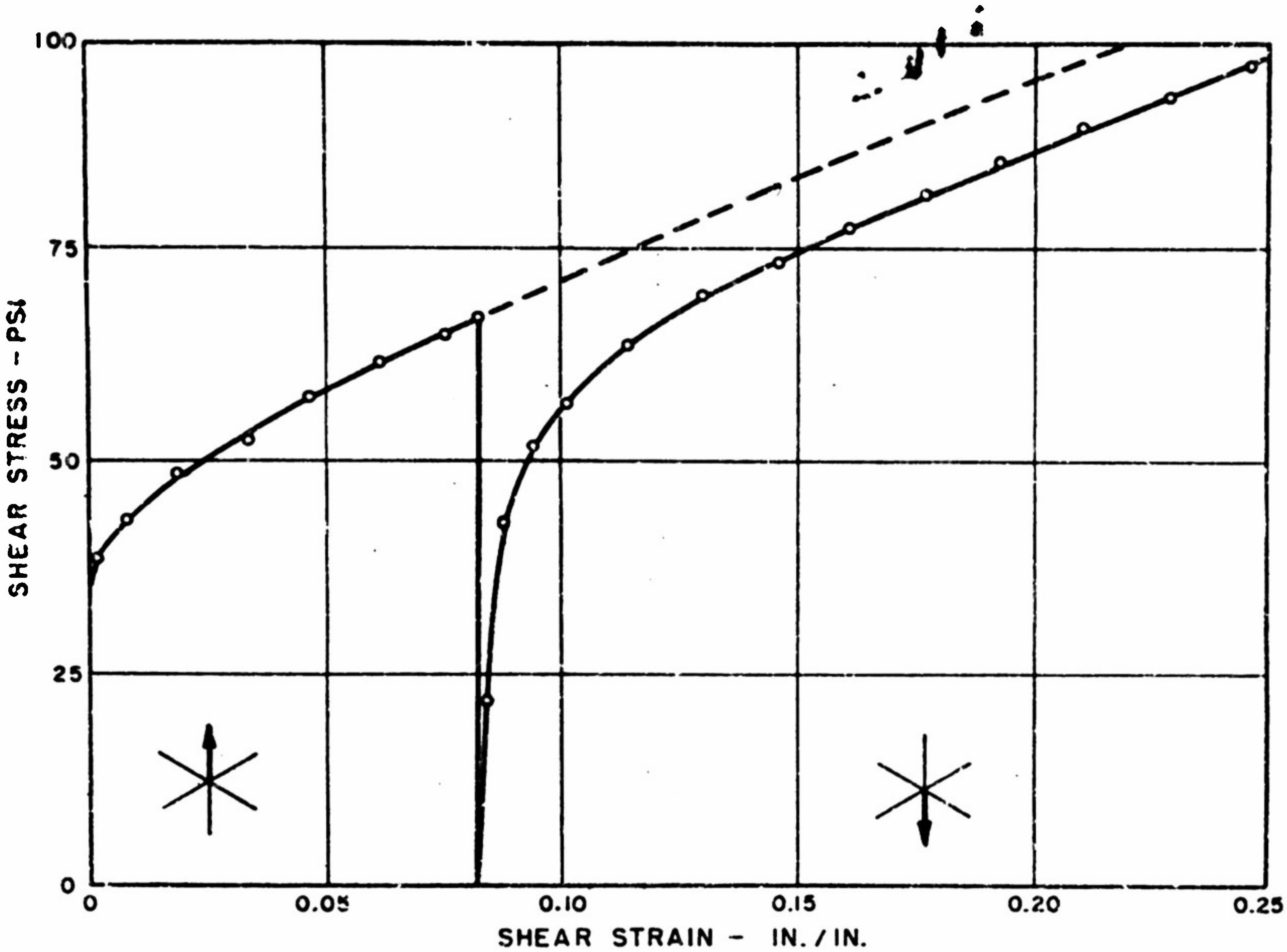

FIG. 1 EFFECT OF REVERSING THE STRAIN DIRECTION DURING TESTING IN SIMPLE SHEAR ON THE STRESS-STRAIN CURVE OF A ZINC CRYSTAL. TEST TEMPERATURE WAS -196°C. DASHED LINE SHOWS NORMAL COURSE OF CURVE HAD STRAIN BEEN CONTINUED IN THE ORIGINAL DIRECTION.

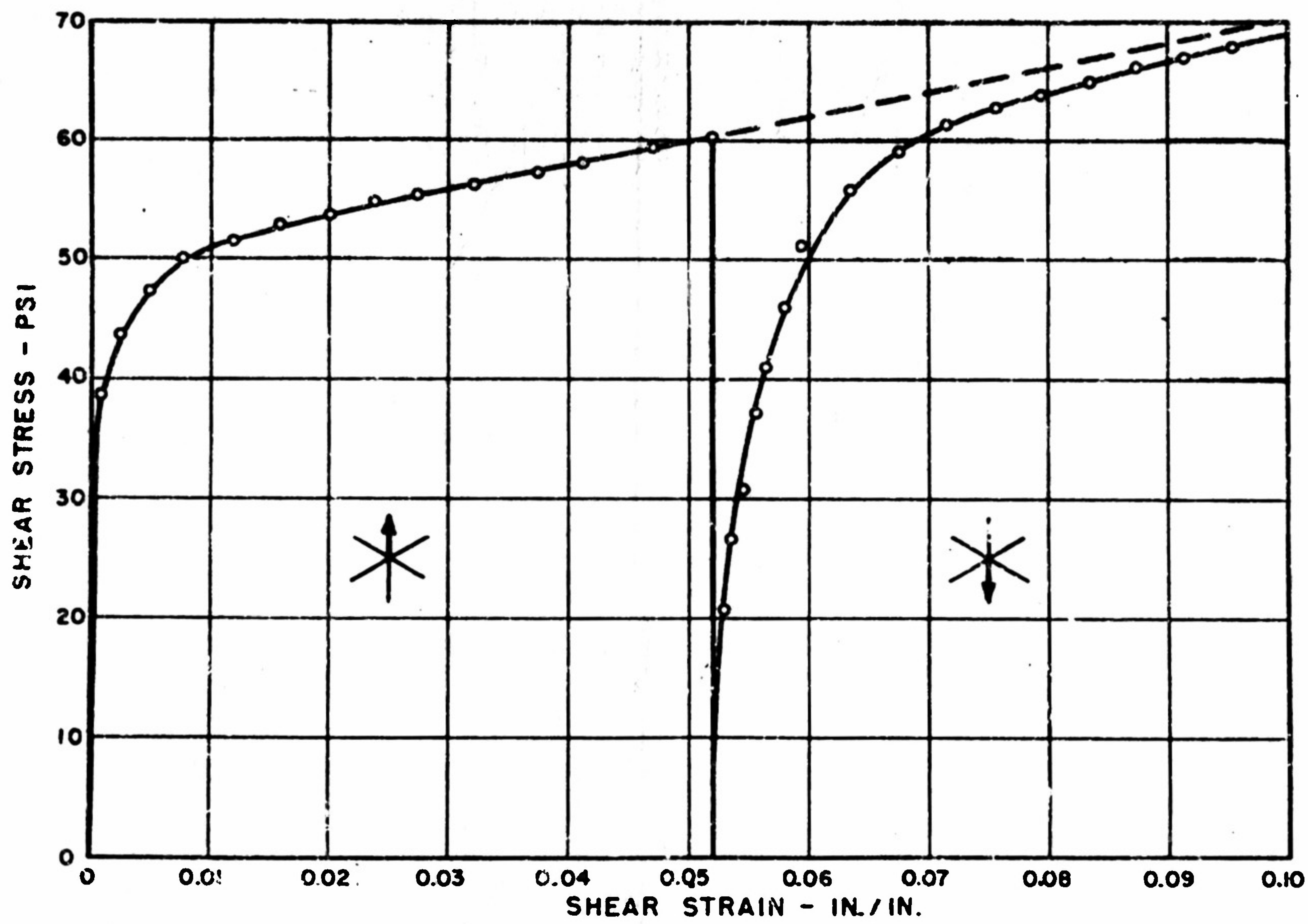

FIG. 2 EFFECT OF REVERSING THE STRAIN DIRECTION DURING TESTING IN SIMPLE SHEAR ON THE STRESS-STRAIN CURVE OF A CADMIUM CRYSTAL. TEST TEMPERATURE WAS -196 °C. DASHED LINE SHOWS NORMAL COURSE OF CURVE HAD STRAIN BEEN CONTINUED IN THE ORIGINAL DIRECTION.

Fig. 3 shows a load displacement curve for a boundary of this type in a zinc crystal. In both directions of movement, the rate of motion was held approximately constant. The critical load for a constant rate of movement was decreased when the direction of movement was reversed.

Macroscopic substructure in a crystal has a marked effect on plastic properties(13). Fig. 4 shows stress-strain curves for a crystal with and without such a network of small angle boundaries. The presence of these subboundaries affects the stress-strain curve in two ways. First, the sharp break in the curve at the yield point characteristic of the nearly perfect crystal is replaced by a more gradual bending over of the curve. Second, the stress level of the curve is raised noticeably. At larger strains, the rates of strain hardening for the two cases are almost identical.

Duplex slip in two directions on the same slip plane (Fig. 4) results in a much higher rate of strain hardening. The yield point was about the same for a crystal sheared along a single slip direction. In agreement with this result is the observation that hardening in a latent slip system may exceed that in the active system. Fig. 5 shows the effects on the stress-strain curve of a zinc crystal tested in simple shear of shifting to a new slip direction 60° from the first after a strain of 0.044 in the original direction. The stress required to cause slip to occur in the new direction was sharply increased relative to that which would have been required to continue slip in the original direction. The test was performed at -196°C to avoid the complication of recovery during the test. This behavior for zinc was in contradiction to the finding of Kochendörfer(6) on aluminum deformed in simple shear. He found that hardening in the active system exceeded hardening in all latent systems. However, his experimental techniques were much more complex than those employed for zinc.

The presence of subboundaries in copper crystals tested in simple shear alters the strain hardening characteristics in a manner consistent with that

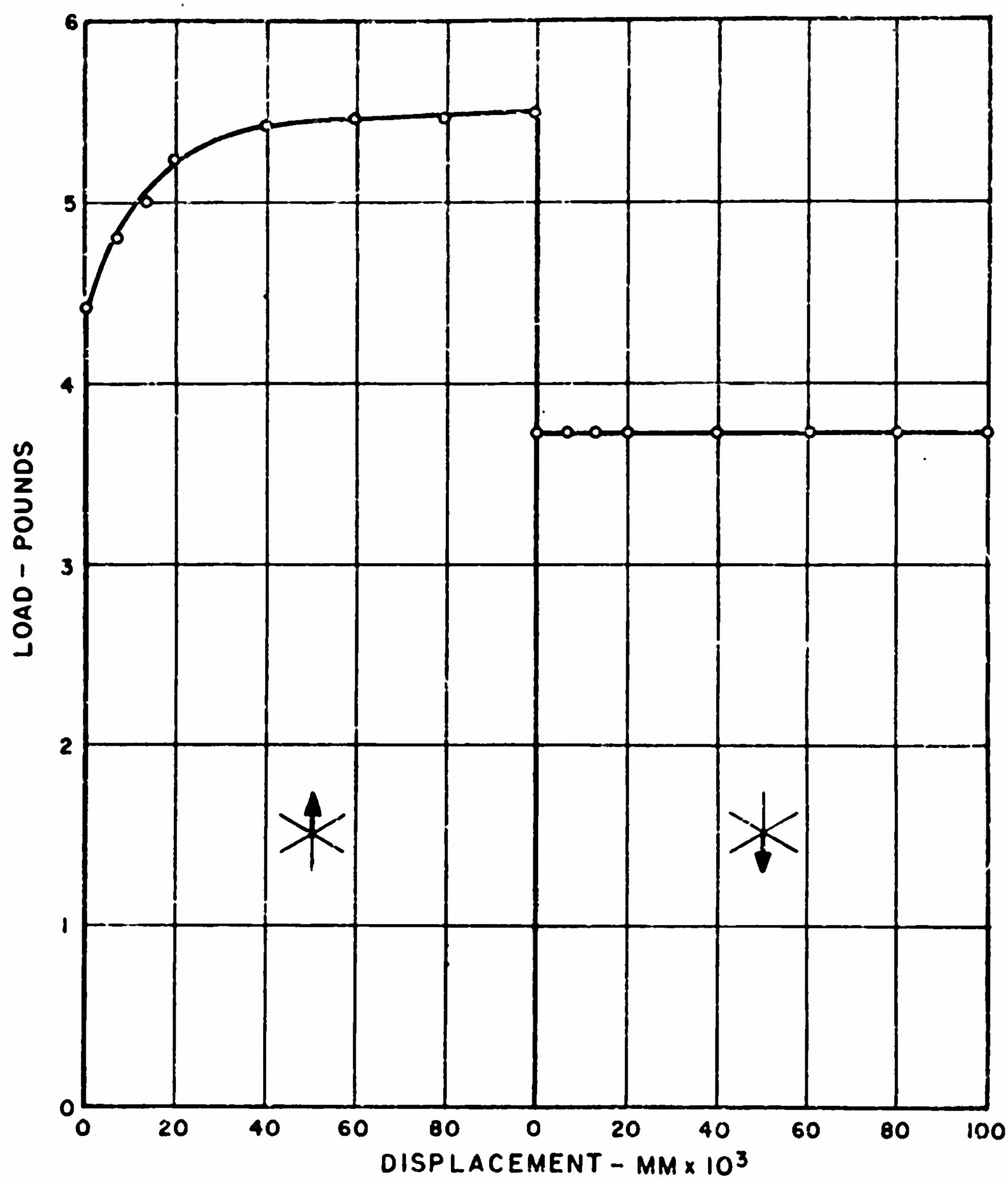

FIG. 3 LOAD – DISPLACEMENT CURVE FOR A MOVING DISLOCATION BOUNDARY SHOWING DECREASE IN LOAD WHEN DIRECTION OF MOTION WAS REVERSED. RATE OF BOUNDARY MOTION HELD APPROXIMATELY CONSTANT. 25°C.

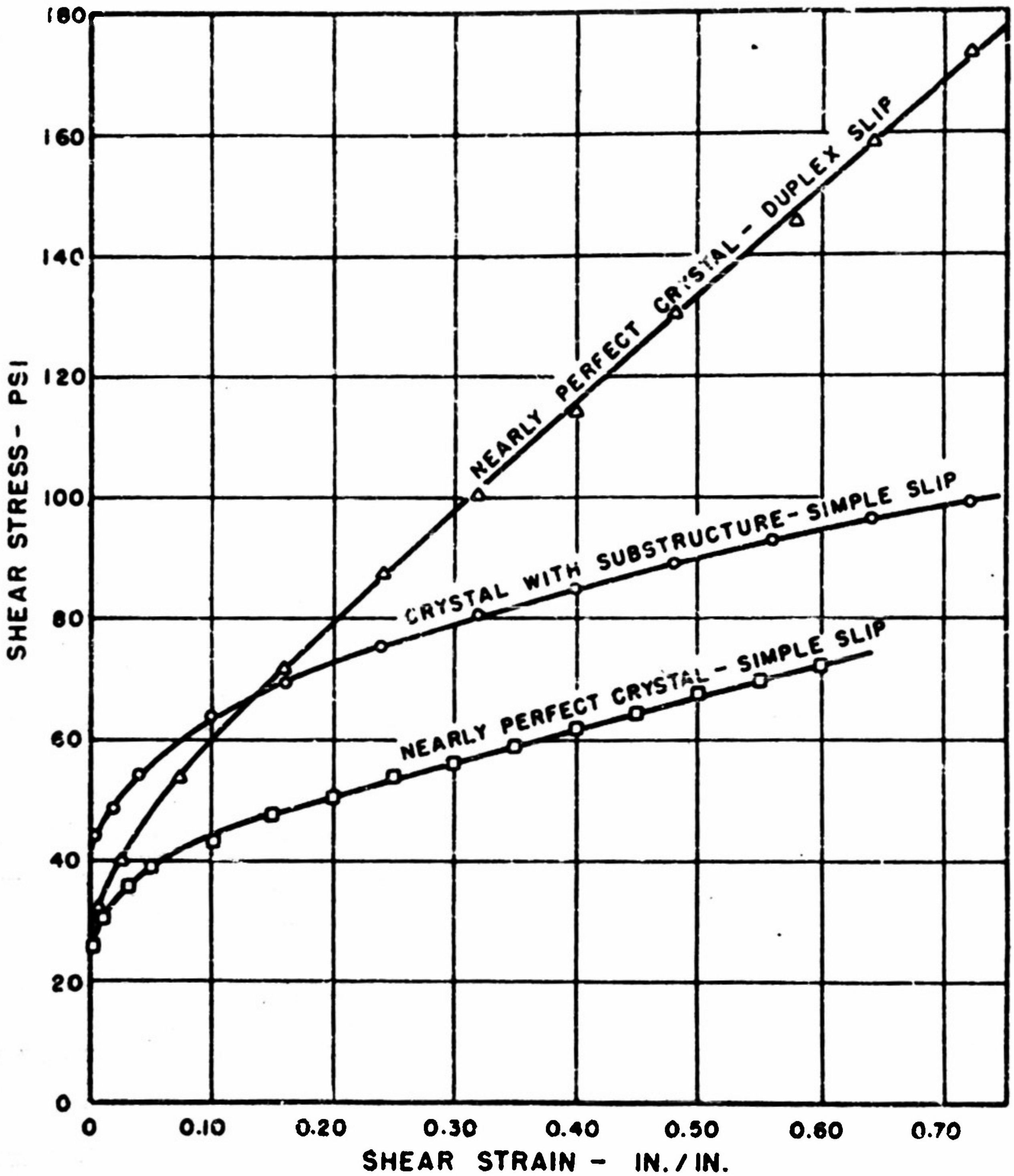

FIG. 4 STRESS-STRAIN CURVES FOR ZINC CRYSTALS SHOWING THAT THE SHAPE OF THE STRAIN HARDENING CURVE FOR SIMPLE SHEAR IS INFLUENCED BY SUBSTRUCTURE AND BY THE NUMBER OF SLIP SYSTEMS OPERATING. TEST TEMPERATURE WAS 25°C.

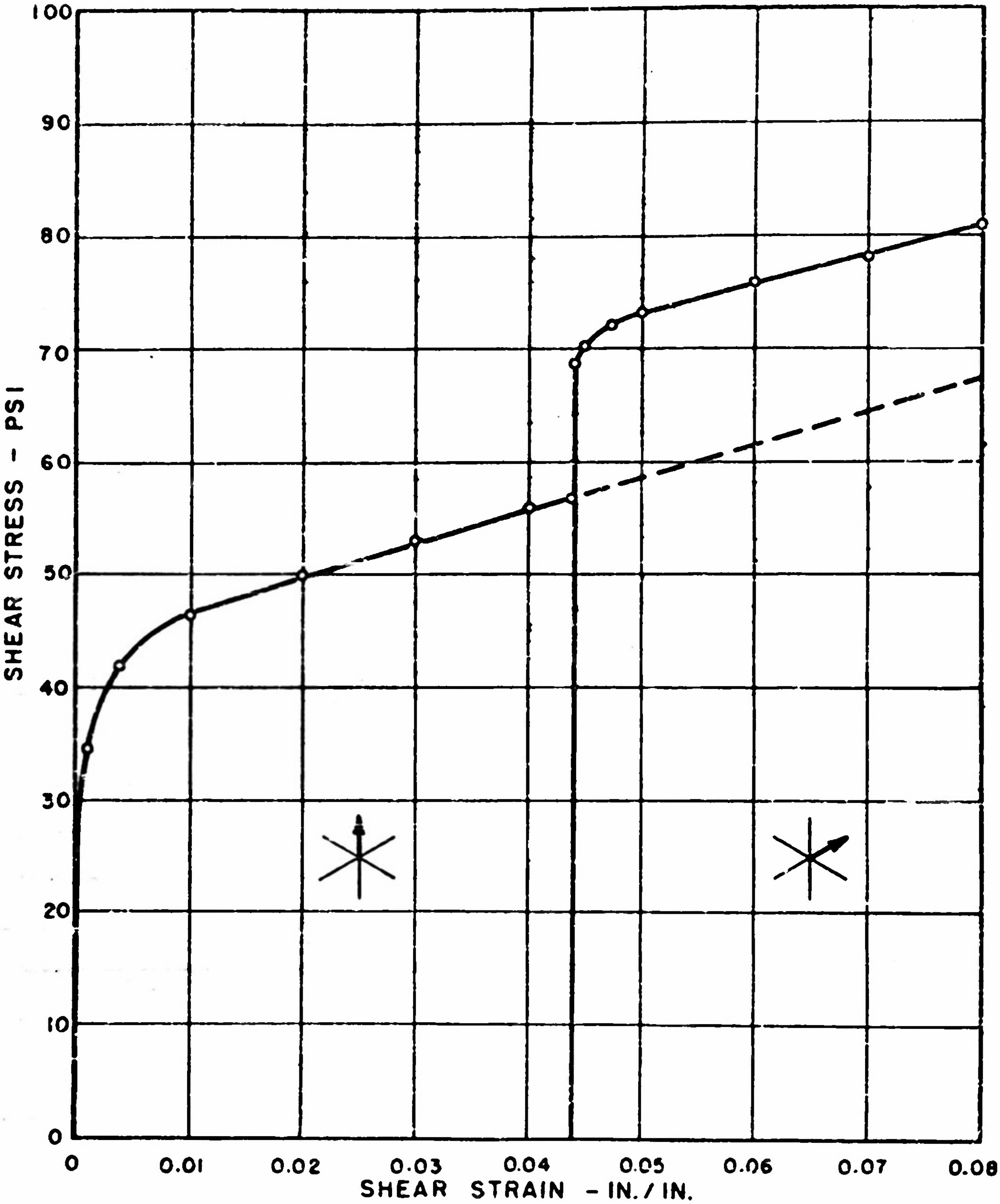

FIG. 5 STRESS-STRAIN CURVE FOR A ZINC CRYSTAL SHOWING EFFECT OF SHIFTING THE STRAIN DIRECTION DURING TESTING TO A NEW DIRECTION 60° FROM THE ORIGINAL DIRECTION OF SLIP. TEST TEMPERATURE WAS -196°C

observed for zinc crystals. Stress-strain curves for copper crystals with varying degrees of internal perfection are presented in Fig. 6. The subboundaries of the extensively polygonized crystal were formed by subjecting the crystal to a bending moment and subsequently annealing at 900°C. That polygonization had occurred was verified by the splitting of reflection spots in x-ray back-reflection photographs. The stress-strain curves for the three crystals form a homologous series, with the level of the curves rising as the extent of the internal boundaries increase. The results of a shear test of a polycrystalline copper specimen are presented for purposes of comparison.

DISCUSSION OF RESULTS

Work hardening of zinc crystals subjected to simple shear on a single glide system can be completely removed upon annealing(11). An additional feature of easy glide is that it does not involve local distortions of a type which give rise to asterism. A theory proposed by Seitz(14) advances the possibility that hardening may be caused by production of large numbers of lattice vacancies by moving dislocations, thus impeding the further motion of dislocation. All current theories require the trapping of dislocations within the crystal in one way or another during plastic flow.

Direct experimental support for the concept that dislocations are obstructed in their movement across a slip plane is provided by observations accompanying the stress induced movement of a dislocation boundary through a zinc crystal at room temperature. Careful measurement of the magnitude of the boundary angle shows that there is a significant decrease in the angle as the boundary moves through the crystal. The implication is that some of the edge dislocations of which the boundary is composed are trapped locally by internal imperfections and prevented from continuing freely with the boundary.

Further experimental evidence for the idea that some of the dislocations

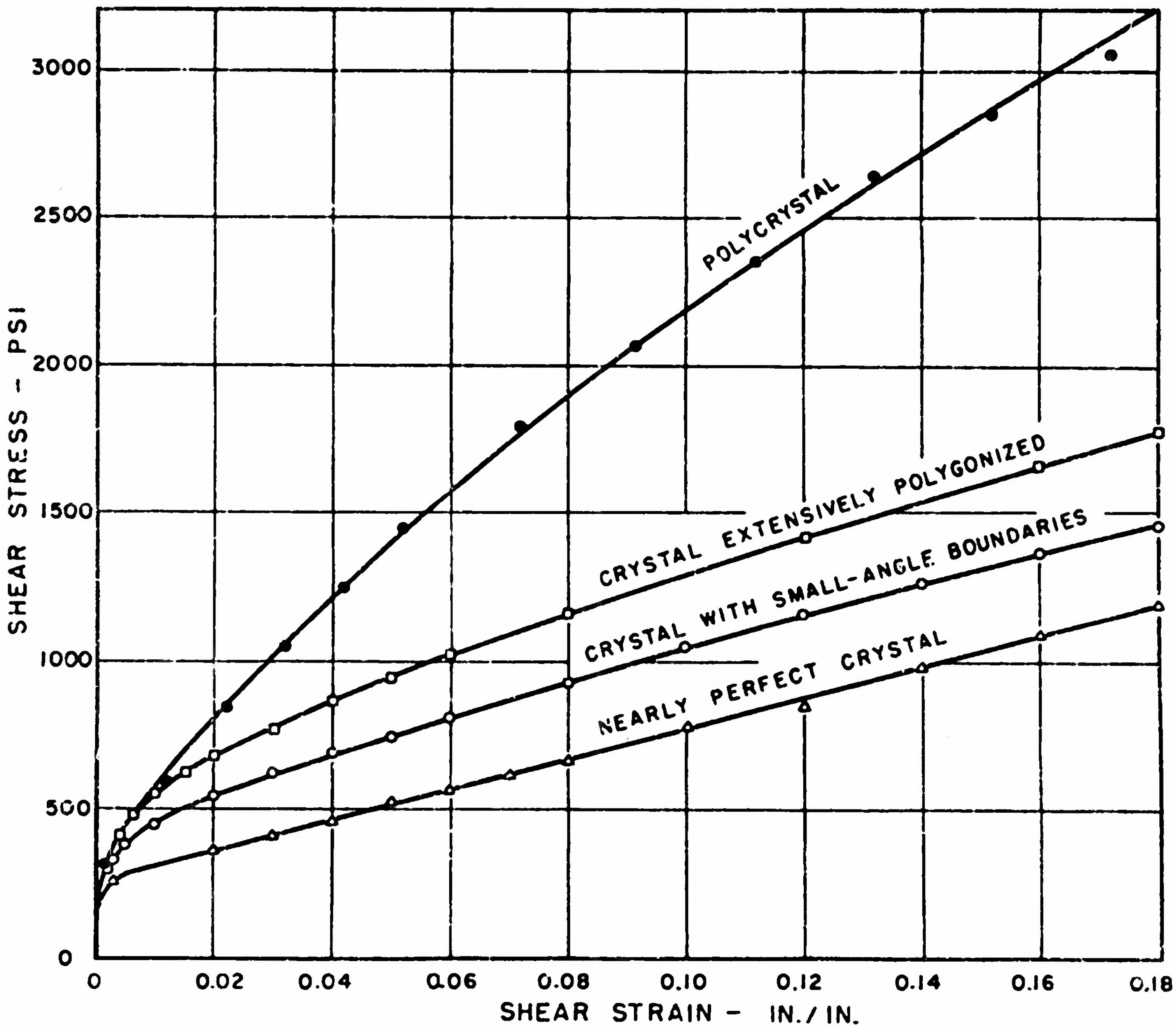

FIG. 6 STRESS-STRAIN CURVES FOR COPPER SPECIMENS TESTED IN SIMPLE SHEAR. THE STRESS AT WHICH YIELDING OCCURS IS INFLUENCED BY THE PRESENCE OF MOSAIC BOUNDARIES. THE RATE OF STRAIN HARDENING IN SINGLE CRYSTALS SUBJECTED TO SIMPLE SHEAR IS SUBSTANTIALLY INDEPENDENT OF SUBSTRUCTURE EXCEPT IN THE SMALL STRAIN RANGE.

moving through a crystal during slip become piled up against barriers is afforded by the tests in which strain direction was reversed. Dislocations piled up against a barrier produce a back stress which is proportional to the number of piled up dislocations. The magnitude of the stress concentration for n piled-up dislocations is given by:

$$\tau_{(back)} = \frac{n\,G\,b}{2\pi r} \qquad (7)$$

where G is the shear modulus, b is the Burgers vector and r is the distance from the barrier. The externally applied stress necessary to start slip in the reverse direction should be reduced in proportion to the magnitude of the back stress.

There seems to be good evidence for the trapping of dislocations at internal barriers during plastic straining of a crystal, and for the existence of back stresses associated with such bound dislocations. However, the view of Kochendörfer that this back stress is the primary cause of strain hardening may not be justified, particularly in view of the fact that latent slip directions are hardened more than the active system in zinc.

Any discussion of strain hardening is incomplete which ignores the lamelar nature of slip. The macroscopic strain hardening measured when a test section of ordinary size is employed must be interpreted to mean that formation of each elementary slip line across the crystal makes growth of the next line a little more difficult. Ruling out the possibility of exhaustion hardening, this implies that the stress concentrations produced by pile-up of dislocations at barriers are sufficiently long range to account for hardening when the spacing of slip lines is many thousands of interatomic distances. For this to be possible the slip lines must contain clusters of dislocations of like sign. Such clusters would tend to align themselves to form dislocation boundaries.

The scale of the clustering is probably influenced by the conditions of deformation. For simple shear strain in hexagonal metals the clustering must be on a small scale because of the lack of asterism. With less uniform strain large scale dislocation boundaries may build up during the strain as a result of macroscopic or microscopic inhomogenieties in applied stress[15]. In these cases strain hardening is more rapid than during simple glide due to an increase in the number and effectiveness of barriers.

Recently, Gay and Kellar[16] have found that cold working of polycrystalline metals resulted directly in the formation of subboundaries within the grains. These subboundaries apparently result from the tendency of edge dislocations to align themselves in stable arrays. Since it was indicated by these investigators that the average size of the subgrains decreased as work hardening increased, the progressive nature of work hardening seems to be due to the more effective trapping of moving dislocations by the progressively developing subboundaries.

The stress-induced movement of dislocation boundaries may provide an additional mechanism for increasing the effectiveness of barriers during straining. From Fig. 7, it may be seen that the critical stress required to move a dislocation boundary at a constant rate through a zinc crystal increases as the magnitude of the boundary angle increases. Furthermore, when two such moving boundaries unite, the load required to move the newly-formed boundary is sharply increased (Fig. 8). Thus the stress-induced merger of subboundaries, creating boundaries of larger angular magnitude, increases the effectiveness of the barriers in blocking dislocation movement.

CONCLUSIONS

A study of single metal crystals tested in simple shear and of stress-induced movement of dislocation boundaries, has lead to the following conclusions on the nature of work hardening:

1. The movement of dislocations through a crystal is impeded by internal barriers, even during simple shear.

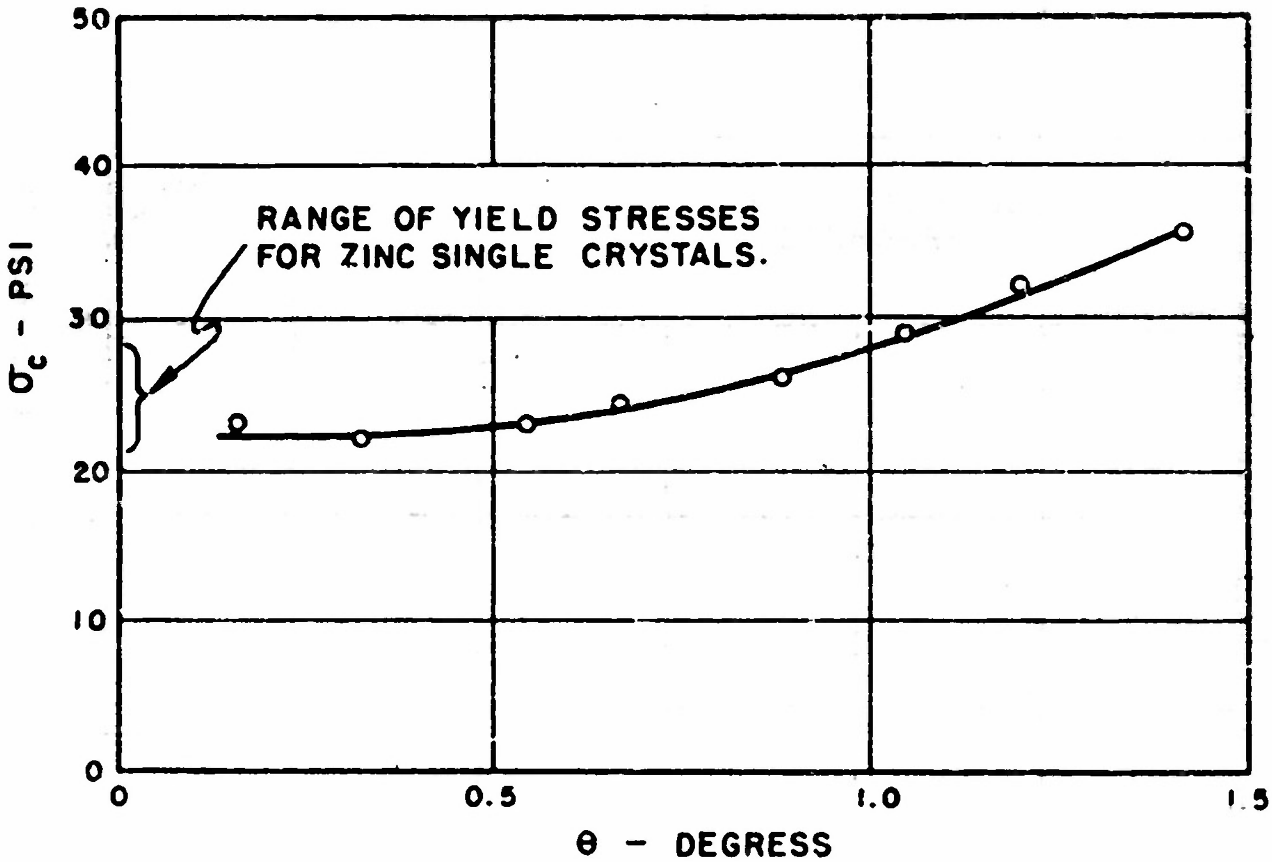

FIG. 7 CURVE SHOWING THAT THE MINIMUM STRESS REQUIRED TO MOVE A DISLOCATION BOUNDARY INCREASES WITH INCREASING BOUNDARY ANGLE. TEMPERATURE 25°C.

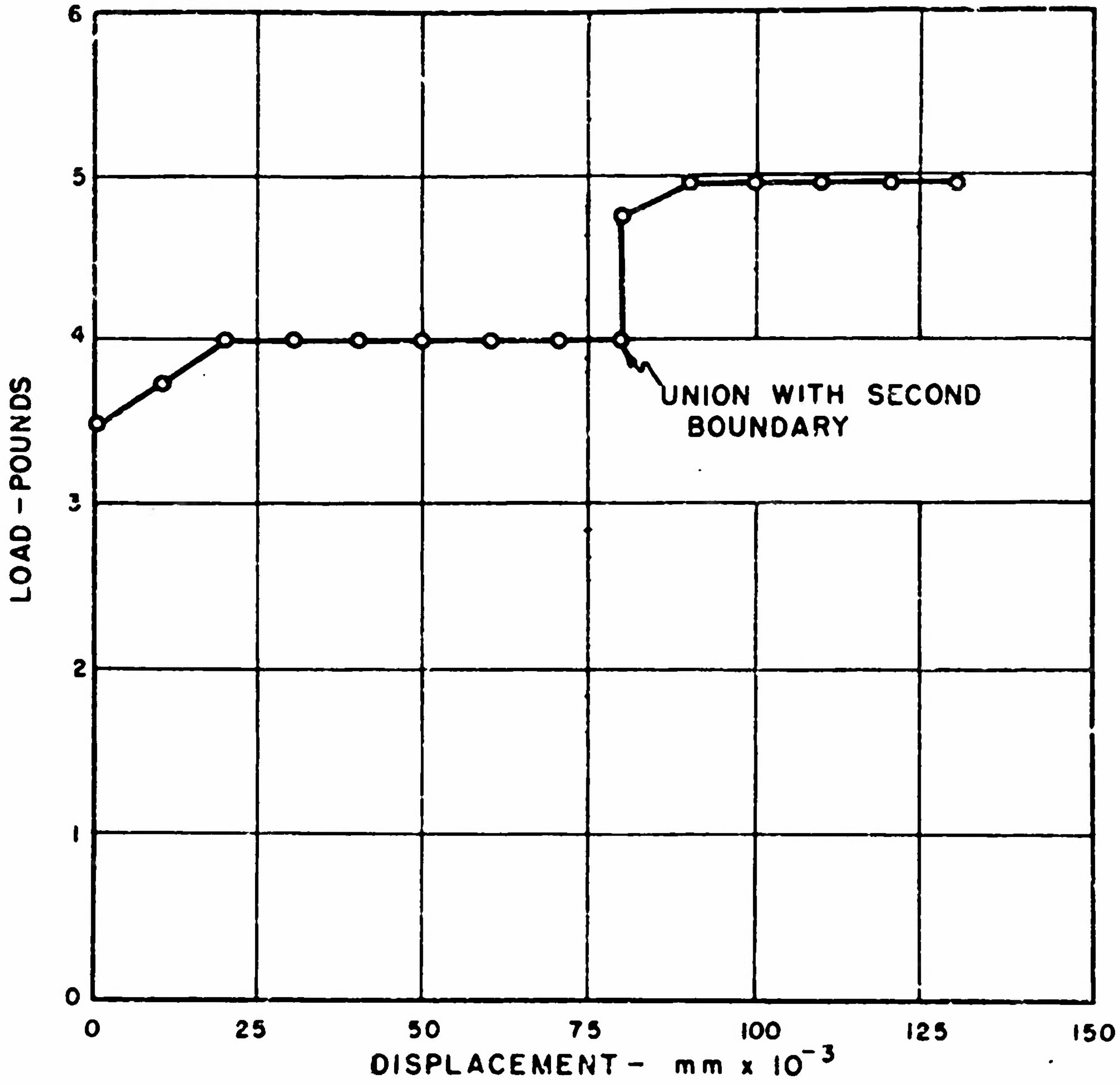

FIG. 8 LOAD - DISPLACEMENT CURVE SHOWING THAT WHEN TWO BOUNDARIES UNITE THE LOAD REQUIRED FOR CONTINUED MOVEMENT IS INCREASED. RATE OF MOTION HELD APPROXIMATELY CONSTANT. 25 °C.

2. Duplex slip in single crystals and the more complex deformation in polycrystalline specimens are accompanied by a progressive formation of dislocation boundaries. These subboundaries act as barriers in the path of moving dislocations.
3. Stress-induced movement of dislocation boundaries leads to union of adjacent boundaries and increased effectiveness in blocking movement of active dislocations.

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