RICE UNIVERSITY

TRANSIENT CREEP IN MOLYBDENUM

by

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ABSTRACT

Incremental loading tests were performed on molybdenum single crystals in direct shear, oriented for slip on the (110)- plane and in the [111]- direction at temperatures from 125°K to 300°K. The resulting transient creep could not be described by the logarithmic creep relation \( \dot{\epsilon} = A/(1+t)^n \) as it can for the close-packed metals; however, the equation \( \dot{\epsilon} = A e^{-B\Delta\epsilon} + C e^{-D\Delta\epsilon} \) was found to fit the data well. Activation volumes calculated from the relation \( v^* = kT \frac{\ln G - \ln \delta_i}{\Delta T_0} \) agreed very closely with those reported by Youngblood (Thesis, Rice University, 1963).

The region characterized by \( A e^{-B\Delta\epsilon} \) was effective only over very small ranges of \( \Delta\epsilon \), with \( C e^{-D\Delta\epsilon} \) subsequently controlling deformation. Difficulty of measurement limited most of the interpretation to \( C e^{-D\Delta\epsilon} \). Analysis is presented which indicates that a Peierls model may be the rate-controlling mechanism for this region.

Studies of orientation dependence were made which indicated that slip in the [111]- direction occurs most easily on the (110)- planes, is slightly more difficult on the (123)- planes, and is substantially more difficult on the (112)- planes.
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I. INTRODUCTION

The mechanisms controlling plastic flow in body-centered cubic (b.c.c.) metals are not known, although many investigators have attempted to determine suitable models. Their investigations may be separated into roughly two phases: low temperature studies (400°K and below) and high temperature studies (300°K and above).

Due to the complexity of the problem, investigators have usually restricted themselves to studying only one temperature region and have otherwise simplified the problem as much as possible. Very close control of test conditions and use of single crystals of known orientations have greatly aided in simplification.

The technique of testing Molybdenum in simple shear in the low temperature region (Youngblood 1963 Thesis) was found to be a very effective method of limiting primary slip-plane and slip-direction uncertainty. Previously, no attempts had been made to eliminate such uncertainties. Youngblood's experiments have led to interesting, though not altogether conclusive, results concerning the mechanisms governing plastic flow. It was, therefore, decided to investigate the low-temperature plastic flow properties of molybdenum in a different manner, in expectation that more conclusive results could be drawn from the two investigations.

Experimentally, there are only four variables which can be measured or controlled directly through testing: $\tau_0$, the applied shear stress; $T$, the absolute temperature; $\varepsilon$, the shear strain, and $\dot{\varepsilon}$; the strain rate (Alefeld, 1962, & Youngblood, Thesis, 1963). Theoretical analyses have

(1)
predicted that the plastic flow will be characterized by the four internal parameters: \( K \), the activation energy; \( v' \), the activation volume; \( \nu \), the frequency factor; and \( \tau \), the effective shear stress. Under various testing conditions, the following relations may be found:

(a) \( \frac{\partial c}{\partial T} \bigg|_{\tau_a} = \text{const.} \) From changing temperature during a creep test.

(b) \( \frac{\partial \tau_a}{\partial T} \bigg|_{c} = \text{const.} \) From changing temperature during a constant strain rate test.

(c) \( \frac{\partial c}{\partial \tau_a} \bigg|_{T} = \text{const.} \) From changing the stress during a creep test, or from changing the strain rate during a differential strain rate test.

(d) \( \frac{\partial c}{\partial \tau_a} \bigg|_{T, \tau_a} = \text{const.} \) From changing the stress during a creep test, and observing the resulting strain rate-strain relation.

These equations are not independent of one another (Alefeld, 1962), and equations relating them may be written. One such equation is

\[
\frac{\partial c}{\partial T} \bigg|_{\tau_a, c} + \frac{\partial c}{\partial \tau_a} \bigg|_{T, \tau} + \frac{\partial c}{\partial \tau_a} \bigg|_{T, \tau_a} + \frac{\partial c}{\partial \tau_a} \bigg|_{T, \tau_a} = 0
\]

Equations (b) and (c) were used by Youngblood to determine values of the internal flow parameters, whereas this investigation utilized information gained from creep tests through (c) and (d). Equation (a) was not used because it was felt that a change in structure may accompany a change in temperature, and that any information gained from a change in temperature test would have to be interpreted in that light.

The temperature range 125°K to 400°K was covered in this investigation in order to cover the entire range of the low-temperature properties of the material. Activation volumes and creep behavior were measured from approximately 1% to at least 60% shear strain for all temperatures.
II. SURVEY OF OTHER RELATED INVESTIGATIONS

Although much work has been done on plastic flow in b.c.c. materials, most of it has been done by methods other than creep. Some creep experimentation has been done at high temperature in the b.c.c. metals, but almost all studies have been in the close packed metals, particularly copper and aluminum. Due to the increased importance of understanding the nature of deformation in b.c.c. metals, some creep studies have begun.

Of the numerous papers on plastic flow in metals, one of the most comprehensive is the recent survey article by Washburn (1960). It mentions what are thought to be the most important factors governing mobility of dislocations in the three most common crystallographic forms. Strain hardening and dislocation multiplication concepts are emphasized due to their importance in explaining material behavior.

Washburn believes, along with many others, that dislocation multiplication and yield stress are closely related, in that yield stress results from growth of slip bands, rather than from their nucleation. Frank-Read sources provide the mechanism for production of dislocations on a single glide plane, but continued slip of the dislocations as they spread out from their sources require the overcoming of more and more obstacles. As overcoming the obstacles becomes more difficult, the dislocations can physically move around them by changing slip-planes. Such a process is cross-slip, and can be initiated within the crystal in several ways:

1. An edge dislocation must climb, through non-conservative motion, to change its slip-plane, but a screw, because it is bound to no particular slip-plane, can cross-slip quite easily. Obstacles to dislocation movement are jogs, piled-up dislocations, and
inclusions in the lattice. Cross-slip by screws over these obstacles is perhaps the most likely in b.c.c. metals because of their high stacking fault energies.

2. Edge dislocations can be made to climb, provided the temperature is high enough for vacancy migration and formation to occur. Whenever a jog is formed in a screw segment, it can often slip along the axis onto the edge portion of the dislocation, thus leaving the screw relatively free. This action can cause formation of long, prismatic loops, as often seen by electron microscopy. If the jogs will not move off the screw segments then these jogs become anchoring obstacles, between which the dislocation must bow out, possibly forming another dislocation source; thus, multiplication.

Strain-hardening is thought to be the formation of tangles of dislocations, causing a cellular structure that is relatively impenetrable to other moving dislocations. This effect is well documented by electron microscopy in all crystal structures.

Gregory and Rowe (1962) have summarized work-hardening mechanisms in b.c.c. metals rather thoroughly, and given some emphasis to their own studies in columbium. Their review is supported by a large collection of electron photomicrographs which show essentially the following: in b.c.c. materials, dislocation tangle formation is a function of strain, temperature and strain-rate. Material highly strained at a low temperature has a tangle arrangement similar to material only slightly deformed at higher temperatures.

Low and Turkalo (1962) made a study of dislocation patterns in 3 per cent silicon iron by electron microscopy. They observed the pinning
of screw segments by jogs to be quite prevalent, especially at low strains. None of their observations could account for what had caused the jogs to form, but they believed it to be due to the ease of cross-slip in b.c.c. metals. They used the ideas of Gilman and Johnson (1962) to explain the different pinning effects of jogs, and how such mechanisms could cause tangling as they observed it.

At the Conference on the Impact of Transmission Electron Microscopy on Theories of the Strength of Crystals, July 1961, Keh and Weissmann reviewed the deformation substructure of b.c.c. metals. They showed that the structure as a function of cold work is similar for all the metals studied at the same homologous temperature. Lawley (AIME Conference, 1963) has corroborated these findings in molybdenum.

Many investigations have been made to determine the slip-plane of b.c.c. metals (Maddin and Chen, 1954, is a good survey article), however, no conclusive results have ever been reached. It is generally believed that slip occurs most readily on the (110)-planes, in the (111)-direction, but that slip can occur on any plane containing a (111)-direction. The ease with which slip can occur on planes other than the (110)-plane gives an indication of cross-slip in b.c.c. metals. Rose, Ferriss and Wulff (1962) investigated Yielding and Plastic Flow in Tungsten Single Crystals, and Ferriss, Rose and Wulff (1962) investigated the Deformation Characteristics of Tantalum. Both of these investigations indicate that slip occurs primarily on the (110)-planes, but that a large amount of cross-slip also occurs. The former article also found some evidence for the reaction \( \frac{a}{2} [111] + \frac{a}{2} [111] = [100] \), as originally proposed by Gottrell (1950).
Hornbogen (1963) found that in pure iron, dislocations form tangles whereas pileups and disassociations have not been observed. Increasing the deformation velocity or lowering deformation temperature changes the structure into a pattern of rather straight screw dislocations. The dislocation pattern is also altered in this manner when certain alloying elements are added. He concludes from his investigation that the adding of alloying elements retards cross-slip, and attributes this to the increased numbers of vacancies caused by the alloying atoms which can cause jog formation in screw-dislocations, thus stopping them.

Keh (1960) made a study of the dislocation arrangements in decarburized vacuum-melted iron deformed at various temperatures. He found a marked temperature dependence on the cellular structure formation. At -196°C many long, straight dislocations were generated during deformation, although major deformation resulted from twin formation. Some evidence was found of dislocation having [100] Burgers vectors. Dislocation pileups were not observed. There was some evidence of cross-slip.

J. C. M. Li (AIME publication "Direct Observation of Imperfections in Crystals") suggests that tangles are formed in iron when dislocations on the primary slip system exert sufficient stress on a secondary system to operate dislocation sources on the secondary system. The reason for tangling is probably the interaction between the dislocations of the two systems. Li concludes from his calculations on external stress required to continue plastic deformation that tangles as a whole must be capable of movement, for if not, sufficient strain cannot be generated. Li suggests that movement of tangles as a whole could explain the absence of a direct correlation between the cell dimensions measured by transmission microscopy.
and the length of surface slip lines.

Kuhlmann-Wilsdorf and Wilsdorf (1962) observed long, prismatic loops and tangles in deformed crystals and theorized on their origin. The basis of the tangles, they reasoned, was the cross-slip of screw dislocations during deformation, and that cross-slip was initiated by stress dilations in the lattice stemming from thermally formed jogs.

The idea of "dislocation uncertainty" was first postulated by Kuhlmann-Wilsdorf (1960). At that time she presented the idea that jogs could be formed in a material due entirely to the motion of the atoms in the lattice. Because the dislocation axis is not well defined, as the result of atomic motion, there may be sufficient movement of the atoms to cause a thermal jog to be formed. It was also postulated that due to the crystal structure of b.c.c. materials this tendency to form thermal jogs probably decreased quite rapidly at moderately low temperatures. A letter to Acta Met (1962) clarified and extended this idea to the production of point defects. Kuhlmann-Wilsdorf believes that jogs, and the accompanying point defects, are formed spontaneously due to dislocation uncertainty, and this is what produces the jogs seen by Low and Turkalo (1962). Long prismatic loops are formed by coalescence of thermally-produced jogs to form deep jogs, which form prismatic loops in the manner described by Gilman and Johnson (1962).

Conrad, Schoeck, Petch, and others, have studied the mechanics of yielding and flow utilizing activation energy, activation volume, and frequency factor concepts. Conrad (1961) suggests that in iron, the mechanism in closest agreement with theoretical calculations and experimental data is overcoming Peierls stress, in other words, that stress inherent in the
otherwise perfect crystal. Schoeck (1961) does not believe a simple Peierls force model too probable, and suggests that movement of jogs intrinsic to the dislocations, and those caused by intersection with other dislocations, are the controlling factor in b.c.c. materials. He also believes that cross-slip of dislocation screw segments is a major controlling factor. Youngblood (Thesis, 1963) has a thorough review of other work using these concepts.

Alefeld (1962) has investigated the mathematical aspects of activation volume and activation energy determination for small stresses and strain rates. Although his work brings out no essentially new concepts, it is a reminder that in thermal activation there is a probability for forward activation (in the direction of applied stress) as well as backward activation, for small stresses and strain rates. He also points out that a true activation volume will differ markedly from that measured when the backward jump probability is not considered.

The foregoing abstracts are but a few of the many, many investigations made previously to determine the yielding and flow characteristics of the b.c.c. metals. The only works included here are those which it is felt will have direct bearing on the experimental observations made in this investigation.
III. SPECIMEN PREPARATION

A. Single Crystals

All of the molybdenum specimens tested in this investigation were zone-refined single crystals grown in the Electron-Beam Melting and Zone-Refining apparatus built by Youngblood (See Youngblood thesis, 1963, for a complete description of the equipment). All creep tested shear specimens were grown with a (110) plane oriented perpendicular to the specimen axis and under a vacuum of between $1 \times 10^{-5}$ mm Hg and $5 \times 10^{-5}$ mm Hg. Due to the uncertainty as to the degree of zone refining necessary to remove impurities, and the uncertainty as to the impurities removed during the course of zone refining (Youngblood thesis, 1963), all crystals utilized in this investigation were zone refined from polycrystalline rod with two passes at a rate of about $5.5$ mm/min., for a total molten time of approximately 1.5 minutes. Further zone refining was not believed to markedly increase purity and was thus not used in this investigation.

Although it is not known what the impurity concentration actually is, it is at least constant for this investigation. Because of the similarity of equipment and methods used, the purity levels of these crystals are believed to be about that reported by Lawley, Van den Sype and Maddin (1962). In light of work done by Youngblood and this investigator, no attempt was made to determine purity by hardness testing or chemical analysis, as all attempts to determine purity by these methods had been shown to be ineffective. Further attempts at carburization of polycrystalline molybdenum with subsequent zone refining met with failure upon introduction of only very small amounts of carbon. Crystals carburized to any
appreciable degree could not be formed into single crystals. Those slightly carburized crystals which were formed into single crystals showed no change in carbon content upon chemical analysis, nor any change in Knoop hardness number. For details of Knoop hardness determinations see Youngblood Thesis, 1963.

Starting material was in the form of polycrystalline rod purchased from the General Electric Company, Dover Wire Division, 200 West Broadway Street, Dover, Ohio, and had the following impurity analysis as purchased:

<table>
<thead>
<tr>
<th>Spectrographysis Analysis</th>
<th>Gases</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al .001</td>
<td>Oxygen 12 ± 5 ppm</td>
</tr>
<tr>
<td>Ca .002</td>
<td>Nitrogen 4 ± 4 ppm</td>
</tr>
<tr>
<td>Si .002</td>
<td>Hydrogen 2 ± 2 ppm</td>
</tr>
<tr>
<td>Fe .001</td>
<td></td>
</tr>
<tr>
<td>Cr &lt; .001</td>
<td></td>
</tr>
<tr>
<td>Ni &lt; .001</td>
<td></td>
</tr>
<tr>
<td>Cu &lt; .001</td>
<td></td>
</tr>
<tr>
<td>W .004</td>
<td></td>
</tr>
<tr>
<td>Mn &lt; .001</td>
<td></td>
</tr>
<tr>
<td>Mg &lt; .001</td>
<td></td>
</tr>
<tr>
<td>Sn &lt; .001</td>
<td></td>
</tr>
<tr>
<td>Co ND</td>
<td></td>
</tr>
<tr>
<td>Ag ND</td>
<td></td>
</tr>
<tr>
<td>Pb ND</td>
<td></td>
</tr>
<tr>
<td>Zr ND</td>
<td></td>
</tr>
</tbody>
</table>

B. Specimen Gripping Procedure

Each specimen was mounted in 3/8 inch pieces of hardened drill rod in the exact manner described by Youngblood (Thesis, 1963), except that Sauereisen cement was used exclusively in all tests. No specimen failure resulted from this cementing technique. The orientation of the crystallographic direction in which the specimen was mounted and subsequently sheared was determined by the Laue back-reflection X-ray technique. The crystal was then rotated from the X-rayed position to the desired orientation and mounted in the specimen holders. It is estimated that the entire operation resulted in the specimen being mounted to within 5 degrees of the desired orientation. In all cases, a [111] crystallographic direction was mounted so as to be in the direction of shear.
IV. EXPERIMENTAL APPARATUS AND PROCEDURE

A. General

Creep tests were performed in an especially built creep machine with a maximum capacity of 1,000 pounds exerted on the specimen. Temperature control was attained by immersing the compression test fixture housing the specimen in an isopentane bath alternately cooled by liquid nitrogen and heated by an electric heater. Strain measurement was made by a differential transformer device held above the bath; its output being recorded on a time base recorder.

B. Creep Testing Machine

A relatively light creep testing machine, having a multiplication ratio of about 20:1 and very low creep arm inertia, was built for this investigation. Figure (IV-1) is a photograph of this machine. The crosshead design is intended to be quite similar to that of an Instron Model TT-C-L tensile testing machine in order to allow complete use of the testing apparatus on either equipment. Loading was effected by pouring lead shot (BB size) into the weight pan at the end of the arm. Loading accuracy of ± 1 gram was easily attainable with this machine.

C. Strain Measuring Equipment

A Sanborn\(^1\) Model 7 DCDT-050 DC Differential Transformer was mounted on the Compression Test Chamber (Youngblood, Thesis, 1963) in the manner shown in Figure (IV-2). The thin-walled stainless steel extension tubes shown held the extensometer above the isopentane bath. It was necessary during tests to carefully shield the extensometer from air currents due to the temperature sensitivity of the device.
The DC differential transformer required a 6v d-c power source for proper operation. A Kepco\(^2\) Model ABC 30-0.3M 0.05% regulated power supply was programed for a constant voltage of 6v d-c.

A Varian Associates\(^3\) Model G-14 Strip Chart Recorder was used to record the d-c output of the differential transformer as a function of time. The recorder provides sensitivities of 1-, 10-, 100-millivolts, and a 1-volt full scale. Full scale zeroing plus one full-scale width zero suppression is provided for by the zero control. The range of the recorder was further incremented by construction of a resistance divider network, or Attenuator, which allowed 1/.2, 1/.4, 1/.6, 1/.8 or 1/1 ratios to be fed into the recorder input. An input resistance of 10,000 ohms was designed into this divider network to provide a constant load source for the extensometer. The recorder had a minimum permissible input impedance of 50,000 ohms, which was at no time exceeded, thus insuring the maximum pen speed of 1 second full scale. Figure (IV-3) gives the complete wiring schematic of the extensometry. The 15 microfarad capacitor connected across the input to the recorder damped out undesirable noise in the circuitry, caused principally by the 60 cycle current in the room power lines. It should be noted that the response time of the recorder was not altered by this capacitor\(^4\).

Calibration of the d-c differential transformer was made with a Boeckeler Mechanical Micrometer\(^5\) controlling the movement of the core.

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(1) Sanborn Company, Waltham 54, Massachusetts.
(2) Kepco, Inc., 131-38 Sanford Ave., Flushing 32, N. Y.
(3) Varian Associates, Instrument Division, Palo Alto, California.
(4) \[ t = RC, \ R = 10 \times 10^3 \text{ ohms}, \ c = 15 \times 10^{-6} \text{ farads} \]
\[ t = 10 \times 15 \times 10^{-3} = .15 \text{ seconds.} \]
The transformer calibrated linearly at 22.354 volts output per inch of core travel between +.060 and -.060 inches from electrical zero.

D. Constant Temperature Bath

An improved heat exchanger, similar to that used by Youngblood (Thesis, 1963) was employed in all tests except room temperature tests, which were performed in air. Isopentane was used as a heat-transfer medium for tests below room temperature, while lubrication oil was used for the test above room temperature. High purity liquid nitrogen was used as a coolant in the tests below room temperature. A Thermatrol\(^6\) Model 1053A operating in proportional mode was used to control the temperature; the on-cycle activating a solenoid valve attached to the liquid nitrogen tank, and the off-cycle actuating a 300 watt heating element immersed in the transfer fluid and part of the heat exchanger. Temperature control with this arrangement was held well within the desired \(\pm\) 2\(^\circ\) range. As an example, the temperature variation at 250\(^\circ\)K was held to \(\pm\) 1/4\(^\circ\)C for 15 minutes during a trial run to determine control capabilities. A resistance bulb thermometer was used as a sensing element with the Thermostat, and a thermocouple was placed in the immediate vicinity of the specimen to check the temperature of the bath.

E. Testing Procedure

Prior to beginning the actual test, calculations were made to determine the number of millivolts per percent of shear strain. During the test, sufficient load was added to the specimen to cause the desired strain (roughly) to take place. At each desired strain, an incremental loading

\[\text{(6) Maker: Hellinkainen Instruments, 1341 Seventh St., Berkely 10, Calif.}\]
test was performed, and the desired data of strain versus time recorded.

After gripping the specimen as previously described, it was then mounted in the loading feet in the same manner as Youngblood (Thesis, 1963, figures 13a and 13b). This assembly was placed in the compression test chamber (Youngblood, Thesis, 1963, figure 14). Figure (IV-2) shows the compression test chamber, with specimen inserted, loaded beneath the cross-head of the creep machine, ready to be tested.

The test was begun by immersing the test fixture below the creep machine cross-head in the constant temperature bath, and waiting until temperature control was established at the proper temperature. The Thermotrol settings and liquid nitrogen flow were adjusted until such control was established. Loading was then begun by pouring the lead shot into the weight pan until approximately 1% strain had taken place. The specimen was then allowed to creep out and an incremental load of 800 psi at the specimen was applied. During the incremental loading operation, the recorder time base was running at 16 inches per minute in order to achieve an enlarged look at the early stages of the ensuing creep. A typical portion of the strain-time information gained in the test is shown in figure (IV-4). From such a curve, all information used in this investigation was gained. The same procedure was repeated at 2%, 3%, 5%, 10%, 15%, 20%, 25%, 30%, 40%, 50%, 60%, 70% and in some cases, higher strains. At the end of the test, the specimen was removed and the total plastic strain measured optically to check it against the recorded plastic strain. In all cases measured, the difference was about 2% less actual plastic strain than that recorded.

In order to keep the differential transformer within its limits of linearity, it was occasionally necessary to "re-zero" the transformer
by moving the core of transformer nearer to electrical zero. This was accomplished by inserting a brass plug into a recession in the top of the core to screw or unscrew its threaded end into the inner extension rod (See figure IV-3).
Core Positioner

DC Differential Transformer

Connecting Cable

6 vdc, 0.05% Regulated Power Supply

Extension Tubes

Attenuator

15 Mfd. Capacitor

Varian Recorder

EXTENSOMETER ELECTRICAL CONNECTION

FIGURE (IV-3)
INCREMENTAL LOADING TEST

Chart Speed = 16"/min
5 mv. = Full Scale

\[ \tau = 28,550 \text{ psi} \]

"Rezeroing"
V. EXPERIMENTAL RESULTS

Incremental loading tests were conducted at 150°K, 200°K, 250°K, 275°K and 300°K. Figures V-1 through V-12 present the logarithm of strain-rate versus strain at 225°K. These curves result from an incremental loading of 800 psi at various strains, and are typical of the results obtained at all temperatures tested. Figures V-13 through V-18 show activation volume as a function of strain. Stress versus strain is shown in figure V-20 as a function of strain rate. The relation of flow stress and crystallographic orientation is shown in Figures V-21 and V-22 for crystals deformed in shear, and in figure V-23 for crystals deformed in tension. Figure V-24 shows what was thought to be internal stress, or the athermal component of stress, as a function of strain.

A lack of literature on creep deformation of molybdenum single crystals made it necessary to obtain experimental results before the adoption of a model could be attempted. After a few trial runs it became apparent that the strain-rate plotted logarithmically versus strain yielded a curve with essentially two linear portions of different slope connected by a short transition region. After expending considerable effort on various methods of analysis, it was decided that the results best fitted the pattern of two concurrently-operating, thermally-activated processes. Mathematically this is expressed by  
\[ \dot{\varepsilon} = A e^{-B\varepsilon} + C e^{-D\varepsilon}, \]
where A, B, C and D are selected to fit the equation to the data. Use of digital and analog computation methods was found not to be feasible for this analysis; thus, it was completed by graphical iteration. The quantities derived by this analysis are listed in Table A. The strain-rate before incremental load
application, $\dot{\varepsilon}_1$, the slopes of the straight portions of the curves, $\alpha$ and $\beta$, and the quantities

$$v^* = kT \frac{\ln c - \ln \dot{\varepsilon}_1}{\Delta T}$$

and

$$v^{**} = kT \frac{\ln(A + c) - \ln \dot{\varepsilon}_1}{\Delta T}$$

are also tabulated. $v^*$ and $v^{**}$ are also plotted versus strain in figures V-13 through V-18. The curves $v^*$ ($\dot{\varepsilon}$) are in very close agreement with those of Youngblood (Thesis, 1963).

The plot of logarithm of strain-rate versus strain showed a marked dependence on $\dot{\varepsilon}_1$, in that a small $\dot{\varepsilon}_1$ generally caused a pronounced initial region. It also appeared that, in general, the more pronounced the initial region, the closer it approached linearity.

The unusual nature of the curves of figure V-1 through V-12 indicates for a given stress a non-linear dependence of the logarithm of strain-rate on strain. The curves of figure V-20 corroborate this indication and, thereby, prove that the unusual results are metallurgical rather than measurement phenomena. As a further check, a copper single crystal was tested in the same manner as molybdenum. It exhibited the creep behavior of a f.c.c. single crystal, giving further evidence that the observations represented a physical phenomenon.

Because of the lack of knowledge of orientation dependence on the deformation properties of molybdenum, single crystals of various orientations were deformed at room temperature. Figures V-21, V-22 and V-23 show the results. Results of the tension tests agree well with those in tungsten and tantalum done by Rose, Ferriss and Wulff (1962) and Ferriss, Rose and Wulff (1962).

Alefeld (1962), through his reemphasis of the fundamental equations governing thermal activation, provided the basis for a series of
experiments expected to give "Internal Stress" versus strain, as shown in figure V-24.

No information is included on the crystal tested at 400°K because the extreme plasticity yielded results unrelatable to the other temperatures tested.

With exception of fractured specimens, all those creep tested were ground about half the diameter perpendicular to the shear direction and along the specimen axis, mounted in bakelite, polished, etched and examined metallographically. The following observations were made:

125°K Specimen: Two specimens tested fractured by cleavage along (100)-planes in region of both grips at approximately 10 per cent strain.

150°K Specimen: Very rough cleavage fracture in one grip, partially on (100)-planes and partially on (110)-planes.

200°K Specimen: Smooth shearing across whole section. Radii of both sides about equal. Very little polygonization.

225°K Specimen: Grips cut into specimen about 1/6 diameter. No polygonization. Irregular, unpitted lines, emanating from grip cut, roughly perpendicular to surface.

250°K Specimen: Specimen evidently polycrystalline, having one large surface grain with two others in the center. Grains clearly outlined by etch pits. Wavy, unpitted lines running both parallel and perpen-
dicular to surface visible in vicinity of gripping cuts and extending about 2/3 across crystal. Grip cuts about 1/10 diameter.

275°K Specimen: Grip cuts about 1/5 diameter. Very little polygonization visible. A high density of roughly parallel, unpitted lines emanate from grip cuts. Some lines start and stop within the crystal. Some lines seen throughout the cross-section.

300°K Specimen: Grip cuts about 1/10 diameter. No polygonization visible. Unpitted, slightly curved lines seen emanating from vicinity of grip cuts, extending almost across crystal. A few closely-spaced, parallel, unpitted lines, perpendicular to surface visible.
<table>
<thead>
<tr>
<th>(\varepsilon)</th>
<th>(T)</th>
<th>(\dot{\varepsilon}_1)</th>
<th>(\dot{A})</th>
<th>(B)</th>
<th>(C)</th>
<th>(D)</th>
<th>(V^*)</th>
<th>(V^{**})</th>
<th>(c)</th>
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<tr>
<td>(%)</td>
<td>(psi)</td>
<td>(10^{-6}) sec(^{-1})</td>
<td>(10^{-6}) sec(^{-1})</td>
<td>(10(^{-6}) sec(^{-1})</td>
<td>(10(^{-23}) cm(^3))</td>
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**NO 189-1 275°K**

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**NO 190-1 300°K**

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| 3.9  | 11,430  | 9.0    | 130   | 250   | 35  | 12   | 139   | 297   | 148  | 19.3 |
| 5.3  | 12,600  | 17     | 380   | 200   | 120 | 43   | 200   | 344   | 112  | 5   |
| 9.0  | 15,500  | 19     | 110   | 135   | 65  | 15.5 | 124   | 226   | 64   | 7.6 |
| 10.1 | 16,650  | 19     | 90    | 16    | 158 | 34.3 | 2.1   |
| 10.9 | 18,250  | 20     | 152   | 18    | 207 | 18   |
| 15.6 | 20,500  | 22     | 212   | 27.2  | 230 | 27.2 |
| 16.4 | 21,100  | 28     | 135   | 110   | 105 | 10.8 | 148   | 240   | 50   | 11.2 |
| 20.4 | 22,200  | 31     | 225   | 94    | 225 | 94   | 221   | 298   | 94   |
| 33.6 | 25,500  | 65     | 363   | 150   | 130 | 0    | 59    |
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Incremental load was in all cases 800 psi.
225^\circ K, \varepsilon = 0.76\%\nT = 2,750 ~\text{psi}
A = 48 \times 10^{-6} ~\text{sec}^{-1}
B = 1000
C = 11 \times 10^{-6} ~\text{sec}^{-1}
D = 170

FIGURE (V-1)
225°K
$\varepsilon = 3.2\%$
$T = 15,500 \text{ psi}$
$A = 170 \times 10^{-6} \text{ sec}^{-1}$
$B = 280$
$C = 100 \times 10^{-6} \text{ sec}^{-1}$
$D = 58$
$225^\circ K$

$\varepsilon = 5.9\%$

$\dot{\varepsilon} = 22,350 \text{ psi}$

$A = 52 \times 10^{-6} \text{ sec}^{-1}$

$B = 300$

$C = 78 \times 10^{-6} \text{ sec}^{-1}$

$D = 24$

**Figure (V-3)**
$225^\circ$K
$\varepsilon = 15.3\%$
$\tau = 32,550$ psi

$A = 155 \times 10^{-6}$ sec$^{-1}$
$B = 115$
$C = 195 \times 10^{-6}$ sec$^{-1}$
$D = 14$
225°K
ε = 20.1%
τ = 34,950 psi
A = 195 \times 10^{-6} \text{ sec}^{-1}
B = 44
C = 155 \times 10^{-6} \text{ sec}^{-1}
D = 3.8
$T = 225^\circ K$

$C = 2.25 \%$

$T = 3,750 \text{ psi}$

$A = 165 \times 10^{-6} \text{ sec}^{-1}$

$B = 4.8$

$C = 115 \times 10^{-6} \text{ sec}^{-1}$

$D = 1.8$

**Figure (V-7)**
225°K
\[ \varepsilon = 25.1\% \]
\[ T = 36,550 \text{ psi} \]
\[ A = 210 \times 10^{-6} \text{ sec}^{-1} \]
\[ D = 64 \]
\[ C = 120 \times 10^{-6} \text{ sec}^{-1} \]
\[ D = 2.25 \]

**FIGURE (V-8)**
225^\circ K
\epsilon_0 = 3.49 \%
\tau = 38.150 \text{ psi}
A = 320 \times 10^{-6} \text{ sec}^{-1}
B = 80
C = 137 \times 10^{-6} \text{ sec}^{-1}
D = 2.1

FIGURE (V-9)
225°K
ε₀ = 41.8%
τ = 39,750 psi

A = 520 \times 10^{-6} \text{sec}^{-1}
B = 66
C = 300 \times 10^{-6} \text{sec}^{-1}
D = 1.0
225°K
$\epsilon_o = 50.1\%$
$\tau = 40,550$ psi
$A = 230 \times 10^{-6}$ sec$^{-1}$
$B = 50$
$C = 320 \times 10^{-6}$ sec$^{-1}$
$D = 0.7$
$225^\circ K$

$\varepsilon_0 = 61.7 \%$

$\gamma = 41,350 \text{ psi}$

$A = 2.75 \times 10^{-6} \text{ sec}^{-1}$

$B = 2.5$

$C = 4.25 \times 10^{-6} \text{ sec}^{-1}$

$D = 0.1$

**Figure (V-12)**
Activation Volume vs Strain 200°K

FIGURE (V-14)
FIGURE (V-15)

Activation Volume vs Strain 225°K
STRESS vs STRAIN
Various II0-Plane, III-Directions
300°K
$\varepsilon = 270 \times 10^{-6} \text{sec}^{-1}$

FIGURE (V-21)
STRESS vs STRAIN
Planes Having a Ill-Direction
300°K \( \dot{\varepsilon} = 270 \times 10^{-6} \text{ sec}^{-1} \)
Shear \( \bigcirc \) Force

\[ \tau \times 10^3 \text{ psi} \]

\[ \varepsilon \text{ (\%) } \]

FIGURE (V-22)
Tensile Tests
Mo. Single Crystals
$\dot{\varepsilon} = 0.25 \text{ min}^{-1}$
$300^\circ\text{K}$
VI. DISCUSSION OF RESULTS

A. General

The principal finding of this investigation was that the results obtained from incremental loading cannot be described by $\dot{\varepsilon} = A(1 + t)^m$, as it can for the close packed metals. Further, by fitting the equation $\dot{\varepsilon} = Ae^{-B\Delta e} + Ce^{-D\Delta e}$ to the data, activation volumes calculated from the expression

$$v^* = kT \frac{\ln C - \ln \dot{\varepsilon}_1}{\Delta \tau a}$$

(1)

were in very close agreement with those reported by Youngblood. Difficulties of measuring parameters in the region characterized by $Ae^{-B\Delta e}$ resulted in very little insight into the mechanisms of this region.

B. Creep Experiments

1. The equation $\dot{\varepsilon} = Ae^{-B\Delta e} + Ce^{-D\Delta e}$

(2)

fits the data well, therefore, implying the possibility of the existence of two simultaneously-operating, thermally-activated processes.

Beginning with the concept that a thermally-activated process may be described by

$$\dot{\varepsilon} = \dot{\varepsilon}_0 e^{-\frac{H_o - vT}{kT}}$$

(3)

Expanding $vT$ about $\Delta e = 0$ gives

$$vT = vT_0 + \frac{\partial}{\partial \Delta e} (vT)$$

(4)

Where

$$\bar{T}_o = T_a - T_{go}$$

$T_{go}$ = Internal stress at $\Delta e = 0$

$\bar{T}_o$ = Effective applied stress at $\varepsilon = 0$
If the above is applied to the formulation of Alefeld (1962) then a very complex form of \( \dot{\epsilon} \) results in which the internal stress and the applied stress are intermixed. Fortunately, however, the strain-rates observed in this investigation are generally above those to which the Alefeld equations are applicable.

Assuming that over the regions where eq. (1) applies, \( \dot{\epsilon} \) is high enough to allow the approximation \( 2\sinh x = e^x \), the following formulation may be developed:

\[
\dot{\epsilon} = \dot{\epsilon}_1 e^{-\frac{H_{01} - v_1 \tilde{T}}{kT}} - \dot{\epsilon}_2 e^{-\frac{H_{02} - v_2 \tilde{T}}{kT}}
\]

where the subscripts 1 and 2 describe the two thermally-activated processes.

Putting (2) into the above equation gives

\[
\dot{\epsilon} = \dot{\epsilon}_1 e^{-\frac{H_{01}}{kT} v_1 \tilde{T}_0 / kT} - \dot{\epsilon}_2 e^{-\frac{H_{02}}{kT} v_2 \tilde{T}_0 / kT}
\]

\[
\dot{\epsilon}_1 e^{-\frac{H_{01}}{kT} v_1 \tilde{T}_0 / kT} = \frac{1}{e^{\frac{H_{01}}{kT} v_1 \tilde{T}_0 / kT}} \frac{\partial}{\partial \Delta \epsilon} (v_1 \tilde{T})\Delta \epsilon = -H_{01} / kT v_1 \tilde{T}_0 / kT
\]

From this equation the constants \( A, B, C \) and \( D \) may be interpreted as

\[
A = \dot{\epsilon}_1 e^{-\frac{H_{01}}{kT} v_1 \tilde{T}_0 / kT}
\]

\[
B = -\frac{1}{kT} \frac{\partial}{\partial \Delta \epsilon} (v_1 \tilde{T})
\]

\[
C = \dot{\epsilon}_2 e^{-\frac{H_{02}}{kT} v_2 \tilde{T}_0 / kT}
\]

\[
D = -\frac{1}{kT} \frac{\partial}{\partial \Delta \epsilon} (v_2 \tilde{T})
\]

2. The data collected experimentally by this investigation contain considerable scatter, due mainly to two factors: the method of determining strain-rate, and the method of determining the constants \( A, B, \)
C and D. Strain-rate was visually determined by taking the slope of the strain-time curve. Determination of the constants A, B, C and D was determined assigning them values, putting these into eq. (1), and fitting the resulting curve to the data. This process was repeated until a reasonably accurate fit of the curve to the data was achieved.

In spite of the scatter in the data, some general trends may be noted:

a. A decreases as $\dot{\varepsilon}_1$ increases
b. B decreases as $\dot{\varepsilon}_1$ increases
c. C increases as $\dot{\varepsilon}_1$ increases
d. D decreases as $\dot{\varepsilon}_1$ increases

The first, or high strain-rate, region where A and B are the controlling constants is quite pronounced for low $\dot{\varepsilon}_1$, and the curves are most nearly linear in this region. As $\dot{\varepsilon}_1$ increases, the first region is suppressed. At large strains steady-state creep sets in, causing D to approach zero, and suppressing A and B.

e. At 125°K, which is below the brittle-ductile transition, there is virtually no evidence of the region $Ae^{-B\Delta\varepsilon}$. This observation is made for the low strain region only, where the effect should be the strongest, because of crystal cleavage at about ten per cent strain.

3. Examination of the second region, where $\dot{\varepsilon} = Ce^{-D\Delta\varepsilon}$: This region is of primary interest due to correlation, as determined by eq. (1), of the work done by Youngblood with the experimentally-determined quantities C and D of this investigation.
From eq. (8), considering one incremental test,

$$kT = v_2 \frac{\Delta \tau}{\Delta \varepsilon} + \tau \frac{\Delta v_2}{\Delta \varepsilon} \quad \text{at } T \text{ const.} \hspace{1cm} (9)$$

By this formulation, $\frac{\partial \tau}{\partial \varepsilon}$ may be found using two different approaches:

a. Assume a Peierls-Nabarro model. $v_2 \neq f(\varepsilon)$, and $\frac{\partial v}{\partial \varepsilon} \bigg|_T = 0$.

Therefore,

$$\frac{\partial \tau}{\partial \varepsilon} = - \frac{kT}{v_2} \quad \text{at } T \text{ const.} \hspace{1cm} (10)$$

From (1) $v_2 = v^*$

b. Assume some model is rate controlling in which

$v_2 = f(\varepsilon)$. Letting $\delta_0 = \delta_{o2} e^{-H_0^2/kT}$

then

$$\frac{\partial \tau}{\partial \varepsilon} = \frac{kT}{v^*} \left[ \frac{1}{\delta_{o2}} \ln \frac{C}{\delta_{o2}} - D \right] \quad \text{at } T \text{ const.} \hspace{1cm} (11)$$

In (b) the value of $\delta_{o2} = 4 \times 10^9$ per sec. and $H_0^2 = 1.12$ ev as determined by Youngblood can be used. Calculation of $\frac{\partial \tau}{\partial \varepsilon}$ by eq. (10) yields a positive quantity, whereas eq. (11) yields a negative quantity. Physically, $\frac{\partial \tau}{\partial \varepsilon}$ must be a positive quantity.

The slope of the stress-strain curve may also be derived, using eq. (7).

At const. $T$, $\frac{\partial \tau}{\partial \varepsilon} \bigg|_T = \frac{d \tau}{d \varepsilon}$

$$\tau \equiv \tau = \tau - \tau_g = \frac{kT}{v^*} \ln \frac{C}{\delta_{o2}} \hspace{1cm} \text{and utilizing eq. (10)}$$

$$\frac{d \tau}{d \varepsilon} = \frac{kT}{v^*} \left[ \frac{d \ln C}{d \varepsilon} - \frac{d \ln v^*}{d \varepsilon} \frac{\ln C}{\delta_{o2}} - D \right] \hspace{1cm} (12)$$

This equation is formulated on the Peierls-Nabarro model.
Calculations based on eqns. (10) and (12) are shown in figures (VI-1) and (VI-2) for 225°K and 250°K. Similar results were obtained for the other test temperatures. Since the term \[ \frac{dlnx_k}{de} \] is extremely sensitive to the exact determination of \( x^k \), the figures should be interpreted qualitatively rather than quantitatively. \( \frac{\Delta T_a}{\Delta e} \) appears to be reasonable.

4. The First Region

a. An interpolation of \( x^{**} \), which is listed in Table A.

\( x^{**} \) was determined from the experimental variables \( A \) and \( C \) by

\[
x^{**} = \frac{kT \ln (A + C) - \ln \delta}{\Delta T_a}
\]  

(13)

This is derived considering that the strain-rate at \( \Delta e = 0 \) is

\[
\dot{\varepsilon} = A \cdot C \equiv \dot{\varepsilon}_0 \cdot e^{-\frac{H_0 - (\nu T)}{kT}}
\]  

(14)

where \( \dot{\varepsilon}_0 \), \( H_0 \) and \( \nu \) are parameters associated with both processes acting together.

Rewriting on eq. (14) gives

\[
kT [\ln (A + C) - \ln \delta] = H_0 + (\nu T)_0,
\]

and assuming, as was done in eq. (1), that \( \Delta T = \Delta T_a \) and that \( \delta_0 \neq f (T, e) \), then

\[
\frac{\ln (A + C) - \ln \delta}{\Delta T_a} = \frac{H_0}{\Delta T_a} + \nu \quad \text{at} \quad \Delta e = 0
\]

and from eq. (13)

\[
x^{**} = -\frac{H_0}{\Delta T_a} + \nu
\]  

(15)

This, \( x^{**} \) represents some activation volume, \( \nu \), in addition to the change in activation energies associated with going from the mechanism controlling \( \dot{\varepsilon}_1 \) to that controlling \( \dot{\varepsilon} \) after the addition of \( \Delta T_a \). Experimentally, \( \nu \) has
not been found, so determination of \( \frac{\Delta H_0}{\Delta r_a} \) is impossible.

b. As may be seen from Table A, for low strains and low strain-rates, \( A > C \), whereas at high strains and strain-rates \( A < C \).

If it is assumed that \( H_{O_1} > H_{O_2} \), and that \( \ln \frac{\delta_{O_1}}{\delta_{O_2}} \) changes very little with strain, then it can be shown that

\[-kT \ln \frac{\delta_{O_1}}{\delta_{O_2}} + (H_{O_1} - H_{O_2}) < \eta(v_1 - v_2) \text{ for low } \epsilon\]

which implies that \( \tau v_1 >> \tau v_2 \). Likewise,

\[-kT \ln \frac{\delta_{O_1}}{\delta_{O_2}} + (H_{O_1} - H_{O_2}) > \eta(v_1 - v_2) \text{ for high } \epsilon\]

which implies that \( \tau v_1 \approx \tau v_2 \)

It thus was reasonable to assume that \( H_{O_1} > H_{O_2} \) because the first region, which is controlled by \( H_{O_1} \) and \( v_1 \), appears most strongly at low strains, and gradually disappears at higher strains. These trends are further supported experimentally, since \( B > D \).

c. It thus appears that the first region is characterized by a higher activation energy than is the second. Since Youngblood finds \( H_{O_2} \) between 1.04 ev and 1.3 ev, \( H_{O_1} \) should be somewhat greater. The only mechanism known to have activation energies of this magnitude is the non-conservative motion of dislocations, which could mean climb of edge dislocations or dragging of jogs in screw dislocations. Non-conservative motion of dislocations requires the migration of vacancies through the lattice to the dislocation lines, which in molybdenum requires about 1.5 ev at room temperature, and increases as for lower temperatures.
It is felt that cross-slip is the most reasonable explanation for the controlling mechanism in the first region, although it is not understood why it should have a high activation energy. This feeling comes primarily from those observations described in detail in Section II, which show that b.c.c. metals do not exhibit much tendency to cross-slip at low temperatures, but have an increasing tendency to do so as the temperature is increased. Observations of this investigation indicate that below the ductile-brittle transformation temperature, the first region is not found, and that as temperature is increased above the brittle-ductile transition, the first region becomes increasingly pronounced. Finally, steady-state creep ensues, in which the two regions apparently merge. It is felt that a lack of cross-slip at temperatures below the brittle-ductile transition would greatly increase the probability of a reaction such as that described by Cottrell (1958) which would give cleavage fracture, such as has been observed.

C. Internal Stress

Alefeld (1962) explained in detail why the approximation of eq.(3),
\[ \dot{\varepsilon} = \dot{\varepsilon}_0 e^{-\frac{H_0}{kT} + \frac{\nu kT}{kT}} \], must be replaced by
\[ \dot{\varepsilon} = 2\dot{\varepsilon}_0 e^{-\frac{H_0}{kT} \sinh \left( \frac{\nu kT}{kT} \right)} \] eq.(16)
at very low strain-rates and effective stress levels. Thus, it appeared that it would be possible to measure the internal stress, i.e. \( \gamma = 0 \), thus \( \tau = \tau_g \), if a low enough strain-rate could be found. For low temperatures, this would require an extremely low strain-rate (10^{-25} per sec at 150^\circ K). Thus, stress strain-curves were run at the slowest strain-rate that could be achieved in the Instron testing machine, using a reasonable gauge length; i.e., 40 x 10^{-6} per sec.
In either eq. (3) or eq. (16), values of \( v^* \), \( H_{o2} \), and \( \varepsilon_{o2} \) must be known to calculate \( \bar{\tau} \). Values of \( H_{o2} \) and \( \varepsilon_{o2} \) found by Youngblood, and the lowest observed values of \( v^* \) observed in this investigation were used to calculate \( \bar{\tau} \). See Table C.

<table>
<thead>
<tr>
<th>( T ) (ok)</th>
<th>( v^* ) ( x10^{-23} )</th>
<th>( \frac{v^*}{kT} ) ( x10^3 )</th>
<th>( \bar{\tau} ) ( x10^3 )</th>
<th>( \bar{\tau} ) ( x10^8 )</th>
</tr>
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<tr>
<td>150</td>
<td>32</td>
<td>1.07</td>
<td>50</td>
<td>34.5</td>
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<tr>
<td>200</td>
<td>30</td>
<td>0.75</td>
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<td>22.0</td>
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<td>225</td>
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<tr>
<td>250</td>
<td>40</td>
<td>0.80</td>
<td>24</td>
<td>16.5</td>
</tr>
<tr>
<td>275</td>
<td>40</td>
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<td>13.8</td>
</tr>
<tr>
<td>300</td>
<td>42</td>
<td>0.70</td>
<td>16</td>
<td>11.0</td>
</tr>
</tbody>
</table>

Figure (VI-3) shows \( \bar{\tau} \) versus temperature for large strains. The intercept for \( \bar{\tau} = 0 \) is at \( 365^\circ K \), indicating this is the temperature at which the internal stress of the material is no longer temperature-dependent. This confirms the findings of Youngblood (Thesis, 1963, Fig. 22) in which he found the temperature independence at 60 per cent strain to occur at about \( 375^\circ K \). By subtracting the values of \( \bar{\tau} \) as computed in Table C from the curves of figure (V-24), it appears that the approximation is good for all strains down to 20 per cent, and that a similar treatment may be applied at any strain.

D. Orientation Dependence

Due to the lack of information on low temperature mechanical
Effective Stress vs Temperature
\[ \epsilon > 50\% \]

FIGURE (VI-3)
properties of molybdenum as a function of orientation, testing was carried out in an attempt to gain some insight into this phase of the problem. The tension tests of figure (V-23) showed about the same result as Rose, Ferriss and Wulff (1962) found in tungsten.

The (110)-plane has two [111] directions in it. The results of crystals sheared with these two directions at various angles to the shear direction are shown in figure (V-21). The lower curve has one slip direction in the direction of shear and one 70.5° away, thus, 144% of the shear force is acting in the [111]-directions. Likewise, the crystal of the middle curve has two [111]-directions at 35.2° from the shear direction, causing a total of 163.2% of the shear force to be acting on them. It seems likely, then, that this crystal would be softer, not harder, than the first crystal. The crystal of the upper curve has one slip direction perpendicular to the direction of applied shear, thus having no resolved force on it, and one slip direction 15.5° away from the direction of shear force application. The slip directions of this crystal should, therefore, be acted upon by 96% of the applied shear force. By ratioing the force applied to the slip systems of the last crystal to that applied on the first, a numerical multiplier of 0.72 is found. By ratioing the actual stress of the last crystal to that of the first, an average ratio of 0.76 is found. Thus, fairly good agreement is found in applying geometrical factors to the crystals having a predominance of stress applied to one of the slip directions, but for the case of equal stress application such geometric factors apparently do not apply. It may thus be concluded that distortion interaction resulting from two equally operative slip directions impedes motion in both directions.
The difficulty of dislocation movement on other planes containing a [111]- direction is shown in figure (V-22). Dislocation motion on the (112)- plane is much more difficult than on the (110)- plane. The difference in stresses for the (123)- and (110)- planes cannot be resolved even if all slip thought to be occurring on the (123)- plane is actually occurring on the nearby (110)- plane.

It may thus be concluded that at 300°K, slip is increasingly more difficult on the (110)-, (123)- and (112)- planes, respectively.
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