UCRL-11194

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THE MECHANICAL BEHAVIOR OF THE INTERNALLY OXIDIZED SINGLE CRYSTALS OF SILVER AND SILVER BASE ALLOYS

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Research and Development

UCRL-11194

UNIVERSITY OF CALIFORNIA

Lawrence Radiation Laboratory
Berkeley, California

AEC Contract W-7405-eng-48

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(Ph.D. Thesis)

August 1965

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THE MECHANICAL BEHAVIOR OF THE INTERNALLY OXIDIZED SINGLE CRYSTALS OF SILVER AND SILVER BASE ALLOYS

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ABSTRACT

August 1965

The presence of oxygen in the amount of 70 ppm in high-purity silver single crystals is found to strengthen the matrix. Oxygen, which is known to dissolve interstitially in the silver lattice, affects the mechanical behavior of these crystals in a manner analogous to that of substitutional solutes. A yield drop phenomenon, observed in oxygen bearing silver crystals oriented for single slip, has been explained on the basis of 'Dislocation Multiplication Mechanism' proposed by Gilman and Johnston. It is suggested that dislocations are pinned by oxygen atoms and this results in the reduction of the number of mobile dislocations.

The considerable increase in strength which occurs on internal oxidation of silver-magnesium and silver-cadmium alloy single crystals has been attributed to the increase in number of dislocation-dislocation interactions. A similarity between the Cottrell-Stokes ratios for pure silver and the oxidized materials suggests that the deformation in both cases is controlled by the same mechanism. The high work hardening rate exhibited by these materials in the early stages of deformation has been attributed to the high initial dislocation density and its subsequent rapid increase with strain.

In partially oxidized alloy crystals the slip sources in the unoxidized zone start operating first. With a further increase in strain these dislocations interact with the subsurface strengthened layer and eventually break through. The macroscopic yield point is associated with the emergence of dislocations at the surface.

Fred G. Parker

I. INTRODUCTION

There is a considerable discrepancy between the theoretically predicted yield strength and the observed yield strength of pure metals. This difference has been attributed to the presence of linear defects called "Dislocations"; they are supposed to be an inherent constituent of any crystal structure. The ease in the mobility or the generation of these defects controls the initial value of the flow stress in solids. Because of the low value for yield strengths, pure metals hardly find a wide-spread application in industry. In recent years, intensive effort has been devoted to the study of strengthening mechanisms in solids and this in turn has led to the development of suitable materials for certain applications demanding a particular combination of properties.

Strengthening mechanisms of solids can be classified under the following headings: (a) elastic strengthening, (b) yield strengthening. (c) solution strengthening, (d) strain hardening, (e) quench or vacancy strengthening, (f) radiation strengthening, (g) martensitic strengthening, and (h) precipitation and particle strengthening.

(a) Elastic Strengthening. This is encountered in small single crystals called whiskers. The high value for the yield strength, which is of the order of 2×10^6 psi, is associated with the difficulty of generating new dislocations because of the perfection of the crystal. However, the bulk crystals do not have these unique strengths because they contain heterogeneities that can lead to dislocation nucleations.

(b) Yield Strengthening. An important measure of strength is the

yield stress — the stress required to move dislocations in a crystal.

Johnston and Gilman[1] have demonstrated that in lithium fluoride the macroscopic yield strength is linearly proportional to the stresses required to move fresh dislocations at a reasonable velocity. Furthermore, the yield strengths are not determined by 1) the stress to pull dislocations from impurity atmospheres, 2) the stress to push dislocations through a forest of other dislocations, nor 3) the stress to operate Frank-Read sources. Rather, they are determined by the frictional resistance of the lattice to the dislocation motion. It is then apparent that in certain types of solids the ease with which dislocations move in the lattice controls the flow stress. Recently much research effort has gone into the study of the factors influencing dislocation mobilities and this has resulted in a better understanding of the behavior of solids.

(c) Solution Strengthening. The increment in yield strength is quite small on solid solutioning but the problem itself is of considerable theoretical interest.

The interactions between moving dislocations and stationary solute atoms were considered by Mott and Nabarro[2] for random solute dispersions, and more recently by Fleischer[3] for non-uniform impurity concentrations. The general picture is that the dislocation line is a flexible string that vibrates and radiates elastic energy as it passes or is "plucked" by solute atoms. This energy must be supplied by additional applied stress. If the solute concentration in a crystal varies, then the lattice parameter will vary and the elastic forces will vary. They can be pictured as impairing the motion of dislocation lines and the stress required to achieve a given velocity.

(d) Strain Hardening. It is well established that once plastic deformation has started the stress required to cause additional deformation increases. This effect has been associated with the increase in the number of intersections between the forest and mobile dislocations and also with the increment in long-range back stresses whose source of origin is still a controversial matter. But there is a limit to the strength level that can be attained by plastic deformation because the dislocation density almost reaches a saturation value of $10^{12}/cm^2$ quite early in the deformation sequence and beyond which no significant gain is made in strength level on subsequent deformation. Consequently, this technique is hardly of any practical importance.

(e) Vacancy Strengthening. It has been known that the yield strength of a metal increases when it is quenched from a temperature close to its melting point rapidly enough to retain the high-temperature vacancy concentration. This effect is considered to be associated with the interaction of moving dislocation with dislocation loops which result from the collapse of these vacancy discs. This technique does not find universal application because in some body centered cubic metals it is very difficult to quench in the supersaturation of vacancies.

(f) Radiation Strengthening. Radiation strengthening is very similar to vacancy strengthening in concept except for the difference in mode of production of point defects. The lattice defects, which are produced on irradiation, cluster to form loops which in turn interact with the mobile dislocations to cause hardening.

(g) Martensitic Strengthening. In general, the hardness of martensite is increased by increasing carbon content associated with increased lattice strain. Recently Schmatz and Zackay[4] have reported additional

strengthening due to a marked refinement in the microstructure caused by deformation of austenite prior to transformation. This unique combination of deformation and transformation is going to be an important tool in attaining desirable strength levels.

(h) Precipitation and Particle Strengthening. It is well known that metals can be hardened by a dispersion of discrete particles of a second phase. Various theories have been developed to explain the increase in yield strength of these materials as well as their subsequent work hardening characteristics. Orowan[5] demonstrated that the shear stress τ_y necessary to force a dislocation through the matrix between incoherent particles of precipitate a mean distance 2ℓ apart on the slip plane is given by

$$\tau_y - \tau^* = \frac{G_b b}{\ell} \qquad (1)$$

where τ^* is the local long range back stresses, G is the shear modulus of elasticity, and b is the Burger's vector. The yield strength should not exhibit a strong temperature dependence if Orowan's mechanism is rate controlling because the temperature dependence of flowstress is related to the temperature sensitivity of the shear modulus and this is quite small. In order to explain the rapid work-hardening observed in dispersion-strengthened materials, Fisher, Hart and Fry[6] extended Orowan's theory. They attribute the increase in flowstress to the rapid increase in backstresses as a result of the elastic interaction between the left-over loops round the particles and the glissile dislocations.

Kelly and Chiou[7] and Byrne et al.[8] suggested an alternative strengthening mechanism on the basis of their work on aluminum-copper alloys containing coherent precipitates. They proposed that strengthening

can result from the atomic scrambling which occurs and the local strains which must be overcome when a dislocation moves through a zone.

Hirsch[9] concluded from Thomas and Nutting's[10] observations on dispersioned-hardened materials that the dislocations can bypass particles by cross-slip leaving dislocation loops around the particles. By utilizing these concepts, Ashby[11] has been able to explain the origin of observed prismatic loops around the silica particles in internally oxidized copper-silicon alloys.

There are numerous ways to produce disperions in metal matrices, but the widely used technique consists of a solution treatment of a suitable alloy followed by an aging treatment. The nature, size and distribution of the precipitating phase can be altered by varying the temperature and time of aging. Only a few systems such as aluminum-copper, copper-beryllium, etc. can be strengthened to a sufficient extent so as to be of any practical importance.

Recently the technique of internal oxidation has been used to an advantage to produce dispersions in suitable alloy systems.[12] This essentially involves the removal of solute atoms from the metal matrix by combining these atoms with oxygen. In general an alloy system should meet the following requirements to be suitable for this treatment. (1) Oxygen must diffuse more rapidly in solvent metal than solute does; otherwise, a compound will not be formed on the interior, but as a surface layer. (2) The affinity of oxygen for solute atoms must be sufficiently greater than for solvent atoms.

Oxygen diffuses through silver,[13] copper,[14] and nickel[15] with a speed considerably greater than the metallic elements. Furthermore its affinity for silver is so small that silver cannot form (solid)

Ag_2O when heated above 190°C in oxygen of normal pressure. This affinity is larger for copper and nickel. Consequently the silver alloy system appears to be the most satisfactory on the basis of the considerations outlined above.

The internally oxidized materials exhibit high creep-resistance at high-temperatures[16] and the recrystallization phenomenon is considerably inhibited by the dispersed particles.[17] Further clarification is necessary concerning this unusual behavior and the microstructure of these materials.

Recently the experiments on the temperature dependence of the flowstress of pure metal single crystals[18,19] have yielded valuable information on the nature of the elementary dislocation processes which determine the flowstress. It is of particular interest to extend those measurements to the internally oxidized materials. In this way information can be obtained on the nature of the interaction between dislocations and the dispersed phases. At first sight this would appear a truly formidable task since the interactions of a large number of dislocations with one another in a pure metal are not understood. However, in considering the internally oxidized materials the increase in strength brings on its train some simplification of the problem since, in certain cases, the obstacle impeding the motion of dislocations is known, and other possible contributions to the flow stress can almost be completely neglected.

Dislocation theory suggests that a thin surface film may exert an important effect on the initiation of slip in crystals if this is due primarily to surface sources. It has been explained by Fisher[20] that a surface Frank-Read source can result from a dislocation segment that is

pinned at one end with the other end terminating at the surface of the crystal. In effect, the stress required to generate dislocation loops at the average surface source should be about half that for generating loops at the average interior source because the interior sources are twice as short. The role of surface sources can be conveniently studied by partial oxidation of single crystals of suitable alloy system and the depth of the zone in which dislocation sources are to be affected can be varied by altering the time of oxidation. In fact, this technique provides a convenient way of studying the effect of barrier size on the extent of easy glide, the rate of work hardening and the critical shear stress.

The effects of substitutional solid solutioning on the characteristics of stress-strain curve for a face centered cubic metal have been studied extensively.[21,22] It will be interesting to study the effects of oxygen, which is known to diffuse interstitially in silver,[13] on the mechanical behavior of the crystals oriented for single slip because such a study will shed some light on the nature of interstitial-dislocation interaction in the metal.

II. EXPERIMENTAL TECHNIQUES

A. Materials

The materials selected for the present investigation were ultra-high purity (99.9999%) oxygen-free (< 1 ppm) silver* and two silver-based alloys containing approximately 1.0 at % Mg and 1% Cd, respectively. The Ag-Mg alloy was prepared by induction melting appropriate quantities of 6-9's Ag and sublimed Mg (99.995%) in a thoroughly outgassed high-purity graphite crucible under a purified argon atmosphere. Essentially the same procedure was followed in preparing the Ag-Cd alloy except that the initial Cd content was increased to about 2 at % in an effort to compensate for the rather extensive loss of Cd due to vaporization during the melting and subsequent crystal growing operations. Samples taken from various portions of the ingots established that the composition was uniform to within ±0.02 at %. The average analyses of the alloy ingots were as follows: Ag-0.98 at % Mg; Ag-1.65 at % Cd.

B. Specimen Preparation

Since previous experience had shown that polycrystalline samples of these alloys became highly embrittled as a result of internal oxidation, all of the specimens used in this work were in the form of single crystals with gauge dimensions of approximately 0.25 in. X 0.30 in. X 1.50 in. (6.4 mm X 0.75 mm X 38 mm). All crystals were grown from the melt in well outgassed reactor-grade graphite crucibles using a modified Bridgman technique. The high-purity silver and Ag-Mg alloy crystals were grown in

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* Supplied by Consolidated Mining and Smelting Company of Canada Ltd.

vacuum (~ 10^{-6} mm Hg) while the Ag-Cd alloy crystals were grown in a purified helium atmosphere to minimize volatization of Cd. Crystals of two predetermined orientations were grown by using conventional seeding techniques. One group of crystals was oriented for multiple glide, i.e., with < 100 > parallel to the tensile axis; the other group was oriented for single slip. The < 100 > oriented crystals were grown so that the flat faces of the crystals were parallel to (110); in the crystals oriented for single slip the flat faces made an angle of approximately 18° with (110).

Portions of each material were first rolled to a thickness of about 1/32" and strips 1/4-inch wide and about 5-1/4 inch long were then machined from the rolled sheet. The dimensions of the strips were such that two strips placed end to end completely filled one of the smoothly finished specimen channels in the graphite crucible. The strips were etched in dilute HNO_3 to remove any surface contamination, thoroughly rinsed in water, dried, and then inserted into the previously outgassed crucible along with a 1-inch long seed crystal of the desired orientation. As many as four crystals, each approximately 10-inches long, were grown simultaneously by passing a molten zone vertically through the charge at the rate of about 8 inches per hour. Seed crystals had previously been prepared in a similar manner by joining polycrystalline rods to suitably oriented spherical single crystals. All crystals of a given composition and orientation were therefore actually grown from a common seed. The orientations of the crystals were determined by the standard Laue back-reflection technique; the seeds as well as the resulting crystals were found to be of good perfection.

The alloy single crystals were next homogenized at 900°C for about 4 days and then furnace cooled at an average rate of about 50°C/hr. The high-purity silver crystals received essentially the same treatment except

that they were annealed at 900°C for only 1 day before being slowly cooled to room temperature. Both the pure Ag and the Ag-Mg alloy crystals were annealed in the same graphite crucible in which they were grown without any additional or intermediate handling. The crucible containing the crystals was placed inside a fused quartz container about 18 inches long; this was then evacuated (to ~10^{-7} mm Hg), filled to about one-third of an atmosphere with purified argon, and sealed off. Annealing was carried out in a vertical resistance furnace with three independently controlled windings. By suitable adjustment of the power input, it was possible to obtain a hot zone approximately 12 inches long within which the temperature was uniform to ±5°C. The crucible was positioned within this zone during the annealing operation. Cooling at a uniform rate was achieved by attaching a clock motor drive to the L and N Speedomax recorder which was used to control the furnace temperature.

Certain modifications in this technique were required with the Ag-Cd alloy crystals in order to prevent virtually complete loss of Cd from the samples as a result of condensation on the cooler parts of the quartz container. The Ag-Cd alloy crystals were first carefully removed from the graphite crucible and cut into sections approximately 3-inches long by spark erosion. A 1/8 inch dia. hole was also drilled by spark machining near one end of each specimen to provide a means for supporting the crystals during the annealing operation. After cutting, the crystals were cleaned by etching lightly in dilute HNO_3. The crystals were then freely suspended inside a holder made of fused quartz and the entire unit was in turn encapsulated under vacuum (~ 10^{-6} mm Hg) in a fused quartz container only 4 inches long. Since it was now possible to keep the walls of the quartz tube at a uniform temperature, no gross redistribution of Cd occurred during the homogenization anneal.

After being annealed, the high-purity Ag and Ag-Mg alloy crystals were also cut into approximately 3-inch lengths and holes were drilled in one end of each specimen by spark erosion. The crystals were then freely suspended in a vertical position and stored in this manner before use. It should be noted that extreme care was taken in handling these crystals at all stages up to and including mounting in the tensile machine. All operations were carried out with the crystals either hanging vertically under their own weight or supported along their entire length on a smooth flat surface. A simple method was developed for transferring a crystal from a horizontal support to a vertical position (or vise versa). For example, with the crystal lying on its support, a pin was inserted through the hole drilled near the end of the specimen, and the supporting tray or block was then carefully tilted into a vertical position, thus allowing the crystal to be lifted off the support by the pin without experiencing any bending moment.

The final step, namely internal oxidation, was carried out at 600°C in flowing oxygen at 1 atm pressure. For this purpose the specimens were first cleaned in dilute HNO_3 and then freely suspended inside a vertical resistance-heated tube furnace identical to that used for the homogenization anneals. The furnace was first brought up to 600°C, the individual windings adjusted to give a uniform temperature zone (±2°C) at least 5-6 inch long, and the specimens then inserted. The crystals were suspended in the furnace on stainless steel pins which fit into a slotted alundum holder and the holder itself was fastened to a rigid stainless steel support rod. All specimens were oxidized for 24 hours; this time

was considerably in excess of that needed for complete oxidation.* After oxidation, the specimens were withdrawn from the furnace and allowed to cool normally in air. However, to determine the effect of partial oxidation on the stress-strain behavior, the crystals were oxidized for varying times at 600°C and were then pulled from the furnace.

C. Mechanical Testing

The tensile tests which were carried out were of three general types: a) conventional stress-strain curves obtained at a fixed temperature and strain rate; b) measurements of the temperature dependence of the flow stress at a constant strain rate, and c) determination of the strain rate dependence of the flow stress at a constant temperature. All tests were performed with standard Instron testing machine. Since the changes in stress involved were sometimes quite small, especially in experiments of type "c", an auxiliary unit was incorporated in the load cell circuit which allowed the zero to be suppressed in ten equal steps. Using zero suppression and operating the machine at its highest sensitivity level, it was possible to obtain a stress sensitivity of the order of 0.1% over the entire load range employed. This corresponded to a load sensitivity of less than 3 g. at the lowest loads and about 75 g. at the highest end of the load range.

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*According to Meijering and Druyvesteyn $t = \frac{T^2 C_2}{2C_1 D_1}$ were T is the diffusion distance in cm, C_1 is the equilibrium concentration of oxygen in the matrix at the surface of the specimen C_2 the solute concentration and D_1 the diffusion coefficient of oxygen with $T = 0.04$ cm, $C_1 = 7.2 \times 10^{-5}$, $C_2 = .01$, $D_1 = 5 \times 10^{-5}$ cm^2/sec, $t = 5$ hours.

Tests were carried out with the crystals immersed in fixed temperature baths. The baths employed and the temperatures they maintained were: liquid nitrogen, b.p. 78°K; n-pentane, m.p. 141°K; ethyl alcohol m.p. 158°K; dry-ice acetone, 194°K; melting ice, 273°K; boiling water 373°K. In addition some tests were made at 323°K using a thermostatted water bath. The bath temperatures were measured with an iron-constantan thermocouple and were found to be reproducible to within ±1°K.

The reversible changes in flow stress with temperature were measured by deforming the crystals alternately at two different temperatures at a tensile strain rate of 2.2×10^{-5} sec^{-1} (which corresponded to a constant cross-head speed of 0.002 in/min). Regardless of their subsequent test history, in fact, all crystals were deformed initially at this strain rate. To determine the strain rate dependence of the flow stress, the crystals were deformed alternately at 2.2×10^{-5} sec^{-1} and 2.2×10^{-4} sec^{-1}. The rate of deformation could be changed instantaneously by a factor of 10 by driving the cross-head first with the variable speed and then with the synchronous motor drive system (or vice versa). Switching from one gear drive to the other was accomplished simply by engaging or disengaging a magnetic chuck.

The usual procedure in carrying out tests of type (b) above was to deform the crystal first at one temperature, partially relax the load, change to another bath, and then to continue the deformation at the new temperature. After changing baths, a standard period of 5 min was allowed for equilibrium to be reestablished before resuming the test. It was necessary to relax the load, particularly when changing from a lower to higher temperature, in order to prevent creep deformation under load. However, it is found[23] that unloading sometimes introduces a small

yield point on retesting at the same temperature, and this effect becomes more pronounced the greater the degree of unloading. To eliminate possible complications due to unloading yield point effects, the load was therefore only partially relaxed, a minimum load of about 70% of the flow stress being maintained on the specimen while temperature changes were being made. Since the change in strain rate was accomplished instantaneously no complications due to unloading were encountered here.

The crystals were held in self-aligning, split, friction-type grips made of stainless steel. Immediately before mounting the crystals were cleaned in dilute HNO_3 and the thickness was carefully measured near the grip ends with a micrometer. Mounting and alignment of the specimens in the grips were greatly facilitated by the use of a special mounting jig which incorporated locating pins for both the grips and the specimens. The fact that easy glide of up to 20% shear strain was observed in some crystals, and that good reproducibility of the critical resolved shear stress was obtained in duplicate runs indicates that serious deformation was avoided during mounting. After the crystals had been mounted in the grips, a tie bar was bolted across both grips, and, with the crystal rigidly supported in this manner, the specimen was removed from the mounting jig and installed in the testing machine. The tie bar was then carefully removed and a small load, sufficient to align the grips, was applied preparatory to starting a test.

The work on high-purity Ag was limited inasmuch as Basinski[24] had already investigated the temperature and strain rate dependence of the flow stress in considerable detail over the temperature range from 4.2° to 373°K. The few tests which were carried out on pure Ag were designed primarily to check the previously published data and to prove out the

experimental techniques. Temperature or strain rate changes were made at strain intervals of about 1% and the tests were continued well into stage III of the stress-strain curve. Since the total elongation obtainable with the internally oxidized alloys was only of the order of 10% or less at the lowest temperatures, the strain increment between changes in temperature or strain rate was usually held to about 1/2% with these materials. It almost always proved possible to carry out at least five or six complete reversals in temperature or strain rate, by which time a constant flow stress ratio had generally been obtained. In all cases cycling was continued until it became apparent that the maximum load had been reached and that fracture of the specimen was imminent.

Since each test normally required several hours, the load calibration was frequently checked after the completion of the run to insure that no serious drift had occurred. The calibration was always found to be quite stable and no corrections for drift were considered necessary.

III. RESULTS

A. Silver-Oxygen System

In the search of interstitial solution effects on the mechanical behavior of silver, two different orientations were employed, namely, < 100 > and the easy glide parallel to the tensile loading axis.

The deformation behavior and the work hardening characteristics for the high purity silver single crystals, oriented for multiple slip, with and without oxygen in solution, are reproduced in Fig. 1. There were no noticeable effects of the presence of oxygen on the amount of total strain and the work hardening behavior, but the yield strength of pure silver was raised from 0.84 Kg/mm^2 to 1.12 Kg/mm^2 by the presence of oxygen. However, the slip line pattern appeared to be identical at equivalent amounts of deformation in the two cases.

The interstitial solution of oxygen in silver single crystals, oriented for single slip, did alter significantly the stress-strain behavior; the results are shown in Fig. 2. The extent of easy glide was considerably increased when oxygen was present. However, the work hardening rate in Stage I was not affected. The initiation of Stage II was delayed, but the rate of hardening in that stage underwent no change. The slip line patterns were quite similar and the total extensions were more or less the same in both cases. The doped silver crystals oriented for single slip, however, exhibited the yield point phenomenon which was absent in silver crystals, containing oxygen and oriented for multiple slip. In order to analyze this behavior in a detailed manner, the temperature dependence as well as the strain rate dependence of the yield point were investigated; the amount of drop

increased with increasing strain rate and a similar effect was associated with decreasing temperature. The parameter m' $\left(\frac{\Delta \ell n\ \dot{\epsilon}}{\Delta \ell n\ \tau}\right)$ was determined by changing the strain-rates by a factor of 10 at 273°K. The results obtained for doped and pure silver crystals are represented in Fig. 3. The value of m' for the crystals saturated with oxygen was always smaller than that for pure silver at equivalent strains.

Moreover, there was no evidence for "Luder's Band" type of deformation in the oxygen-bearing crystals.

B. Silver-Magnesium-Oxygen Alloy

For investigating the hardening effects associated with internal oxidation of dilute silver alloys, single crystals containing approximately 1 at.% Mg were used. Two different orientations were employed in the study; one produced multiple slip and the other led to single slip. The mechanical behavior of silver-magnesium-oxygen alloy crystal oriented for multiple slip is shown in Fig. 4. The yield strength (measured at 0.1% offset) exhibited a tremendous increase on oxidation. The strength level was raised from 1 Kg/mm^2 to about 30 Kg/mm^2, but it was achieved at the expense of ductility; the total strain was reduced to almost half on oxidation. A very high initial work hardening rate was the characteristic feature of the deformation behavior, but this rate tapered off to an insignificant value at large strains. Furthermore, the slip seemed to occur on a very, very fine scale because no slip bands were visible even after 10% of tensile strain. The critical resolved shear stress exhibited a strong temperature dependence as shown in Fig. 5. The value increased from 11.7 Kg/mm^2 to 17.9 Kg/mm^2 in changing from 273°K to 77°K. However, the crystals showed a stronger

temperature dependence of the critical shear stress below 150°K.

The stress-strain behavior of the internally oxidized silver-magnesium alloy single crystals, oriented for single slip was considerably different from that of the crystals oriented for multiple slip. The easy glide stage exhibited by the unoxidized crystal (Fig.6) was conspicuously absent in the oxidized crystals, as shown in Fig. 7. In fact, the oxidized crystals exhibited very peculiar deformation characteristics; the rate of work hardening was very high initially and after attaining a peak load of approximately 134.5 Kg, the load continued to drop on further straining. The deformation was observed to be localized in a certain region and necking of the specimen occurred quite early in the deformation sequence. A few slip bands were found to be clustered on the necked region towards the end of the deformation.

Following the procedure of Cottrell and Stokes[18] a large number of measurements were made of the change in flow stress accompanying a change in temperature during deformation of these single crystals. In polycrystals and single crystals of pure metals the ratio of the flow stress at a temperature $T(\tau_T)$ to that at some standard temperature, eg. 273°K (τ_{273}) is found to be independent of deformation at large deformations. This ratio, for the oxidized alloy crystals, was found to be more or less constant. The variation of this ratio with temperature is shown in Fig. 8. The results of Neighbours et al.[25] on the variation of shear modulus of silver with temperature (Fig.9) were utilized to make necessary correction in the ratio. A strong temperature dependence of the flow stress is well documented by the high values for the Cottrell-Stokes ratio. Similar experiments were carried out with internally

oxidized single crystals of Ag-1 at.%Mg, oriented for single slip, but the results were not reliable because of the peculiar deformation characteristics. Furthermore, there was no evidence for work-softening in the internally oxidized materials.

C. Silver-Cadmium-Oxygen Alloy

In the study of the hardening effects coupled with internal oxidation of dilute silver-cadmium alloys, crystals of two different orientations were used, namely < 100 > and the easy glide along the loading axis.

The deformation characteristics of the crystals oriented for multiple slip, as shown in Fig. 10, were identical with those for Ag-Mg oxygen alloy crystals of the same orientation except for the differences in the strength levels. The tensile strength for this alloy system was 13.50 Kg/mm^2 as compared to 30.2 Kg/mm^2 for Ag-Mg-O alloy. However, the internally oxidized silver-cadmium crystals oriented for single slip showed normal deformation behavior (Fig. 11). The yield strength of the unoxidized crystal was raised from an insignificant value to about 13.6 Kg/mm^2 (measured at 0.1% offset) on oxidation. The easy glide region exhibited by pure silver cadmium crystal (Fig.12) was completely absent in the oxidized crystals. Furthermore, the rate of work hardening was quite high initially and then it tapered off. And slip appeared to occur on a very, very fine scale because no slip bands were visible even up to the fracture stage of the material.

The Cottrell-Stokes type of experiments were also carried out on the internally oxidized crystals oriented for multiple slip. The results of these experiments are shown in Fig. 13. The ratio of the flowstresses

at two oscillation temperatures was found to be more or less constant. The strong temperature dependence of the flowstress exhibited by these materials is documented by the high values for the ratio.

D. Experiments Involving Partial Oxidation

To study the effects of partial oxidation on the stress-strain behavior, single crystals of silver-magnesium and silver-cadmium alloys, both oriented for single slip were oxidized for 1/2 and 1-1/2 hrs. at 600°C in flowing oxygen at one atmosphere pressure. The deformation characteristics are reproduced in Figs. 14 and 15 for the partially oxidized silver-magnesium and silver-cadmium crystals respectively. The mechanical behavior was completely altered on partial oxidation. The silver-magnesium crystals showed almost no work hardening after the initial yielding, but the yield strength (measured at 0.1% offset) exhibited a considerable increase on varying the time of oxidation. The slip bands were distributed on a very fine scale in the Ag-Mg crystals oxidized for 1/2 hr., but no bands were visible in the crystals oxidized for 1-1/2 hrs. The partially oxidized silver-cadmium crystals exhibited similar effects but the yield points were absent. They showed regions of positive work hardening after a few percent strain. It appeared that ductility of these materials was not impaired considerably on partial oxidation.

IV. DISCUSSION

Before attempting an interpretation of the results several pertinent features of the silver-oxygen system should be noted. First of all, it has been shown that the solubility of oxygen in silver is proportional to the square root of the pressure,[26] which implies that the oxygen molecule is dissociated at the surface and that oxygen goes into solution as the atomic species. It is also believed that the oxygen atoms occupy interstitial positions in the silver lattice based on the fact that the activation energy for diffusion is only 11.0Kcal/mole as compared to 45Kcal/mole for self diffusion of silver and 30 to 50 Kcal/mole for diffusion of known substitutional solutes (e.g.,Zn) in silver. The ionization state and hence the effective radius of oxygen in silver is unknown. Oxygen has a covalent radius of 0.73 Å and a crystal or ionic radius of 1.4 Å. The atomic radius of silver is 1.44 Å while the "radius" of the octahedral interstice in the silver lattice is 0.64 Å.

Finally it should be pointed out that at an oxygen pressure of 1 atm. Ag_2O becomes thermodynamically unstable above 190°C. However, no studies seem to have been carried out on the kinetics of precipitation of Ag_2O from oxygen-saturated silver.

A. Silver-Oxygen System

The amount of oxygen introduced at any given temperature and pressure can be calculated from the experimental solubility relationship given by Eichenauer and Muller[26] namely,

$$\log \ell = 1.403 - \frac{2593}{T} + \frac{1}{2} \log P \quad \ldots \qquad (2)$$

where ℓ is the solubility in $cm^3 O_2(S \cdot T \cdot P)/100$ gm Ag, T is the absolute temperature and P is the partial pressure of oxygen in mm Hg. The amount of oxygen that could be introduced into the samples amounted to 70 ppm at 600°C under normal oxygen pressure.

The presence of oxygen in high purity silver single crystals, oriented for multiple slip, affected only the yield strength value. And this a real effect because the formation of oxide coating on the surface was eliminated by saturating the samples with oxygen at 600°C and by chemical polishing of the crystals in dilute HNO_3 before testing. This increase in yield strength of the oxygen-bearing crystals could very well be associated with the interaction between the distortion produced in the lattice by an oxygen atom and the mobile dislocations. However, no attempt was made to correlate this increment in yield strength with the strain associated with different positions of the oxygen atom in the silver lattice because there is no certainty about the ionization state of oxygen atom in the lattice.

Even though there is convincing evidence for the interstitial diffusion of oxygen, however, the effects of solution of oxygen in silver crystals, oriented for single slip, are similar to those for substitutional alloying elements. The increase in yield strength, extension of easy glide region and the subsequent delay in the beginning of the linear hardening region are the features of the oxygen-saturated samples that are common to dilute solid solutions of the substitutional type. The lengthening of Stage I and the delay in the initiation of Stage II could be attributed to the increase in shear stress necessary to activate the secondary slip system. This seems logical because the yield strength of silver does show an increase in the presence of oxygen.

The most dramatic effect observed in the oxygen-bearing silver crystals, oriented for single slip, was the existence of yield drop phenomenon in the stress-strain behavior. However, this feature was absent in the oxygen-saturated crystals oriented for multiple slip. The appearance of yield points in solids has been mostly associated with either of these mechanisms:

1. Cottrell's Mechanism.

This was proposed by Cottrell and Bilby[27] to explain the presence of yield drops in the stress-strain behavior of steels. They associated the drops with the breakaway of dislocations from the impurity atmospheres which could have been formed during thermal and mechanical treatments.

2. Dislocation Multiplication Mechanism

This alternative mechanism was suggested by Gilman and Johnston[28] and Johnston[29] on the basis of their work on LiF. They attributed the yield drops to the difference in the number of mobile dislocations before and after the macroscopic yield. It was emphasized that dislocation pinning by impurities was a necessary requirement to produce yield drops, and this feature is common to both the mechanisms, but they differ considerably in detail. The Cottrell mechanism envisages the break away of dislocations from the atmospheres whereas the second mechanism involves rapid multiplication of mobile dislocations. However, the immobile dislocations are supposed to play an indirect role by reducing the number of mobile dislocations.

The appearance of the yield drop in the presence of oxygen implies that there is a binding force between a dislocation and an oxygen atom. On the basis of the features, pertinent to silver-oxygen system, outlined

above, it is reasonable to assume that an oxygen atom may occupy an octahedral site in the silver lattice. This can lead to strain in the matrix because of the difference in size of the oxygen atom and the size of the available interstice. This strain can be relaxed by the migration of oxygen atoms to the dislocation cores. The accumulation of these atoms around the dislocations will result in decreasing the density of mobile dislocations - a necessary requirement for the observance of yield point phenomenon.

The observed increase in yield drop in the oxygen-saturated crystals, with increasing strain-rate can easily be explained in terms of "Dislocation Multiplication Mechanism". The strain rate can be expressed as

$$\epsilon^{*} = b\ n\ v \ldots \qquad (3)$$

where b is the Burgers vector, n is the number of mobile dislocations which is a function of the strain, v is the dislocation velocity which at constant temperature is a function of the stress.

Moreover, the dependence of dislocation velocities on shear stress can be written as

$$v \ \alpha \ (\tau)^{m} \qquad (4)$$

There has been no attempt to interpret this exponent 'm' physically but it was well documented by Johnston's work[29] that the smaller the value of 'm', the larger the yield drop. Assuming that the number of mobile dislocations, which is dependent on the thermal and mechanical history of the specimen, are the same prior to macroscopic yield at various deformation rates, then the dislocations would have to move faster at higher strain rates. Furthermore, the drop in stress after the yield is a sensitive function of the velocity and number of mobile dislocations and this in turn is affected by the strain-rates. The

higher is the deformation rate, the greater is the change in dislocation velocities after the yield. This consequently results in bigger drops.

The experimental value of 'm', which was obtained by extrapolating the value of 'm' to zero strain was smaller for the oxygen-bearing silver crystals in comparison with that for the silver crystals without oxygen. It is very difficult to attach any physical significance to this difference, but all the observations appear to be consistent with the conditions specified by Gilman and Johnston for the appearance of yield drop.

Before going into the relevant discussion pertaining to the internally oxidized silver-magnesium and silver-cadmium systems, it will be pertinent to discuss in detail the predicted temperature dependence of the yield strength as well as of the flow stress on the basis of proposed strengthening mechanisms for the dispersion strengthened systems.

As indicated earlier, the Orowan mechanism involves the formation of dislocation loops about the particles. Hence the temperature sensitivity of the loop production is a function of the size of the particles. Considering the particles are of radius r(cm), then the dislocation length to be formed is $2\pi r$(cm). Consequently the energy involved is

$$E = 2\pi r \frac{Gb^2}{2} \qquad (5)$$

$$= 3 \cdot 2 \times 10^8 \text{ rev.}$$

Even for the particles of 10Å size, the thermal fluctuations will not be of any help in the formation of loops because the activation energy is too high. It is, therefore, reasonable to assume that the flow stress as well as the yield strength of a material should exhibit little or no temperature dependence if the Orowan's mechanism is rate controlling.

The strengthening produced by coherent particles has been associated with the atomic scrambling which occurs and the local strains which must be overcome when a dislocation moves through these zones. Let ν be the energy of the interface produced when the particle is sheared along the vector b, i.e., the Burger 's vector of the dislocation moving the matrix corresponding to the smallest repeat distance in the matrix. At 0°K the stress required to move the dislocation through particles of radii is r separated by a distance d is given by

$$\tau_c \, bd2r = \pi r^2 \nu \qquad (6)$$

hence

$$\tau_c = \frac{\pi \nu r}{2bd}$$

The only parameter which can exhibit temperature sensitivity is the surface energy; but the creation of surface involves breaking of bonds and this process is not very temperature sensitive. Consequently if the shearing of particles is rate controlling, the yield strength should show little or no temperature dependence.

The cross slip mechanism for dispersion strengthening is similar in detail to the Orowan's mechanism except that it involves recombination of partial dislocations before the dislocation can avoid the particles by cross-slip. This process would be expected to be temperature sensitive because the values of activation energies are small. However, magnitude of the activation energy is a strong function of stacking fault-energy of the matrix material, the state of internal stresses and size of the obstacle.

All these mechanisms outlined above involve a dislocation-particle interaction of some sort; the particles are considered to play a direct

role in strengthening of the matrix. However, strengthening of the solids can also result from the increase in number of dislocation -- dislocation interactions. In this case the particles are assigned an indirect role and increase in dislocation density is attributed to their presence.

B. Silver-Magnesium-Oxygen Alloy

The important features of the mechanical behavior of the single cyrstals oriented for multiple slip are summarized below:

1. The internal oxidation of these crystals results in a considerable increase in yield strength, but this treatment does result in decrease of ductibility.
2. A strong temperature dependence of the flow stress as well of the critical resolved shear stress is exhibited by the system.
3. A very high rate of work hardening is observed in early stages of deformation, but this tapers off to a very small value after a few percent of strain.

All these characteristics of the mechanical behavior appear to be similar to those for aluminum-copper alloys containing θ and θ' precipitates.[8] However, the critical resolved shear stress (measured at 0.1% offset) exhibits a strong temperature dependence as compared to that for aluminum-copper alloys containing discrete particles.

In order to facilitate interpretation of these results, a structural model will first be presented for this alloy. The proposed model is similar to that recently suggested by Darken[30] for the Ag-Al-O system, its principal feature being that during internal oxidation small clusters of magnesium and oxygen atoms, perhaps no more than 20-30A° in diameter, are formed in the matrix. The evidence supporting this model will now be reviewed.

The NaCl-type structure of MgO can be simply derived from that of silver by substituting magnesium atoms for silver and placing oxygen atoms in the octahedral interstices in the fcc silver lattice. Moreover, the lattice constants of MgO and silver (at room temperature) differ by less than 3%. This close structural similarity suggests it is probable that coherency exists between these two phases, especially when the Mg-O clusters are extremely small.

The failure to observe any discrete particles during transmission electron microscopy of Ag-2 at.% Mg oxidized at 650°C[31] suggests that the clusters of magnesium and oxygen atoms formed during oxidation must be less than 30A° and probably smaller.

The small size of the Mg-O clusters formed during oxidation may be understood on the basis of the following considerations: Oxygen diffuses extremely rapidly in silver[26] compared to magnesium. At 600°C, the two diffusion coefficients probably differ by four or five orders of magnitude. Because of the strong chemical affinity between the two solute elements, oxygen atoms diffusing in from the surface rapidly seek out and combine with magnesium atoms before the latter have had an opportunity to migrate many atomic distances. Once this happens, the distribution of magnesium and oxygen in the silver matrix is effectively 'frozen in' since pairs and larger aggregates of magnesium and oxygen atoms are immobile. The growth of these clusters can only take place by thermal dissociation of magnesium and oxygen atoms within a cluster, and then having both constituents diffuse individually to other clusters in the vicinity. This process should be very sluggish because of the large binding energy between magnesium and oxygen atoms.

Essentially the cluster size will be determined by the initial

distribution of magnesium atoms in solid solution prior to oxidation. Since silver and magnesium have nearly the same, atomic radii, extensive clustering would not be expected.

The ionic nature of the bond between magnesium and oxygen atoms suggests that oxygen atom may be ionized once it jumps into an interstitial site adjacent to a magnesium atom. The oxygen is now trapped in an interstitial position, not as a neutral atom but essentially as an O^{--} ion. A very large matrix strain would therefore result. This strain might be reduced by forming larger MgO clusters.

It can be concluded from this discussion that the clusters are of the order of 20A° in diameter; assuming them to be spherical, the computed mean-planar spacing between the particles is about 100-120A°.

On the basis of the high temperature sensitivity of the flow stress as well as of the critical resolved shear stress exhibited by this alloy, the possibility of shearing of particles by dislocations is ruled out. If the yield strength had been controlled by this process, then it would have shown a little or no temperature dependence. This is further supported by the following arguments. The activation energy for forcing a dislocation through a cluster is composed of two terms:

1. The increase in energy of a dislocation because of the change in shear modulus, and
2. The energy required to break the bonds.

Large differences between the shear moduli of silver and magnesium oxide result in a very high value for the first term. It is, therefore suggested that a dislocation will be strongly repelled by these clusters. Similarly high temperature sensitivity of the critical shear stress exhibited by this alloy system rule out the possibility of the Orowan's

mechanism as a controlling factor.

The experimental evidence favour either by-passing of obstacles by cross-slip mechanism or strengthening of the matrix as a result of increase in number of dislocation-dislocation interactions. The stacking fault energy of silver is about 25 ergs/cm^2 and, therefore, the partials will be widely separated. Consequently, the activation energy for the cross-slip process may be too high for it to be thermally activated. In order that this process is sensitive to thermal fluctuations, the particles or clusters should produce very large stress-concentrating effects so that separation between the partials of leading dislocation is considerably reduced. However, it is not possible to visualize the origin of such large stress concentrations from the clusters in the system under consideration. It is, therefore, suggested that cross-slip of dislocations will be quite infrequent at the temperatures of investigation and the flow stress will be controlled by the intersection mechanism. Essentially, Mitchell, etal[32] arrived at a similar conclusion from their work on the Al-Cu system; the increase in strength and strain-hardening for the system has been shown to arise from the high initial dislocation density and its rapid increase with strain.

The conclusion outlined above is further supported by the absence of work softening in these materials. This implies that the deformation structure obtained at a low temperature is quite stable at high temperature. In other words, the presence of clusters in the matrix hinders the rearrangement of dislocations by cross-slip.

The constancy of the Cottrell-Stokes ratio with strain and the absence of any appreciable difference between the ratios obtained on

single crystals and polycrystalline samples suggest that the distribution of the dislocations cannot play an important part in determining the values of the ratio.[24] If the flow stress in this alloy system is controlled by the intersection mechanism, the ratios should be the same for pure silver and the internally oxidized material. However, the values of the Cottrell-Stokes ratio for this material differ slightly from that for pure silver, for example $\left(\frac{\tau_{77}}{\tau_{273}} \frac{G_{273}}{G_{77}}\right)$ is 1·24 for internally oxidized Ag-Mg alloy where as it is about 1·06 for pure silver. These small differences may stem from the following reason: The absence of a sharp yield point in these materials introduces an uncertainty in the flow stress value at the new testing temperature. In the experiment, the flow stress at 0.1% offset was taken as a measure of τ_T. But this additional deformation can alter the dislocation structure, thereby resulting in a slightly higher value for τ_T and consequently leading to higher $\frac{\tau_T}{\tau_{273}}$ ratios.

The rapid hardening observed in this material implies an unusually high rate of dislocation generation within the matrix and this demands that initial dislocation density should be quite high. This suggestion, of course, is consistent with high values of the critical shear stress exhibited by this alloy. But it is not known precisely how the particles may operate as sources; it is possible that they punch out prismatic loops, or shear loops may generate directly from the particle-matrix interface.

The decrease in work-hardening rate can be interpreted in terms of equivalence between the annihilation rate and the multiplication rate of dislocations, thereby producing a little damage in the material on further straining. This, however, can occur only if the dislocation

spacing within the cells is quite small so that they can annihilate each other without having to move large distances.

The occurrence of slip on a very fine scale implies that each slip source does not contribute significantly to strain, but a lot of slip sources have to operate to produce total strain. This could result if each slip source work hardens very rapidly and the operation of a new source becomes imperative to produce further deformation.

The unusual mechanical behavior exhibited by the internally oxidized silver-magnesium alloy crystals, oriented for single slip, can be associated with the rapid multiplication of dislocations after the upper yield point and thereby leading to continuous drop in load on further straining. This suggestion implies that the dislocations are strongly pinned by these clusters. This, of course, is supported by the observations of Papazean and Hummel[33] on the internal friction behavior of these materials.

C. Silver-Cadmium-Oxygen Alloy

The essential features of the mechanical behavior of this alloy system are more or less identical to those for the silver-magnesium-oxygen system and can be explained on similar lines. However, the values of the Cottrell-Stokes ratio for this material do not differ from that for pure silver, for example $\frac{\tau_{77}}{\tau_{273}} \cdot \frac{G_{273}}{G_{77}}$ is 1.07 for internally oxidized Ag-Cd alloy where as it is about 1.06 for pure silver. Furthermore these ratios appear to be more or less constant with strain. This implies that the long range back stresses are also dependent on the spacing of dislocation forest and its validity is based on the following reasoning: If the intersection mechanism is rate-controlling, then flow stress (τ) of a metal below a critical

temperature T_c, is given by

$$\tau = \tau^* + \frac{1}{Lb} \psi^{-1} \left[U - kT \ln \frac{NAb\nu}{\gamma} \right] \qquad (7)$$

where τ^* is the long range backstress, L is the spacing of the forest dislocations, U is the activation energy of the intersection process, N is the number of points of contact per unit volume between forest dislocations and glide dislocations, A is the area swept out per successful intersection, b is the Burger's vector, ν is the frequency of vibration of the segment of the glide dislocation undertaking intersection, k is Boltzmann's constant, T is the absolute temperature, and $\dot{\gamma}$ is the shear strain rate.

The ratio of flowstresses at two different temperatures but for the same dislocation structure can be written as

$$\frac{\tau_{T_1}}{\tau_{T_2}} \cdot \frac{G_{T_2}}{G_{T_1}} = \frac{\tau^*/G_{T_1} + \frac{1}{G_{T_1}} \frac{1}{Lb} \psi^{-1} \left[U - kT_1 \ln \frac{NAb\nu}{\dot{\gamma}} \right]}{\tau^*/G_{T_2} + \frac{1}{G_{T_2}} \frac{1}{Lb} \psi^{-1} \left[U - kT_2 \ln \frac{NAb\nu}{\dot{\gamma}} \right]} \qquad (8)$$

For the ratio to be constant with strain, τ^* should vary as 1/L. This conclusion is quite significant because it implies that the particles play an indirect role in strengthening of the matrix.

The absence of easy glide region in the internally oxidized silver-cadmium crystals oriented for <u>single slip</u> suggests that slip occurs simultaneously on a number of slip systems. This can only happen if the stress necessary for propagation of strain on the primary slip system and in the secondary slip system are of the same order of magnitude.

The differences in the strength level achieved in the two systems can be attributed to the differences in the particle size in the two cases. The bonding between cadmium and oxygen atoms is not as strong as the bonding between magnesium and oxygen atoms. This ease in dis-

sociation of the CdO molecule and the surface energy considerations will result in bigger particles in the silver-cadmium-oxygen system. In spite of the same volume fraction of the dispersion in the two cases, the obstacles will be distributed on a much finer scale in silver-magnesium-oxygen system. This finer distribution will lead to a different dislocation substructure and thereby to those observed differences on the two systems.

D. Partial Oxidation Experiments

The partial oxidation of these alloy crystals produces a dispersion within a narrow region interiro to the surface. This second phase particles indirectly serve as obstacles to the motion of dislocations. The depth of the affected zone can be computed from the diffusion equation suggested by Meyering et al.,[12] for $\frac{1}{2}$ hour oxidation time, the zone should be 1.16×10^{-2} cm deep and the zone depth should be 2×10^{-2} cm when the oxidation time is 1-1/2 hours.

There are two possibilities for the initiation of slip: (a) the slip starts from interior of the crystal and propagates towards the surface, and (b) the slip sources start operating at the surface and then the loops expand inwards. If the slip had initiated at the surface, the stress-strain curve would have exhibited rapid work hardening in the early stages of deformation. However, this feature was absent in the mechanical behavior of the partially oxidized materials. These observations suggest that it is likely for the slip to start from the interior and then propagate outwards. However, a considerable amount of work-hardening can result from the interaction of expanding loops with the dislocation substructure in the oxidized zone. This implies

that the stress necessary for the source to produce a step on the surface will depend on the depth of the affected zone; the greater is the depth, the higher is the stress value. Consequently the macro yield strength of the material will also depend on the depth of this zone. Once an expanding loop has broken through the oxidized zone, more loops will follow through without any difficulty, thereby resulting in a little or no work hardening. This conclusion is consistent with the experimental observations on slip distribution and work hardening rate.

V. SUMMARY AND CONCLUSIONS

The interstitial solution of oxygen in silver affects the mechanical behavior of single crystals; initiation of stage II is delayed and easy glide region is extended oxygen bearing crystals oriented for single slip exhibit yield point phenomenon. It has been concluded that these effects result from pinning of dislocations by oxygen atoms.

The deformation characteristics of silver-magnesium-oxygen and silver-cadmium-oxygen alloy crystals are similar to those for aluminum-copper alloys containing incoherent precipitates. The strengthening of these materials arises from the increase in number of dislocation-dislocation interactions. The flow stress has been shown to be controlled by the intersection mechanism. The dispersed phase has been assigned an indirect role in strengthening of the matrix. The differences in strength levels attained in the two systems have been shown to arise from the differences in dispersion size.

Each slip source does not appear to contribute significantly to the total strain. Moreover, motion of dislocations over large distances is restricted by the substructure resulting from the presence of a second phase in the matrix.

The slip sources near the surface appear to be effectively pinned in partially oxidized silver-magnesium and silver-cadmium alloy crystals.

ACKNOWLEDGEMENTS

The author wishes to acknowledge the advice and the guidance of Dr. L. Himmel and Professor E. R. Parker.

This work was performed under the auspices of the United States Atomic Energy Commission through the Inorganic Materials Research Division of the Lawrence Radiation Laboratory.

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FIGURE CAPTIONS

Fig. 1 Stress-strain curves for (a) pure silver crystal and (b) silver crystal containing oxygen; [100] being the loading axis.

Fig. 2 Stress-strain curves for silver crystals (a) without oxygen and (b) with oxygen.

Fig. 3 Variation of m' with strain (a) pure silver (b) silver doped with oxygen.

Fig. 4 Stress-strain curve for internally oxidized silver-magnesium alloy crystal; [100] orientation.

Fig. 5 Critical resolved shear stress of silver-magnesium-oxygen alloy crystals as a function of temperature; tensile axis is along [100] direction.

Fig. 6 Stress-strain curve for silver-magnesium alloy crystal oriented for single slip.

Fig. 7 Load-elongation curve for silver-magnesium-oxygen alloy crystal oriented for single slip.

Fig. 8 Ratio of the flow stress at T(τ_T) to flow stress at 273 °K (τ_{273}) plotted against temperature for silver-magnesium-oxygen alloy crystals oriented for multiple slip (a) before correcting for the change in shear modulus with temperature (b) after.

Fig. 9 Variation of shear modulus of silver with temperature.

Fig. 10 Stress-strain curve for silver-cadmium-oxygen alloy crystal of [100] orientation.

Fig. 11 Stress-strain curve for silver-cadmium-oxygen alloy crystal oriented for single slip.

Fig. 12. Stress-strain curve for silver-cadmium alloy crystal oriented for single slip.

Fig. 13 Ratio of the flow stress at T(τ_T) to flow stress at 273 °K (τ_{273}) plotted against temperature for silver-cadmium-oxygen alloy crystals oriented for multiple slip (a) before correcting for the change in shear modulus with temperature (b) after.

Fig. 14 Stress-strain curves for partially oxidized silver-magnesium alloy crystals oriented for single slip (a) 1/2 hour oxidation time (b) 1-1/2 hours oxidation time.

Fig. 15 Stress-strain curves for partially oxidized silver-cadmium alloy crystals oriented for single slip (a) 1/2 hour oxidation time (b) 1-1/2 hours oxidation time.

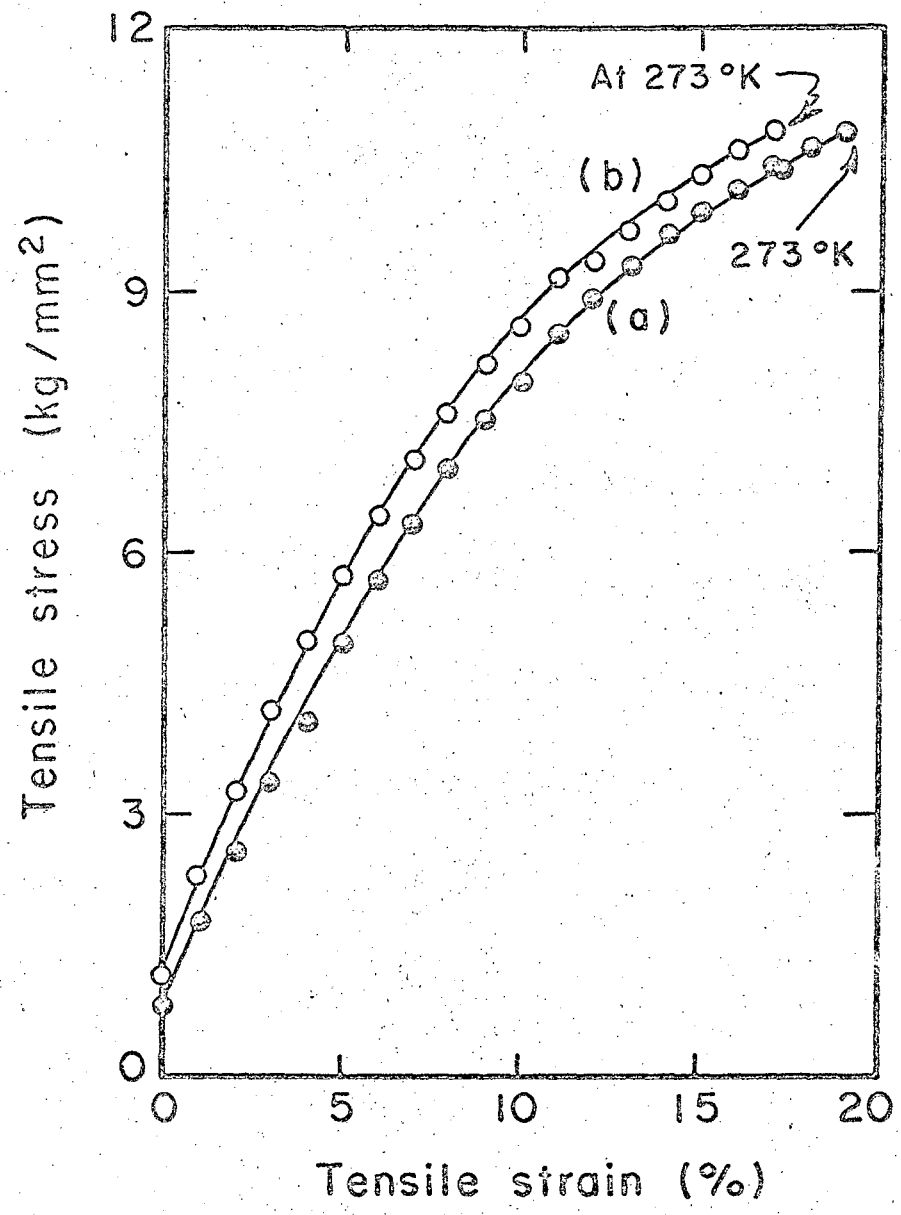

Fig. 1

MU-35652

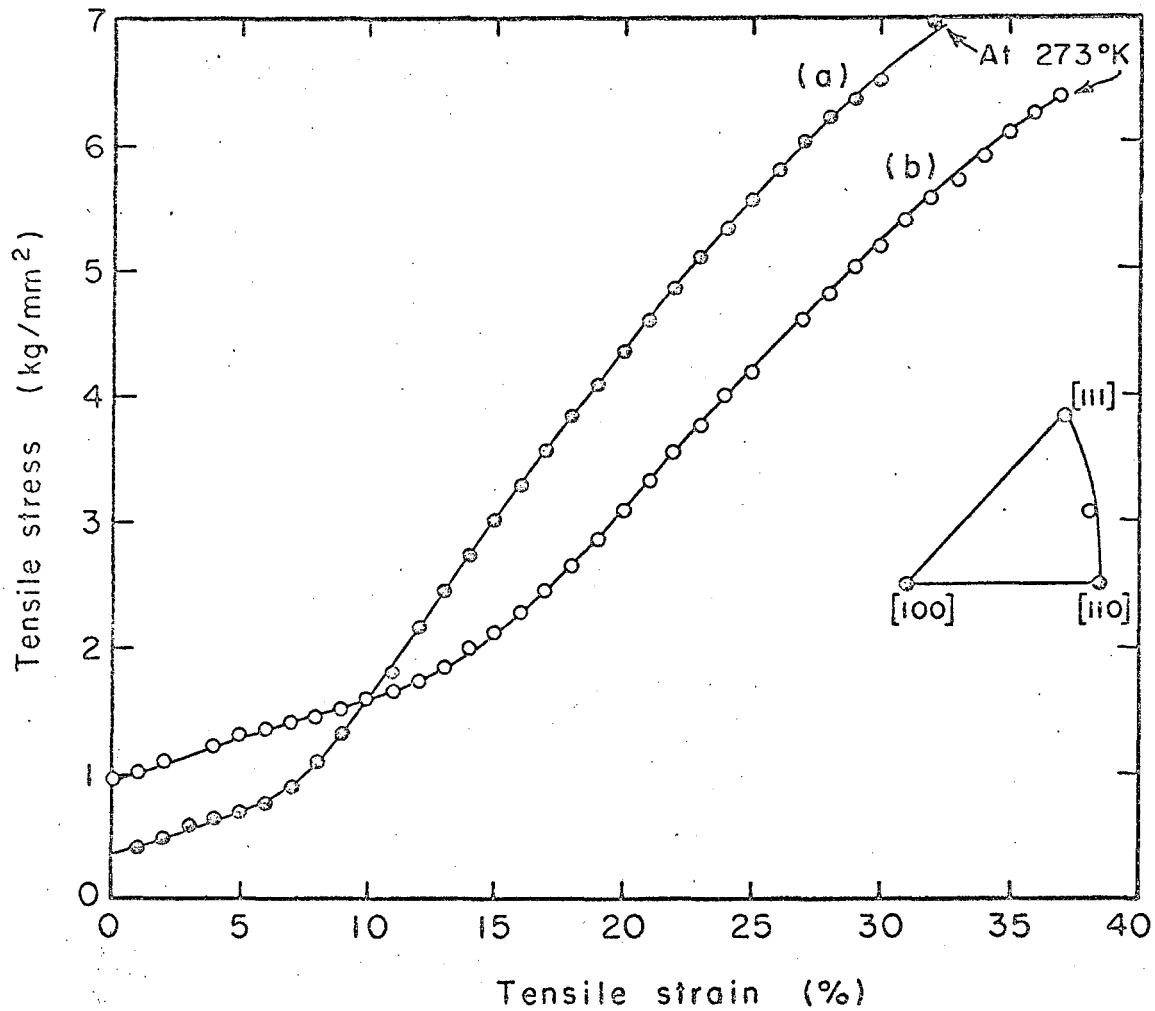

Fig. 2

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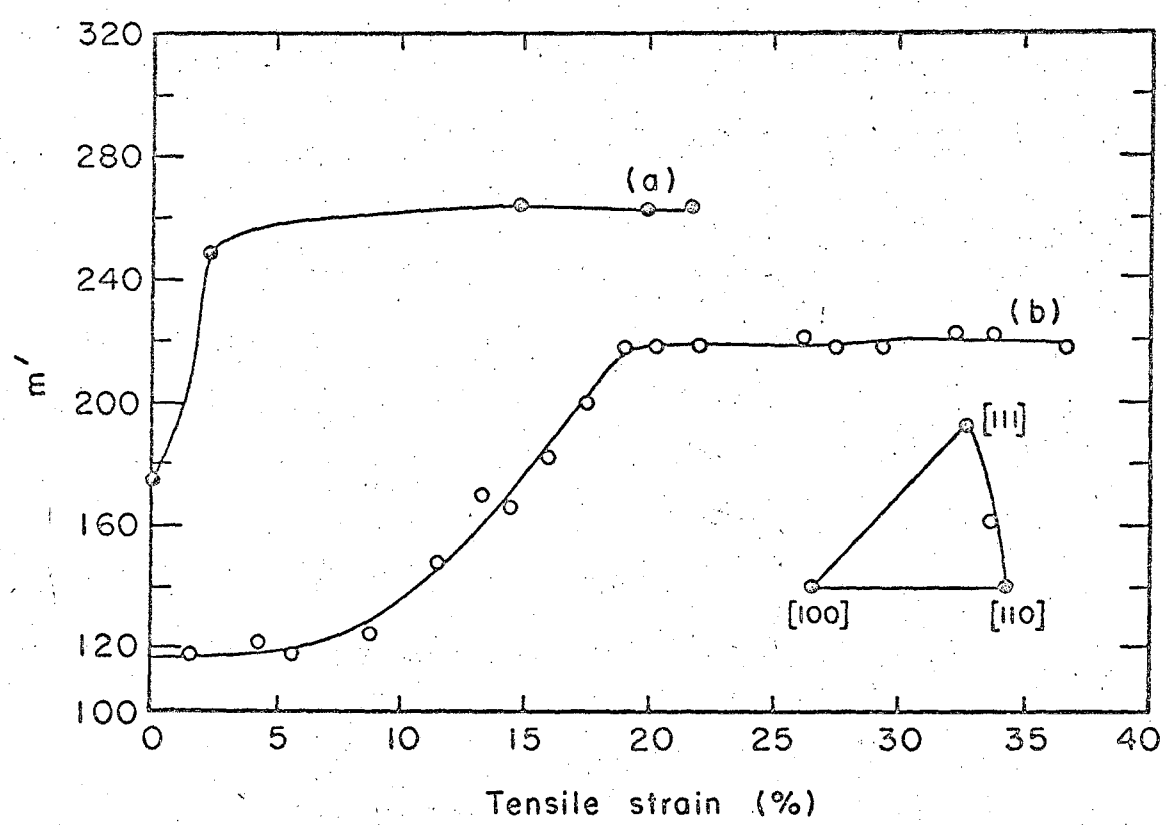

Fig. 3

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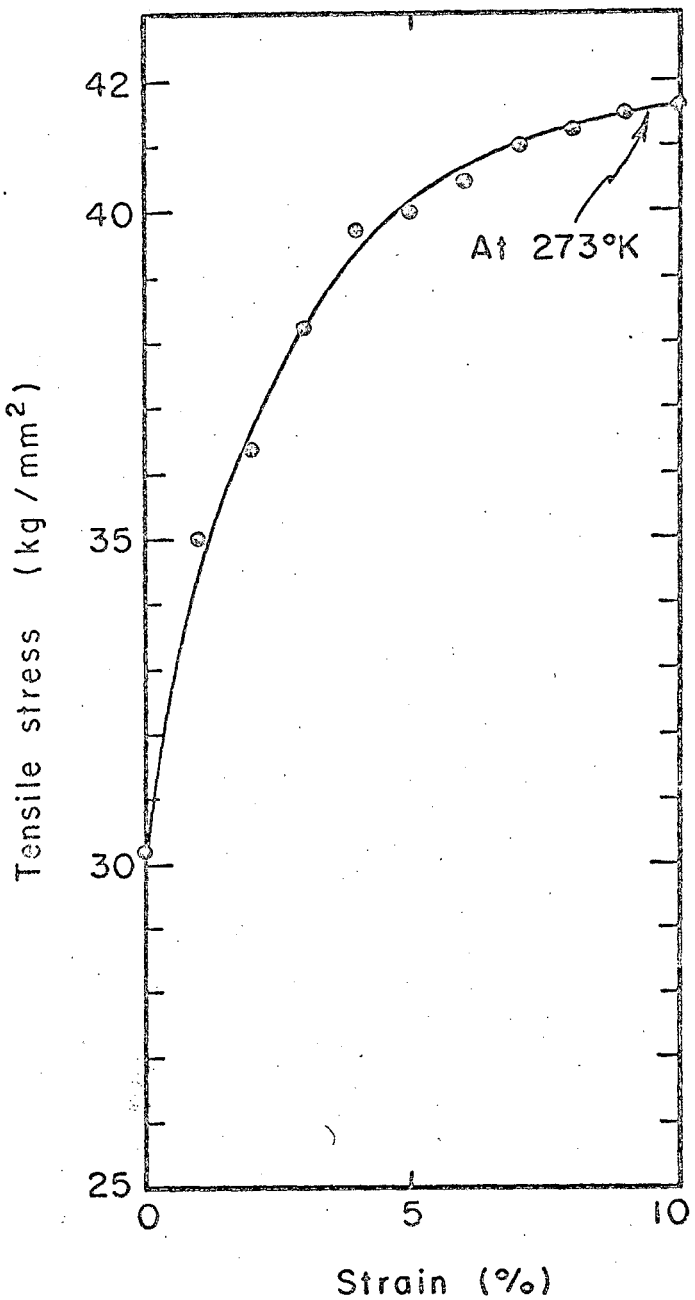

Fig. 4

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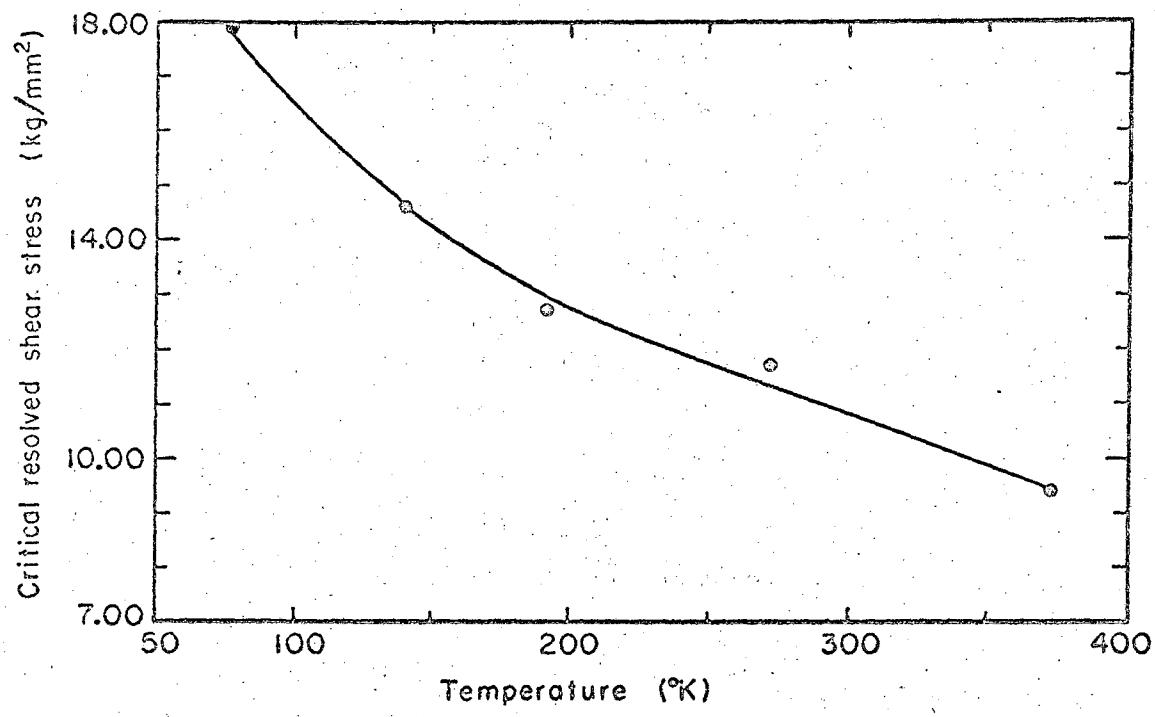

Fig. 5

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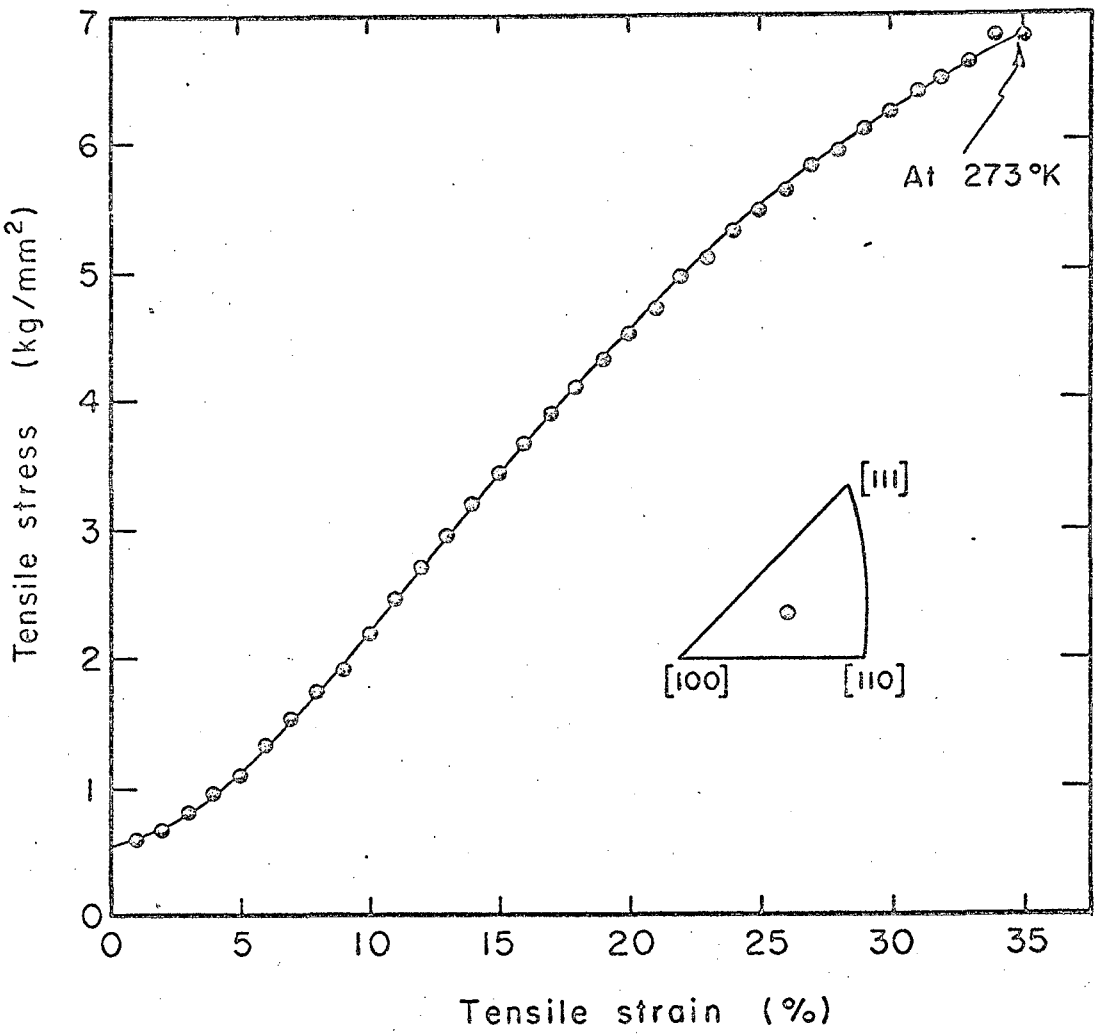

Fig. 6

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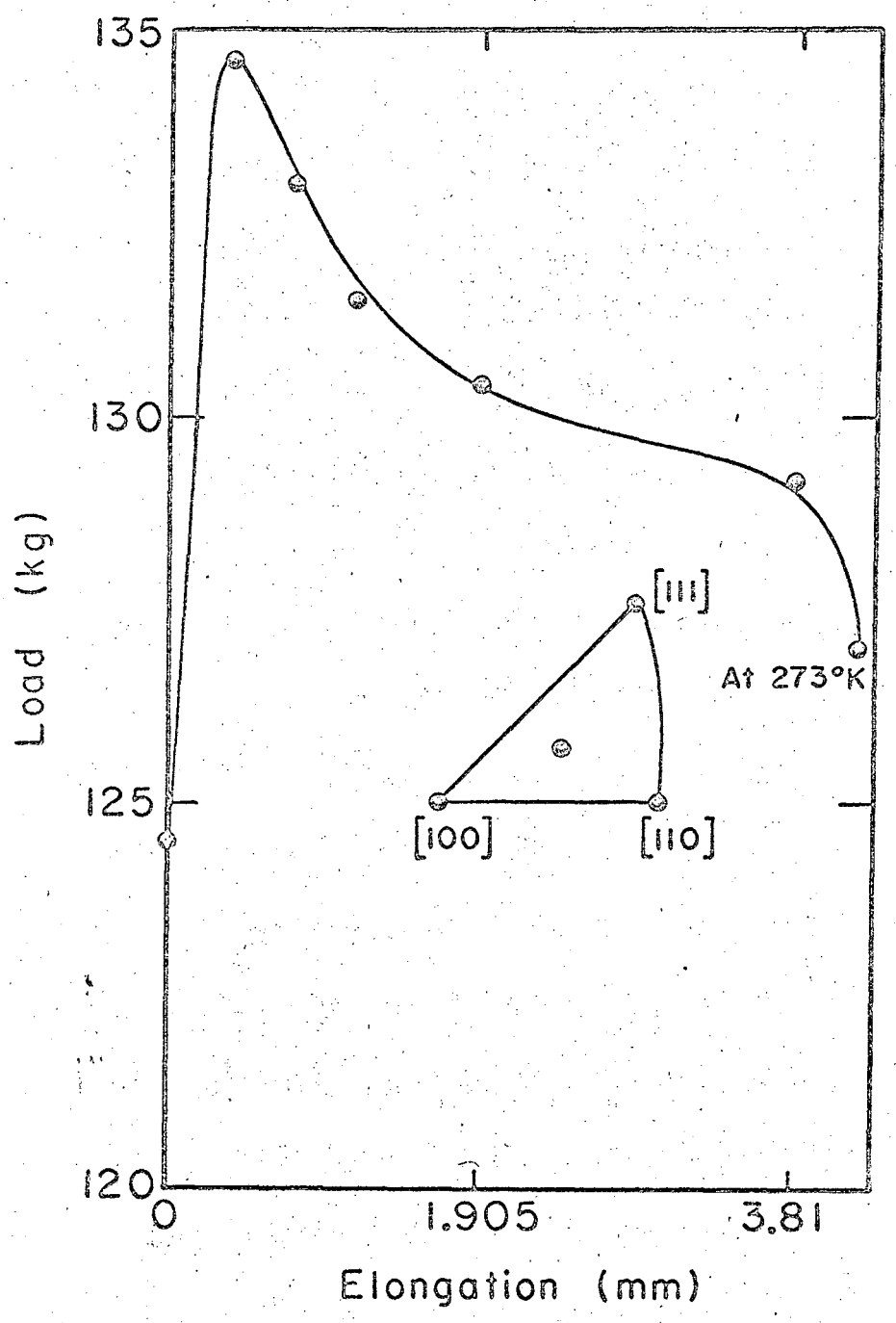

Fig. 7

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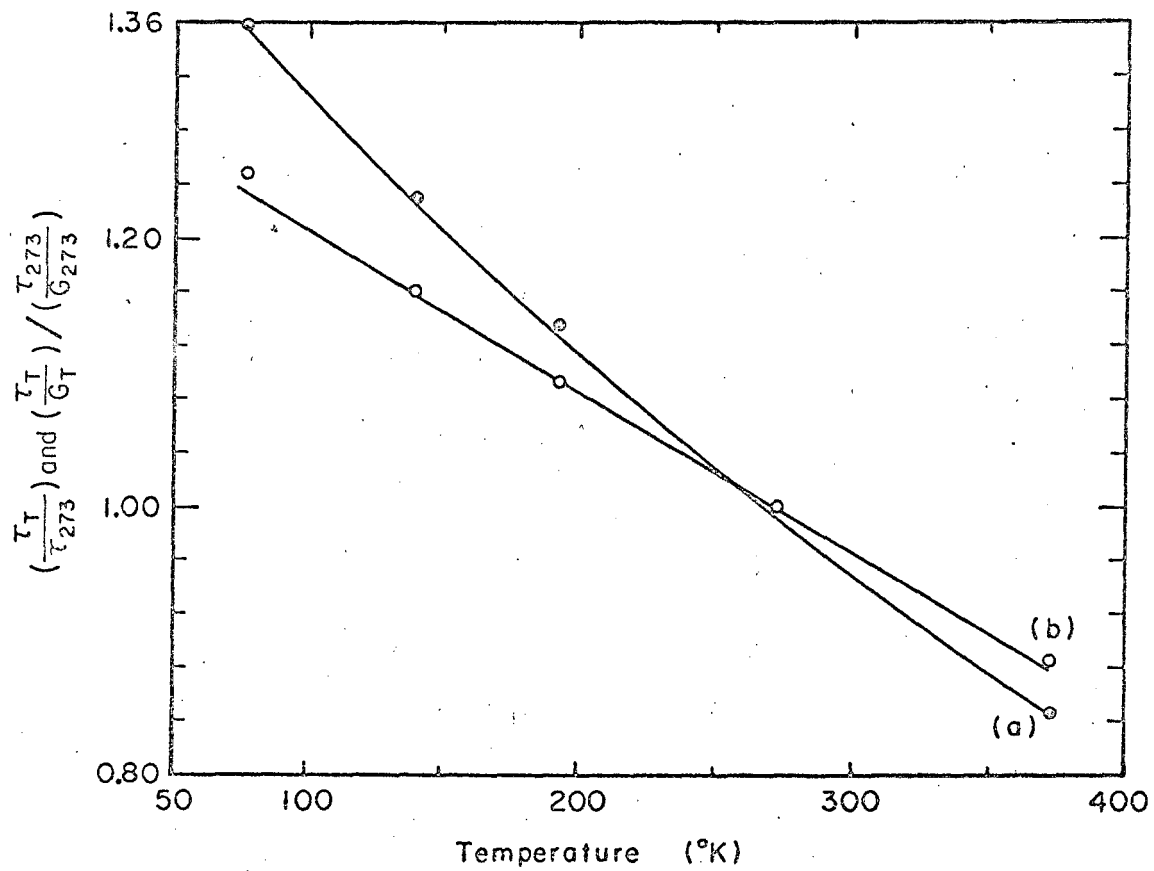

Fig. 8

MU-35648

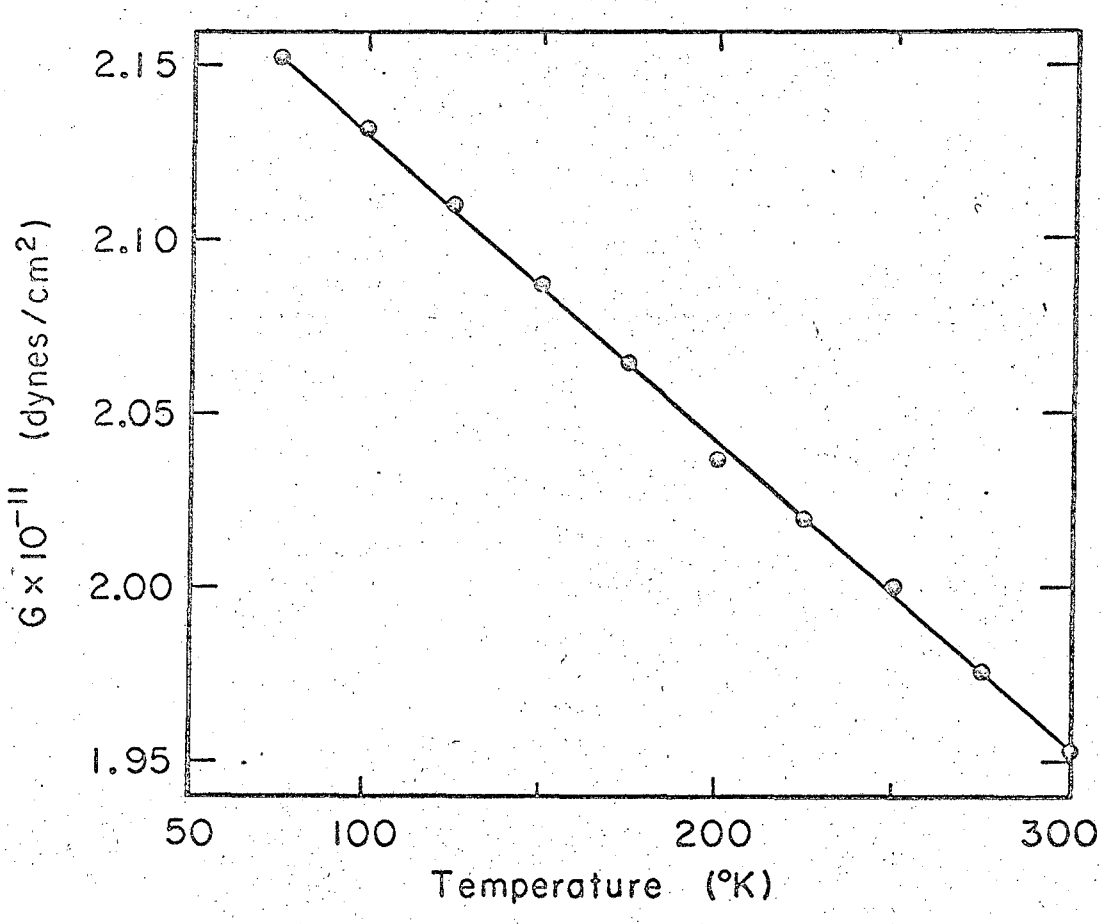

Fig. 9

MU-35646

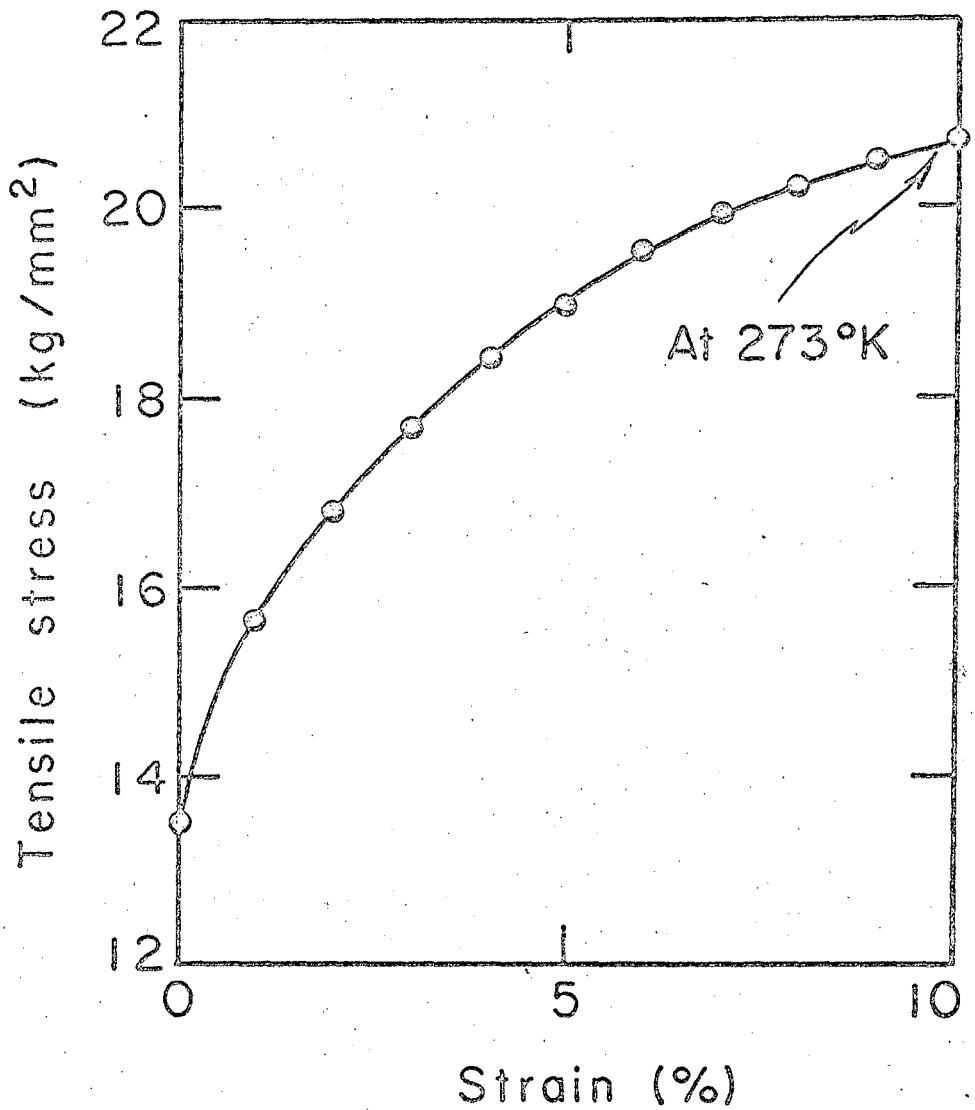

Fig. 10

MU-35644

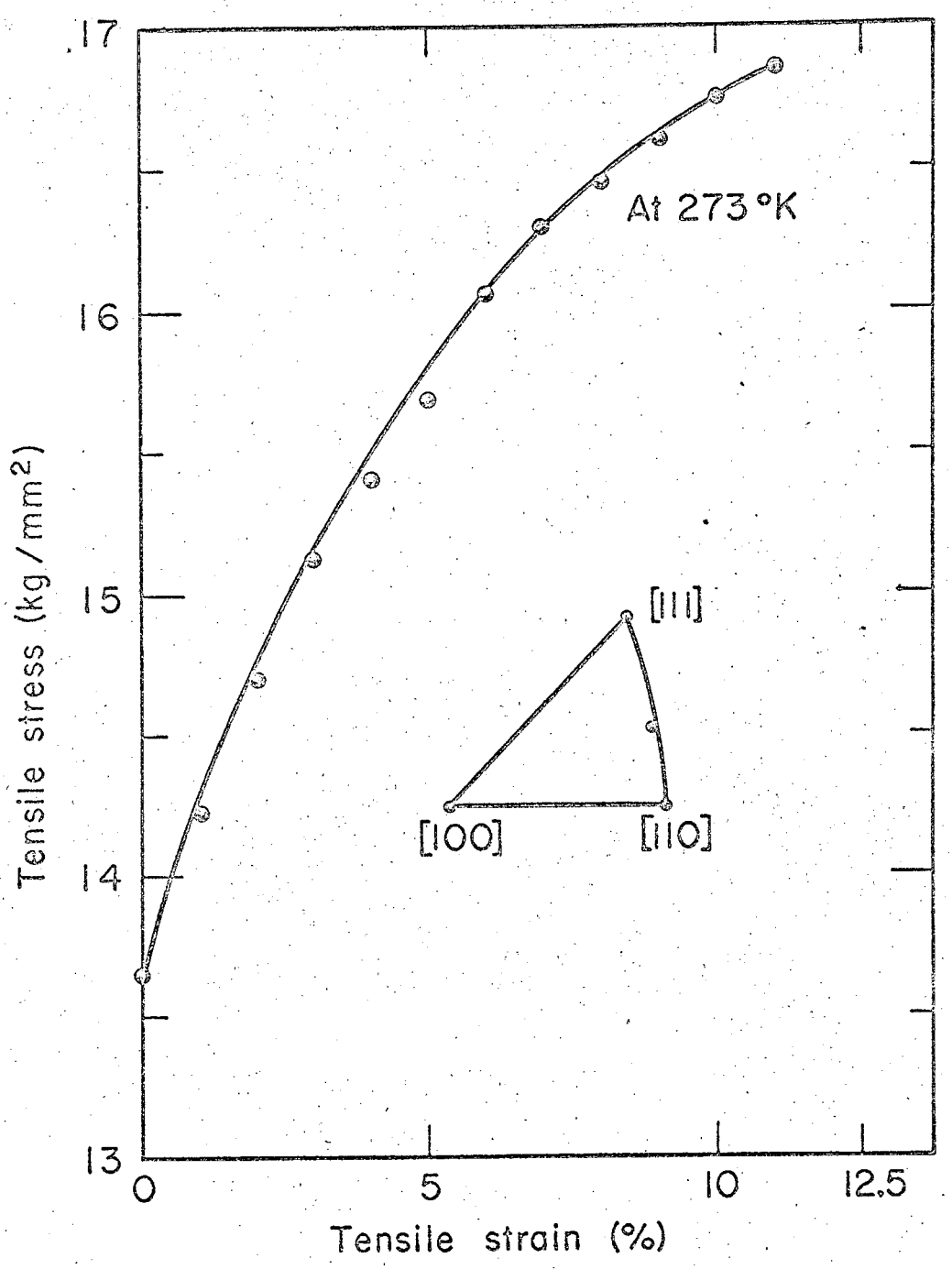

Fig. 11

MU-35771

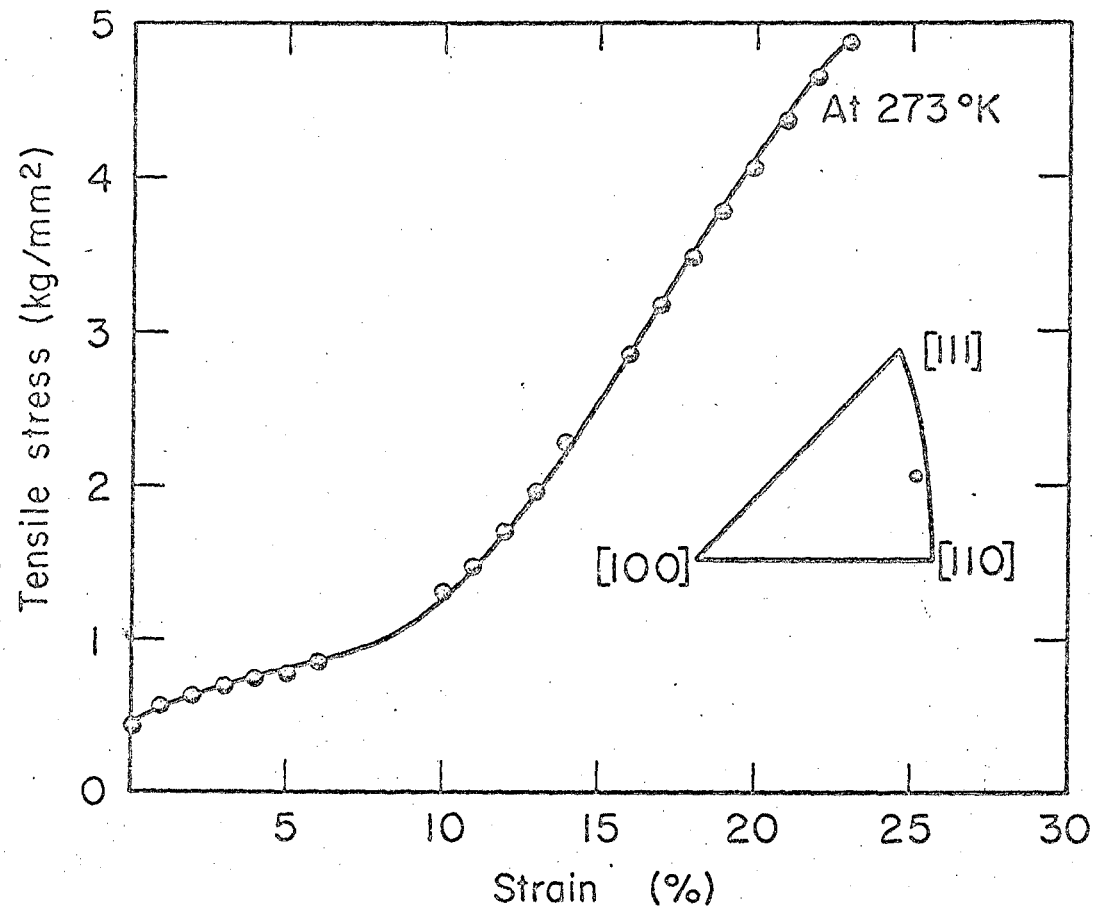

Fig. 12

MU-35768

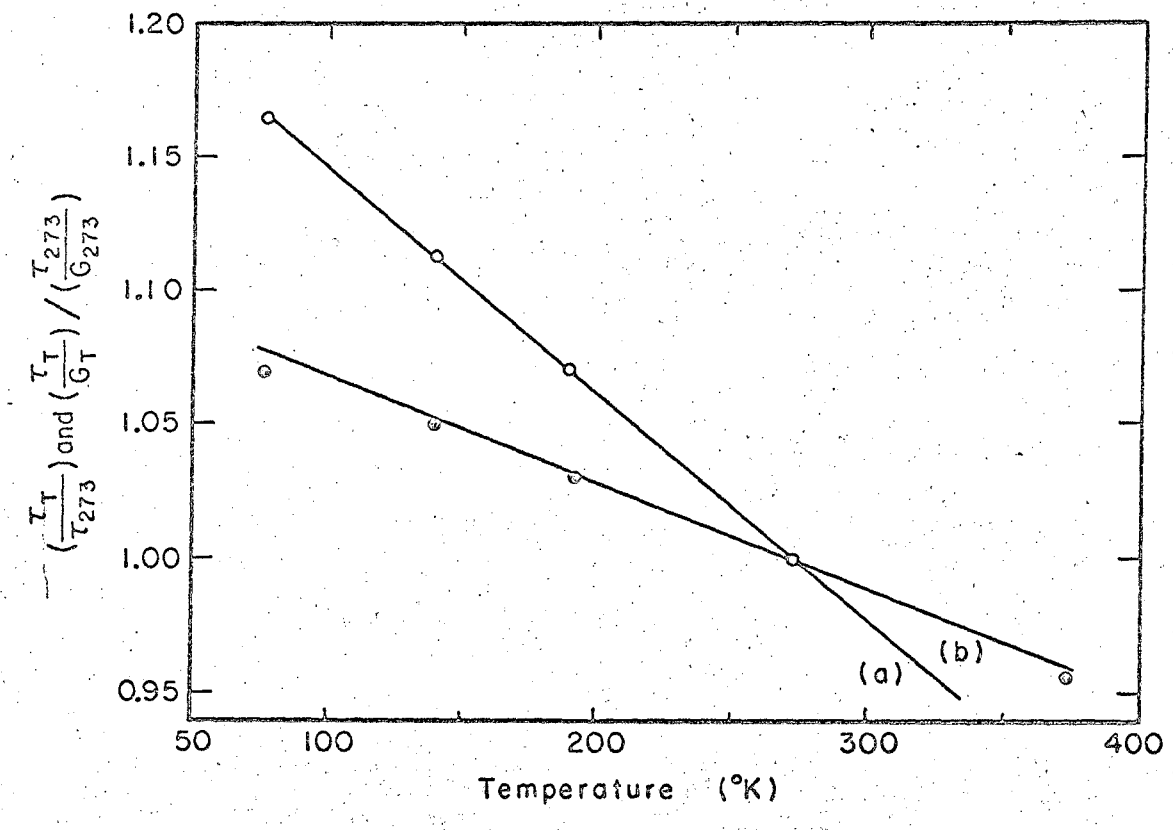

Fig. 13

MU-35647

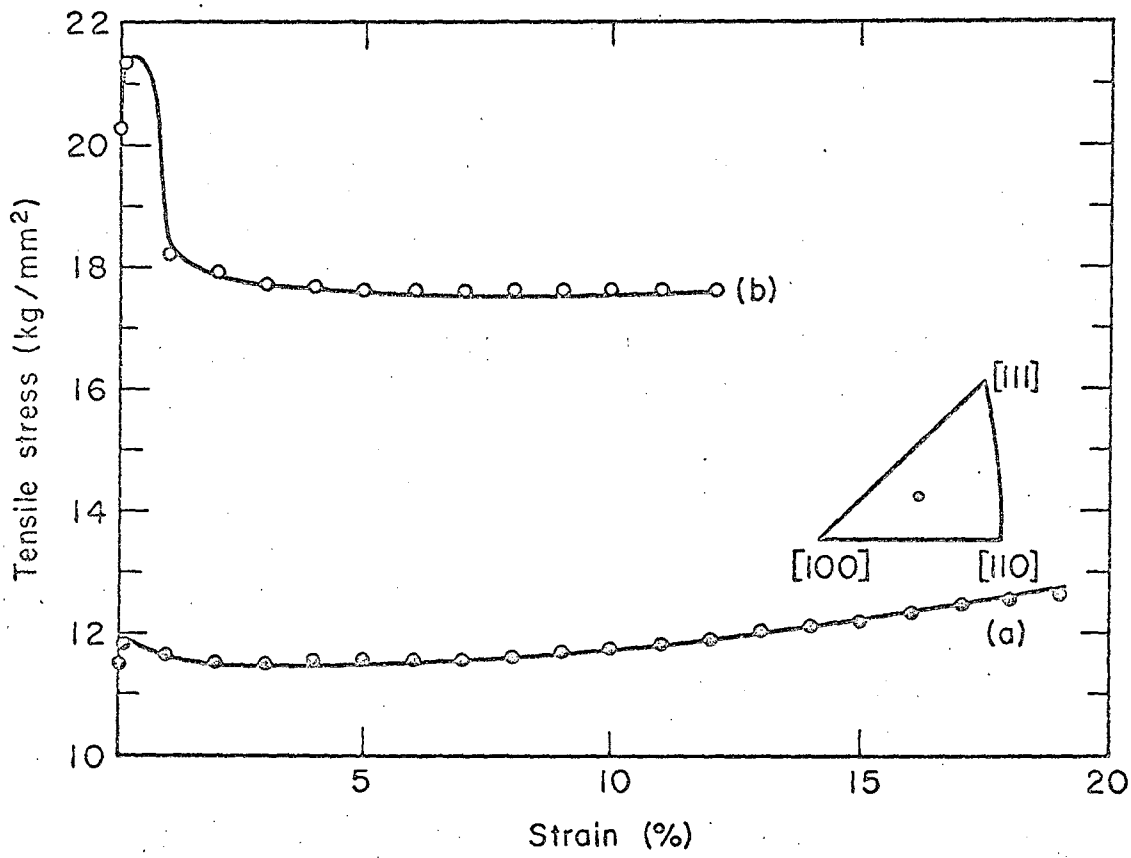

Fig. 14

MU-35770

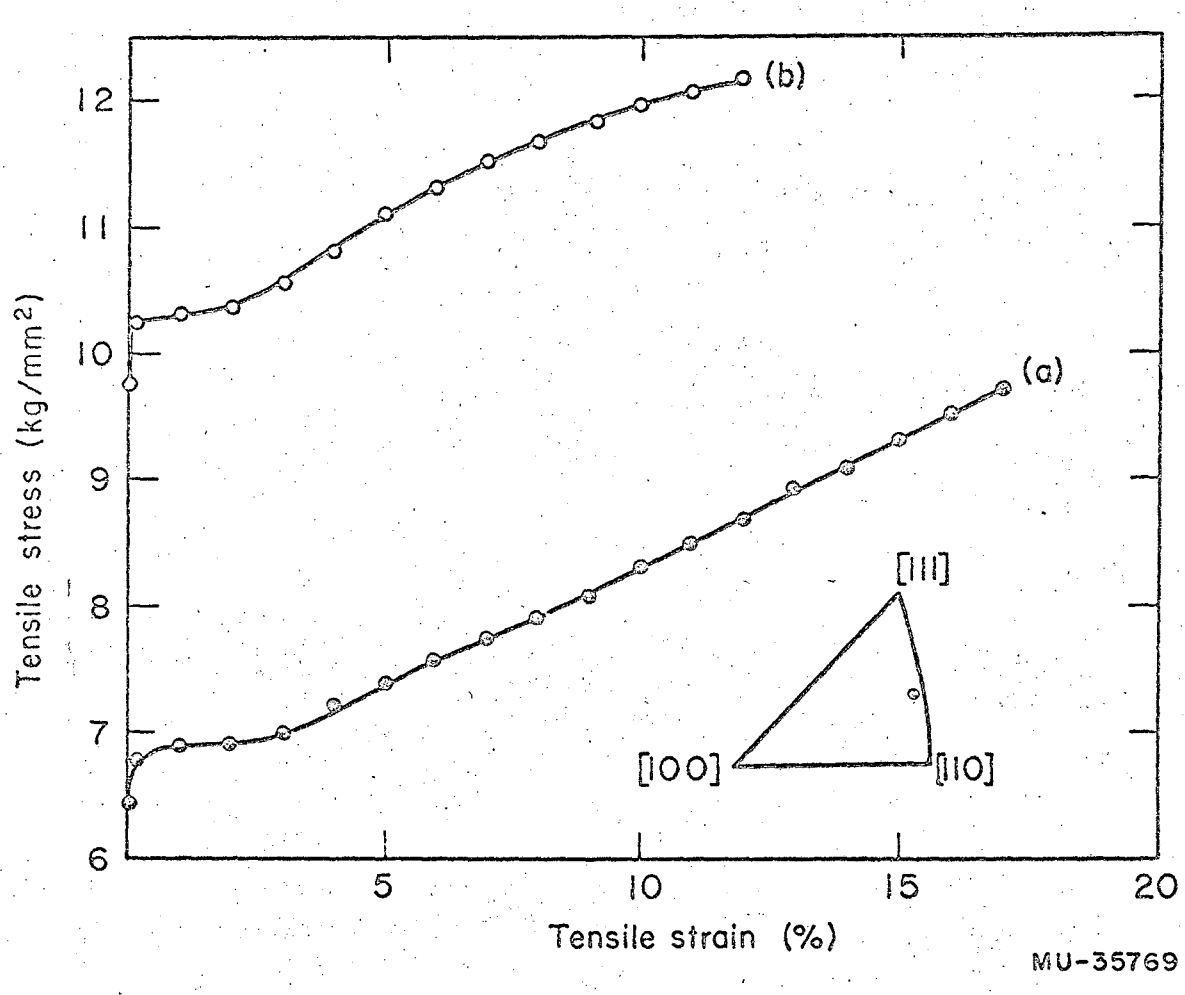
(b)
(a)
[111]
[100]
[110]
Tensile stress (kg/mm²)
Tensile strain (%)
12
11
10
9
8
7
6
5
10
15
20
MU-35769

Fig. 15

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