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THE MECHANICAL BEHAVIOUR OF SINGLE CRYSTALS OF CERTAIN FACE-CENTRED CUBIC METALS

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[Plates 1 and 2]

Single crystal rods of gold, silver and nickel have been prepared from the melt by a modified cooling method. The three metals differ in that the gold is very pure and has no surface contamination; the silver is very pure and the surface normally shows contamination; the nickel is markedly less pure. These three features find expression in the results.

The stress-glide relationship has been investigated over the whole range to fracture, special attention being paid to the region of small glide. At very small glides the results show certain variations attributed to metastable conditions. For pure metals at low temperatures a region of easy glide has been established, which is absent in nickel. Considerations of the variation of hardening with temperature at constant glide show that, at moderate glide, rise of temperature at low temperature promotes hardening, but rise of temperature at high temperature leads to decrease of hardening. From these results it is concluded that in metals without notable impurity glide is due to two distinct processes, first the initiation of glide planes, which temperature agitation tends to check, and secondly to glide proceeding on planes already initiated, which temperature agitation tends to promote.

The surface contamination of silver single crystals which takes place by exposure to air at room temperature raises the critical shear stress somewhat, but has a much more marked effect on the hardening after some tens of per cent of glide has taken place, which is attributed to diffusion of the impurity atoms, probably oxygen, taking place during the travel of the dislocations. This is distinct from normal diffusion, which is negligible at room temperature. The results on the effect of surface contamination explain abnormalities previously noted with silver.

The close connexion of asterism and hardening is exhibited in the results obtained with gold and silver. In particular, silver in the region of easy glide shows very small asterism, even at large glide.

The cube root of the breaking stress of all metals in single-crystal form has been found to show a linear relationship with temperature.

Introduction

The hardening of single crystals of metals under deformation is a complicated process which, from the first onset of permanent glide to the very considerable strains that precede fracture, presents a series of unsolved problems. The general theory of deformation by the propagation of dislocations leaves uncertain the origin of the dislocations which are postulated, both as regards position and circumstances of local stress, and gives no account of the influence of temperature. In particular, the part played by surface conditions is not clear, nor the conditions that lead to twinning nor the interaction of one set of possible glide planes on another.

Face-centred cubic metals offer certain attractions as materials for research on the plastic properties of single crystals. Many of them have high melting-points, and so permit investigation over a wide range of temperature; they show hardening to a remarkable degree, as is clear from the work of Sachs & Weerts (1930) on gold, silver and copper and of Osswald (1933) on copper and nickel; and they show multiple glide without twinning, which is one simplifying factor. Measurements made hitherto on face-centred cubic metals have paid little attention to the early stages of glide, being concerned with the general features of the shear stress/glide curves and, in most cases, have been restricted to atmospheric temperature. The experiments on single crystals of gold, silver and nickel described in the following pages were carried out with the object of throwing light on the mechanism of glide from the inception up to the stage that leads to the very marked hardening shown by these metals and final fracture.

THE PREPARATION OF CRYSTALS

Attempts were made to prepare single crystals of silver and nickel by maintaining wires at high temperature by the passage of an electric current and superimposing a moving temperature gradient by a small travelling electric furnace, a method which was successful with molybdenum and alpha iron (Andrade 1937; Andrade & Chow 1940). The results were not very satisfactory, although short lengths of single crystal were so produced. Other attempts to prepare the single crystals by the method of Andrade & Roscoe (1937), now generally used for many metals of comparatively low melting-point, were likewise unsuccessful. It was then decided to adopt the method of slowly lowering a tubular container of the molten metal from a furnace, which leads to progressive solidification from the bottom (Bridgman 1923, 1925). The chief objection to this method is the difficulty of removing the very soft single crystals from the crucible. This difficulty has been avoided by the use of a new form of crucible, consisting of two halves separated by an axial plane. Crucible and furnace are contained in a high vacuum, which avoids both oxidation of the graphite crucible and any possible attack on the metal by gas.

With gold and silver, graphite is in every way a suitable substance for the crucible, and has been used by previous workers. The form adopted by us is shown in figure 1: there are two half cylinders, of the form shown in *AA*, that fit together; *P* is a plug that closes the lower end with a friction fit; *R* and *R* are well-fitting graphite rings which can be slipped on to hold the crucible together. *P*, which is shown enlarged at *P′*, is bored with a conical axial hole, so that the molten metal terminates in a point where a seed crystal can readily form. With nickel, graphite crucibles cannot be used, since the carbon is soluble in the metal. When half-tube crucibles of pure alumina were first tried the metal adhered to the walls. This difficulty was overcome by the introduction of a layer of powdered alumina between the metal and the walls. The two halves of the crucible, the bore of which was greater than that of the desired single-crystal wire, were laid horizontally and packed with the powder; a straight nickel wire was then pressed gently into the powder to an appropriate depth and the other half-crucible placed in position, the two halves being then held together by carbon rings. So surrounded, the nickel wire retained its form on melting and after cooling could be easily removed from the crucible. The powder was brushed off with a camel-hair brush, very little being left adhering.

The vacuum furnace is shown in figure 2. *T* is a tube of pure alumina, the top half of which is wound with a molybdenum heating coil covered with alumina cement. *Y* is an outer steel case, the space *X* being filled with powdered alumina. *T* is fastened to a base-plate *B*, on which stands the flanged glass cylinder *E*, which can be evacuated through *O*. The joints are sealed with apiezon compound. The cylinder is closed at the top by a brass plate Q, on which is mounted a small geared-down motor M, the vertical shaft of which passes through the plate Q and carries a spur wheel which drives the spur wheel G. A gas-tight joint is made by the synthetic rubber seal S; the shaft and the apiezon lubricant are kept cool by a stream of water through a surrounding jacket. With the help of a small pulley and the phosphor-bronze wire W the crucible C is lowered at a rate of 10 cm./hr.

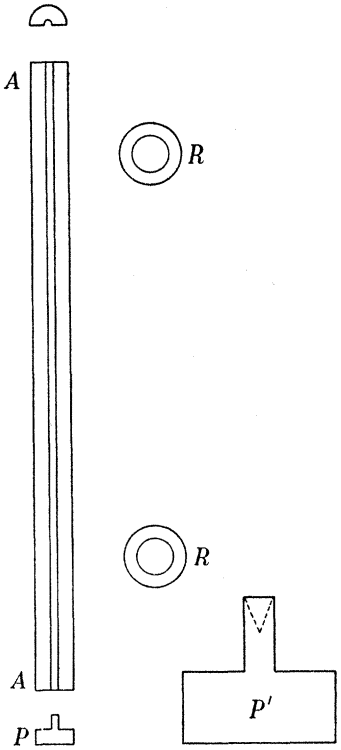

FIGURE 1. Graphite split crucible.

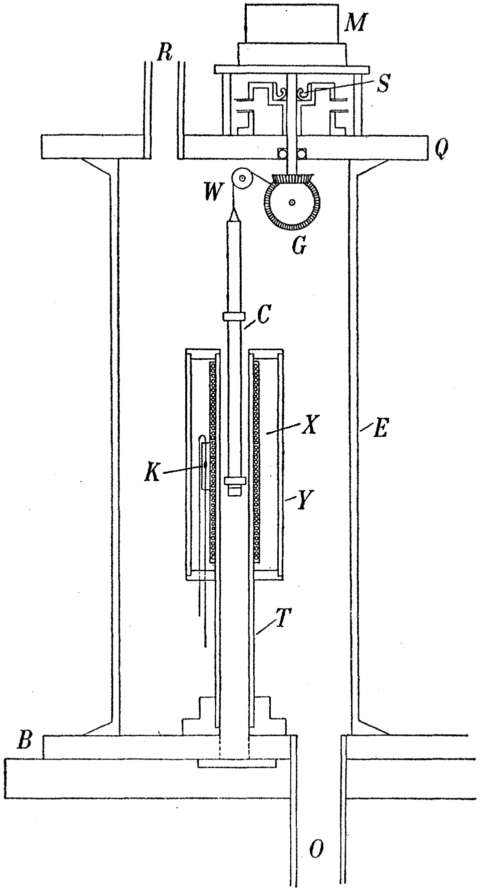

FIGURE 2. Vacuum furnace with moving crucible, for making single-crystal rods.

A Metro-Vac O3 oil diffusion pump backed by a rotary pump maintains the pressure, measured by a Pirani gauge attached at R, at about 0·0001 mm. of mercury. This seems to be considerably lower than the pressure hitherto used in such work. A high degree of evacuation is particularly desirable with silver, which readily takes up oxygen.

The furnace temperature was kept constant to within ±2° C at the highest temperature (1500° C), and somewhat better at lower temperatures, by the method described by Zabel & Hancox (1934). The thermocouple K (figure 2) controls a galvanometer mirror, light reflected from which acts on a photocell, which in turn governs the current through a thyratron.

STRESS-STRAIN MEASUREMENT

The stress measurer, shown at B in figure 3 and in more detail at figure 3a, was of the Polanyi bent beam type. The ends of the beam F rest on parallel cylindrical rods rigidly connected to a block carried by the screw A, which passes through the screw head

H resting on the block *V*, *V* being rigidly attached to the plate *P*. This steel plate, which carries the whole extension apparatus, was supported by a strong girder system. The force is applied to the specimen through a steel rod *R*, pinned to a block carrying a steel point *D*, which rests on the middle of the beam; this rod *R* ensures that the extensometer beam is well away from the jacket used to maintain the temperature of the specimen *S*. Two mirrors, *M*, *M′*, attached as shown in figure 3*a*, serve to measure the bending of the beam; a beam of light falls on *M*, is reflected to *M′* and thence back to *M*, finally forming an image of a cross-wire on a metre scale at a distance of about 1 m. from the specimen. The magnification thus effected of the movement of *D* is about 500. Steel beams of different stiffness were used on the same crystal, according as a smaller or a larger force was being applied (see p. 181).

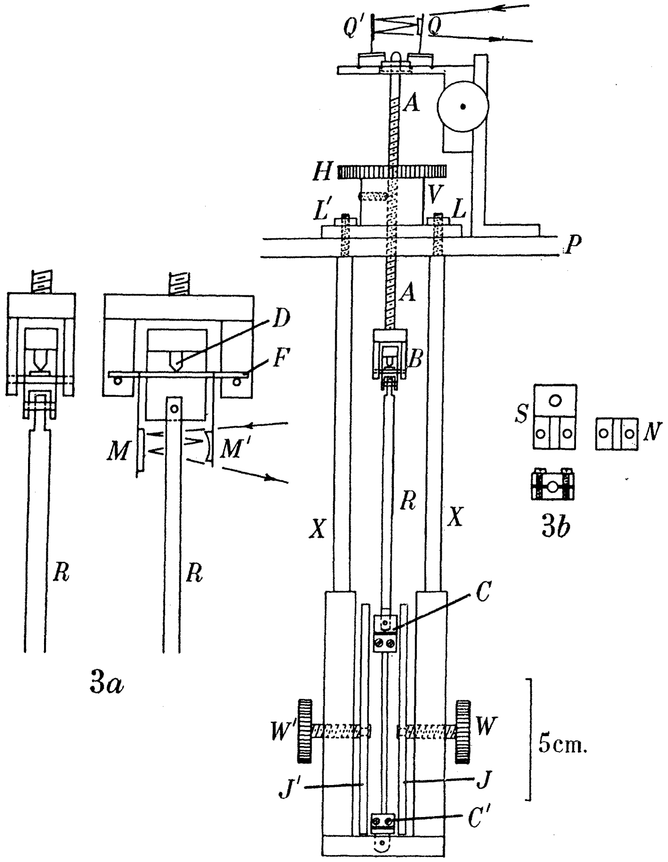

FIGURE 3. Stress-strain apparatus for use with single-crystal rods.

The load was applied by turning *H* so as to raise the rod *A*. To know the extension of the specimen the movement of the upper end of *A* was measured by two optical levers, *Q*, *Q′*, arranged opposite one another, one end of each being supported by a platform attached to *A* and the other by a plate attached to *P* through a rack and pinion. By adjusting the height of this plate at appropriate intervals an extended range of movement—up to 4 cm.—can be measured at the high magnification provided by the lever system.

The supporting rods *X* and the rod *R* are made of stainless steel tube, which is a relatively poor conductor of heat.

Both the stress and strain indicators of the optical system were carefully calibrated, allowance being made in the extension measurements for the yield of the beam. For this part of the calibration the specimen was replaced by a steel wire.

Since the position of the cross-wire on the scale can be read to 0·1 mm. and the optical magnification is about 500, a change of length of 0·0002 mm. can be detected, which is an extension of 1 part in 250,000 on a length of 5 cm. With the metals used a variation of 1° C in temperature gave a change of length just on the border of detection, so that, as the worst temperature fluctuations were slow changes of less than 2° C, their effect can be neglected. Calculations showed that the thermal effects of applying the loads were also negligible.

A special method of mounting the single-crystal wire was devised with the object of minimizing strain in handling. The part of the apparatus below the plate *P*, which could be freed by removing the nuts *L*, *L′*, was placed horizontally on a table. The steel clamps *C*, *C′*, shown in detail in figure 3*b*, consisted of a part *S*, which could be attached to the frame by a pin, provided with a semi-cylindrical groove and an attachment *N*, secured by screws, with a corresponding semi-cylindrical groove. The radius of the cylinder was that of the single-crystal rod, but each groove was slightly less in depth than the radius, so that when the screws were tightened the attachment *N* gripped the wire a little before the two steel pieces came into contact. This was found to give a satisfactory grip. The clamps were firmly held in an accurately alined position by means of the two plates *J*, *J′*, controlled by the screws *W*, *W′*. The single-crystal rod was then transferred from the half-crucible, placed horizontally, to a small brass platform, provided with a *V* groove to receive it. The platform was carefully lowered between the jaws *J* and *J′* until the ends of the crystal rod just rested in the grooves *C* and *C′*; the rod was then secured by the gripping piece, *N*, at each end. During all the preliminary operations, such as taking the X-ray pictures, the jaws *J*, *J′* were retained in position, being released only when everything was ready for the extension of the crystal. In this way deformation, before loading, of the very soft rod was avoided, a very important point when small glide is being investigated.

Experimental procedure

After the specimen had been mounted, by the method described, in the extensometer, a back-reflexion Laue photograph was taken on a Raymax set, the target being generally tungsten, the voltage 50 kV, and the film distance 3 cm. A typical photograph from a gold crystal is shown in figure 4, plate 1. In the early stages of the work pictures were usually taken at three or more different points of the wire, to make sure that it was a single crystal, but later this precaution was found to be unnecessary.

When the extensometer rods *X*, *X* had been secured to the plate *P* (figure 3) and the plates *J*, *J′* released, a small load was applied by turning the head *H*, the stress and strain being read by the optical devices already described. Periodically the load was increased, stress and strain being read at each stage. A steel beam of high sensitivity was used for the small extensions; at an appropriate stage this was replaced by a stiffer beam. X-ray photographs were also taken at different stages, the crystal holders being clamped by the plates *J*, *J′* and the extensometer removed for the purpose. Such occasional interruption of a stress-strain series of experiments on one crystal was found experimentally not to affect the course of the stress-strain curve.

For the higher temperatures the lower part of the extensometer, up to about the mid-point of the rod *R*, was surrounded by an electric furnace, the temperature of which was controlled to be constant within 2° up to 600° C. At these temperatures the crystals were, in general, in contact with air, but a method is described later by which, in certain cases, the wires were heated in an inert gas. For temperatures below atmospheric the lower part of the extensometer was enclosed in a rectangular copper box, which was surrounded by a solid CO_2-alcohol mixture or by liquid oxygen. At these low temperatures the crystals were, in general, maintained in an atmosphere of carefully dried argon, which issued in a slow stream from a pressure cylinder. Before entering the copper box it passed through a copper spiral immersed in the cooling medium, to bring it to the appropriate temperature. The top of the box was closed by a lid, through a hole in which the rod *R* passed without contact. Since cold argon is much denser than air and was continually supplied to the box there was little possibility of air entering through the small clearance round the rod.

The results are exhibited not as extension against axial stress, but as glide against resolved shear stress, the glide being the relative displacement of two parallel glide planes unit distance apart, and the resolved shear stress the stress on the glide plane resolved in the operative glide direction. The angles χ and λ, which the glide plane and the glide direction respectively make with the wire axis, are found from the analysis of the X-ray back-reflexion photographs.

Gold

The metal was supplied as spectroscopically pure by Messrs Johnson and Matthey; it contained less than 10 parts in a million of foreign metal, the chief impurities being silver and copper. The crystals were all prepared *in vacuo*, as already specified, with a furnace temperature of 1170° C.

The hardening curves are exhibited in two diagrams, one showing the initial stages of glide, up to 1·4 %, on a large scale (figure 5) and the other (figure 6) showing the whole range up to fracture, which, remarkably enough, took place at roughly the same glide, viz. 90 %, at all temperatures.

The curves for low glide show in two cases, 17 and 414° C, two very well-marked bends, one following the other, within the first 0·2 % glide. Such sharp bends were first recorded by Greenland (1937, figures 8 and 9) in his work on mercury carried out in the laboratory at University College, London. In the curves for gold at −80 and 330° C the first bend is sharply marked but the second less so.

The critical shear stress is not a very precisely determined quantity; as pointed out by Andrade & Roscoe (1937), if we assume that a formula of Becker type governs the rate of flow, then the critical shear stress should be defined as the stress required to produce an arbitrary small rate; fairly large changes in the rate, however, correspond to relatively small changes in the stress. In our investigation the readings were taken 2 min. after the application of the load concerned; a subsidiary experiment showed that the stress/strain curves would not have been significantly changed if this interval had been increased by a factor of ten times.

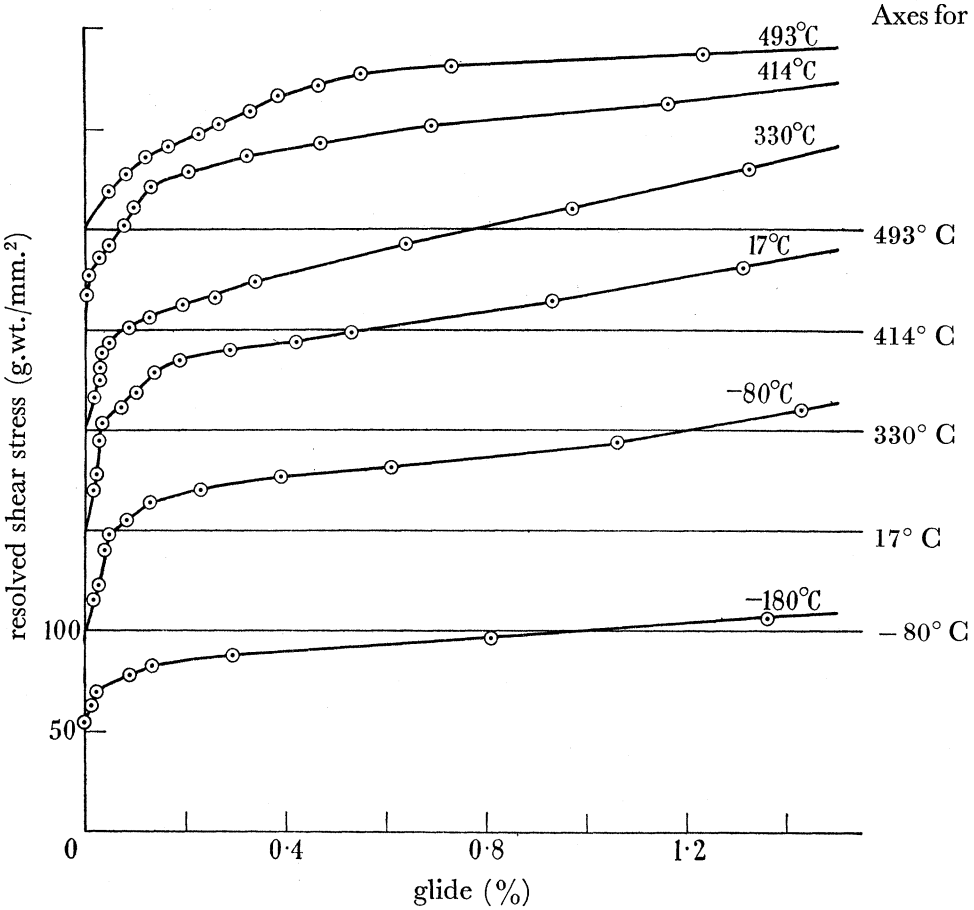

Figure 5. Resolved shear stress/glide curves for gold; small glide.

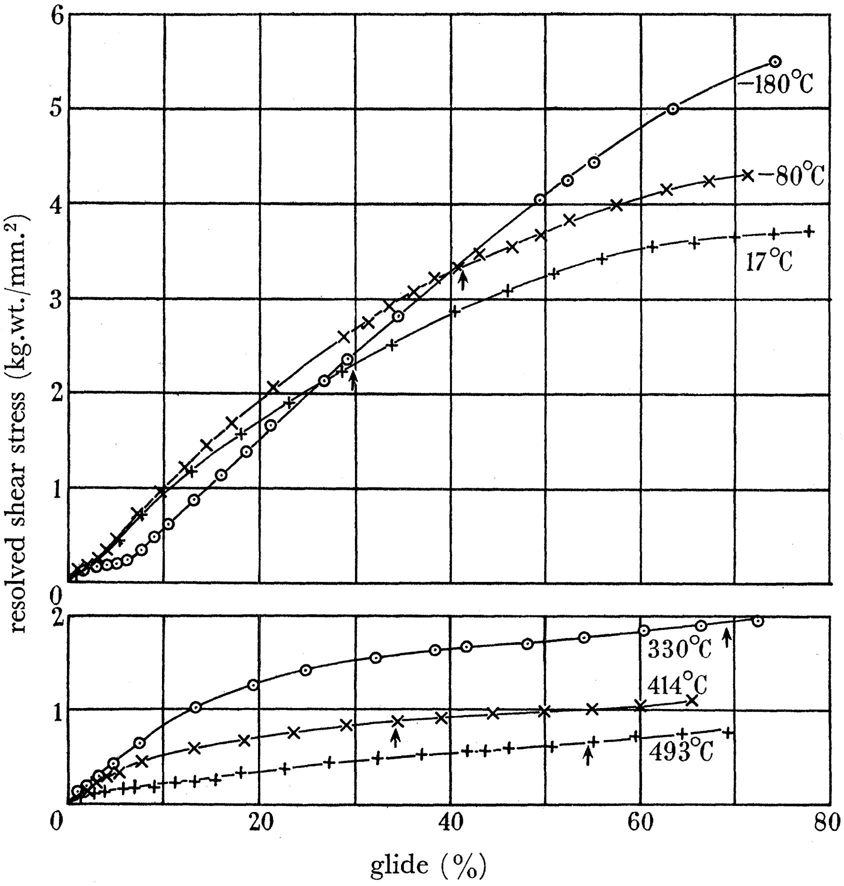

Figure 6. Resolved shear stress/glide curves for gold; large glide.

Since the critical shear stress is very sensitive to small impurities* and is, in general, highly structure-sensitive, it is not a quantity for which an exact value characteristic of the metal can be expected. However, if we take as the critical shear stress the value at which the slope of the shear stress/glide curve first rapidly diminishes we obtain a figure which, within a few per cent, appears to have a physical meaning.

TABLE 1. GOLD

temperature (° C)	critical shear stress (g. wt./mm.^{2})
−180	59
−80	48
17	50
330	37·5
414	27·5

With gold, values obtained in this way are given in table 1, and shown, against temperature, in figure 27, together with those for nickel. The value given for gold at 17° (Sachs & Weerts (1930) is 91 g.wt./mm.^{2}, which is considerably in excess of our value. No estimate of the purity of the metal is given, and the authors were mainly concerned with the very large changes of critical shear stress caused by alloying with silver.

X-ray pictures were taken at various stages, and from them the positions of the wire axis were derived. Figure 7 shows, in the standard projection, the initial position of the axes of the different wires. The movement of the axis caused by glide confirmed that, as often recorded, the glide elements were the (111) plane, and the [110] direction for which the resolved shear stress was greatest. The records of the movement of the axis enabled us to deduce the stage of glide at which it became symmetrically placed with respect to two possible glide systems, the original system (the one on which glide had been taking place) and the one that we shall call the new system; this stage is indicated by the arrows

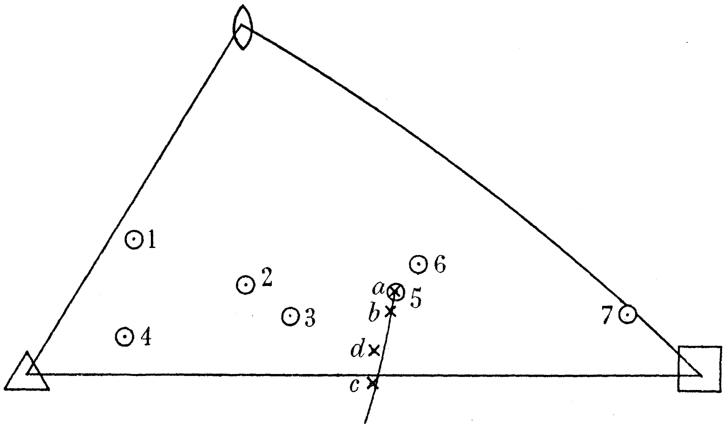

FIGURE 7. Initial position of wire axes of gold single crystal rods in the standard projection. Movement of axis with glide is shown for one case by crosses *a*, *b*, *c*, *d* are successive positions: the circular arc through *a* shows the path to be theoretically anticipated. The numbers 1 to 6 indicate rods used for following temperatures respectively: −180, −80, 17, 330, 414, 493° C.

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* Cf. Greenland (1937) on mercury containing very small amounts of silver.

in figure 6. There is no abnormal change in the course of any of the curves at the marked point. Actually the glide continues on the original system for a small further extension, although purely geometrical reasoning would lead one to expect double glide. This points to a virtual hardening of the new (non-active) system which exceeds the hardening of the active system. When some percentage glide has taken place beyond the symmetrical position of the two systems, double glide sets in, without any apparent effect on the course of the stress-glide curve, and the wire axis returns to the symmetrical position. The travel of the wire axis for one particular gold crystal are shown by the crosses on figure 7 lettered in sequence. Like results as regards double glide were obtained with our silver and nickel crystals, the curves for which are likewise marked with arrows at the symmetrical position. A similar behaviour was noted earlier by Sachs & Weerts (1930) on gold and silver alloys and by Osswald (1933) with copper and nickel alloys. (Figure 6 in Osswald's paper shows, for a copper-nickel alloy, a typical course of the wire axis, during glide, in the standard projection.) Figure 8, plate 1, shows the typical low magnification appearance of glide packets in a heavily strained gold crystal and figure 9, plate 1, the appearance, at high magnification, of a gold crystal in which triple glide has taken place.

The surface of the unstrained gold crystals shows, at high magnification (×1800), a laminar structure of which an example is given in figure 10, plate 1. The curvature of the lines is due to the fact that they are intersections by the (100) planes of lamellar growth with the surface of the specimen, in which small surface irregularities are imposed on the general cylindrical form. The average spacing, as measured in the photograph, is about 1·5 μ, the actual separation of the privileged planes being about 1 μ. This is roughly the same as the separation of the planes of the substructure discussed by Graf (1942) and by Andrade & Randall (1950), who showed that the planes were revealed, but not caused, by strain. Earlier, Straumanis and Stranski and his collaborators established a natural spacing of this order for hexagonal metals (for references see Andrade & Randall 1950).

Silver

The silver was Messrs Johnson and Matthey's spectroscopically pure metal, containing less than 10 parts in a million of impurities. The crystals were prepared by the method used for the gold crystals, with a furnace temperature of 1100° C. Two different samples of metal were used, both to the same specification; the results obtained with crystals grown from sample I are shown in figures 11 and 12, with crystals grown from sample II in figures 13 and 14, figures 11 and 13 showing the early part of the glide, up to 1·5 % only, and figures 12 and 14 showing the behaviour over the whole glide range. The character of the early extension varies; at 530° C the curve shows the abrupt bend at two places, to which attention has already been directed in the case of gold and mercury, while at other temperatures there is a more or less sharp single bend, followed by regions in which the resolved shear stress is roughly constant for a glide of 1 % or so. The critical shear stress is taken as that at which the slope of the curve, initially constant, first begins markedly to diminish. This critical shear stress is set down against temperature in figure 15, which shows a remarkable rise at high temperatures, quite contrary to the recorded behaviour of most other metals. However, Miller & Milligan (1937) found a rise of the critical shear stress of both silver and aluminium over a certain range of temperature; their values for silver are shown in figure 15, and agree well with ours.*

To elucidate this abnormal behaviour, the load-extension curve was taken with a specimen alternately at 17 and 390° C, with results shown in figure 16. The first determination at 390° C shows a marked rise in the stress as compared with the first determination at 17° C, but the determination at 17° C, after cooling from 390° C, shows still higher stress. The next measurements at 390° C show about the same stresses as at the second 17° C experiment, but the third 17° C experiment gave increased stresses. The 390° C experiment

FIGURE 11

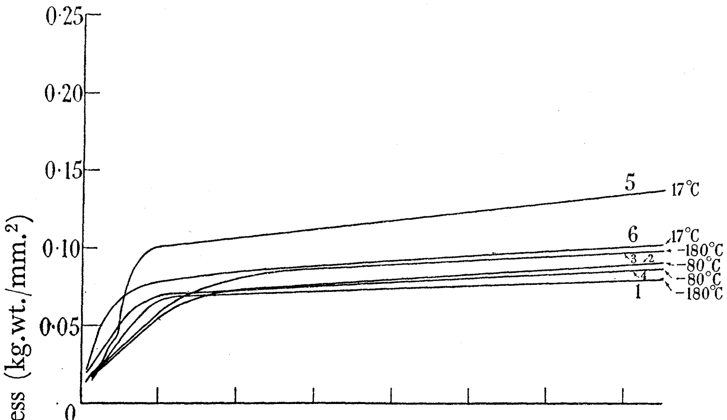
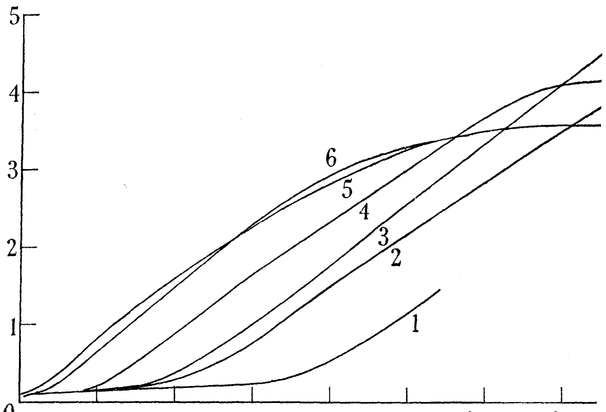


FIGURE 12



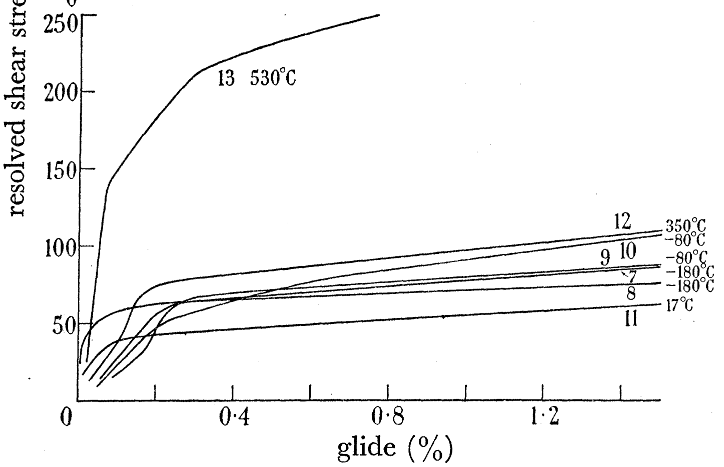


FIGURE 13

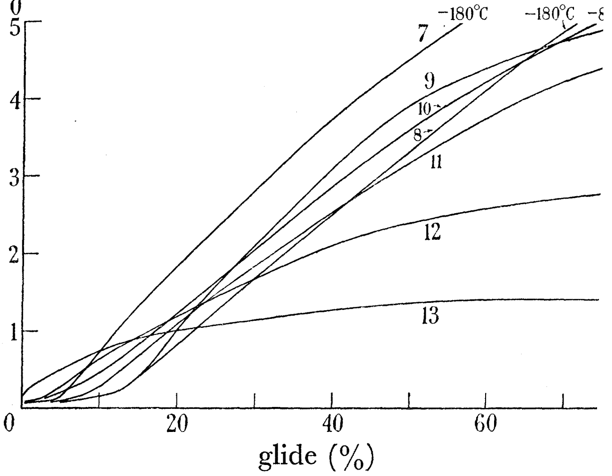


FIGURE 14

FIGURE 11. Silver single crystals, metal sample I. Resolved shear stress/glide curves: small glides. 1 and 2, in contact with liquid oxygen −180° C; 3, in argon at −180° C; 4, in argon at −80° C; 5 and 6, in air at 17° C.

FIGURE 12. Silver single crystals, metal sample I. Resolved shear stress/glide curves: large glides (for significance of 1–6, see figure 11).

FIGURE 13. Silver single crystals, metal sample II. Resolved shear stress/glide curves; small glides.

FIGURE 14. Silver single crystals, metal sample II. Resolved shear stress/glide curves; large glides. At −180 and −80° C in argon; at higher temperatures in air.

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* The value which they found at 100° C has been omitted, since the authors say that the crystal in question was accidentally bent before testing, which would cause the measured stress to be too high.

after this showed stresses not very different from the preceding 390° C experiment, and alternating cycles at 17 and 390° C tended to bring the stresses to a limiting value. All

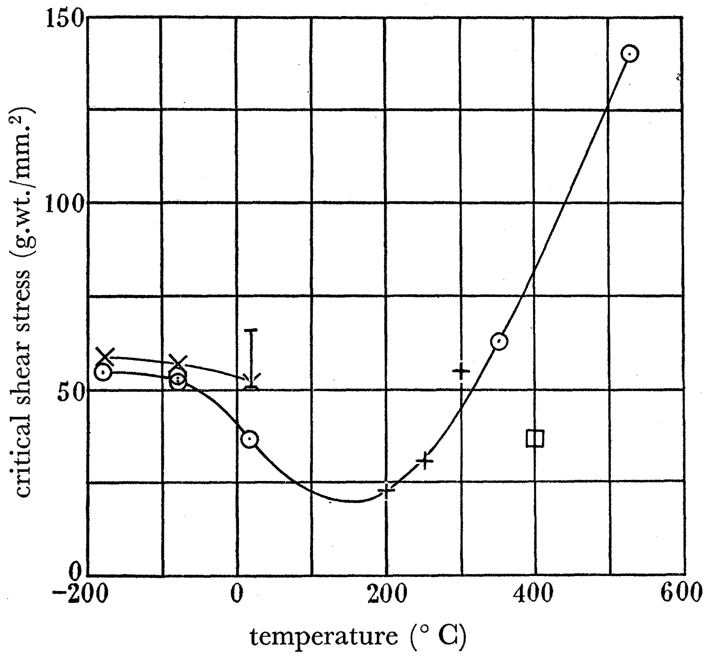

FIGURE 15. Silver single crystals: critical shear stress against temperature, showing effect of invisible surface oxide layer. All points at room temperature and higher are for crystals in air, except □, which is for crystal in krypton-xenon mixture. × Crystals from metal sample I. ⊙ Crystals from metal sample II. + Results of Miller & Milligan. The vertical line at 17° C indicates the range of values found by Sachs & Weerts for critical shear stress of silver at room temperature.

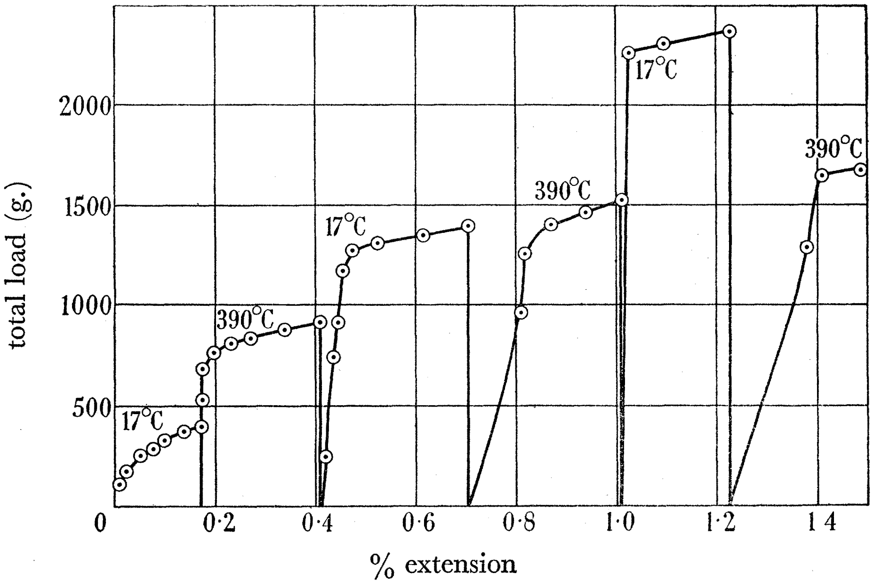

FIGURE 16. Effect of heating in air, with formation of oxide layer, on stress-strain curve for silver.

these results can be explained on the supposition that the heating in air gives rise to an oxide film which causes a rise in the resolved shear stress necessary to produce glide. Roscoe (1936), working in the University College laboratory, showed with cadmium that an oxide film has this hardening effect, the stresses increasing as the film is thickened, and similar surface effects have been recently established (Andrade & Randall 1948; Andrade 1949; Andrade, Randall & Makin 1950; Harper & Cottrell 1950; Phillips & Thompson 1950). The results expressed in figure 17 confirm this conclusion. After an experiment to establish the stress necessary for standard glide the crystal was heated in air and tested after cooling; the necessary stress was found to have increased to about three times the original value. Subsequent heating *in vacuo* at 800° C for half an hour diminished the stress, as would be expected, since much of the film should have been dissipated by the treatment. It should be emphasized that no oxide film could be seen on the surface of the crystal at any stage, so that it must have been very thin.

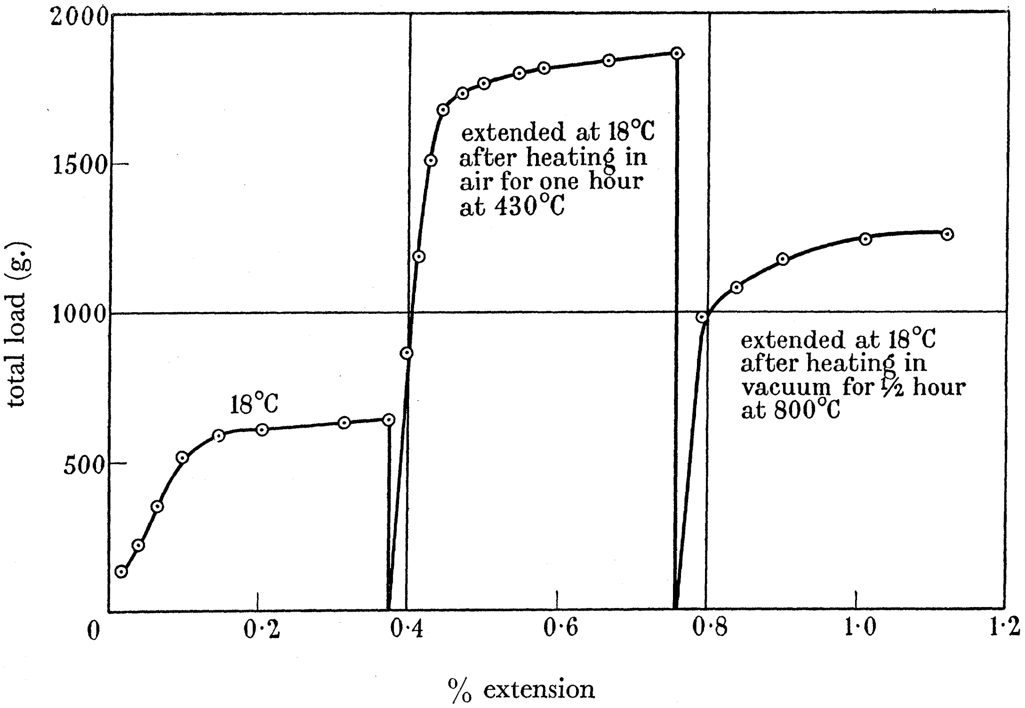

FIGURE 17. Hardening due to formation of oxide layer by heating in air and softening due to evaporation of oxide layer by heating at 800° C *in vacuo*.

To place this hardening effect of an invisible oxide film beyond doubt, an experiment was done in which the crystal was heated out of contact with air. To carry out the experiment *in vacuo* would have meant devising a stress-strain measuring device which could be included in the exhausted vessel, a troublesome if by no means impossible task. The method adopted was to surround the wire with a mixture of xenon and krypton, used because such a mixture is considerably cheaper than pure xenon. Even at 400° C krypton is considerably denser than air at atmospheric temperature. The part of the extensometer containing the wire was sealed off, leaving just enough clearance round the rod *R* to admit free movement. The only method of escape for the inert gas was thus diffusion against gravity through this narrow opening. The stress-glide curve for these conditions is shown in figure 18, compared to that in air at 350° C, and the single point in figure 15 shows the critical shear stress, which is below that at 17° C, whereas for a crystal tested in air at 400° C it was much higher. It is possible that a little oxygen contamination reached the wire and that the point for a clean surface should be even lower, but the effect of oxygen is clearly established.

The fact that the stress-strain curves for certain of the silver crystals show an increased slope in the region just below the critical shear stress, a feature which gold crystals never show, can probably be attributed to oxide-layer effects.

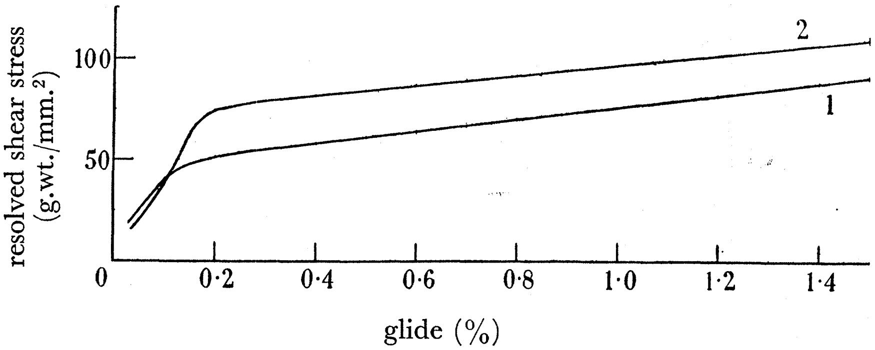

Figure 18. Resolved shear stress/glide curve for a silver crystal at 400° C in krypton-xenon mixture (curve 1), compared with that of a like crystal at 350° C in air (curve 2).

Since the film that produced such considerable mechanical effects was invisible it seemed possible that, although the crystals were prepared *in vacuo*, a sufficient oxide layer might be formed by the action of the atmosphere at room temperature. To test this point a crystal was extended under conditions that avoided such action. The crystal, made in the usual way, was fixed in the holders of the extension apparatus as previously described, and then heated in a highly evacuated silica tube at 900° C for 2 hr., which cleans the surface by vaporization. After the temperature had fallen to atmospheric, dry argon was admitted and the whole cooled to liquid-air temperature. The crystal, in its holder, was then transferred to the extensometer, where it was surrounded by an atmosphere of argon at liquid-oxygen temperature. This procedure ensured that the crystal was in contact with the air for a few seconds only, and then at a very low temperature, so that any attack by oxygen was very unlikely. The whole was then allowed to reach room temperature, the crystal being maintained in the dry argon atmosphere, and a stress-strain experiment carried out in the normal way. The result is shown in figure 19; after 20 % glide has taken place the stress necessary to cause further glide is about one-tenth that required for a normal crystal similarly extended. The critical shear stress is lowered, but only by 25 %; the striking feature is the great diminution of work hardening. The comparison curve, 18, is not taken from the earlier experiments, but was carried out with a crystal prepared at the same time as 17 and treated in exactly the same way, except that 2 hr. before the measurements were made the argon was replaced by atmospheric air. The curve 18 lies fairly close to the curve for 17° C in figure 12 and somewhat above that of figure 14; all three show the same remarkable contrast to the behaviour of the clean crystal.

A few experiments were done on the thermal etching of silver crystals, following the experiments of Andrade & Randall (1950). Figure 20, plate 1, shows at high magnification a crystal which has been maintained near its melting-point in air, with the result that mirror-bright crystal planes have been bared. The surface of the wire is not cylindrical to a high degree of accuracy, which probably accounts for some of the irregularities shown by the contours at this magnification. The mirror surfaces, which are planes, are traversed by faint lines, running upwards, in the photograph, from left to right at an angle of about 30° with the horizontal. These are probably steps in the baring of the (111) plane, running along the glide direction. The normal separation of the planes revealed is about 1·1 μ. Figure 21, plate 1, shows the glide lamellae in a heavily strained single crystal, where the normal separation is about 0·7 μ. Once more it would appear that there is in the unstrained crystal a fundamental substructure, with planes separated by about 1 μ, which is revealed by straining.

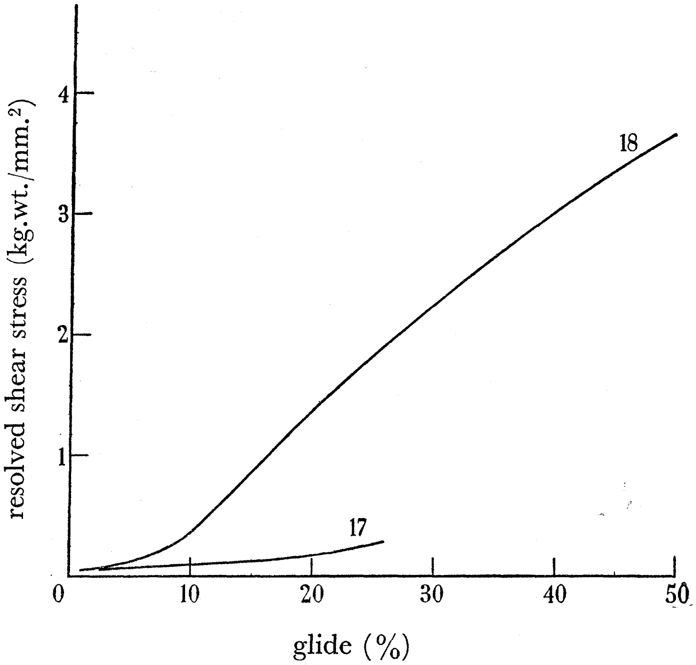

FIGURE 19. Contrast between behaviour of crystal (18) with clean surface that has been exposed to the air for 2 hr. at atmospheric temperature and clean crystal (17) in argon. Both tested at 20° C.

Nickel

The metal, supplied by Messrs Johnson and Matthey, was inferior to the gold and silver in purity; we were given an assurance that there was certainly less than 0·1 % total impurity, and that the chief foreign metals were calcium, copper, magnesium and silicon. Owing to the difficulties which have already been described the surface of the crystals was not as satisfactory as with silver and gold. The presence of adhering alumina particles sometimes gave powder rings in addition to the single-crystal spots in the X-ray photographs and the contrast between spots and background intensity was not great, although this was increased by using copper in place of tungsten as the target.

The crystals extended uniformly, showing the characteristic flattening and slip lamellae. The typical appearance after extension is shown in figure 22, plate 1, both at low and at high magnification. The appearance of the unstrained crystal surface at high magnification is shown in figure 23, plate 1, and of the heavily strained crystal, showing double glide, in figure 24, plate 1. Allowing for the inclination of the planes, the normal separation of the privileged planes comes out to be about 2μ in the case of the unstrained crystal and roughly the same in the case of the strained crystal, confirming the general results discussed in considering the results with gold and silver.

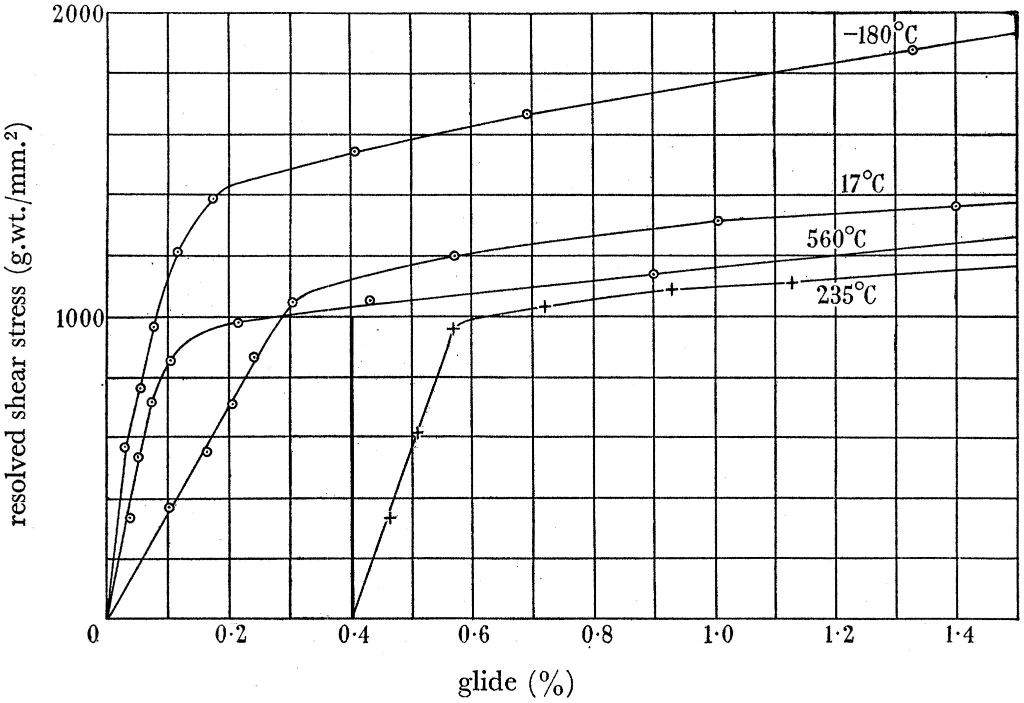

FIGURE 25. Resolved shear stress/glide curves for nickel: small glide. The origin has been displaced to the right by 0·4 % for the 235° C curve, to avoid overlapping with the 560° C curve.

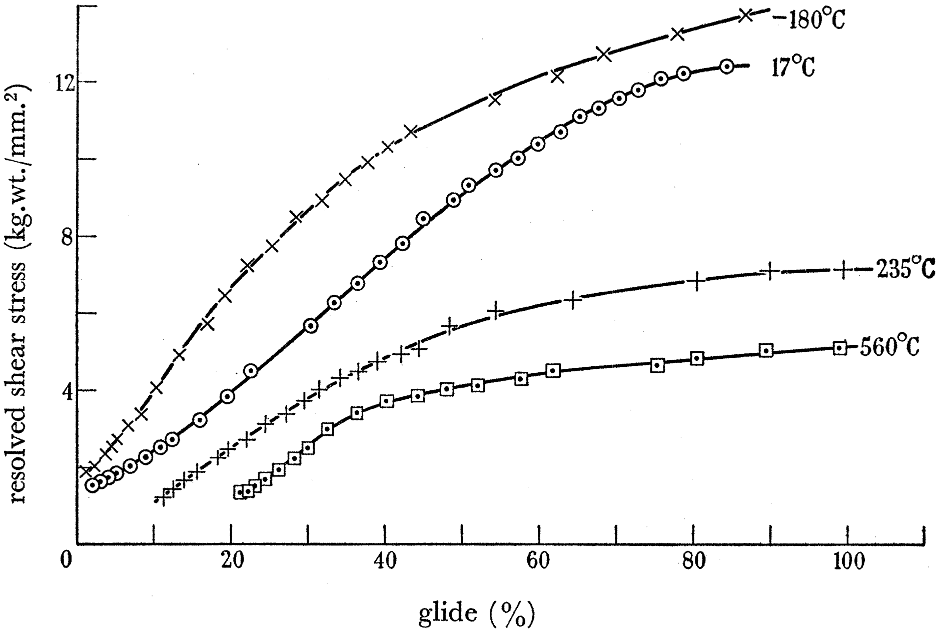

FIGURE 26. Resolved shear stress/glide curves for nickel: large glide. The origin has been displaced to the right by 10% for the 235° C curve and by 20% for the 560° C curve, to avoid overlapping.

The stress-strain curves at high and low glide are given for four temperatures in figures 25 and 26. It will be seen that up to 20 % glide the behaviour at 17, 235 and 560° C is much the same. The critical shear stress is fairly well marked at 17 and 235° C, less well marked at the other two temperatures. It is estimated as in table 2, and shown against temperature in figure 27. The crystal extended at 560° showed a grey surface film after the experiment; the other three crystals remained bright, but, in view of the results obtained with silver, it cannot be assumed that there was no surface contamination. However, the variation of the critical shear stress shows nothing resembling figure 15, so that no very serious surface effects appear to be in question.

Osswald (1933) gives four determinations for the critical shear stress of nickel at room temperature, viz. 713, 445, 606, 545, the average being 580, which is markedly less than our value.

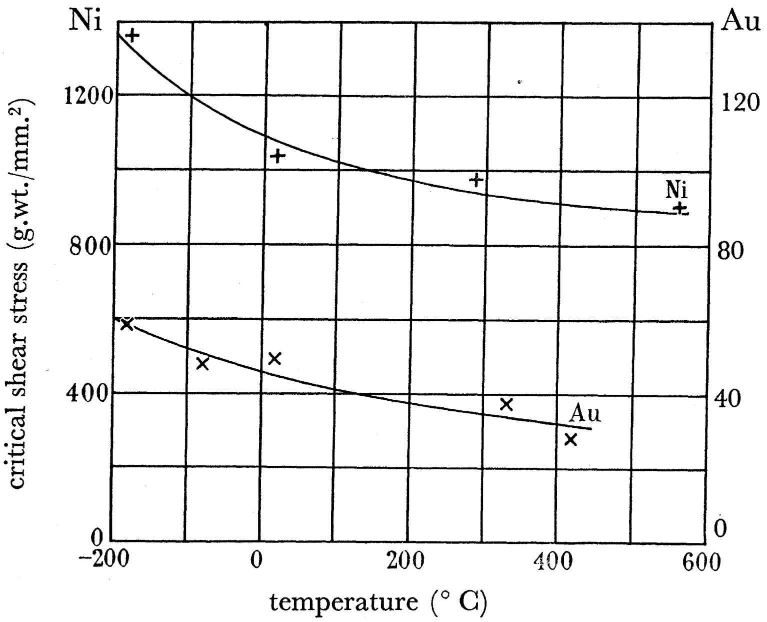

FIGURE 27. Critical shear stress for nickel and gold at different temperatures.

TABLE 2. NICKEL

temperature (° C)	critical shear stress (g.wt./mm.^{2})
−180	1360
17	1040
235	970
560	900

THE VARIATION OF BREAKING STRESS WITH TEMPERATURE

Over the whole range of temperature, practically all the crystals—gold, silver and nickel—used in these experiments broke at about the same glide, lying between the extremes of 80 and 110 %. The breaking stress (resolved shear stress) σ_f increases very rapidly as temperature falls; different determinations at one temperature show a fair consistency. We have established as an empirical rule that the cube root of the breaking stress is linearly related to the temperature. In figure 28, $\sigma_f^{\frac{1}{3}}$ is shown against absolute temperature T, expressed as a fraction of the temperature of melting T_m. Besides our results for silver, gold and nickel the results for aluminium (Boas & Schmid 1931), cadmium* (Boas & Schmid 1930), zinc (Fahrenhorst & Schmid 1930) and magnesium (Schmid 1931) are given. The breaking stress is not a quantity that can be very precisely determined, and there is a certain scatter in the points; nevertheless, the results seem to be significant. All the lines cut the melting-point line ($T/T_m = 1$) at a finite value of σ_f, as is to be expected, except that for magnesium, which indicates a zero value for breaking stress at about 70° C below the melting-point. Magnesium, the slope of the line for which is markedly greater than for the other metals, differs from the other two hexagonal metals (zinc and cadmium) in that at higher temperatures a new system of pyramid face translation planes is added to the basal planes. This would have the effect of making the operative shear stress higher than that calculated.

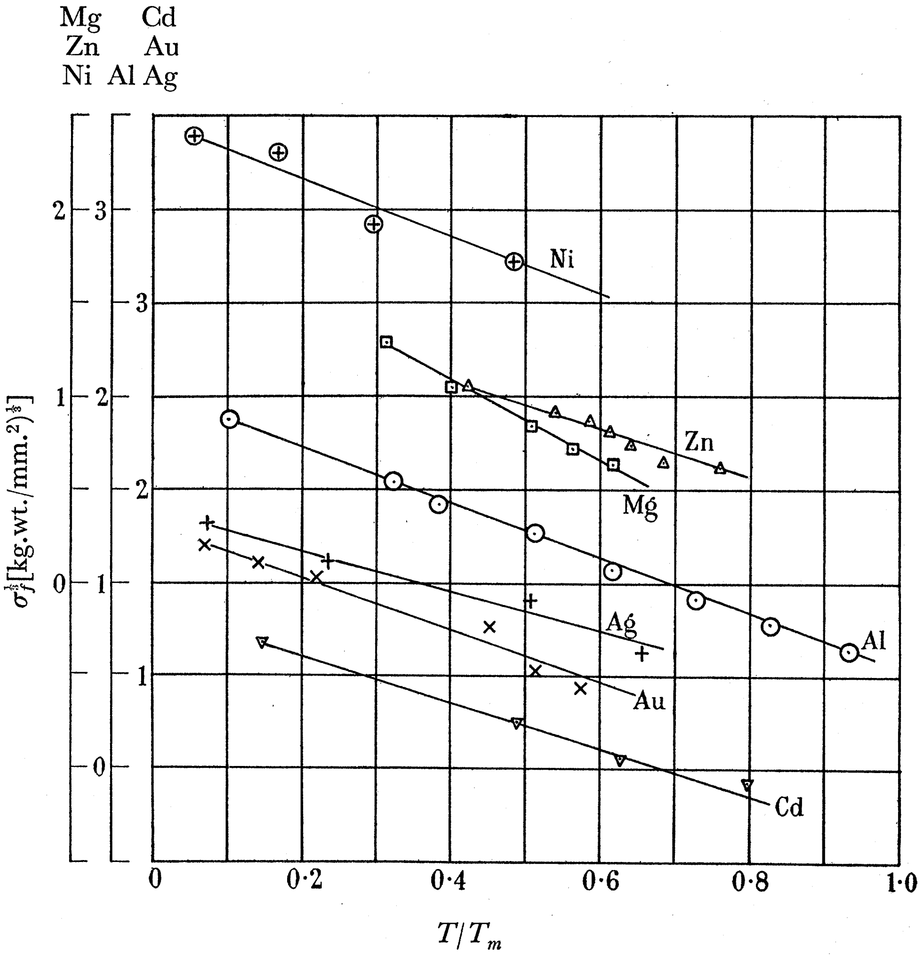

FIGURE 28. Cube root of resolved shear stress at fracture, σ_f, against T/T_m, where T is the temperature and T_m the melting point of the particular metal, both absolute. Three different scales for the ordinates are given, the metals to which each scale pertains being marked at the top.

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* The result for −235° C is omitted, because it is excepted by the authors from the results at other temperatures.

It may be noted that aluminium, the other cubic metal, breaks at about the same amount of extension over a wide range of temperature, as was the case with our gold and silver crystals. The crystals of the hexagonal metals, where the behaviour is complicated by extensive twinning at lower temperatures, show much greater glide at break for high temperatures than for low.

At room temperature, Sachs & Weerts found for the breaking stress of silver and gold 5·6 and 3·6 kg.wt./mm.^{2} respectively, as against our values of 4·5 and 3·6. For glide at breakage their values of 95 and 70 % compare with ours of 95 and 90 %. For nickel Osswald found 8 kg.wt./mm.^{2} as against our value of 12·4; it will be remembered that his critical shear stress for nickel was also markedly less than ours. He found for the glide at breakage 30 % as against our 85 %. There was clearly some difference in the metal specimens; we have already pointed out that available nickel is markedly less pure than silver or gold.

Asterism

X-ray photographs of gold and silver crystals* which have been subjected to high shear stress exhibit asterism, of which examples are shown in figure 29, plate 2. It was found, as has been established before (e.g. for compression, Taylor 1928; Burgers & Louwerse 1931; for extension, Andrade & Tsien 1937; Andrade & Chow 1940), that the extent of the different streaks of the asterism corresponded to a rotation of crystal planes through a single small range of angles about an axis (the 'axis of asterism') in the glide plane perpendicular to the glide direction, planes of all angles within that range appearing. Each photograph was placed on a Greninger chart and the angular range corresponding to the elongation of the spots estimated, which was possible to within about half a degree for each spot, so that the average angular range determined for all the spots should be somewhat better than this.

In figure 30 is shown the hardness (resolved shear stress supported without appreciable flow) against spread of asterism for gold at the different temperatures. It will be seen that at each temperature hardness is roughly proportional to spread of asterism. For -80 and 17° C the points lie approximately on one line; the stress-glide curves of these two temperatures are not very different. With increasing temperature the slope of the line decreases. These results indicate that at a given temperature the hardening, but not the glide, is proportional to the spread of asterism. They also show that at high temperatures there can be large spread of asterism and comparatively little hardening, whereas for sodium Andrade & Chow found that the spread of asterism was proportional to the hardening, independent of temperature. The conditions for the two metals are, however, very different; the gold is very pure, while the sodium, although a pure specimen for this particular metal, may have contained 1 part in 10^4 of potassium; the greatest glide in the sodium measurements was 5 %, for which the spread of asterism lay between 5 and 10°, according to temperature, while for gold the glide extended to over 40 % and the spread of asterism at 5 % glide was 1 or 2°.

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* With nickel no satisfactory photographs of asterism were obtained, as explained in the section on nickel.

The explanation of the spread of asterism with comparatively little hardening at high temperature is probably to be sought in the fact that, as already found with other metals (Tsien & Chow 1937; Andrade & Chow 1940), the spacing of the glide planes is markedly coarser at high temperatures with gold and silver. This means that the glide is concentrated in comparatively few regions where it would be large, which necessarily entails large asterism.

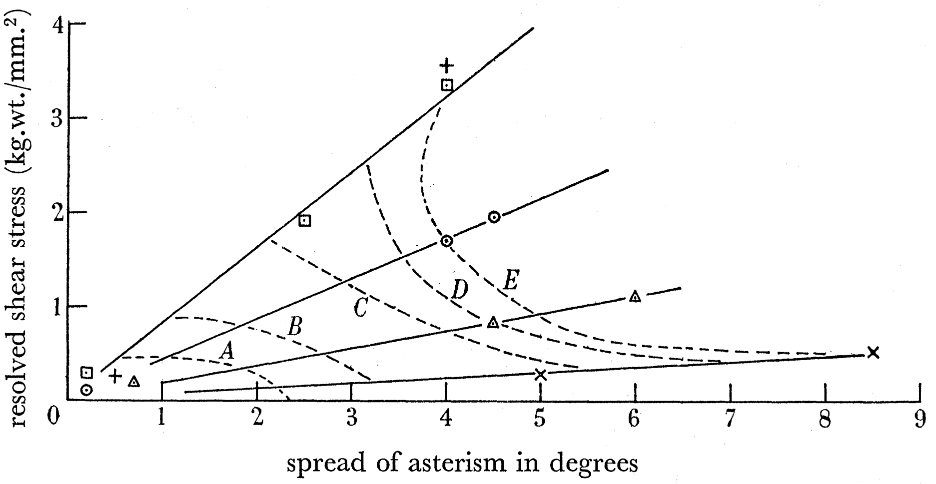

FIGURE 30. Hardness, as measured by resolved shear stress, against spread of asterism, for gold. ⊡ $-80°$ C; + 17° C; ⊙ 330° C; △ 414° C; × 493° C. The broken lines are lines of constant glide: *A*, 5%; *B*, 10%; *C*, 20%; *D*, 30%; *E*, 40%.

It is clear that there is a close connexion between hardening and asterism, but further work is required with other metals before the connexion can be made fully clear. Particularly striking as evidence of the close connexion between hardening, as distinct from glide, and asterism, are the three photographs shown in figure 29, plate 2, which pertain to silver at $-180°$ at glides 0, 35 and 54% respectively, the specimen being that for which the stress-glide curve is shown as 1 in figure 12. At 35% the crystal was still in the region of easy glide and there is hardly any sign of asterism; at 54% distinct hardening has already taken place and asterism is well marked.

It has been contended by Kochendoerfer (1941) that asterism in metal crystals is due entirely to non-homogeneity of the applied stress, this conclusion being the result of X-ray studies of the deformation of naphthalene crystals under tension and under shear. The generalization from naphthalene to metallic crystals would seem to require some justification. Burgers & Lebbink (1945) have justly pointed out that the absence of slip markings on the surface of the deformed naphthalene crystals suggests that the glide planes must be very closely spaced in this substance, so that the distortion on each glide plane would be small and the asterism very small.

DISCUSSION

The samples of the three metals used in these investigations, although all of face-centred cubic structure, have each certain peculiarities, the consequences of which are instructive. The silver and gold are of high purity, containing less than 1 part in 10^5 of foreign metal but differ in that, unless great precautions are taken, the surface of the silver becomes coated with a very thin oxide layer, which exercises a great influence on the mechanical properties, while the gold is free from such complication. The nickel is much less pure, and, in accordance with other experience, the behaviour of the crystals is simpler than that of the very pure metals; the low- and high-temperature stress-glide curves do not cross, because the region of easy glide, which is so marked at low temperature with the pure metals, is absent. Further, at small glides the behaviour of the nickel with temperature is more regular.

Considering first very small glide in the region from 0 to 0·4 %, the most general case would appear to show first a steep, approximately linear rise of stress with glide; then a region of lesser slope and finally a region where the increase of glide with stress is relatively very large. Typical examples, with reference to the case of mercury, have already been given of this behaviour and of the other case, where there is only one abrupt change of slope. This latter is particularly clearly exemplified by nickel, the relatively impure metal.

In one or two instances the crystal was tested for elastic recovery by the removal of load at an early stage, but unfortunately no such test was made in the cases where there is a linear relationship between stress and glide in the earliest stage (e.g. gold at 17° C, figure 5). It has been questioned whether there is an elastic range with single crystals. Greenland (1937), however, found elastic recovery up to an extension of about 0·4 % with his very soft mercury crystals. Miller & Milligan (1937) state that with single crystals of aluminium and silver at room temperature (and, in general, below recrystallization temperature) there is a definite elastic range. The experiments of Chalmers on the microcreep of tin show, with his extremely delicate measurement of deformation, a rate of creep extrapolating to zero with no load. However, if an elastic range is interpreted as a range within which, if the stress is applied for a short time and then removed, there is complete, or nearly complete, recovery, then Chalmers's experiments establish such a range with single crystals of pure tin, for two different methods gave a mainly linear stress-strain curve for 'zero time' up to a strain of $1{\cdot}5\times10^{-5}$. Various workers have found values of the elastic coefficients s_{11}, s_{12}, s_{44} (usual notation—see, for example, Schmid & Boas, *Kristallplastizitaet*) for crystals of different cubic metals by both dynamic and static methods.* These enable Young's modulus E to be found with a rod of any crystal orientation, from the formula

$$\frac{1}{E}=s_{11}-2\{(s_{11}-s_{12})-\tfrac{1}{2}s_{44}\}\,(\gamma_1^2\gamma_2^2+\gamma_2^2\gamma_3^2+\gamma_3^2\gamma_1^2),$$

where γ_1, γ_2, γ_3 are the direction cosines of the wire axis with respect to the three cubic axes. The initial straight-line part of our load-extension curves yields for gold and silver at the temperatures given the following rough values of E: gold (17° C), $1{\cdot}3\times10^{11}$; silver (530° C), $1{\cdot}3\times10^{11}$; while, taking the values for s_{11}, s_{22}, s_{44} found by Röhl (1933) by a dynamical method, we calculate with the angles pertaining to our experiments: $7{\cdot}4\times10^{11}$,

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* For references see Schmid & Boas, *Kristallplastizitaet*, p. 338 (*Bestimmung elastischer Parameter von Kristallen*), and further, for example, R. H. V. M. Dawton, 1938, *Proc. Phys. Soc.* **50**, 483 (Na); S. L. Quimby & S. Siegel, 1938, *Phys. Rev.* **54**, 293 (Na); O. Bender, 1939, *Ann. Phys.* **34**, 359 (Na, K).

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$3{\cdot}1\times10^{11}$ respectively, the units being dynes/cm.^{2}.* Hence the values of E given by assuming that our initial straight lines represent elastic deformation are of the right order, but somewhere about, on the average, a third of those calculated from constants found dynamically.

Masing (1948) has pointed out that, on G. I. Taylor's theory of rows of positive and negative dislocations acting on one another, there will be, in addition to the ordinary yield of the perfect lattice, a yield due to the interaction of the dislocations, since the restoring force due to the dislocations does not reach the maximum, succeeded by plastic deformation, until a certain shear strain has been exceeded. This elastic deformation has to be added to the elastic yield of the perfect lattice, so that the effective elastic coefficient $\gamma_{\text{eff.}}$ is given by

$$\gamma_{\text{eff.}}=\gamma_{\text{r.}}+\gamma_{\text{el.}}$$

where $\gamma_{\text{r.}}$ pertains to reversible glide of dislocations past one another and γ_{el} is the ordinary elastic shear strain. If l is the distance between successive dislocations in a line, d the perpendicular distance between two lines of dislocations of like sign and λ the atomic step caused by the passage of one dislocation, and if each dislocation moves a small distance ξ, then the reversible strain is

$$\gamma_{\text{r.}}=\frac{\lambda\xi}{ld},$$

and the stress

$$\tau=\frac{G\lambda}{d}\,F(\xi),$$

where G is the shear modulus and

$$F(\xi)=-\frac{2\xi}{l}+\sum_{m=-\infty}^{m=+\infty}\left\{\tanh\frac{\pi ml}{d}-\tanh\frac{\pi}{d}\,(ml-\xi)\right\}.$$

Masing gives the variation of F with ξ/l, which measures the strain, for various values of l/d, from which it is possible to prepare curves showing τ against $\gamma_{\text{eff.}}$. The curves† in figure 21 are constructed with $G=10^{11}$ dyne/cm.^{2}., $\lambda=10^{-8}$ cm., $d=10^{-4}$ cm. If $G=a\times10^{11}$ dynes/cm.^{2}, $\lambda=b\times10^{-8}$ cm., $d=c\times10^{-4}$ cm., then $\tau=ab/c\,10^7F$ and the values of τ shown in the diagram must be multiplied by ab/c, which will be of the order 1. However, what is now in question is the ratio of the slope of the line A, which corresponds to the perfect lattice, and of curves B and C, which represent the effect of the addition $\gamma_{\text{r.}}$, and this is independent of the multiplying factor. The purpose of this brief discussion is not to obtain quantitative results, for which the data are insufficient, but to indicate that the elastic interaction of dislocations will give an effective value of E which is lower than that for the perfect lattice, the exact value depending upon the nature and distribution of imperfections in the crystal, as measured by l/d. The assumptions on which the theory is based are also clearly too artificial to correspond very closely to reality.

The behaviour for small glides should be, then, very dependent upon the density and distribution of dislocations in the specimen, which accounts for the fact that it differs so much from specimen to specimen, especially with very pure metals. Experiment leaves

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* For the calculated values at the high temperature, it has been assumed that the single-crystal value of E varies with temperature to the same extent as the polycrystalline value, which must be, in any case, a good approximation.

† These curves were kindly prepared for us from the data by Mr D. A. Aboav.

no doubt that minor structural irregularities in metal single crystals very much affect the very early stages of deformation. For example, Dehlinger (1933), Dehlinger & Gisen (1934) and Gisen (1935) have investigated aluminium single crystals, of various degrees of purity, made in two methods, (1) by cooling from the melt, the method used by us, (2) by recrystallizing in the solid state, with strain and subsequent heating. They found that sharp yield occurred with those made by the second method, but not with those made by the first method, which they attributed to a more pronounced mosaic structure in the second case. Gisen, in particular, found the difference of behaviour in the two cases well marked with aluminium of 99·998 % purity and supported by X-ray intensity measurements the assumption of more marked mosaic structure in the crystals grown in the solid state. The deformation in all these experiments was measured by the diminution of diameter, which is very insensitive compared to our method, the scale of glide in the curves being about $\frac{1}{50}$th that in ours, so that detailed comparison is not possible. This work does, however, indicate that variations in the small departures from perfection which all single crystals show may reasonably be assumed to account for the different types of initial behaviour which our curves display. An increase of slope shown by, for instance, the curve for silver at 350° C, only occurs with this one metal and may be due to the surface effects to which, as we have shown, it is particularly susceptible.

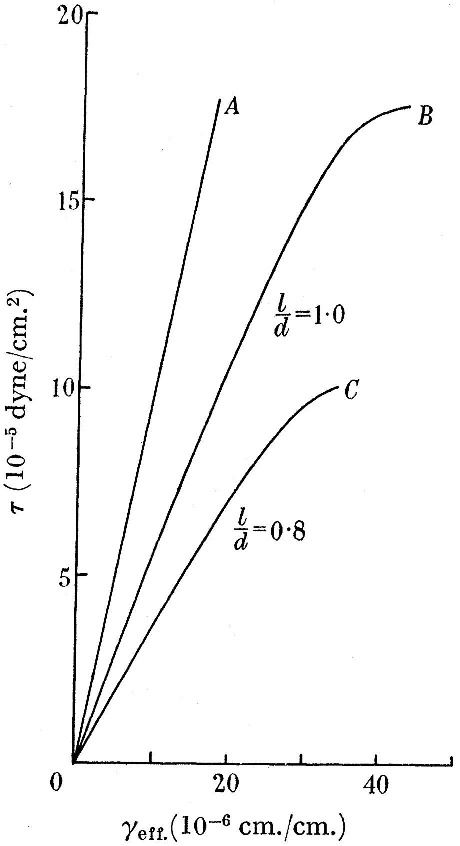

FIGURE 31. The effect of interaction between dislocations on elastic yield: stress-strain curve for (*A*) perfect crystal, (*B*) crystal containing dislocations with $l/d=1\cdot0$, (*C*) crystal containing dislocations with $l/d=0\cdot8$, where l = distance between successive line dislocations in a plane, d = distance between two planes of dislocations of like sign.

Our general conclusion, then, is that with pure metals elastic deformation may occur within regions of glide not exceeding 0·05 % or may not, the deciding factor being minor structural features of a metastable nature. For instance, it may be supposed that glide tends to start at dislocations located on the surface of separation between the planes of the 'block structure' revealed by our thermal etching and by the experiments of other workers to which reference has been made, these surfaces being separated by about 1 μ. If, under low stress, a dislocation runs right across the specimen on one of these planes, minor considerations will determine whether it arrives at the boundary with a high or a low velocity; with a low velocity it may stop, with an advance of but one atomic spacing between the neighbouring planes; with a high velocity there will be reflexion, resulting in a step of several atomic planes and, as the result of waves sent out from the surface disturbance, the initiation of glide on other planes. It may be noted that glide of one atomic spacing on planes separated by 1 μ would lead to a shear strain of about 0·03 %, which is the order of glide within which the behaviour of pure metals, as exemplified by gold, is irregular.* It is possible that the second stage in the most general case, which shows the two sharp bends in the stress-glide curve, represents the initiation of glide by a relative advance of an atomic spacing or so on all the glide planes which are liable to develop ultimately into visible glide lamellae.

The initiatory stage or stages being complete at something like 0·2 % glide, there are two alternative methods of development of the stress/glide relationship. In the one case, exemplified by all the nickel curves and by, for instance, gold at 330 and 414° C and silver at 530° C the slope never increases markedly with increasing glide; there may be a long approximately linear stretch, as with nickel at 17° C, followed by a region in which the slope diminishes, or the slope may continually diminish.† In the other case, exemplified by all the low-temperature silver curves and by the low-temperature gold curves, especially that for −180° C, there is a region where the rate of increase of glide with stress is large, followed by a region of hardening, where this rate diminishes. The region of large rate of increase will be called here *region of easy glide*: this region of easy glide appears to be absent with aluminium at atmospheric temperature, the only likely temperature for which results are available (Karnop & Sachs 1927; Taylor 1927; Miller & Milligan 1937). Both our nickel and the aluminium in question are relatively impure. For copper and nickel, containing impurities of the order of 0·1 %, Osswald's measurements, while not precise at low glide, show no region of easy glide. With a copper-nickel alloy containing about equal parts of the two metals, however, Osswald's curves show an initial region of about 10 % where the glide was very easy and Sachs & Weerts's curves show a rather larger region of easy glide for a silver-gold alloy, these experiments being at room temperature. No previous attention seems, however, to have been directed to this feature.

The following generalities, then, seem indicated for easy glide. It takes place for very pure metals, being hindered by metallic impurities of the order of 0·1 % and also by surface contamination, for, although normal silver at 17° C shows a small region of easy glide,

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* Heidenreich & Shockley (1948) interpret their electron microscope photographs as indicating a *maximum* relative displacement of adjacent laminae in aluminium of 2000 Å. This, however, is after considerable strain has taken place, and gives no indication of the minimum displacement.

† Only in the case of nickel at −180° C is there an approximation, and that very rough, to the parabolic form demanded by Taylor's theory (1934*a*).

this region is very much greater for silver with a really clean surface (see figure 19), an effect which will be discussed later. It is hindered by high temperatures. It can take place for solid solutions when the metals are in roughly equal proportions. The easy glide must be associated with the development of the elementary glide planes formed in the first stage of glide, probably both by further glide on the elementary planes, and, more particularly, by the creation of packets of very close associated glide planes, of the kind demonstrated by Heidenreich & Shockley (1948). That atoms of impurity tend to stop the free run of dislocations accords well with accepted views. The suppressing effect of rise of temperature on the region of easy glide is more remarkable. It may be explained either by supposing that temperature agitation itself, by removing certain atoms from the regularity of potential alternations postulated when the run of dislocations is discussed, may hold up the run, or by supposing that the very few foreign atoms that exist in the purest metal migrate to dislocations and that this migration is accelerated by high temperature, in the manner discussed by Cottrell (1948).

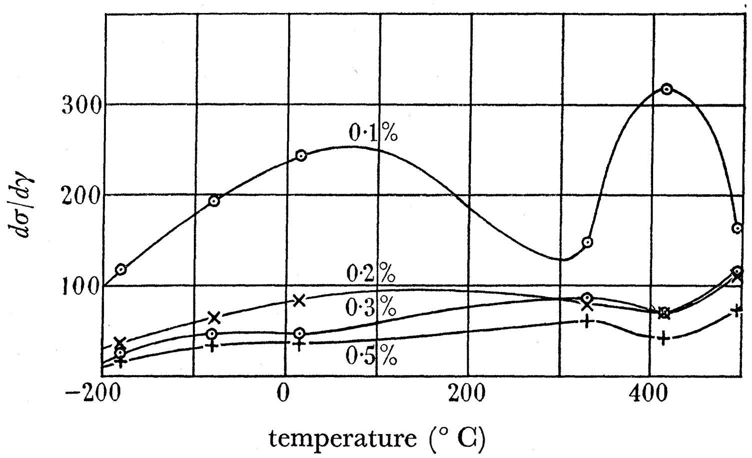

FIGURE 32. Variation of $d\sigma/d\gamma$ at constant glide with temperature, for small glides, up to 0·5 %. Gold. σ in g.wt./mm.^{2}, γ in per cent.

Whether easy glide takes place before it or not, there is finally a large region of glide in which marked hardening takes place, a hardening which becomes less marked towards the end of the glide curves, say after 50 % glide.

To make clear the existence of the physically different stages in the glide of these cubic crystals we take the results for gold, since with this metal the complications of surface contamination and marked impurity are not present. We adopt a new way of exhibiting the glide properties, namely, by plotting the rate of increase of resolved shear stress with glide $d\sigma/d\gamma$ (derived from the stress-glide curves), at constant glide, against temperature. Thus what is expressed is, in general, the effect of temperature on strain hardening at various stages of glide. The results are expressed in three diagrams: figure 32 for small constant glides, 0·1 to 0·5 %; figure 33 for medium glides, 1 to 10 %; figure 34 for large glides, 20 to 50 %. It is to be noted that the scale in figure 32 is very much reduced compared to that in the other two diagrams, since the hardening in the early stages is so large.

The general conclusions are that in the early stages of glide, up to about 0·5 %, the variation is irregular, which agrees with our general conclusions that at this stage the variable crystal imperfections have a very large effect, finding expression in the different types of behaviour to which attention has been drawn. From 1 to 5 % glide the curves show clearly, at low temperatures, a rise of hardening with temperature, succeeded, after a flattish maximum, by a fall at high temperature, behaviour which can be explained by two effects of temperature, one to hinder glide and the other to promote glide, the first effect being predominant at lower temperatures and the second predominant at higher temperatures. For large glide, 20 to 50 %, the hardening falls steadily with temperature.

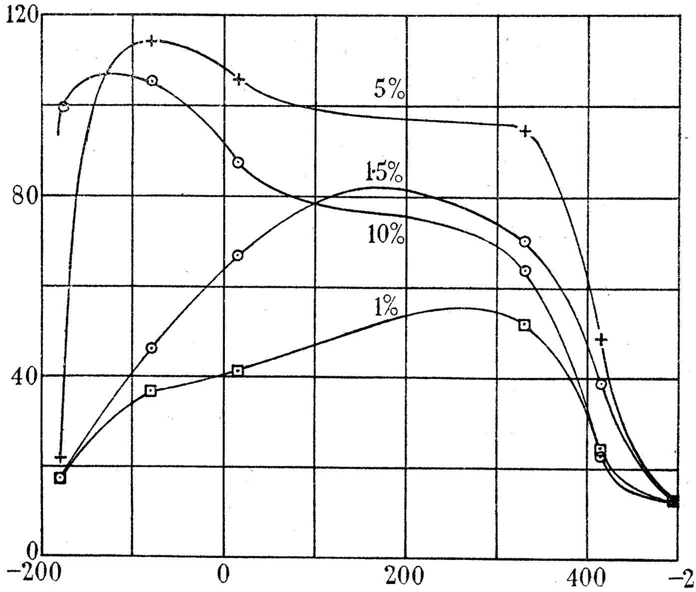

FIGURE 33. Variation of $d\sigma/d\gamma$ at constant glide with temperature, for medium glides, 1 to 10 %. Gold.

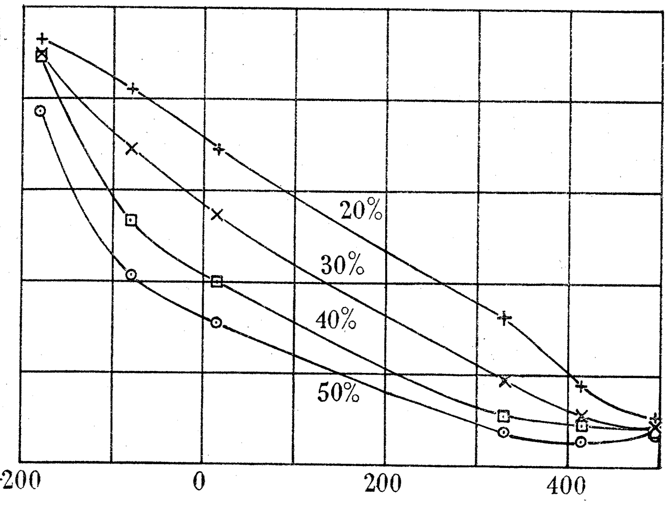

FIGURE 34. Variation of $d\sigma/d\gamma$ at constant glide with temperature, for large glides, 20 to 50 %. Gold.

A general explanation is that temperature agitation tends to hinder the initiation of glide, but that, the glide once developed on a single glide plane, temperature agitation promotes the development of glide on neighbouring close planes, of the type shown in the photographs of laminar slip bands given by Heidenreich and Shockley. Thus once all the pilot slip bands have become active, which is the case at large glides, temperature agitation promotes further glide in their neighbourhood, but the first development of the pilot bands is hindered by temperature agitation. Some such general picture will account for the region of easy glide shown by pure metals at low temperature and for the results expressed by the curves of figures 32 to 34. It also accords with the wide separation between glide surfaces which is evident at temperatures from 1000 to 2000° C with molybdenum (Tsien & Chow 1937; Andrade & Chow 1940), and, in general, at high temperatures. At these temperatures glide proceeds readily in the neighbourhood of the few planes that are initiated, but glide on new planes is not easily initiated.

Experiments, including some in which glide in gold is to be initiated at low temperatures and continued at high temperatures are to be carried out shortly at the Royal Institution, with the object of confirming the views on glide here put forward.

Curves prepared for nickel show, except in the first 0·5 % of glide, a steady decrease of $d\sigma/dy$ with temperature, corresponding to the absence of the region of easy glide with the relatively impure metal. The silver curves are too much complicated by the surface effect to offer evidence on this point.

The effect of the very light surface contamination consequent on exposing a clean silver surface to air at atmospheric temperature presents interesting features. As already remarked when the results shown in figure 19 were described, while the surface layer leads to an increase of the critical shear stress by some 30 %, the increase of the resolved shear stress at 20 % glide is some ten times. This remarkable result we attribute to a diffusion of the foreign atoms—possibly oxygen ions—into the interior of the metal when deformation takes place. The diffusion of oxygen into static silver at atmospheric temperatures is insignificant, but it may well be that the propagation of dislocations from the surface along the glide planes carries with it foreign atoms, which migrate to the locality where dislocation is gravest and effectively stop further propagation. This again is a point that it is hoped to investigate further at the Royal Institution. Roscoe (1936) in his experiments on the effect of surface oxidation on cadmium single crystals found the increase of strength at zero glide to be somewhat greater than at 300 % glide, but the general behaviour of hexagonal and face-centred metals differs so markedly in many respects that this difference is not surprising. For one thing, while for silver the glide remained constant at a given stress, with cadmium a standard rate of glide (0·5 % per second) was adopted in finding the stress at a given glide.

One general lesson of this work is that gold is a metal particularly suited for work designed to elucidate the mechanical behaviour under the simplest possible conditions. Probably platinum is equally apt for this purpose.

We desire to express our best thanks to Mr Ronald King, Assistant Director of the Davy Faraday laboratory, for much helpful discussion. The gratitude of one of us (C.H.) is due to the Department of Scientific and Industrial Research for a maintenance grant during the period of this investigation.

Description of plate 1

Figure 4. Back reflexion Laue picture for single crystal of gold.
Figure 8. Gold single crystal, heavily strained: 69 % glide at 493° C (magn. × 15).
Figure 9. Gold single crystal, heavily strained, showing triple glide (magn. × 1800).
Figure 10. Gold single crystal, unstrained, showing fundamental laminar structure (magn. × 1800).
Figure 20. Thermal etching of unstrained silver single crystal at about 900° C (magn. × 100).
Figure 21. Glide lamellae in a strained silver crystal (magn. × 1800). 90 % glide at 25 ° C.
Figure 22. Nickel crystals after heavy straining: (*a*) 65 % glide, at 20° C (magn. × 15); (*b*) 84 % glide at 20° C (magn. × 1800).
Figure 23. Surface of unstrained nickel crystal (magn. × 1800).
Figure 24. Heavily strained nickel crystal, showing double glide (magn. × 1800).

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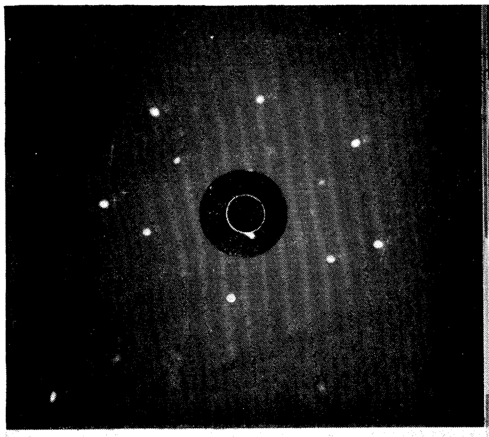

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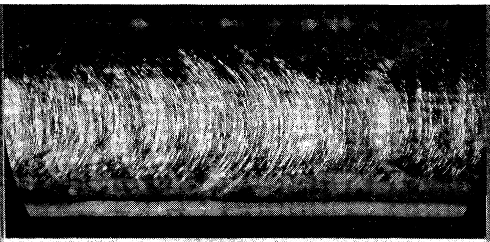

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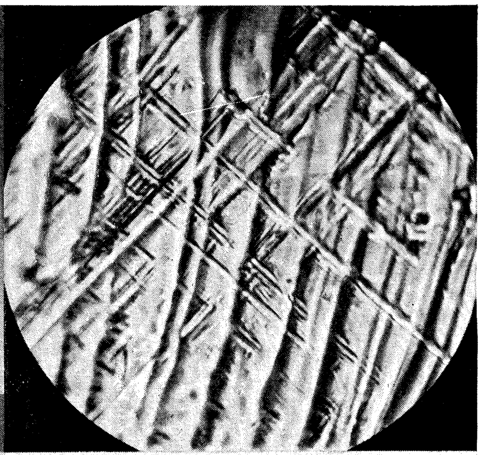

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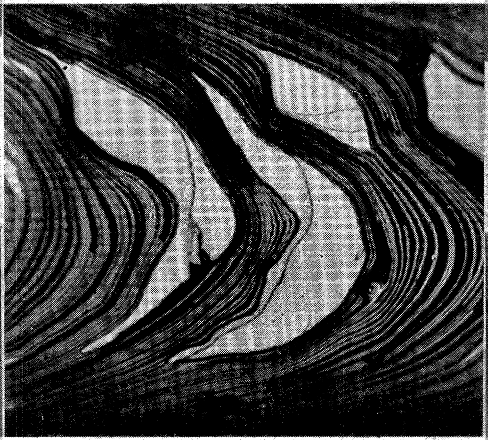

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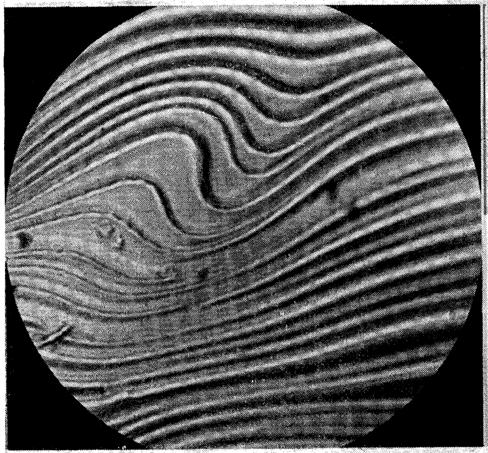

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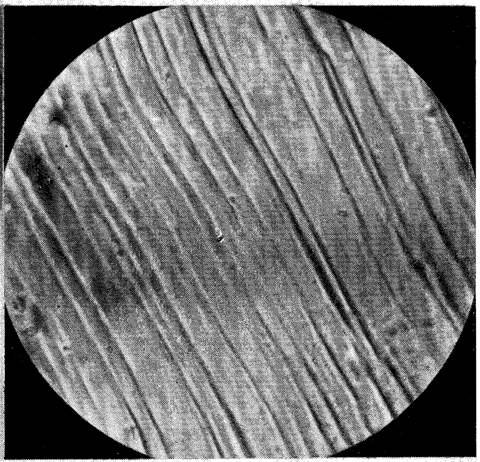

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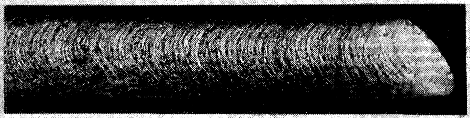
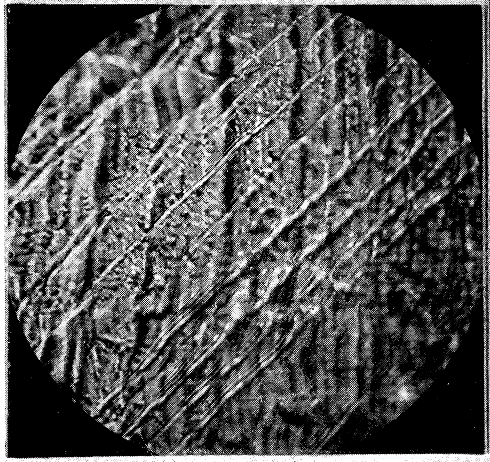


22*a*



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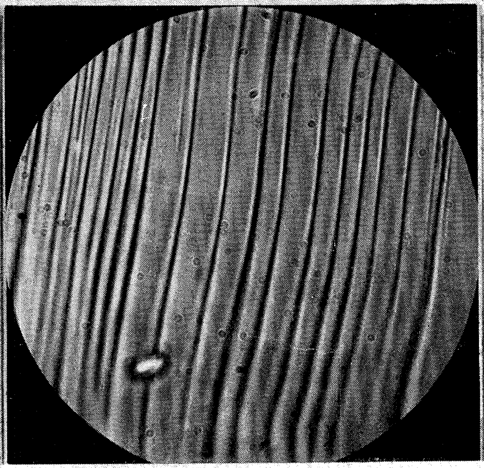

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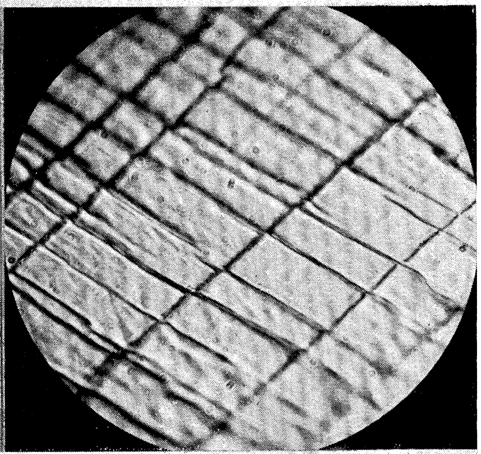

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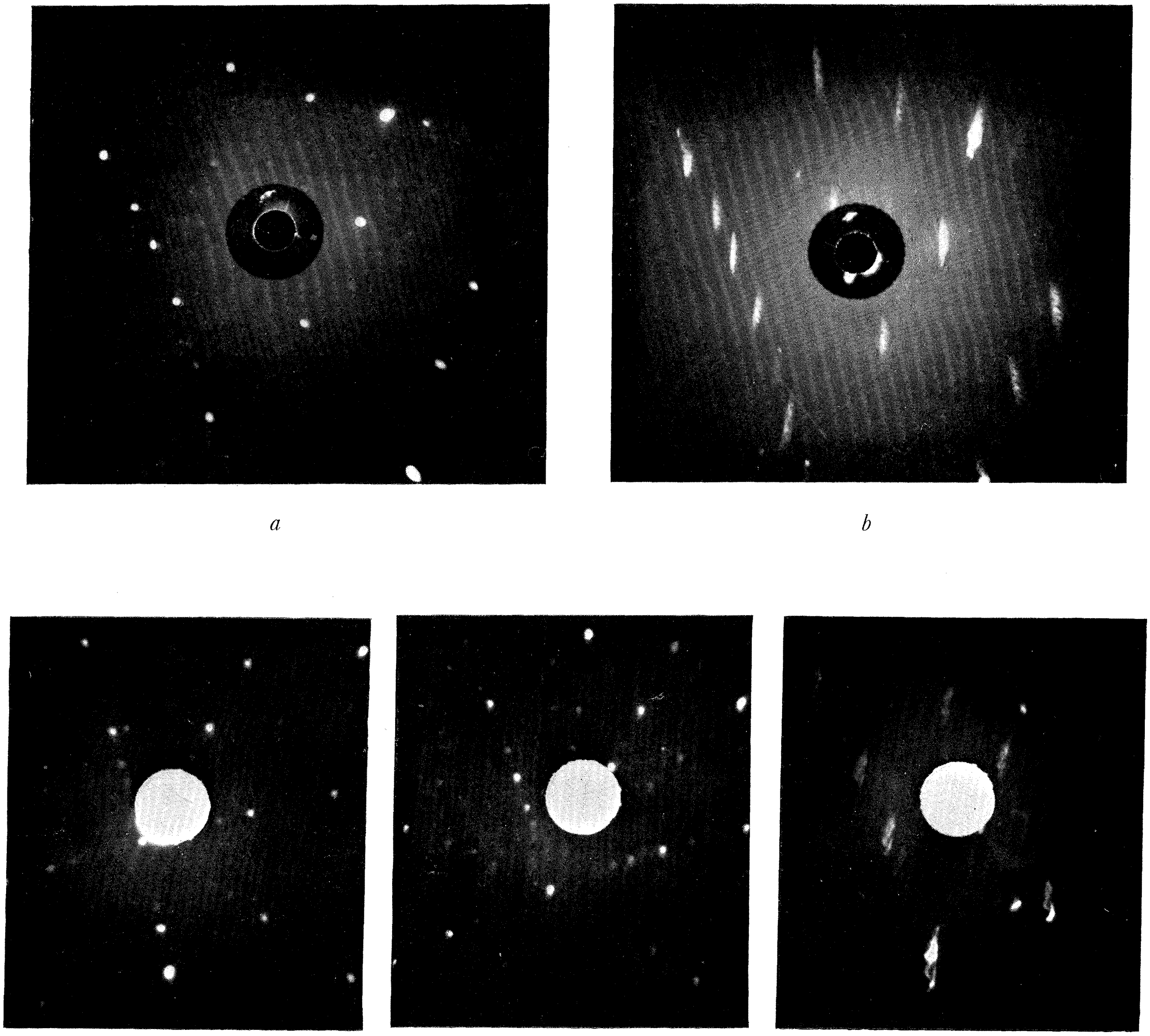

a *b*

c *d* *e*

FIGURE 29. Asterism. Gold at 493° C: (*a*) unstrained, (*b*) glide 13 %. Silver at −180° C: (*c*) unstrained, (*d*) glide 35 %, (*e*) glide 54 %. Compare figure 12, curve 1 of which refers to this crystal.

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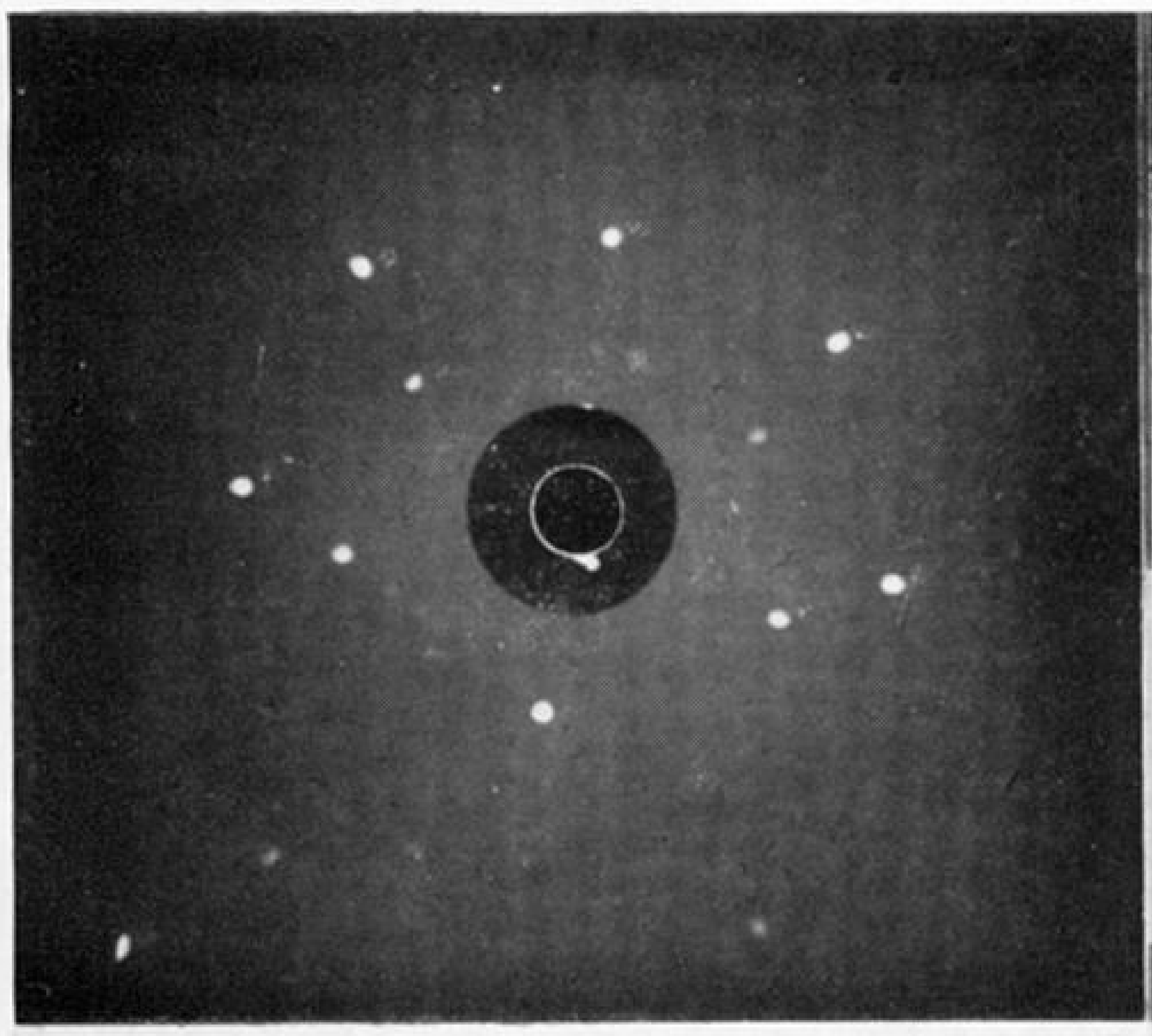

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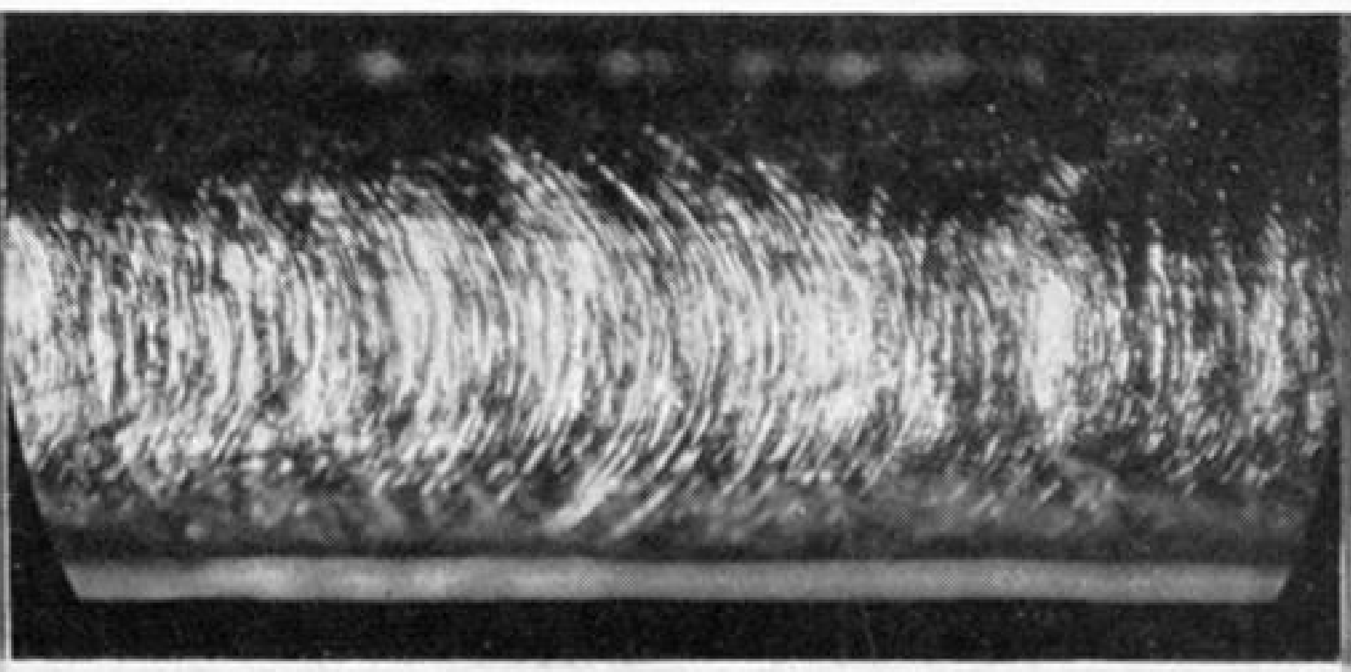

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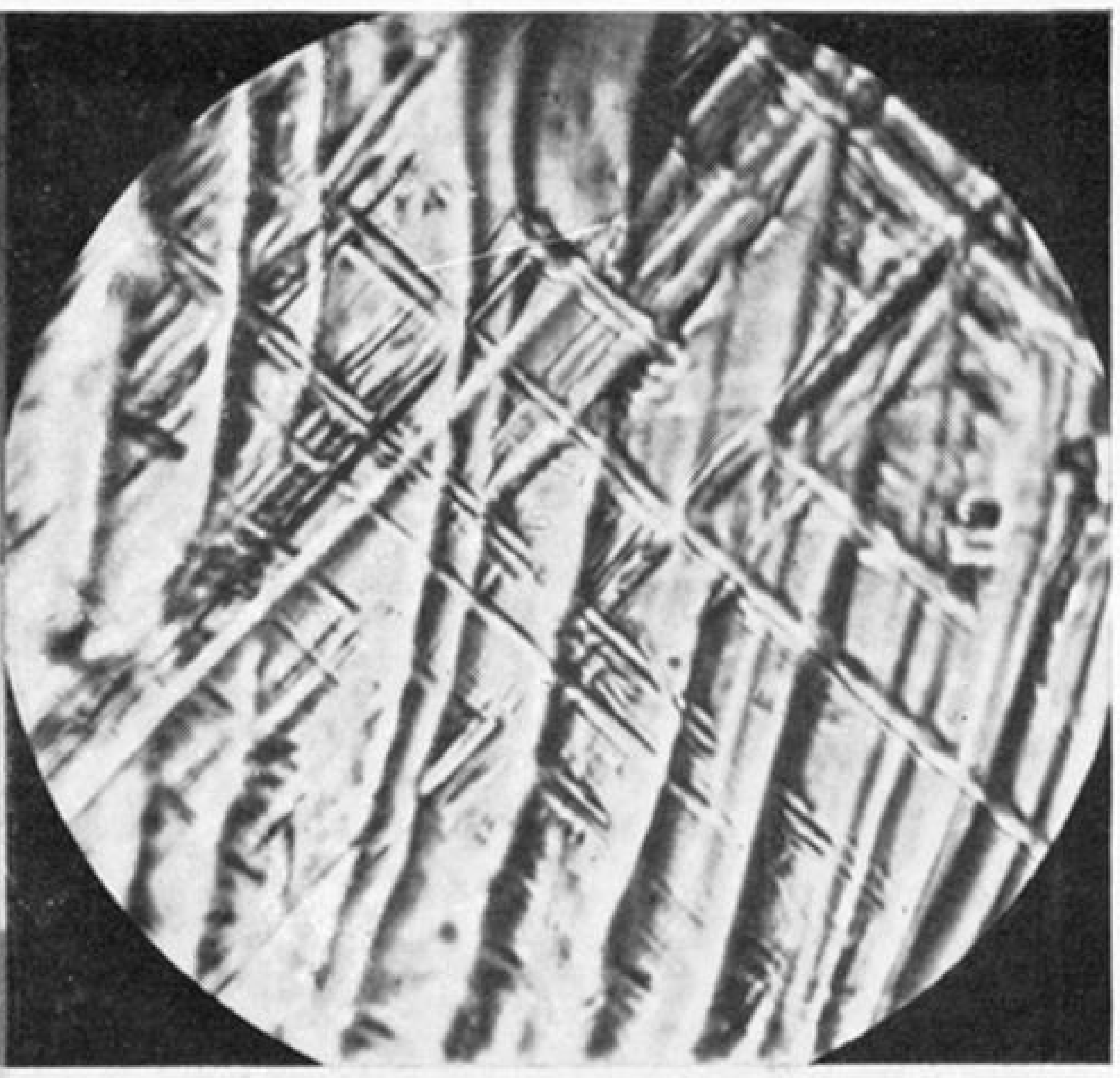

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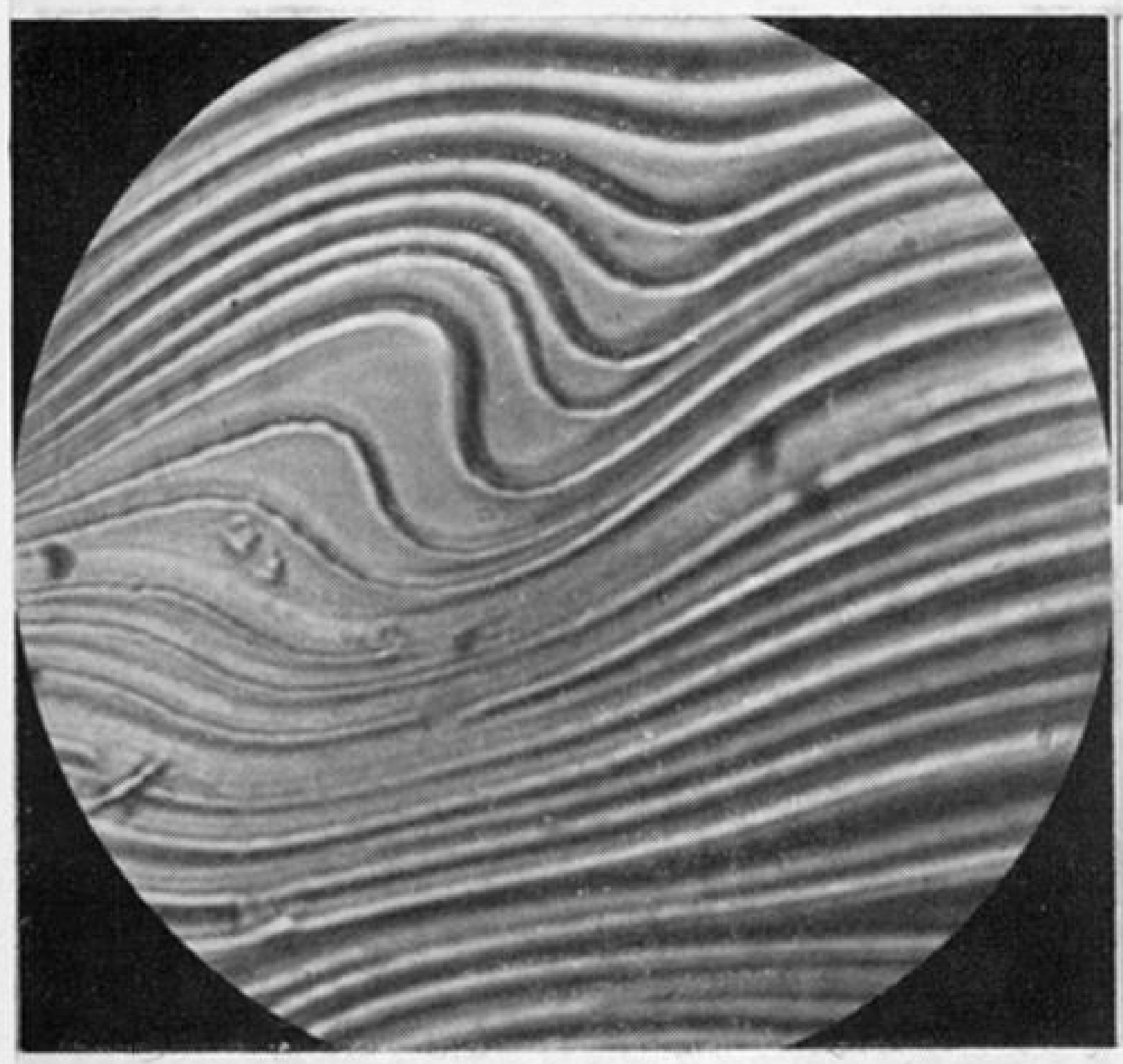

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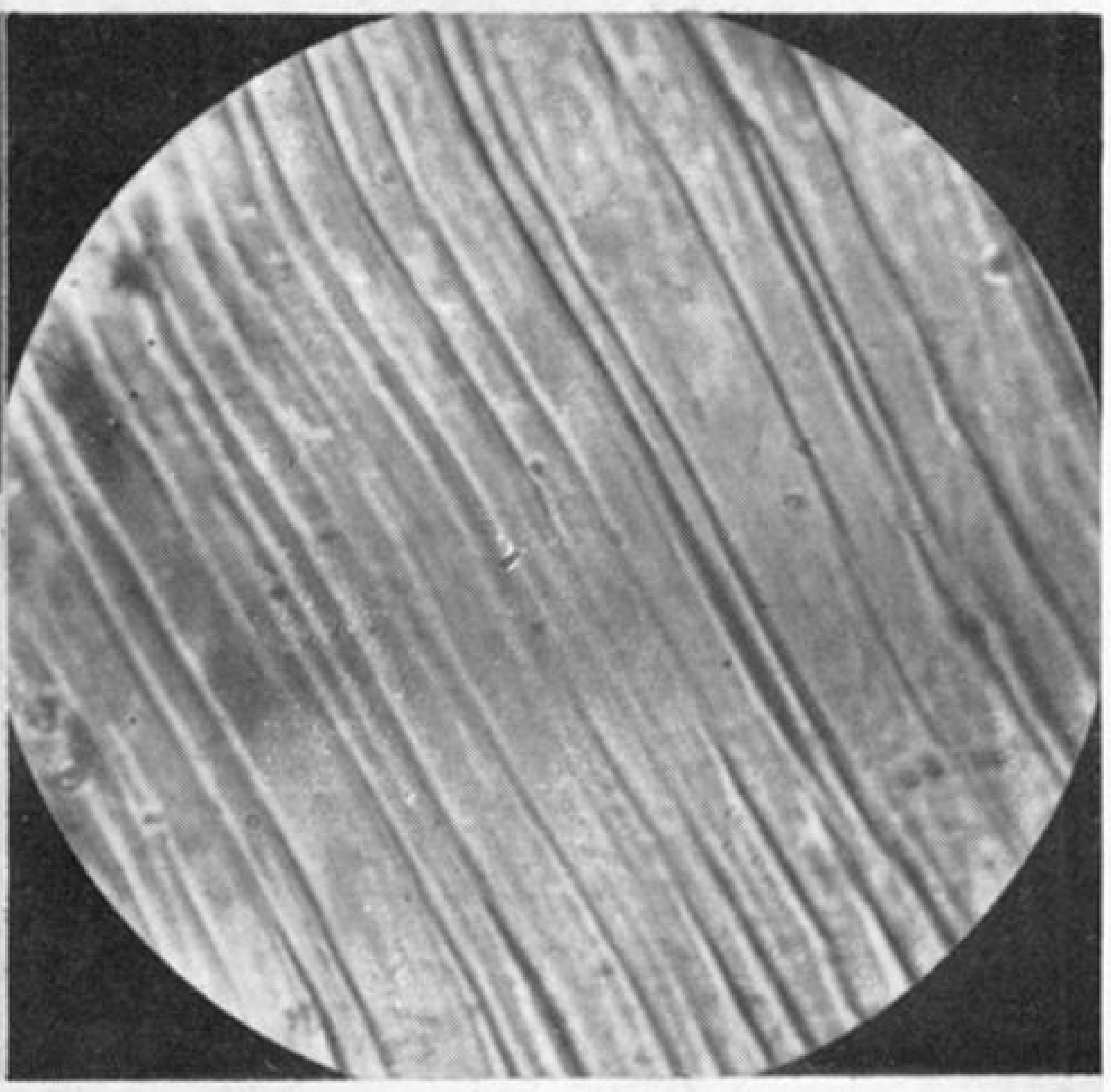

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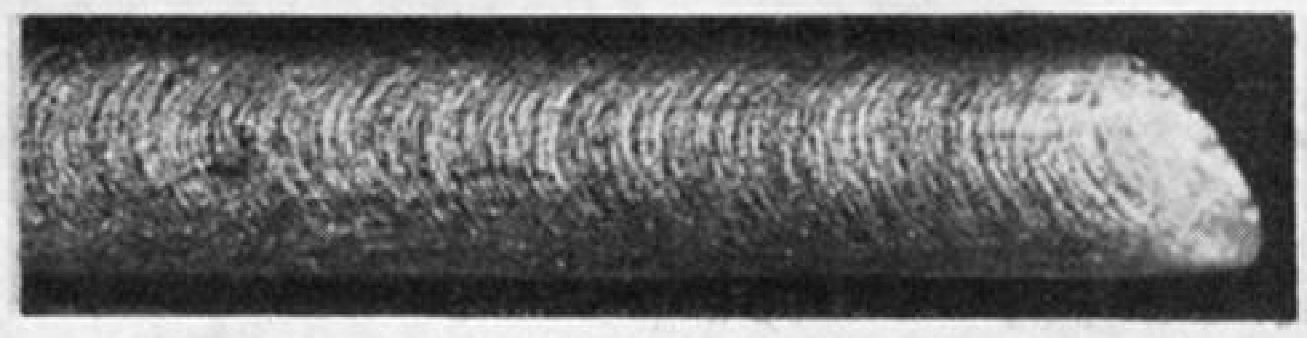

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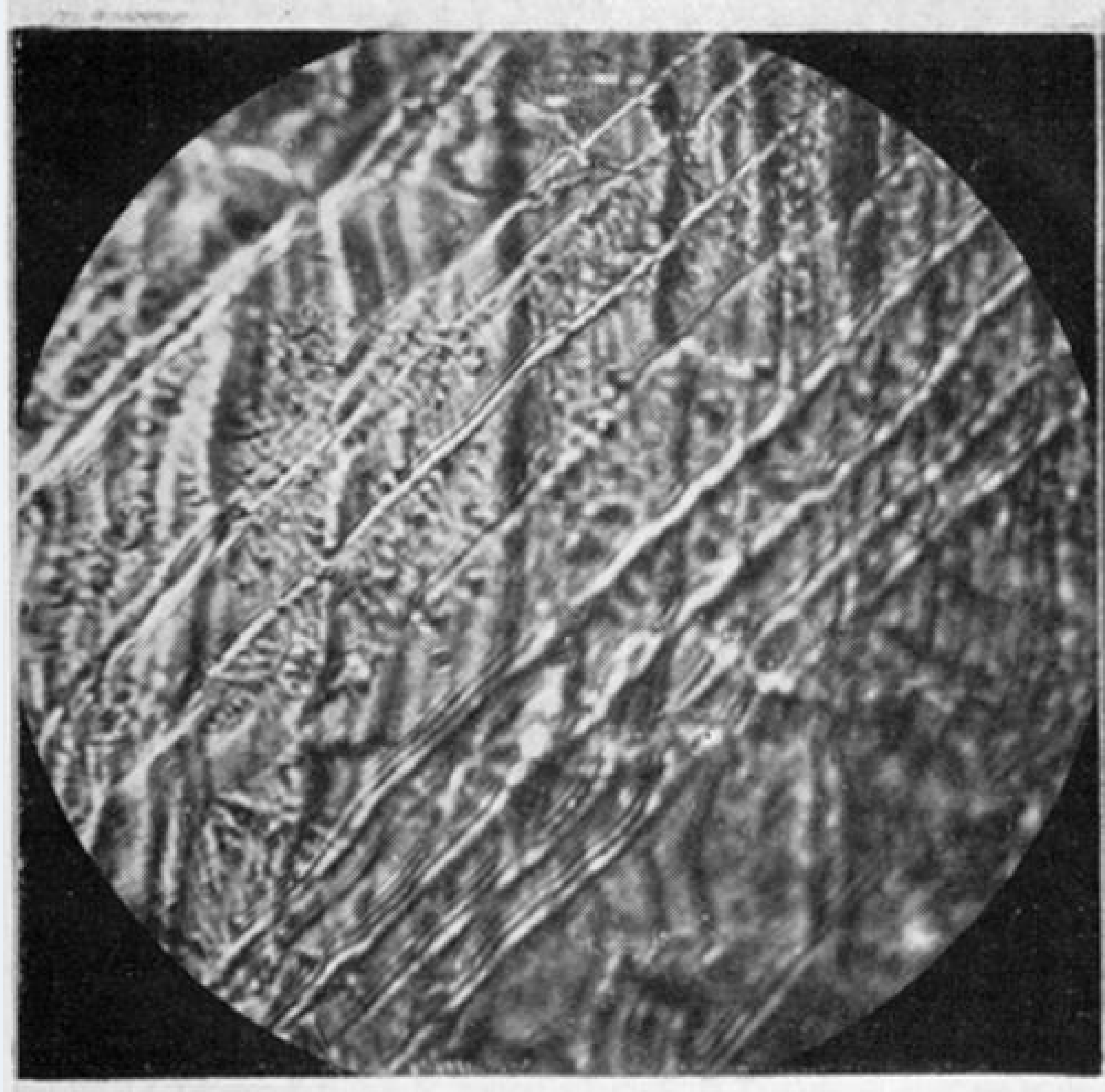

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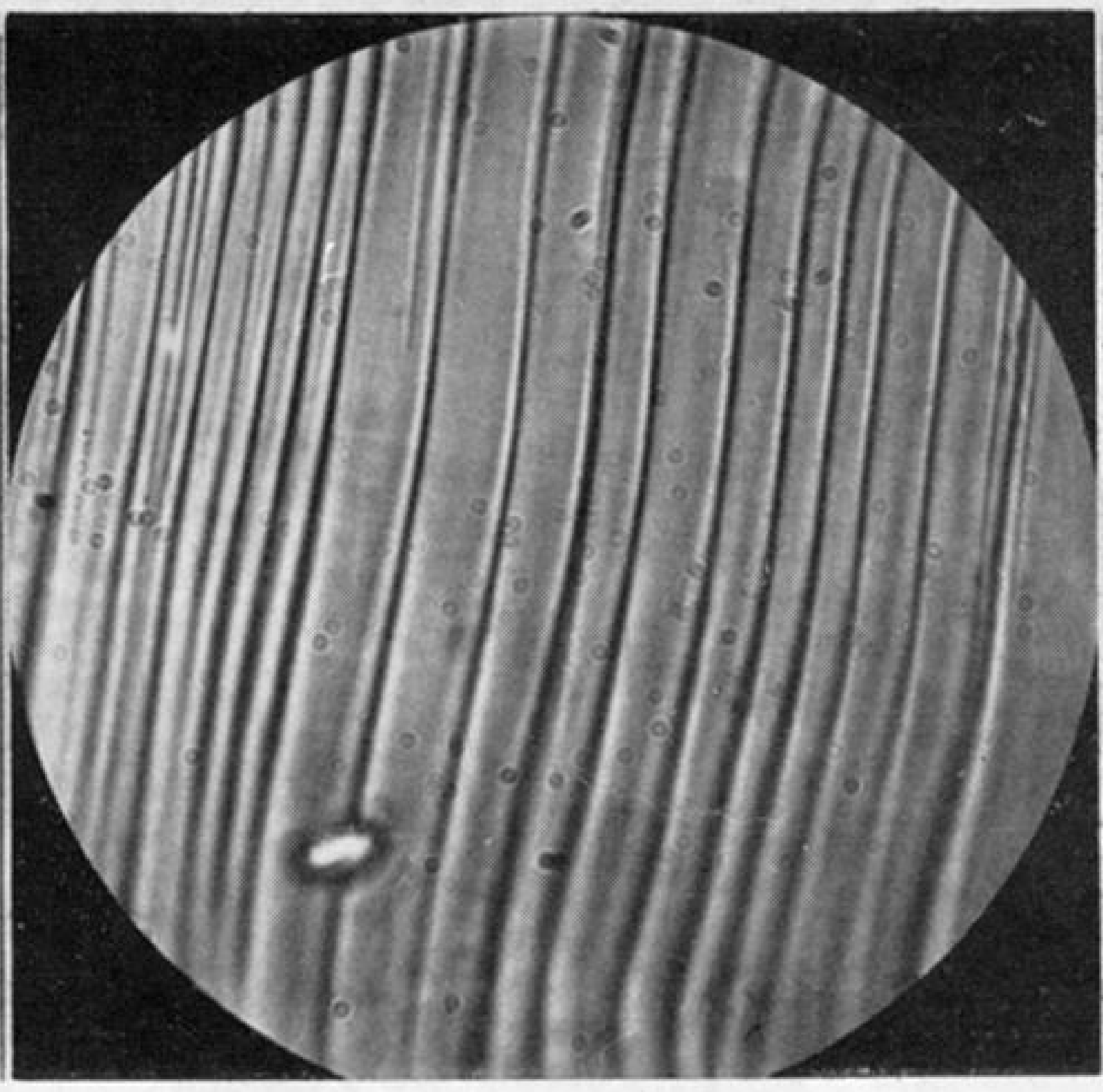

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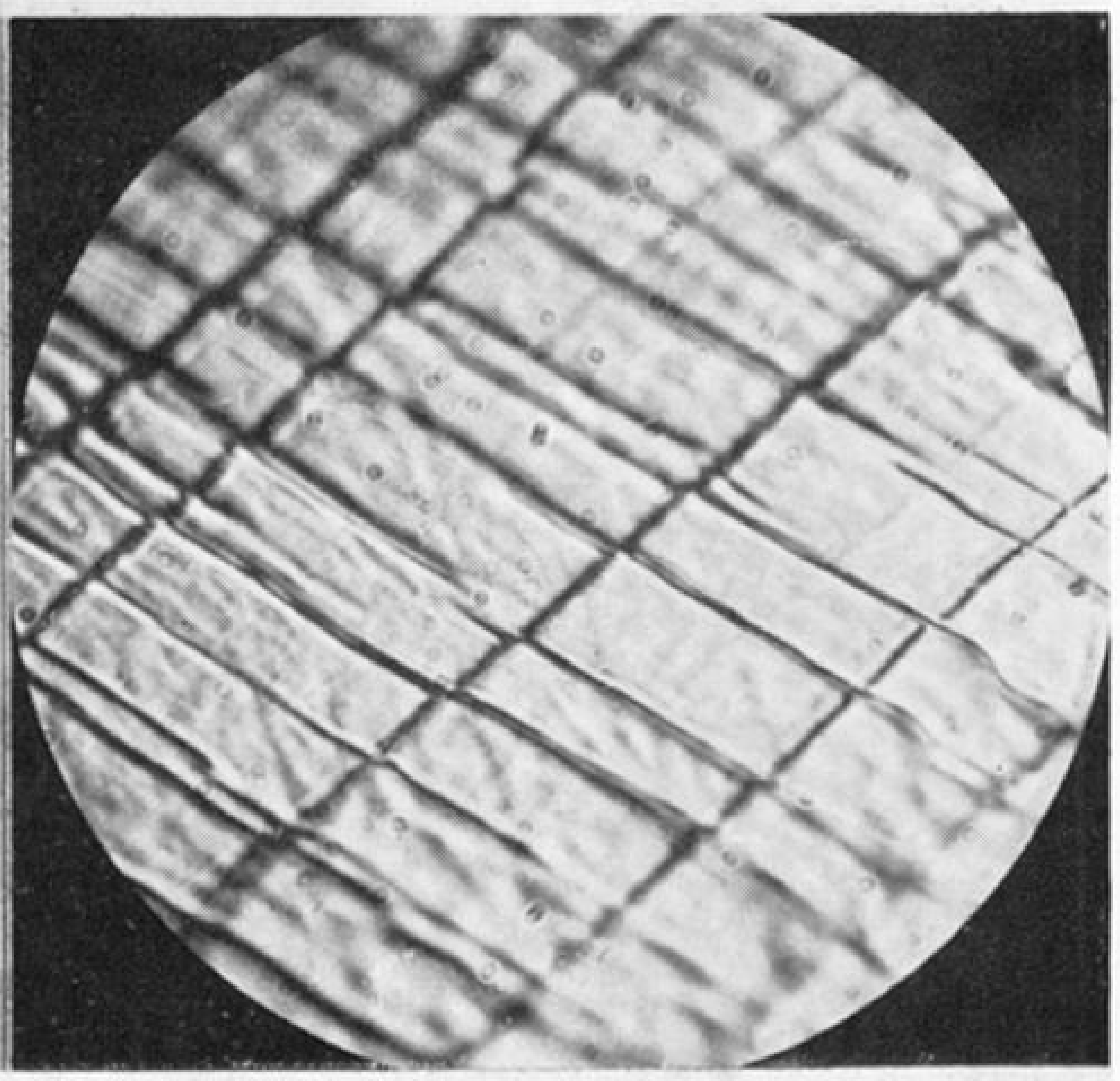

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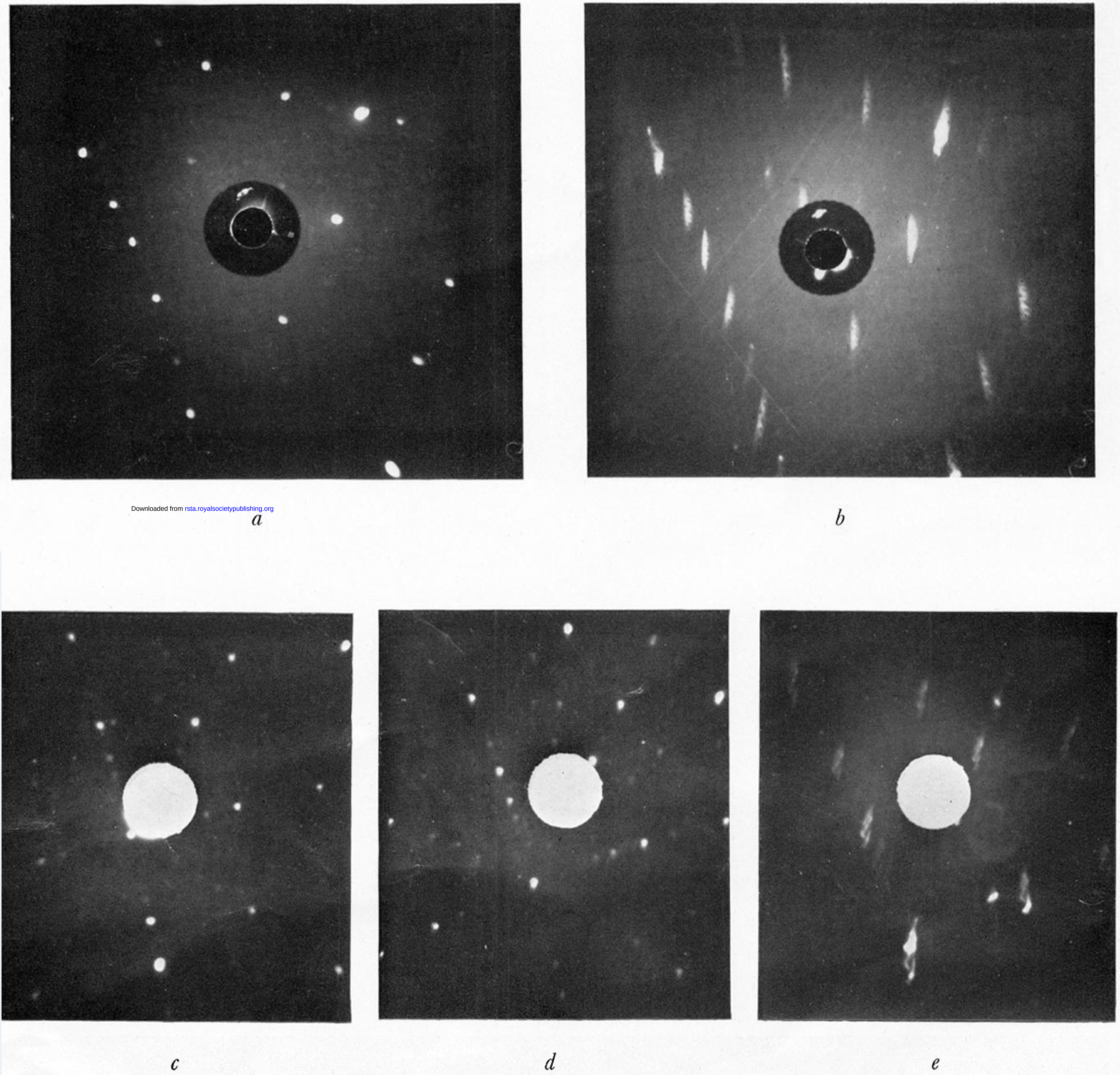

FIGURE 29. Asterism. Gold at 493° C: (*a*) unstrained, (*b*) glide 13 %. Silver at $-180°$ C: (*c*) unstrained, (*d*) glide 35 %, (*e*) glide 54 %. Compare figure 12, curve 1 of which refers to this crystal.