

PHYSICS OF THE STRENGTH OF CRYSTALLINE MATERIALS

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Present-day technology makes greater and greater demands on the strength properties of materials. A physical theory of strength has not yet been developed, although the last 30 or 40 years have seen the development of a new branch of molecular physics which is concerned with the study of the deformation and fracture of solid materials. In this review we examine present physical ideas about the strength of crystalline materials, the reasons for the low strength of real materials, and also the main ways of increasing their strength.

1. THE MICROSCOPIC THEORY OF STRENGTH

Estimates based on the interatomic interaction forces in crystals lead to high values of the strength. Various methods of calculation¹⁻⁷ give for the value of the theoretical strength of crystals expressions of the type

$$\sigma_T = K \sqrt{\frac{ES}{a}} \quad (1)$$

where K is a coefficient of the order of unity, E is the modulus of elasticity, S is the coefficient of surface tension, and a is the crystal lattice parameter. The actual strength is smaller by a factor of 100 to 1000.

An analogous discrepancy exists between the practical flow limit and the theoretical resistance to shear. J. Frenkel⁸ showed that the critical shearing stress at which a lattice becomes mechanically un-

stable is $G/2\pi$, where G is the shear modulus; this is some orders of magnitude larger than values obtained in tests on plastic crystals ($10^{-5}G$). Frenkel's theory is based on the idea that for the resistance to shear to be overcome an atomic plane must be displaced by half the distance between two positions of stable equilibrium. It follows from this that the resistance to shear must be of the order of magnitude of the shear modulus G . More detailed calculations made later by a number of authors⁹⁻¹² who took into account the possibility of the existence of stable intermediate positions of the lattice led to somewhat smaller values of the limiting resistance, about $G/30$, which is still incomparably larger than the shear strength of actual metal monocrystals.

A sharp discrepancy between theory and experiment is also found in brittle solids, for which, as is well known, Hooke's law holds right up to fracture, whereas near the theoretical strength there should be a marked departure from proportionality between stress and strain.

Two approaches are used in the search for the causes of these differences. The first is based on the hypothesis that low strength is a characteristic property of an ideal crystal, and that the discrepancy between theory and experiment is a consequence of inadequacy of the existing theory of the ideal crystal. Doubts are raised as to the legitimacy of extrapolating the elastic properties of a crystal to large displace-

ments of the atoms, and as to the constancy of the elastic modulus during the process of deformation. In analogy with the case of fusion, for example, the comparative ease with which slip occurs in crystals could be due to a breaking of bonds of long-range order only, with the bonds of short-range order remaining almost unchanged.^{13,14,191} Therefore the demand that in rupture the stress must reach the order of magnitude of G might actually be excessive. Most workers, however, see the cause of the serious contradiction between theory and experiment not in inadequacies of the atomic theory of the crystal, which has given the correct solution of the problem of the strength of an ideal crystal lattice, but in the fact that an actual crystal does not correspond to the ideal concept on which the theory is based. The difference between the theoretical and actual strengths arises because in the calculation of the theoretical strength, determined by the interatomic binding forces, one assumes a simultaneous fracture over some surface that crosses the entire specimen, whereas in a real crystal simultaneous resistance to the external forces is offered by only a small fraction of the atoms adjacent to such a surface. Both in the case of brittle fracture and in that of plastic shear such behavior of the real crystal is due to the effects of defects of various kinds in the crystal structure and the redistribution of the stresses near such defects. It is assumed that the breaking of the atomic bonds in the lattice occurs at local stresses corresponding to the theoretical strength, while the average stress over a cross section can be much smaller.

Thus the contradiction between theory and experiment can be overcome by renouncing the ideal model of the crystal lattice and studying the real crystal with its many defects of structure which are formed both in the process of crystallization and also during the deformation of the crystal.

A strict quantitative theory of the strength of solid materials on the basis of ideas about real crystals has not yet been developed, but the available experimental information is in satisfactory agreement with such ideas.

2. DEFECTS IN THE STRUCTURE OF A REAL CRYSTAL

Various hypotheses about the nature of the defects of the crystal lattice have been proposed from time to time by a number of distinguished authors.¹⁵⁻¹⁹

There is no exact classification of the defects. From a purely geometrical point of view defects in real crystals can be classified in terms of the number of dimensions in which the inelastically deformed region is of macroscopic extent; in the remaining dimensions its extent is microscopic, of the order of a few interatomic distances.

In this case vacant sites, interstitial atoms, and

impurity atoms are regarded as zero-dimensional defects, and boundaries between crystallites and blocks of a mosaic are two-dimensional defects; the one-dimensional defects include chains of vacancies and other zero-dimensional defects, and also a special distortion of the ideal crystal structure called a dislocation.

Displacement of dislocations can occur under the action of much smaller external stresses than would be required for two parts of the crystal to be shifted simultaneously relative to each other over the whole surface separating them. This peculiarity of dislocations has provided an explanation of values of the yield point of crystals that are low as compared with the shear modulus; this fact was indeed the original basis for the assumption that defects of this type exist in real crystals.²⁰⁻²²

By now a number of experimental proofs have been found for the presence of such defects in certain crystals, by observations of growth patterns^{23,24} and etching patterns on the surfaces of deformed and undeformed crystals, and by studies of the properties of the boundaries of crystallites with small angles of disorientation and polygonization of the grains.²⁸⁻³³

The most direct proofs of the existence of dislocations are the electron-microscope observations of images of individual atomic planes, i.e., the direct study of crystal lattices and their imperfections.³⁴⁻³⁷

Studies from many points of view on the role of defects and their origin, interactions, and transformations are necessary for an understanding of the deformation and fracture of crystals.

3. THE EFFECTS OF MICROSCOPIC CRACKS

In considering the question of the influence of defects on the strength of crystalline materials one is most often concerned with microscopic cracks at the surface or in the interior of the body.

An attempt to explain the discrepancy between the observed and theoretical values of the strength by the presence of cracks was made by Griffith in his theory of the brittle fracture of amorphous materials.³⁸ He assumed that real materials contain large numbers of small cracks, which can act to concentrate stresses and bring them up to the value of the theoretical strength. The process of fracture is then the increase of the length of a crack until the specimen is separated into two parts. Certain substances from the surrounding medium can penetrate into the cracks, lower the surface energy, and thus further diminish the strength.^{39,40}

Griffith studied the conditions of propagation of a crack in an elastic medium when the material is acted on by a system of external forces. The calculation for an infinitely extended thin homogeneous plate with a crack through its thickness showed that the strength

against rupture is connected with the length $2l$ of the hypothetical crack by the relation

$$\sigma_c = \sqrt{\frac{2ES}{\pi l}}, \quad (2)$$

where E is Young's modulus and S is the coefficient of surface tension.

Several authors^{41,42} have considered the problem of the strength of a body having a crack of elliptical shape, taking into account the interatomic forces and the location of the atoms at the edges of the crack, and have obtained similar results, differing from the Griffith formula only in the numerical coefficient.

It was found that for amorphous materials and polycrystalline metals broken under tensile stress the length of the embryonic cracks must be extraordinarily small (for example, for glass 2.6×10^{-6} cm, for α iron 7.8×10^{-6} cm), whereas for single crystals and plastic metals the cracks must be large (for example, in a single crystal of zinc 0.55 cm).

Thus the theory of Griffith gives satisfactory results only for the case of brittle fracture of amorphous and polycrystalline materials and is inadequate for the explanation of the low strengths of plastic materials, including metallic single crystals.

Several papers are devoted to the theoretical study of the development of actual microscopic cracks in crystalline materials.^{39,43,44} In contrast with the ideas of Griffith about the rounded shape of the edge of a crack in an amorphous substance, Rebinder³⁹ and Frenkel⁴³ believe that actual microscopic cracks in crystalline substances must have sharp ends, at which the radius of curvature of the section of the crack is zero. Therefore the width of real microscopic cracks is a variable quantity, which ranges from a maximum value in the more developed part of the crack to zero at the end points. This corresponds to a gradual mutual approach of the two surfaces separated by the crack until they reach a distance equal to the lattice spacing.

In his expression for the energy Griffith did not take into account the fact that the surface energy in a narrow space must be smaller than in a wide one. The sharpened shape of the edge of the crack leads to a decrease of the surface tension at the edge of a microscopic crack, from a certain maximum value in the wider part to values near zero at the very edge.

Such a shape of the microscopic cracks follows from the very nature of brittle fracture of solid bodies, with a gradual "cutting through" of the body by the spreading crack, and corresponds to the "self closing" of the cracks after removal of the load.

The shape of the microscopic cracks, especially at their edges, is of great importance for the theory of strength against brittle fracture, since the conditions at the edges of a crack determine its degree of sta-

bility. Griffith's incorrect choice of the geometry of the cracks caused one of the main shortcomings of his theory—the impossibility of the existence of stable microscopic cracks in a loaded solid body.

According to Griffith the curve $W(2l)$ of the dependence of the free energy on the length of an embryonic crack at a given load has one extremal point, a maximum, which corresponds to an unstable state of the crack (Fig. 1). It follows from Fig. 1 that with increase of the critical stress σ_c the critical length $2l_c$ of an embryonic crack becomes smaller. If the length of the crack is larger than the critical value ($l > l_c$) the crack will grow with the speed of sound, and if its length is smaller than the critical value ($l < l_c$) it must close up.

In his curves of free energy against crack length Frenkel found not only a maximum but also a minimum, corresponding to a stable state of the cracks (Fig. 2). As the stress is increased these extremal points approach each other. When the stress reaches a critical value, which is the limiting strength, the points come into coincidence, and there is catastrophic lengthening of the cracks, resulting in fracture. In this theory, unlike that of Griffith, the critical stress is not due to the previous existence of a crack of some finite length.

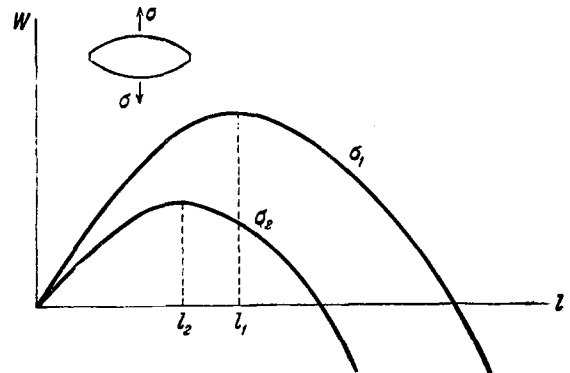


FIG. 1. Dependence of the value of the free energy ($W = 4lS - \pi l^2 \sigma^2/E$) on the length $2l$ of a hypothetical embryonic crack (Griffith³⁸).

Stress $\sigma_2 > \sigma_1$; l_2 and l_1 are the critical sizes of cracks for the stresses σ_2 and σ_1 .

In his calculation of the variation of the elastic energy Griffith used the solution of the plane problem for a plate of infinite extent with an embryonic crack through it. The results of his calculation do not apply to ordinary three-dimensional bodies, in which we can suppose that embryonic cracks through the body are out of the question. Because of mathematical difficulties no solution has as yet been obtained for the three-dimensional problem of a body with a crack, even for simple types of stress.

B. Ya. Pipes⁴⁵ used ideas of dimensions and similarity to get an estimate of the brittle strength of such a body. He found the formula

$$\sigma_c = \frac{3}{2} \cdot \sqrt{ES \cdot \frac{L}{\omega}},$$

where L is a linear dimension of the body and ω is the area of the crack. This formula holds for the volume case, in which the linear dimensions of the crack are small in comparison with those of the body. The results of Pipes' calculation can be used successfully to explain the increased strength of thin homogeneous filaments. The important point here is that the numerical value of the limiting strength is smaller in the case of a surface crack than in the case of an internal crack.

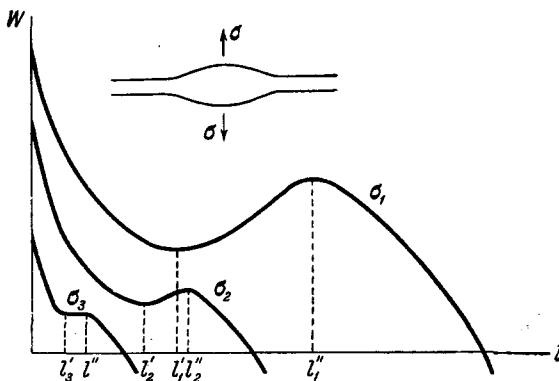


FIG. 2. Dependence of the value of the free energy (W) on the length l of real microscopic cracks (Frenkel⁴³).

Stresses $\sigma_3 > \sigma_2 > \sigma_1$; l'_3, l'_2, l'_1 are the sizes of stable cracks for the stresses $\sigma_3, \sigma_2, \sigma_1$. Fracture occurs at a stress σ_c for which $l'_c = l''_c$.

On the question as to whether the dangerous crack defects are located inside the material or on the surface of the stressed specimen there have been various arguments.

In the classic experiments of A. F. Ioffe⁴⁵ it was shown that in water the strength of rock salt increases from 0.5 to 160 kg/mm² and reaches values close to the theoretical prediction. This large increase of the strength of a rock-salt crystal under water is explained by the removal, by solution in the water, of the surface layer of the crystal, in which all the dangerous defects are concentrated.

Orowan⁴⁷ used attachments that left the edges of the mica free from stress, and found that the strength of laminar sheets of mica against splitting was thus increased by about a factor of ten (from 35 to 320 kg/mm²). He drew the conclusion that the strength of

the mica depends on the damage done to its edges during the splitting.

In 1932 S. N. Zhurkov⁴⁸ described ways of obtaining high-strength specimens of glasses by etching the surfaces with hydrofluoric acid.

Griffith showed experimentally that the technical strength of glass and quartz can be greatly increased if the specimens are prepared at a high temperature and are kept from contact with other solid bodies, i.e., if conditions are such as to hinder the occurrence and propagation of surface cracks.

In experiments by A. V. Stepanov⁴⁹ on the stretching of rock-salt crystals observations in polarized light showed that shears occurred mainly at artificially produced scratches on the surface of the crystal. These experiments provide a basis for the assumption that natural cracks that occur on the surface of a crystal through various accidental causes are the centers around which the first plastic shears appear in a strained solid body; these shears then lead to the development of microscopic cracks and to fracture.

Many other experiments have also shown that the presence of surface cracks greatly decreases the strength of both amorphous and crystalline brittle bodies. Despite the fact that these observations are not a final proof, it is quite probable that Griffith's cracks in brittle materials are mainly on the surface.

Andrade and other authors^{50,51} assert that by methods of etching and dusting they have been able to locate lines on glass and rock-salt surfaces, and that these lines are Griffith's cracks.

4. EFFECT OF SIZE. STRENGTH OF FIBER CRYSTALS

It is unlikely, however, that previously existing surface cracks play an important part in viscous or fatigue fracture of metals. The relatively low fracturing stress of metals is most easily explained by the presence of microscopic cracks inside the metal itself. One fact that certainly provides proof for the active influence of volume defects on the strength of crystals is the existence of a size effect in fracture. For a given material the mean strength of specimens against rupture increases with decrease of the size, and in thin filaments the strength is much larger than the usual technical strength.

It was suggested by A.P. Aleksandrov and S.N. Zhurkov⁴⁸ that the increased strength of thin filaments is an effect of a statistical nature. In their opinion the dependence of the strength of a thin fiber on its diameter is due to the presence in the material of inhomogeneities that differ in their effects on the strength and are distributed through the entire volume,

and that the relation of strength to diameter is given by an equation of the form

$$\sigma_c = a + \frac{b}{d+c}, \quad (3)$$

where a , b , c are constants and d is the diameter of the fiber. The results of known experiments with thin filaments^{48,52,53} are interpreted as a proof that with decrease of the total volume of the specimen there is a decrease of the total number of defects, which leads to a smaller probability that the more dangerous defects will be present in the specimen.

With regard to the experiments with glass fibers it must be noted that very recently very interesting observations have been published^{54,55} which indicate that there is a dependence of the strengths of fibers on a number of factors besides the size. The properties of fibers, including their strength, are determined by the conditions of their formation (the method of preparation, the temperature of the vitreous mass, the degree and rate of drawing) (Fig. 3). Otto⁵⁴ prepared glass fibers of various diameters under controlled, approximately identical conditions, and did not find any dependence of the strength on the diameter. Bartenev and Tsepkov⁵⁵ called attention to the fact that in glasses the influence of changes of the diameter or thickness of specimens depends on technological factors--the plastic deformations and thermoelastic stresses arising in the process of shaping the glass--while the dependence of the strength on the length or the size of the surface can be explained by a statistical theory.

The high strength of thin fibers is particularly strongly manifested in studies of the mechanical properties of fiber-shaped metal crystals. In 1952 Herring and Galt⁵⁶ showed that tiny fibers of tin (about two microns in diameter and a few millimeters long), which had been described as early as 1946 and are known by the name of "whiskers,"⁵⁷ have extraordinarily high strength properties. In recent years there have appeared a large number of papers devoted to the studies of the conditions of growth and the properties of fiber crystals.⁵⁸ Several different descriptions of the process of the growth of whiskers from the solid phase have been proposed,^{59,60} and there have been theoretical and experimental studies of the growth of fibers in condensation from the gaseous phase,⁶¹⁻⁶⁴ in the process of chemical reduction of metallic halides at high temperatures,⁶⁵⁻⁶⁷ and in the precipitation of the solid substance from solutions of salts.⁶⁸⁻⁷⁰ Most authors assume that in accordance with the mechanism of Frank⁷¹ whiskers are formed through crystallization on the nonspreading base of a single screw dislocation in each whisker. The line of this dislocation must lie along the axis of the resulting crystal fiber. There are indirect evidences that fibers of aluminum oxide⁷²

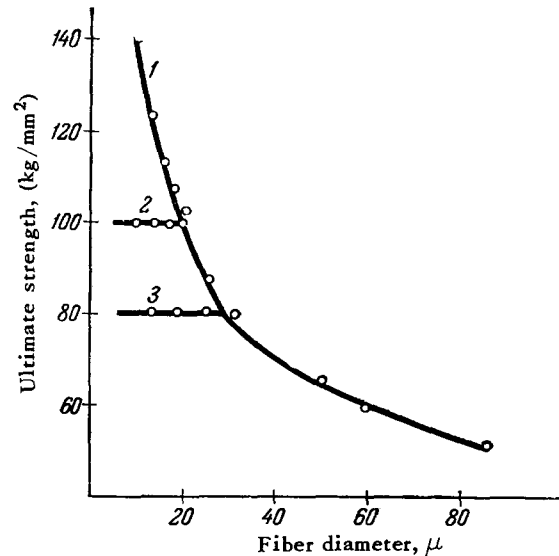


FIG. 3. Dependence of the tensile strength of glass fibers pulled from nonalkali glass on their diameters (Bartenev and Tsepkov⁵⁵).

- 1 -- dependence for fibers with variable degree of pulling out, $\alpha = D^2/d^2$ (D is the feeder diameter, d the diameter of the fiber); 2 -- dependence for fibers with constant degree of pulling out, $\alpha = 1.6 \times 10^4$; 3 -- dependence with $\alpha = 0.8 \times 10^4$

and of mercury⁷³ each contain a single dislocation. On the other hand, for fibers of iron, copper, nickel, zinc, silicon, and manganese⁷⁴⁻⁷⁷ there is no sure foundation for such an assumption. In a single set of fiber crystals, all grown together, one commonly finds a wide variety of forms of growth.^{78,79} The spiral shapes are of great interest; they are assumed to arise because of precession of the dislocation in the growing crystal. A remarkable property of high-strength whiskers is the optical perfection of their surface faces. Whiskers are most often single crystals.⁸⁰ The fiber is usually oriented along an axis of high order. Some experimental facts (growth of one fiber crystal from a lateral surface of another, data on the plasticity of whiskers, etc.) are a basis for the assumption that a fiber crystal is not entirely perfect. It is possible that there are many different dislocations in each such crystal.^{74,75,81}

As studies in recent years have shown,^{74-76,82,83} the strengths of fiber crystals prepared by various methods come close to theoretical values. For example, according to the data of Brenner,⁸³ the tensile strength of an iron whisker of diameter 1.6 microns is 1340 kg/mm^2 , or 60 to 80 times the tensile strength of ordinary single crystals of iron. The maximum tangential stresses in fiber crystals of iron, copper, and silver are from 80 to 1200 times the values in massive

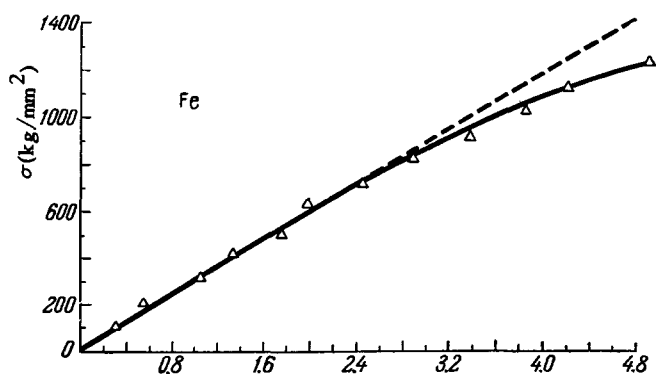


FIG. 4. Diagram of stretching of an iron whisker produced by reduction of the halogen compound FeBr_2 (abscissas are values of the elongation in percent; data from Brenner⁸³); axis of whisker parallel to the $[111]$ direction; diameter of whisker 1.6 microns; tensile strength 1340 kg/mm^2 ; maximum tangential stress 364 kg/mm^2 ; Young's modulus $E_{[111]} = 29,000 \text{ kg/mm}^2$; elastic deformation 4.8 percent; fracture was brittle.

crystals. Within experimental error Hooke's law holds up to strains of 2 percent.⁸⁴ In whiskers of iron elastic deformation can be as large as 4.8 percent, whereas in ordinary crystals it does not exceed 0.01 percent. The elastic moduli of fiber crystals do not differ from the usual values (for iron $E_{[111]} = 3 \times 10^6 \text{ kg/cm}^2$).

The fracture of fiber crystals can occur as brittle fracture, (for example, in the case of iron, Fig. 4) or can be preceded by plastic deformation (in zinc, copper, silver, cadmium, Fig. 5). In the latter case the deformation curves show a sharply marked elastic limit (peak with abrupt drop). The appearance of such breaks in the stretching curves of body-centered and face-centered cubic crystals is usually explained as due to the presence of impurities which form Cottrell's "atmosphere."⁸⁵ Some authors believe that the purity of whiskers is high enough to exclude the effect of Cottrell clouds.

Earlier experiments by P. I. Garber⁸⁶ on the twinning of large calcite crystals led to the assertion that the appearance of peaks in deformation curves can best be explained as arising from the destruction of the homogeneity of the single crystal when the first traces of plastic deformation occur. Similarly, also in the case of fiber crystals the presence of a peak does not necessarily mean the blocking of dislocations by impurities, and can be explained as a consequence of the destruction of the homogeneity of the crystal under the action of the applied load; that is, dislocations can arise when the flow limit is exceeded for the ideal crystal, in which previously there were no dislocations.

Fiber crystals of copper, silver, cadmium, rock salt, lithium fluoride, etc., have plastic deformations of more than 35 percent. The appearance of traces of slip

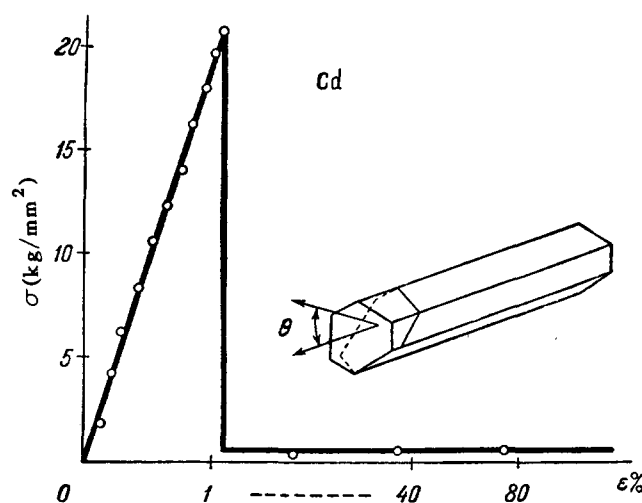


FIG. 5. Diagram of stretching of a fiber crystal of cadmium, grown from the vapor phase (Coleman, Price, and Cabrera⁷⁵); purity 99.999 percent; axis of specimen parallel to (1010) plane; angle between axis of specimen and sixth-order axis, $\theta = 30^\circ$; two faces in (1010) plane, four in (0111) plane; diameter of whisker 2.5 microns; length of whisker 535 microns; upper yield point (elastic limit) 27.8 kg/mm^2 ; flow stress 1 kg/mm^2 ; Young's modulus $2.78 \times 10^5 \text{ kg/cm}^2$; elastic deformation 1 percent; fracture was plastic.

sharply lowers the resistance of the specimen to deformation. For example, the flow stresses in cadmium and zinc whiskers⁷⁵ are smaller than the upper flow limit (elastic limit) by a factor of 30 to 80.

Because of the nonuniform distribution of slip bands along the axis of a whisker there is a possibility of cutting out of a deformed whisker sections that are free from slip. A new test with such a section of a whisker leads to still higher values of the elastic limit; evidence of this is found in the results of reference 76, which are reproduced in Fig. 6. Evidently in addition to the effect of size, to which the observed increase of the flow limit of small sections is usually ascribed, an even more important factor in such cases may be the fact that the weakest places in the specimen have been used up in the first test.

In many cases the stretching curves of whiskers have a very complicated shape, with alternate rises and drops of the flow limit; these are evidently due to overcoming of some sort of obstacles to slip, whose nature is still unknown (Fig. 7).

It is very difficult to explain the plastic deformation of fiber crystals. The formation of slip traces may be evidence of imperfections in the structure of the whiskers--the presence of dislocations that arose during the growth process. On the other hand, such inhomogeneities as dislocations can arise as the result of shear under high stresses. The experimentally observed decided lowering of the upper yield point of

silicon whiskers with increase of the temperature⁷⁴ indicates that in such a specimen there were many dislocations (a network of dislocations) before the deformation, together with a high concentration of obstacles to their possible motions.

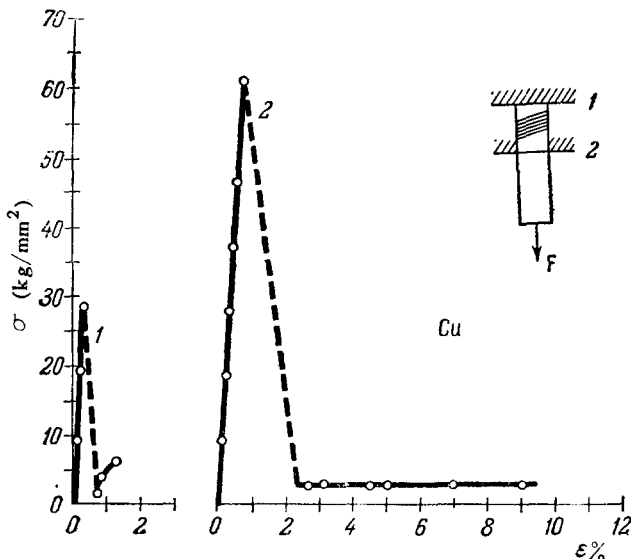


FIG. 6. Diagram of stretching of a fiber crystal of copper (Brenner⁷⁶): 1 -- initial curve (crystal attached in position 1); 2 -- curve obtained after removal of the region with the slip bands (crystal attached in position 2).

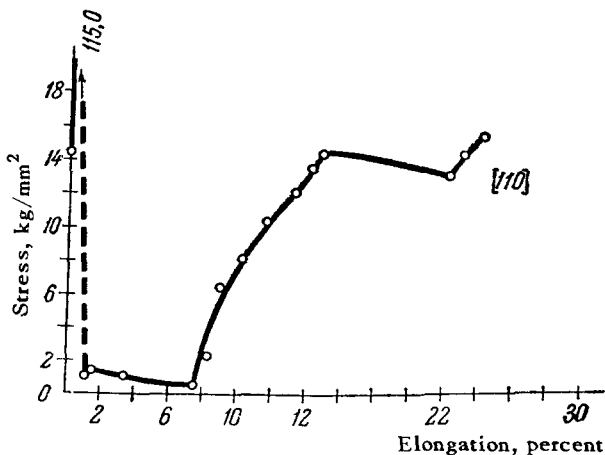


FIG. 7. Diagram of stretching of a fiber crystal of copper with successive segments of hardening and weakening (Brenner⁷⁶).

The strength of a fiber crystal depends strongly on the diameter and length. In almost all cases the strength increases with decrease of the diameter or length. The dependence of strength on diameter obeys laws of the form

$$\sigma_c = a + \frac{b}{d},$$

where a and b are constants and d is the diameter of the fiber; for copper and iron whiskers of a definite length⁸³ the equations are

$$\sigma_c = 36 + \frac{410}{d} \text{ kg/mm}^2 \text{ и}$$

$$\sigma_c = -50 + \frac{1630}{d} \text{ kg/mm}^2 \text{ (Fig. 8).}$$

A considerable increase of strength with decrease of length is observed in experiments in which successive tests terminating in rupture are made on a single specimen. For example, the strength of a fiber crystal of iron of diameter 4.71 microns and original length 4 mm was 99 kg/mm², but a crystal 1 mm long left after the second rupture had a breaking strength of 423 kg/mm². As was mentioned before, in this case the increase in strength can be ascribed to the fact that the weakest places had been used up in the previous tests of the long specimen.

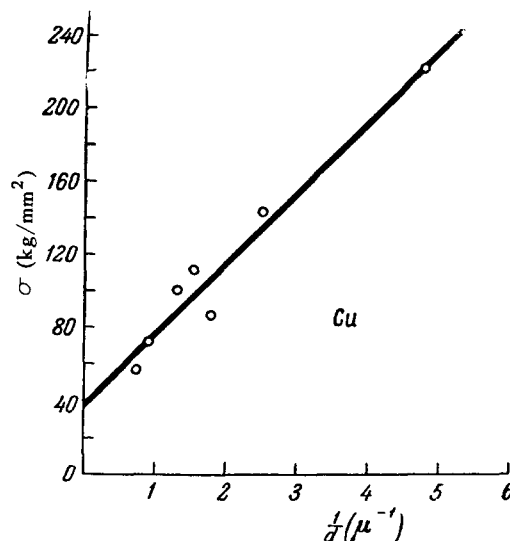


FIG. 8. Dependence of the strength of copper whiskers on the quantity $1/d$ (d is diameter of whisker) (Brenner⁸³). Axes of crystals mostly parallel to $[111]$ direction.

The large spread of the values of the strength as function of diameter (Fig. 9) indicates that in these crystals there is some sort of statistical distribution of defects. It is very hard to make a statistical analysis on the basis of test results available so far, because of the lack of a sufficient amount of data on the nature of the defects and their orientations and localization.

The number of surface defects of a given effectiveness should vary in direct proportion to the diameter, and the strength should be inversely proportional to the diameter of the whisker. In just the same way, if there is a certain number of internal sources of dislocations, then the length of the most extended source will vary directly as the diameter, and the strength again inversely as the diameter. Therefore an experimentally established linear relation between the mean strength and $1/d$ cannot be used to determine the location of the defects, and does not suffice to settle the question of the relative roles of surface and internal defects.

The determination of the nature of this very high strength of fiber crystals is a matter of great interest. For this purpose parallel studies have been made on the strength of whiskers and of specimens of the same dimensions cut from massive crystals. It was found that the strength of cut specimens having transverse dimensions of 50 to 25 microns⁷⁴ does not differ from the strength of fiber crystals of silicon of the same dimensions ($\sigma_c = 2.2 \times 10^{10}$ d/cm²). It remains unknown, however, whether this agreement will also hold at smaller diameters of the specimens.

It is very hard to evaluate the effect of impurities on the strength of fiber crystals. This fact is due to the absence of adequate data on the chemical composition of such crystals. It is known that in some cases the purity of whiskers is not exceptionally high (for example, the average impurity content is about 3×10^{-5} for copper whiskers grown from CuI and 10^{-4} ; for iron whiskers grown from FeBr₂; the smallest measured values of the residual resistance for zinc and copper indicate⁸⁷ an impurity content of 10^{-4}).

The available data on the properties of fiber crystals grown by electrolysis from salts of various purities leads to the belief that impurities impair the mechanical properties.

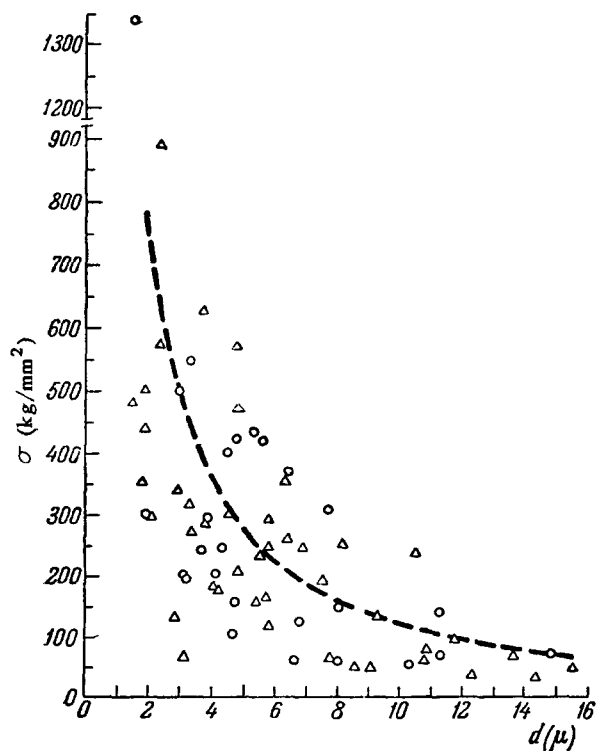


FIG. 9. Dependence of strength of fiber crystals of α iron on the diameter (Brenner⁸³):

Axes of crystals mainly parallel to the [100] direction; lengths 1 to 4 mm; rate of loading 0.2 to 1.0 g/min.

The conditions under which fibers are grown have a decided influence on their strength. With precipita-

tion from solution or reduction of halides one gets a greater spread of strengths than in cases in which fiber crystals are grown under more "pure" conditions, namely by deposition in vacuum or in an inert atmosphere. Fiber crystals with surface defects (pitting, outgrowths, corroded areas, etc.) as a rule have low strength.

Films of oxide increase the strength of ordinary crystals. It is also well known that the thicker the film or the smaller the diameter of the crystal the greater the strengthening effect.⁸⁸ Continuous films of oxide on fiber crystals do not cause much change of the strength. Moreover, as Brenner⁸³ has shown, the result of oxidation of iron whiskers is to lower the strength, and in the opinion of the author this is due to the formation of boundary dislocations.

The available data on the growth and properties of fiber crystals are insufficient for a judgement as to the causes of their high strength. At present it is unknown whether the high strength is due to the small size, to the perfection of the surface, to special perfection of structure, or to a combination of these factors.

It is most probable that the high strength of whiskers can be related to the absence of defects capable of causing inhomogeneous distributions of stress. This may be due either to special features of the growth of these crystals, or simply to their small sizes.

5. STATISTICAL THEORY OF STRENGTH AGAINST BRITTLE FRACTURE

Statistical methods have been applied to the solution of quite a number of problems of strength. N. N. Davidenkov⁸⁹ has proposed the hypothesis that the effect of size on the conditions for the occurrence of brittle fracture of steel parts must be regarded as a statistical effect. In a long series of papers⁸⁹⁻⁹³ N. N. Davidenkov and his collaborators have studied in detail the phenomenon of the brittleness of steel and the dependence of the brittle strength of metal specimens on their size, and have shown that with decrease of the size of the specimen there is an increase of the mean value of the brittle strength and a lowering of the critical temperature for brittleness.

Weibull,⁹⁴ and independently also Ya I. Frenkel and T. A. Kontorova,^{95,96} have considered the problem of the effects of inhomogeneities existing in real crystalline materials on the value of the brittle strength. The two theories, which differ from each other in their mathematical apparatus, start with the common assumption that the strength of a specimen is determined by the most "dangerous" of all the inhomogeneities present in it. Total fracture of the specimen occurs when the average stress reaches the "local strength" of the weakest "defective" place. The statistical distribution of defects over the whole volume of the specimen brings about a dependence of the brit-

the strength of crystalline bodies on their dimensions.

In its general form the formula of Weibull for the brittle strength under stretching, twisting, and bending has the following appearance:

$$\sigma_{br} = \frac{A}{V^{1/n}}, \quad (4)$$

where A is a constant depending on the character of the state of stress and the nature of the material, and n is a constant for the material that takes into account the character of the distribution of "defects."

According to Weibull, in a hypothetical absolutely homogeneous solid there is no dependence of the brittle strength on the volume ($n = \infty$). The experimental data⁹⁷ have given quite satisfactory confirmation of the main conclusions of Weibull's theory.

Ya. I. Frenkel and T. A. Kontorova took as a parameter characterizing the degree of "dangerousness" of a defect the value of the brittle strength (σ_{br}) that a specimen would have if the source of its fracture were the given defect. They assumed a Gaussian law for the distribution of the value of the local brittle strength and found a relation between the strength and the volume, for the special case of large values of the volume, in the form

$$\sigma_{br} = \sigma_0 - \sqrt{A \log V + B}, \quad (5)$$

where σ_{br}^* is the most probable value of the brittle strength for specimens of the given volume, σ_0 is the value of the brittle strength corresponding to the defects most frequently encountered in the given material (i. e., the defects of "average" dangerousness), and A and B are constants depending on the state of stress and the material.

Existing theories of brittle fracture explain the lack of agreement between the values of the theoretical and practical strengths, but cannot explain some other features of the fracture process. They do not take into account the influence of plastic deformations preceding fracture, no of diffusion processes, which lead to important structural changes in a metal.

Great difficulties are encountered in the explanation of the phenomena of static and dynamic fatigue and of the kinetics of fracture. In particular, the observed simultaneous development of cracks of different sizes in a material subjected to a constant stretching load calls for a special explanation, which has not been suggested so far. The theory of brittle fracture stands in need of additional considerations on the kinetics of the growth of cracks.

6. ORIGIN OF EMBRYONIC CRACKS

The physical nature and origin of the embryonic cracks that lower the mechanical strength of crystalline materials have not yet been explained. Evidently there is no single mechanism of the formation of such cracks. B. Ya. Pines⁴⁵ suggests that during the very process of crystal-

lization of polycrystalline metals and alloys regions of "insecure contact of the grains" are formed, which can play the role of cracks. Other probable sources of cracks are particles of precipitated substances (for example, carbides in steels) and interphase boundaries formed as the result of phase transitions in the solid state. In this case cracks can arise on account of the stresses caused by inadequate cohesion between the precipitate and the matrix, and also on account of differences in their elastic and plastic properties. A. V. Stepanov⁹⁹⁻¹⁰² assumes that cracks do not arise spontaneously, but mainly as a result of plastic deformation preceding the brittle fracture. In experiments on the strength of rock salt Stepanov¹⁰² found that by hindering or enhancing the plastic deformation he could change the practical strength of the salt by a factor 30, getting values both above and below the usual value (from 2000 to 80 g/mm², whereas the normal value is 500 g/mm²) (Fig. 10).

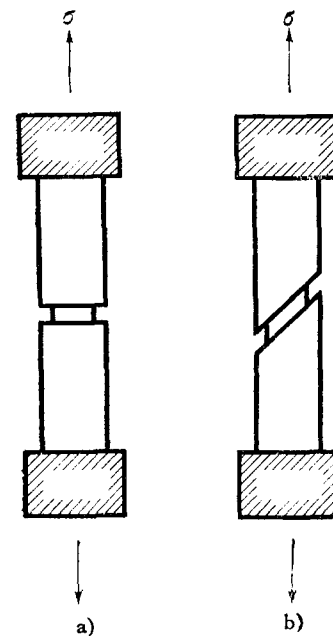


FIG. 10. Two experiments of A. V. Stepanov¹⁰² on the strength of crystals of rock salt. The slip plane (011) is oriented at an angle of 45° with the direction of the tension. a) crystals with strength up to 2000 g/mm²; b) crystals with strength as low as 80 g/mm².

Stepanov's hypothesis has also been confirmed by work of other investigators.^{103, 104} N. N. Davidenkov and E. M. Shevandin¹⁰³ showed that distortions facilitating premature fracture arise because of the emergence of shears on the surface. In experiments by M. V. Klassen-Neklyudova¹⁰⁴ fracture of dry rock salt occurred with the beginning of intense formation of slip planes, and corresponded to the beginning of creep in wet salt. The idea of the development of a brittle crack from distortions caused by plastic deformation

was later extended by N. N. Davidenkov to the case of metals.¹⁰⁵

One of the important conclusions that follows from the many studies of the physical nature of brittle fracture is that brittle fracture of a metal is practically impossible without being preceded by plastic deformation.

X-ray studies of the surfaces formed in brittle fracture of iron and steel^{106, 107} have confirmed the conclusion that brittle cracks develop from distortions produced by plastic deformation. The magnitude of this plastic deformation, as determined quantitatively from the diffuseness and intensity of the $K\alpha_1$ interference line, reaches values as large as 5 percent, according to the data of S. O. Tsobkhallo.¹⁰⁶ The depth to which the plastic deformation extends from the surface of the brittle fracture is a few hundredths of a millimeter (0.03 mm).

Analysis of x-ray interferences also shows that the plastic deformation of steels in brittle fracture occurs through a considerable fragmentation of the original grains. A similar decrease of grain size occurs also in viscous fracture, so that despite the different manner of breaking brittle and viscous fracture are both accompanied by the same process of block formation, which is typical for plastic deformation. The difference between these types of fracture lies only in the magnitude of the plastic part of the deformation. The brittle state corresponds to test conditions that are favorable for the production of large distortions, so that the first shears give rise to damage capable of leading to fracture.

Although in the opinion of A. V. Stepanov the primary defects existing in the crystal before the experiment are not the cause of its premature fracture, they undoubtedly influence the strength of the crystal.¹⁰⁸

In work on the artificial production of shear A. V. Stepanov has studied the mechanism of action of a primary defect and the relations between primary defects and those arising from deformation. Experiments with scratches drawn along rock-salt crystals, and also with point damage to the surface,^{109, 110} showed that growth of a primary defect is possible if it is stimulated by plastic deformation.

According to Stepanov brittle fracture of a crystal occurs by the following pattern: a surface defect or a severe deformation gives rise to a shear, which in turn leads to the development of an original defect or the production of a new one, and this defect is then responsible for the fracture.

Later, on the basis of properties common to all crystals--anisotropy of the elastic properties and the possibility of mechanically oriented processes in crystalline media--A. V. Stepanov suggested that production of shears and fracture should be regarded as the result of loss of the ideal crystal-lattice property of elastic stability.¹¹

In the case of crystalline media mechanical-orientation processes consist of the appearance of reoriented crystals under the action of external mechanical forces. Since the free energy of a crystal subjected to the action of external forces is a function of the angle between the direction of the force and the axis of the elastic anisotropy, in the general case the positions of the atoms in the crystal will not correspond to an equilibrium state. It is just this that gives the possibility of mechanical-orientation effects in a crystal.

Transition to an equilibrium form of the atomic system can occur through regrouping of the particles, leading to reorientation of part or all of the volume of the crystal. A. V. Stepanov assumes that in the case of artificial production of shears all of the necessary energetic conditions for the formation of embryonic shears resulting in fracture are fulfilled.

Stepanov's arguments about fracture as an effect of the loss of elastic stability of the crystal lattice are purely hypothetical and stand in need of experimental tests.

The nature of the distortions caused by plastic deformations and the degree of their dangerousness depend strongly on the experimental conditions, the temperature, the rate of deformation, and so on. Therefore changes of the experimental conditions can cause corresponding changes in the strength and the plastic properties. A good example of this is found in the data of S.N. Zhurkov and T.P. Sanfirova¹¹² on the endurance time and the rate of viscous creep, which were taken over a very wide range of rates of deformation. Their experiments led to a general law for the creep and the endurance time under a load. At a given temperature the dependence of the time τ till fracture and the rate of creep v on the stress σ is given for a large group of metals and alloys by relations of the type

$$\tau = Ae^{-\alpha\sigma}, \quad (6)$$

$$v = A_1 e^{\alpha_1\sigma}, \quad (6a)$$

in which the quantities A and α , A_1 and α_1 vary with temperature according to definite laws and thus fix the temperature dependence of the endurance time and the rate of creep (Fig. 11).

The observed dependences of τ and v on the temperature (Fig. 12) can be written in the form

$$\tau = \tau_0 e^{-\frac{U_0 - \gamma\sigma}{kT}}, \quad (7)$$

$$v = v_0 e^{-\frac{Q_0 - \gamma_1\sigma}{kT}}, \quad (7a)$$

where Q_0 and U_0 are the activation energies of the processes of quasiviscous creep and fracture, and γ and γ_1 are quantities giving the extent to which the stress affects the barrier heights. It has been found that the energy barriers are equal ($Q_0 = U_0$), which in

the opinion of the authors indicates a close interconnection of the processes of plastic deformation and fracture, such that the speed of one process is determined by that of the other.

In the case of fracture of plastic materials, which is even more premature than that of brittle materials, the strength of crystals is determined by their resistance to plastic deformation. A change of the strength is impossible without change of the plastic properties of the crystal. For example, the effect of A.F. Ioffe can be regarded as the result of the dissolving of distortions that appear on the surface of the crystal because of previous plastic deformation.

Ordinarily a definite plastic deformation precedes fracture. This deformation has a twofold effect on the crystal lattice of the metal. The appearance of a zone of shearing smooths out irregularities in the distribution of the stresses localized in regions whose size is much larger than that of the coherence blocks. Such a leveling action of plastic deformation can lead to a checking of the development of microscopic cracks, which has been observed in a number of studies.^{113,114} The local character of the development of intracrystalline microscopic cracks permits the supposition that at

first the plastic deformation blocks the development of microscopic cracks, but subsequently facilitates it through hardening. The fracture of plastic crystals can occur according to the scheme: concentration of stresses--development of slip--embrittlement of a plastic crystal complex--abrupt growth of a microscopic crack. The unblocking and growth of microscopic cracks is influenced by fluctuation processes occurring at the ends of the cracks; these processes depend very strongly on the temperature, the duration of the action of the stresses, and their concentration.

Together with its favorable effect on the practical strength, plastic deformation leads to the formation of ultramicroscopic destructions of the continuous structure and the development of surfaces of separation between the blocks, so that resistance to fracture is weakened.

In our view¹¹⁵ microscopic cracks are formed as the result of structural disruptions accompanying plastic deformation. It is well known that plastic deformation leads to destruction of the homogeneity of single crystals and of the crystallites of polycrystalline aggregates. The formation of slip bands, twinning interlayers, faults, and so on is regarded only as a

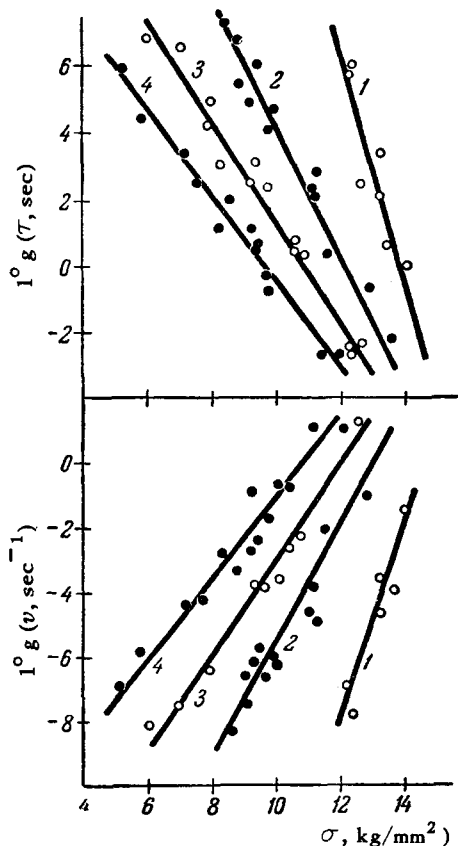


FIG. 11. Dependences of endurance time τ and rate of creep v of platinum on the stress σ (Zhurkov and Sanfirova¹¹²):
1 -- 100° C; 2 -- 300° C; 3 -- 400° C; 4 -- 500° C.

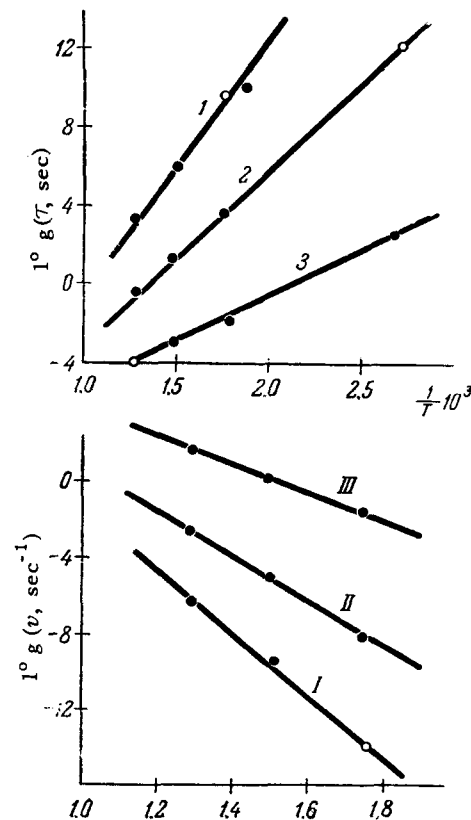


FIG. 12. Dependences of endurance time τ and rate of creep v of platinum on a quantity proportional to the reciprocal temperature $1/T$ (Zhurkov and Sanfirova¹¹²):
1 -- $\sigma = 7$ kg/mm²; 2 -- $\sigma = 10$ kg/mm²; 3 -- $\sigma = 13$ kg/mm²;
I -- $\sigma = 6$ kg/mm²; II -- $\sigma = 9$ kg/mm²; III -- $\sigma = 12$ kg/mm².

loss of homogeneity and is not associated with the microscopic cracks and fractures that undoubtedly also result. The assumption that at all stages of the process of plastic deformation solid bodies remain macroscopically compact cannot be regarded as justified. The available experimental information shows that in deformation there is destruction of not only the homogeneity but also the compactness of crystals.

Evidences of structural disruptions in solid materials in the process of plastic deformation is also

provided by direct observations, such as decreases of the density of deformed materials, the appearance of ultramicroscopic scattering of light in transparent rock-salt crystals,¹¹⁶ the production of microscopic pores in iron specimens when they are stretched and rolled in liquid nitrogen,¹¹⁷ the formation of fractures as a consequence of nonuniform shear along a single slip band in single crystals of beryllium¹¹⁸ (Fig. 13), the localization of fractures and cracks within slip bands¹¹⁸⁻¹²⁰ (Fig. 14), and so on.

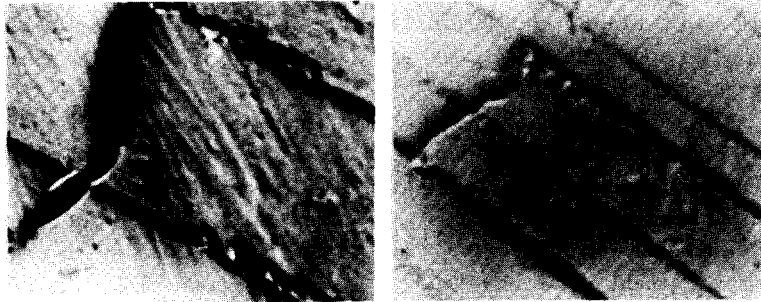


FIG. 13. Traces of bands of basal slip in a single crystal of beryllium (99.7 percent) (Garber, Gindin, Kovalev, and Shubin¹¹⁸). Load axis oriented at angle 45° with basis plane (0001). Cut surface parallel to plane of shear (1100): a) after reduction by 3.4 percent at 196°C ; b) after reduction by 3.9 percent at 20°C . The slip-band traces are crossed by microscopic cracks along the plane of the prism of the second kind (1120). $\times 10,000$.



FIG. 14. Traces of bands of basal slip in a single crystal of beryllium (99.7 percent) (Garber, Gindin, Kovalev, and Shubin¹¹⁸). Reduction by 4.7 percent at 400°C . Microscopic cracks are localized along slip bands. $\times 10,000$.

The accumulation of latent energy in plastic deformation can be explained in terms of structural disruptions. If we assume that as a result of the deformation part of the surface of the blocks may be free, then the latent energy will include the surface energy of the

free part of the surfaces of the blocks. In favor of such an assumption we have the data on the similarity of the dependences of the number of blocks and of the latent energy density on the deformation, and the fact that both these quantities reach saturation at the same time as the deformation is increased.

The ratio γ of the free surface area to the total surface area of the blocks can be represented, under conditions of saturation, by the following formula:

$$\gamma = \frac{\alpha}{\beta} \cdot \frac{l \cdot \rho Q}{S}, \quad (8)$$

where $\alpha = v/l^3$, v is the volume of a block, $\beta = F/l^2$, F is the surface area of a block, l is a linear dimension of a block, ρ is the density of crystal, Q is the latent energy of the plastically deformed crystal at saturation per unit mass of the specimen, and S is the coefficient of surface tension of the undeformed crystal at a boundary with vacuum. For copper we calculate from Eq. (8) a value $\gamma = 0.5$. Thus a large fraction of the surface of the coherence blocks must be free from contacts with neighboring blocks.

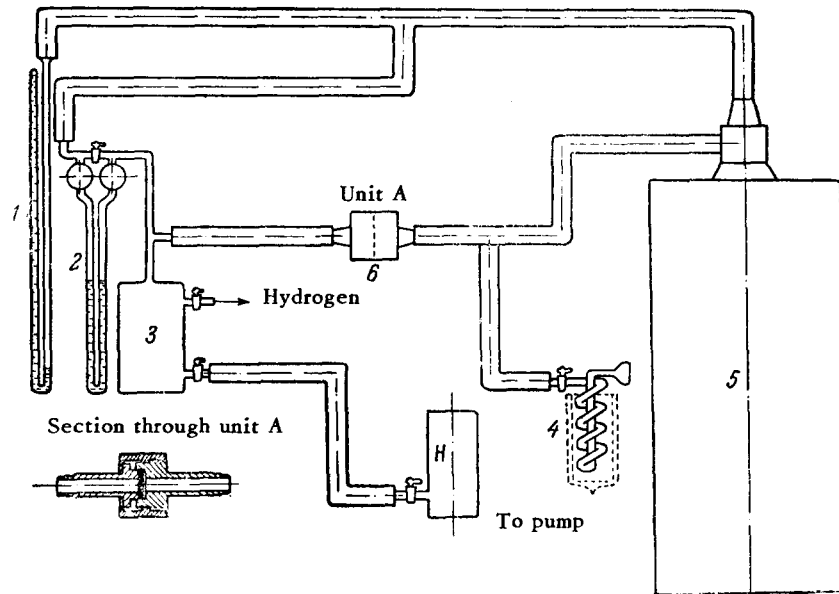


FIG. 15. Diagram of apparatus for testing porosity of specimens: 1 -- absolute mercury manometer; 2 -- differential oil manometer; 3 and 5 -- bulbs; 4 -- trap cooled by liquid nitrogen; 6 -- chamber containing specimen under test.

The part of the latent energy of plastic deformation that goes into other structural distortions, such as defects on the atomic scale (vacancies, interstitial atoms), dislocations, and residual stresses, can be estimated from the results of a number of studies,^{121,122} and is not more than 10 percent. Consequently, almost all of the latent energy of plastic deformation is localized in the form of the surface energy of the parts of coherence blocks that are not making contact.

As an explanation of the fact that a deformed solid body remains macroscopically intact while there is microscopic structural disruption it has been suggested that along with the breaking of contacts there is also renewal of contacts. The subsequent reestablishment of compactness, necessary for further plastic deformation, can be thought of as the result of the fixing of molecular bonds at the new contacts or as the consequence of a process similar to sintering.^{115,123,124}

The process of local disruption can be accompanied by partial or complete reestablishment of contacts. Partial reestablishment of contacts should be found in cases of slip, when there is breaking up of the coherence blocks and loss of their orientation, with decrease of density, scattering of light in transparent crystals, and other effects. In intertwined parts of a crystal the contacts are evidently completely reestablished. At the twinning faces, in elastic twins, in accommodation zones, and in wedge twins there is only partial reestablishment of the contacts. Partial healing of microscopic cracks will go on more or less actively, depending on the conditions of the experiment. For example, adsorption from the surrounding

medium onto newly formed surfaces can hinder such healing. If the contacts are not reestablished, or if the new contacts are not strong enough, fracture will occur at lower stresses.

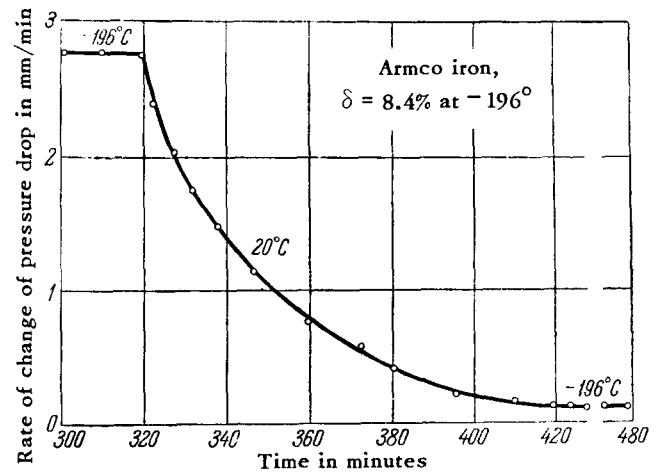


FIG. 16. Decrease of rate of change of pressure drop in flow of hydrogen through a porous partition; specimen of iron deformed in liquid nitrogen and then annealed at 20°C (Garber, Gindin, Kogan, and Lazarev¹¹⁷).

Recovery of compactness at new contacts has been observed in a study of the flow of gaseous hydrogen through a thin partition of iron that had been subjected to deformation in liquid nitrogen.¹¹⁷ A diagram of the apparatus is shown in Fig. 15. The plastic deformation of iron at low temperatures leads to the formation of very minute pores. It was found that at tempera-

ture -196°C the porosity does not change with time. After warming to room temperature there is a rapid decrease of the porosity, which is shown by a decrease of the rate of change of the pressure drop, as seen in the diagram of Fig. 16. In these experiments with iron the porosity was determined from the formula¹²⁵

$$\eta = \frac{\ln \frac{\Delta p_1}{\Delta p_2}}{k\tau}, \quad (9)$$

where k is an apparatus constant, τ is the time of flow of the hydrogen, Δp_1 is the initial pressure drop across the oil manometer 2, and Δp_2 is the final pressure drop across the manometer. These experiments with iron serve as a proof that when the temperature is raised the contacts between the microscopic blocks are rapidly reestablished and the chinks produced in the process of deformation are closed up. The closing up of the microscopic pores is associated with low-temperature recrystallization, but the latter process cannot by itself lead to the formation of new contacts. For the closing up of pores, on the other hand, formation of new contacts is necessary, and growth of the grains is not essential.

Another striking example of the reestablishment of contacts in the process of plastic deformation is the macroscopic shear of beryllium crystals in the basal plane,¹²⁶ which is observed under conditions of uniaxial compression at 400°C . In such specimens the slip is externally reminiscent of the model of pure shear along an atomic plane (Fig. 17). Two parts of a crystal are shifted relative to each other along the basal plane (0001) in the direction $[11\bar{2}0]$, and in some cases the magnitude γ of the relative shift on the traces of concentrated slip reaches the value $\gamma = 200$. The traces of concentrated slip with large relative displacement can also be regarded as regions where compactness is temporarily lost in the process



FIG. 17. Specimen of a single crystal of beryllium (99.7 percent) after compression at 400°C (Gerber, Gindin, Kogan, and Lazarev¹²⁶).

Original orientation: basal plane (0001) parallel to plane of picture; the angle between the axis of compression and the binary axis $[11\bar{2}0]$ is 60° ; concentrated slip along basal plane with relative slip $\gamma = 200$ facilitated after intertwinning of specimen throughout its volume.

of deformation, after which cohesion reappears as the result of reestablishment of contacts.

In brittle metals the formation of blocks and microcracks through structural disruptions precedes the appearance of slip bands or occurs simultaneously with their appearance. For example, the residual deformation of a beryllium crystal associated with the formation of blocks of mosaic structure before the appearance of slip bands resolvable by the electron microscope can be 3 percent at -196°C . Use of a higher temperature for the test leads to a decrease of the part of the deformation that precedes the formation of bands and the localization of breaks and cracks in the slip bands.¹¹⁸

The localization of fatigue cracks near slip bands is confirmed by many authors. Thus breaks and cracks formed in the fragmentation of grains into blocks during the process of slip can be regarded as the main cause of the lowering of the strength of solid materials.

7. DISLOCATIONS AND MICROSCOPIC CRACKS

During the last decade conceptual models of the mechanism of the formation of embryonic cracks and of the fracture of metals have been developed in the framework of the dislocation theory. The dislocation theory offers several schemes for the formation of cracks. All of these schemes are also based on the fact that crystalline fracture is always preceded by plastic deformation. The main dislocation mechanisms for the formation of cracks and for fracture are described by the theory of "accumulation of dislocations at obstacles"¹²² and the closely associated theory of blocked slip.¹²⁷

In the opinion of Zener¹² slip planes can be concentrators of stresses. Overstressing occurs in front of the edge of a spreading slip band and is due to the relaxation of shearing stresses inside the band. Stopping of a slip band at an obstacle can cause further relaxation and growth of the stress to the value of the theoretical strength, which leads to the production of

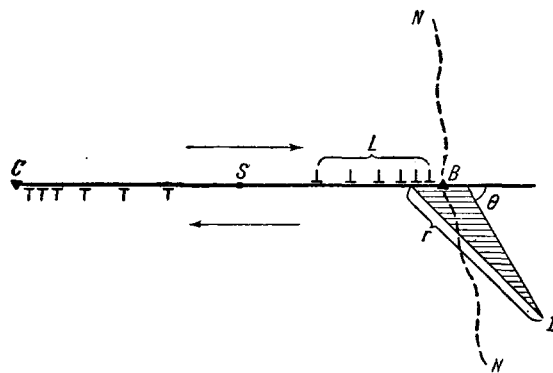


FIG. 18. Scheme of formation of a microscopic crack BD through accumulation of dislocations at the end of a slip plane BC (at a grain boundary) (Stroh¹²⁹).

a crack and its further development. Zener identifies the stopping of a slip band at an obstacle with the formation of an accumulation of dislocations, with their density increasing with approach to the obstacle.

The idea of the formation of cracks through accumulations of dislocations has received further development in papers by Mott,¹²⁸ Stroh,¹²⁹ Petch,¹³⁰ and other authors. The required concentration of stresses is produced near accumulations of dislocations in slip planes. A necessary condition for the accumulation of groups of dislocations is the presence of barriers that block the propagation of the slip. Such barrier may be grain boundaries or fixed dislocations.

It is assumed that under the action of applied shearing stress τ the dislocations generated by a Frank-Read¹³¹ source S shift freely along the slip plane BC until they reach the barriers B and C (Fig. 18), where there is accumulation of the group of dislocations. Jacquet¹³² believes that he has succeeded in observing blocked groups of dislocations on the surfaces of etched specimens (Fig. 19). As



FIG. 19. Accumulation of dislocations at a grain boundary in brass (Jacquet¹³²).

the result of the accumulation of dislocations near the ends of lines of slip, large stretching stresses will arise, analogous to the stresses at the ends of Griffith cracks [cf. Eq. (2)], which lead to the formation of an embryonic crack.

The magnitude of the normal stress on the plane BD at the distance r from the vertex of the accumulation of dislocations is given by the formula¹²⁹

$$\sigma = \tau \sqrt{\frac{L}{r}} \cdot f(\theta), \quad (10)$$

where L is the linear measure of the extent of the region of shear with displacement of dislocations of a given sign, approximately equal to half the length of the line of dislocations, τ is the shearing stress along the slip band BC , and $f(\theta)$ is a function depending on the orientation of the plane BD .

The critical shearing stress required for the formation and growth of a crack to the length of a Griffith crack can be found from the condition that the total energy W of the formation of the crack be a minimum; using the relation (10), we get

$$\frac{\partial W}{\partial (2l)} = \frac{\partial}{\partial (2l)} \left\{ -\frac{\pi(1-\nu)\sigma^2(2l)^2}{8G} + 4Sl \right\}, \quad (11)$$

where $\pi(1-\nu)\sigma^2(2l)^2/8G$ is the elastic energy and $4Sl$ is the surface energy of a crack of length $2l$; ν is Poisson's ratio, and G is the shear modulus.

The condition for production of a crack then takes the form

$$\tau = \alpha G \sqrt{\frac{b}{L}}, \quad (12)$$

Further growth of the crack and complete fracture occur under the action of the stresses normal to the plane BD .

Depending on the properties of the material and the conditions of the deformation the concentration of stresses near the head of the accumulation can lead either to the formation and development of a crack and finally to fracture, or to overcoming of the obstacle by the accumulation of dislocations and thus to plastic flow. Formation and propagation of a brittle crack are possible only under conditions of hindered plastic deformation. The occurrence of plastic deformation will decrease the concentration of stresses in front of the accumulation of dislocations and exclude the possibility of the formation and development of a brittle crack.

Thus if the sources of dislocations are not blocked, for example by impurities, and can generate new dislocations, brittle cracks cannot be propagated. This corresponds to the fact that in metals with face-centered cubic lattices, in which blocking by impurities is small and the flow limit is low, one finds viscous fracture with the formation of a large number of microscopic cracks, which then merge into one crack leading to fracture.

In this form the dislocation theory has been useful for the description of the brittle and viscous fracture of polycrystalline metals. Brittle fracture occurs in cases in which a slip line intersects a grain (this underlines the absence of ideal brittle fracture), or when the shearing stress exceeds the critical stress

$[\tau = G(b/L)^{1/2}]$ required for the formation of a crack, or when an incipient crack is not checked by plastic flow from outside around its end and the stretching stress is large enough to cause its further propagation. In the latter case the crack will grow with a speed close to that of sound.

If there is development of cracks, we must expect plastic fracture when a large number of disordered cracks are formed which combine with each other. Extinction of the stressed zone in the development of a crack occurs through the displacement of dislocations that arise from Frank-Read sources situated around the region. In this case the crack can enlarge only to a distance comparable with the distance between sources of dislocations in the material ($\sim 10^{-4}$ cm).

According to the dislocation theory embryonic cracks (Griffith cracks) in metals arise as a result of plastic deformation, and fracture can occur only when the shearing stresses reach a critical value. It is also believed that tangential stresses are responsible for the formation of cracks, whereas their enlargement is due to the normal stresses acting on the entire specimen.

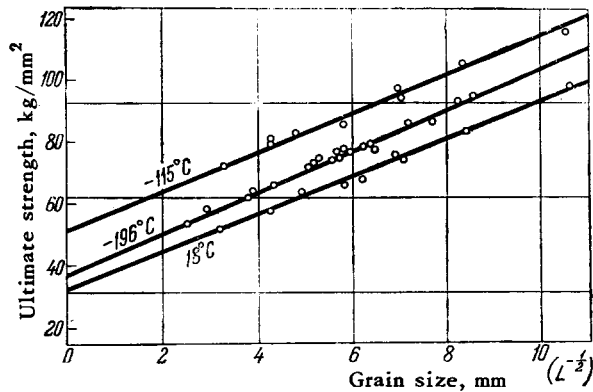


FIG. 20. Dependence of the limiting strength of Armco iron on grain size (Petch¹³³).

It also follows from the theory of dislocations that both in the case of brittle fracture and in that of plastic fracture we must expect similar dependence of the breaking stress on the size of the grain structure. In fact, if the length of a slip line is equated to the diameter of a grain and we identify the Griffith cracks with the slip planes, by using the relation between the shearing stress τ and the pulling stress σ in the linear problem ($\tau = \frac{1}{2}\sigma$) we can get this dependence in the form of the following equation:

$$\sigma = k \cdot d^{-1/2} + \sigma_0, \quad (13)$$

where $k = \{6\pi SG/(1-S)\}^{1/2}$, $L = d$ is the diameter of a grain, and σ_0 is a term that allows for the resistance to the motion of dislocations because of the presence

of dissolved atoms and dispersed precipitations.

The experimental data of Petch¹³³ for iron and soft steel give good agreement with the conclusions of the theory (Fig. 20). A characteristic fact is that in brittle and viscous fractures the curves have the same slope; this evidently indicates the action of the same mechanism—the formation of accumulations of dislocations at grain boundaries.

Some features of hardening are caused by the formation of an “atmosphere” of dissolved atoms around dislocations, which also hinders their motion. In some cases the breaking away of a dislocation from the atmosphere can lead to consequences similar to those that follow when an accumulation of dislocations breaks through an obstacle. Dissolved impurities are of particular importance, since they help to lower the surface tension at the edges of incipient separation and thus facilitate the formation of cracks.

In the case of single crystals the idea of the accumulation of dislocations at obstacles encounters a number of difficulties, and is essentially inapplicable. First, it is impossible to imagine in a slip plane barriers of sufficient strength to hold back an accumulation of dislocations and cause the increase of the stresses to the values required for fracture. Second, it is hard to explain the growth of cracks in cases in which a slip plane coincides with a cleavage plane of a single crystal (for example, in zinc, cadmium, bismuth, and other metals with hexagonal or rhombohedral lattices). Even if accumulations of dislocations were formed in such crystals, they could not give rise to pulling stresses normal to the slip plane, and consequently could not cause the additional stresses required for fracture.

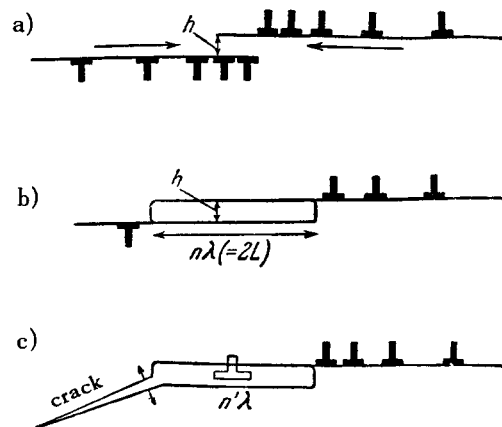


FIG. 21. Formation of a small embryonic crack through the annihilation of edge dislocations of opposite signs (Fujita¹³⁶): a) accumulation in parallel slip planes at separation h of two rows of dislocations of opposite signs; b) flat crack of height h and length $2L = n\lambda$ as the result of the annihilation of n pairs of dislocations; c) the flat crack corresponds to a single dislocation with Burgers vector $n'\lambda$.

Therefore quite a number of authors¹³⁴⁻⁻¹³⁷ look for other dislocation mechanisms that might be used to describe processes of fracture of single crystals. Orowan,¹³⁴ Kochendorfer,¹³⁵ and Fujita¹³⁶ have independently proposed the interesting mechanism of the formation of a thin flat embryonic crack by the annihilation of dislocations. An embryonic crack can be formed as a consequence of the combination of groups of edge dislocations of opposite signs. The cavity so produced is, however, unstable for the same reason that a Griffith crack is: it can close up with the production of a ring edge dislocation. Energetic stability exists only for such cavities of size 10^{-6} cm, which cannot lead to fracture of the crystal, since they do not produce a high enough concentration of stress.

To solve the difficulty Fujita suggests that we consider the possibility that groups of dislocations combine with an excess of dislocations of one sign. The result is not a simple crack, but a hollow edge dislocation with a Burgers vector given by $n'\lambda$, where n' is the number of excess dislocations in the cavity and λ is the Burgers vector of a single dislocation (Fig. 21). Replacing the resulting cavity of rectangular cross section by an elliptical hollow dislocation in the medium, with semiaxes $a = L$ and $b = (ah/2)^{1/2}$;^{*} Fujita calculated the distribution of tangential and normal stress over the surface of the cavity. The maximum pulling stresses act near the edge of the cavity in a direction close to the perpendicular to its surface (or to the slip plane), and are given by

$$\sigma_{\max} = \frac{Gn'\lambda}{2\pi(1-\nu)} \cdot \frac{1}{b}, \quad (14)$$

where G is the shear modulus and ν is Poisson's ratio.

When the Burgers vector $n'\lambda$ is taken comparable with the quantity b , the pulling stresses are close to $G/10$. With this mechanism the condition for fracture can be expressed by the equation

$$n'\lambda = b = \left(\frac{Lh}{2}\right)^{1/2} \quad (15)$$

The value of $n'\lambda$ cannot increase without bound, since along with its increase there is also an increase of the maximum tangential stresses, which facilitate the exit of "captured" dislocations from the other side of the cavity.

Thus it follows from Fujita's calculations that a hollow edge dislocation with cavity diameter 10^{-6} cm and Burgers vector of the order of several times the lattice constant is energetically stable and produces overstressing that reaches the strength of the ideal crystal at relatively small external loads.

According to Fisher¹³⁷ there is another possible mechanism by which plastic deformation can lead to

* The radius of curvature at the ends of the major axis, $r = b^2/a$, is taken equal to $h/2$.

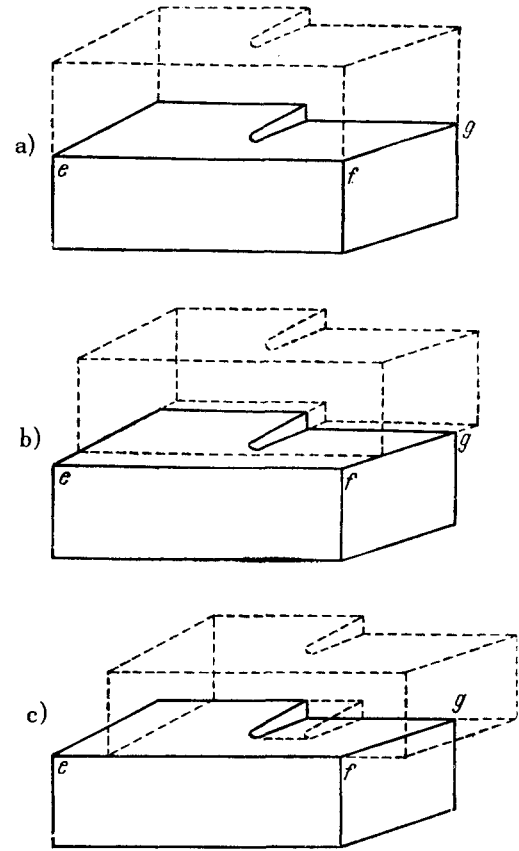


FIG. 22. Model of the formation of a thin cavity at the intersection of a slip plane by a screw dislocation (Fisher¹³⁷).

the formation of cracks. He has proposed a model of the formation of a very thin cavity at the intersection of a slip plane by screw dislocations. Here there is no need to assume that the motion of the dislocations is restrained by any sort of obstacle. Figure 22 shows the configuration of the crystal before the slip (Fig. 22a), after shear in the plane of the screw dislocation (Fig. 22b), and after shear perpendicular to the plane of the screw dislocation, as a result of which a crack has been formed (Fig. 22c). For the propagation of the crack it is necessary that there be sufficiently large normal pulling stresses, in accordance with the Griffith criterion [cf. Eq. (2)]; otherwise the crack can close up with the formation of two edge dislocations. The successive stages of the formation of the crack as shown in Fig. 22 concretize the role of the stresses tangential and normal to the slip plane.

A hydrostatic pressure superposed on the pulling stress can lead to the closing up of the crack formed at the intersection of screw dislocations, and consequently can increase the plasticity. An increase of plastic properties under conditions in which pulling and hydrostatic compression act together has been observed experimentally by Bridgman.¹³⁸ It can be expected that pulling in all directions at once would enhance the brittleness of a crystal.

When many identical screw dislocations in succession intersect a stationary screw dislocation at right angles, traces are produced in the form of compact chains of vacancies, forming a continuous layer of vacancies.

Green¹³⁹ tries to extend Fisher's analysis to the case of arbitrary intersections of dislocations in which the stationary dislocation has a component of its Burgers vector perpendicular to the slip plane of the movable dislocations. When a series of similar moving dislocations successively intersects a stationary screw dislocation, a close-packed layer of vacancies is formed, independently of the direction of the chain.

Thus fracture can be regarded as originating close to the place of localization of the vacancies, where cracks are formed which they are propagated, in accordance with the Griffith criterion (2), under the action of normal pulling stresses.

A paper by Stroh¹⁴⁰ discusses a different model of the formation of embryonic cracks in single crystals of hexagonal metals, in which cleavage along the slip planes is usually observed. Stroh has considered a mechanism of the origin of cracks and fractures that takes into account the concentration of stresses at the end of an interrupted wall of dislocations.¹³⁴ Such a bounded wall of dislocations can be formed from a complete wall passing through a crystal if part of it is held up by some obstacle and the other part is pushed farther by an applied stress (Fig. 23). If a vertical wall of dislocations that in its motion intersects an obstacle, for example a boundary of a block of the mosaic, is broken off, the ends of the wall may be joined by a crack lying in the slip plane.

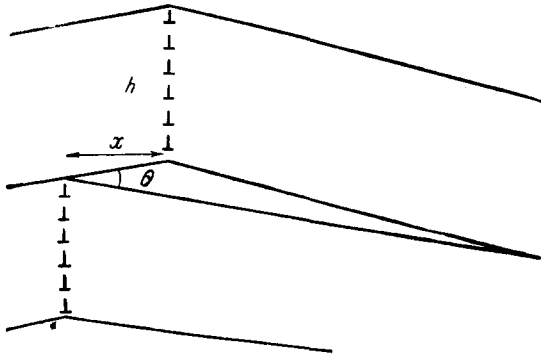


FIG. 23. Model of the origin of a cleavage crack at the end of a wall of dislocations broken apart inside a crystal (Stroh¹⁴⁰). x -- distance between the two parts of a vertical wall of dislocations; h -- distance between adjacent dislocations in the wall; $\theta = b/h$ is the angle of disorientation by the wall of dislocations; b is the Burgers vector.

The barrier required to stop part of the wall of dislocations is not as high as would be needed to stop the head dislocation of an accumulation, which is pushed

on by the others behind it. The barrier can be a small-angle edge, corresponding to a disorientation of 1° or even less. Cracks of this type arising in a single crystal of zinc under the action of a compressing load (Fig. 24) have been described by Gilman.¹⁴¹

The scheme proposed by Stroh corresponds to the condition for fracture

$$\sigma_s \cdot \sigma_n \cdot L = SA \ln \frac{B\theta}{\sigma_s}, \quad (16)$$

where σ_s and σ_n are the tangential and normal stresses applied to the slip plane, L is the extent of the wall of dislocations, θ is the angle of disorientation produced by the wall, S is the surface energy, and A and B are constants depending on the elastic moduli of the crystal. Equation (16) is derived on the assumption that the dislocations can be displaced under any arbitrarily small stress. It does not take into account the important resistances to the motion caused by intersecting dislocations, impurities, the Peierls-Nabarro force (σ_0), and so on.

Neglecting the weak dependence of the quantity $\ln(B\theta/\sigma_s)$ in Eq. (16) on the magnitude of the tangential stress σ_s , and taking into account the presence of the forces σ_0 , Stroh obtained the dependence of the strength of a specimen on its orientation in the form of the relation

$$\sigma_s - \sigma_0 = K \frac{\cos \chi}{\sigma_n}, \quad (17)$$

where K is a constant for a given diameter of the specimen, independent of the orientation of the crystals, and χ is the angle between the axis of the specimen and the slip plane.*

The linear relation predicted by Eq. (17) agrees satisfactorily with the experimental data on the strength of single crystals of zinc and bismuth.¹⁴² For example, for zinc an analysis of the experimental data gives $K = 6.2 \times 10^3$ (g/mm^2)² and $\sigma_0 = 55$ g/mm^2 , which agree with the calculated $K = 8.1 \cdot 10^3$ and the flow limit.

Recently E. D. Shchukin and V. I. Likhtman¹⁴³ have made an interesting theoretical study of the brittle fracture of single crystals of zinc. They assume that the origin of microscopic cracks and their development to dangerous size is due to previous plastic deformation, which because of nonuniformity of the formation of shears leads to sharp local concentrations of stresses.

Nonuniformity of the strain appears in the form of incomplete shear, which occurs owing to the presence

*It is assumed that the extent L of the wall of dislocations is connected with the specimen diameter D and the angle χ by the relation $L = D/\cos \chi$.

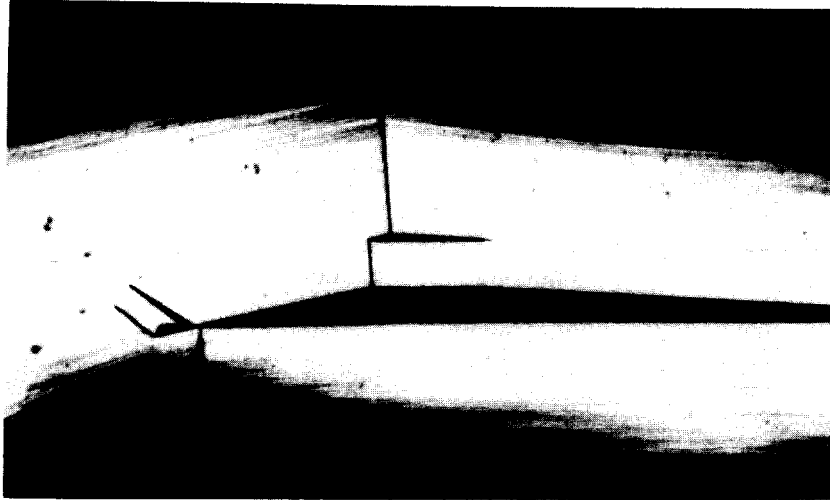


FIG. 24. Two cleavage cracks in a single crystal of zinc which were formed under the action of a compressing load (Gilman¹⁴¹).

of a sufficiently strong obstacle in the slip plane.** The incomplete shear is regarded as the formation of an accumulation of n dislocations with total extent¹²⁷

$$L = \frac{nGb}{\pi\tau(1-\nu)} \quad (18)$$

and the formation of a crack is treated as the result of fusion of dislocations. Shchukin and Likhtman show that cracks of sizes up to

$$2l = \frac{\beta\tau^2 L^2}{ES}, \quad (19)$$

can occur in the crystal, where L is the extent of the slip trace, τ is the applied tangential stress in the slip plane, and β is a dimensionless coefficient of value not very different from unity.

They obtained a new relation

$$\sigma_{\text{crit}} \cdot \tau_{\text{crit}} = \text{const} = K^2, \quad (*) \quad (20)$$

which can be defined as the condition of the constancy of the product of the normal and shearing stresses acting in the slip plane in brittle fracture.

Equation (20), unlike Sohnke's law ($\sigma_{\text{crit}} = \text{const}$) takes into account the plastic deformation that always precedes brittle fracture.

The dependence of the normal and tangential stresses in fracture at the basal plane on the orientation of

single crystals of zinc is described by the equations:

$$\sigma_{\text{crit}} = K \sqrt{\tan \chi}, \quad (21)$$

$$\tau_{\text{crit}} = K \sqrt{\cot \chi}, \quad (22)$$

where χ is the angle between the axis of the specimen and the slip plane. The experimental values of $\sigma_{\text{crit}}(\chi)$ and $\tau_{\text{crit}}(\chi)$ are in good agreement with the calculated values corresponding to Eqs. (21) and (22).

8. EFFECTS OF THE SURROUNDING MEDIUM ON THE MECHANICAL STRENGTH OF SOLID MATERIALS

The medium surrounding a specimen has a great effect on its strength. In early experiments of A.F. Ioffe with a solution of rock salt effects of the medium on processes of deformation and fracture of solid bodies were observed. P. A. Rebinder and his co-workers^{39,145,146} have made broad studies of the effects of the medium on the mechanical properties of various materials, and in particular the effects on their strengths.

The physical and chemical actions of the external medium can manifest themselves in the following main directions:

1) dissolving of the solid body with removal of the most dangerous surface defects, which must lead to an increase of the strength;

**Unfortunately the authors do not disclose the physical nature of these obstacles.

*) $K = \delta \sqrt{\frac{ES}{l}}$, $\delta \approx 1$.

2) adsorption.

P. A. Rebinder discovered in 1928 effects of adsorption in facilitating deformation and decreasing strength. These effects are due to the fact that by lowering the surface tension surface-active substances assist the production of embryonic shears and cracks, and by penetrating into surface microscopic cracks and being adsorbed on their walls such substances facilitate the further development of such defects. The phenomenon of adsorption is one of great generality. For example, the lowering of the strength of a metal as the result of direct chemical or electrochemical action of the medium is also due to a considerable extent to processes of adsorption. Adsorption effects are of a kinetic nature and therefore manifest themselves especially strongly in processes of creep, in tests of duration and of strength against fatigue, i. e., in cases when the time of interaction is long.

Large decreases of strength are caused by surface-active liquid metal media, which lower the surface tension at the boundary with the metal to values close to zero. As an example we may mention that the tensile strength of single crystals of zinc is lowered to one-tenth by the application to their surfaces of a layer of thickness as little as a micron of a liquid surface-active metal (for example, mercury at room temperature¹⁴⁷). The plasticity so characteristic of metal

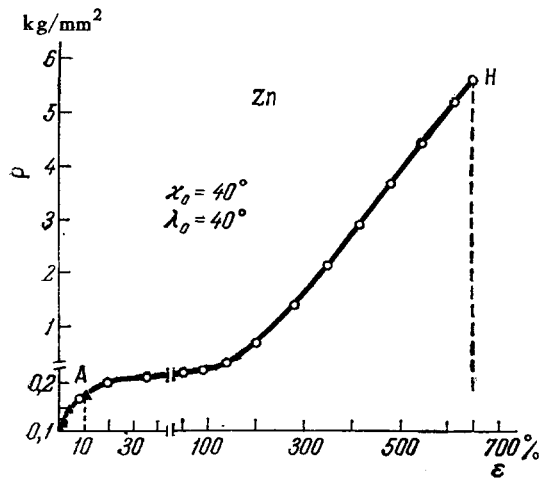


FIG. 25. Dependence of the true stress [$p = p_0(1 + \epsilon/100)$] on the fractional elongation ϵ for amalgamated (▲) and non-amalgamated (○) single crystals of zinc (Rozhanskii¹⁴⁷, Pertsov, Shchukin, and Rebinder¹⁴⁷):

χ_0 -- angle between axis of specimen and slip plane (0001);
 λ_0 -- angle between axis of specimen and direction of shear during slip. Rate of extension 15 percent per minute. A and H are the points corresponding to the rupture of the amalgamated and nonamalgamated crystals.

single crystals cannot manifest itself, and the fracture becomes brittle (Fig. 25). A study of axially cut specimens has shown that the decided decrease of the

strength of amalgamated single crystals of zinc is due to the formation of cracks both on the surface and throughout the volume of the crystal. The volume cracks arise as a consequence of the diffusion of the mercury into the zinc at room temperature and the lowering of surface tension on separation surfaces that arise internally.

The sharp decreases of strength and plasticity that a number of authors^{148,149} have observed in polycrystalline refractory and heatstable alloys affected by small quantities of low-melting impurities or by a liquid metal medium are also due to lowering of the interphase surface tension at newly formed surfaces.

P.A. Rebinder¹⁵⁰ has sought a relation between these phenomena and the probability of formation of embryonic shears and destructive cracks, in dependence on the value of the interphase surface tension S_{12} and the applied stress p , and obtained as the condition for fracture

$$S_{12} < \frac{kT}{\delta^2} + \beta p \delta, \quad (23)$$

where δ^2 is the average area of a block of dimension δ or the square of the average distance between embryonic defects of the structure, and β is a dimensionless factor characterizing the state of stress. For S_{12} much larger than kT/δ^2 (for this it is usually sufficient that S_{12} be not less than 1 erg/cm²) it follows from the condition (23) that the ratio of the strength P_0 in vacuum to the strength P_a in a given surface-active medium is equal to the ratio of the respective surface energies:

$$\frac{P_0}{P_a} = \frac{S_0}{S_{12}}. \quad (24)$$

Single crystals and polycrystalline metals are affected in quite a different way by readily adsorbed surface-active media that produce a comparatively small lowering of the surface tension of solids; examples of such media are solutions of organic acids and alcohols. Whereas strong adsorption-active media in the form of thin films or impurities of low-melting metals can cause the appearance of catastrophic brittleness and lowering of the strength by a large factor, the adsorption of organic substances makes crystals plastic and lowers the yield point.¹⁵¹

There are interesting papers^{39,44,152} in which the authors try to describe the mechanism of the action of adsorbed layers on the strength and deformation of solid materials. Shil'krut⁴⁴ assumes that the appearance of brittle fracture or plastic flow under the action of a surface-active medium is determined by the lowering of the surface tension and the ratio between the speed of deformation and the speed of two-dimensional migration of molecules (or ions) of the medium inside the microscopic cracks.

The effects of surface-active media on strength fit in well with ideas about dislocations. The emergence of dislocations onto the surface of a crystal is obviously hindered by surface tension. Therefore a small decrease of the surface tension (in the case of application of organic substances) can diminish the accumulation of edge dislocations at the surface and facilitate the motion of dislocations from sources inside the crystal to its surface. This must lead to a lowering of the yield point and to increased plasticity of the solid material.

According to the dislocation theory of accumulations the force acting on unit length of the leading dislocation reaches the value $n \tau b$ under the action of an external shearing stress τ , where b is the Burgers vector and n is the number of dislocations in the accumulation.¹⁵³ Even for small values of τ the leading dislocations can fuse and form the hollow nucleus of an embryonic crack.

From the dislocation point of view the effects of surface-active substances on strength can be explained by the penetration of these substances into the region of formation of an embryonic crack at an accumulation of dislocations. It is assumed that the transport of the substance from the surrounding medium occurs not only by regular diffusion, but also along the lines of the dislocations and especially along hollow dislocation nuclei. There is then a sharp decrease of the surface tension, which facilitates the formation and development of microscopic cracks both at the surface and in the interior. The lowering of the strength of iron when it is saturated with hydrogen, which in this case plays the role of a surface-active impurity,¹³³ can be explained in a similar way.

It must be remarked that in our opinion the penetration of surface-active substances into cracks and breaks between fragments of blocks must lead to similar results.

9. DEPENDENCES ON TEMPERATURE AND DURATION OF STRESS

The fracture of a solid body cannot be regarded as an instantaneous event that occurs immediately when the limiting load is reached. The idea of limiting strength agrees with the picture of the fracture of a brittle material proposed by Griffith, in which no consideration is given to the question of the duration of the fracture process after the stress has reached the critical value.

In any solid body there are always weak, defective places. With increase of the stresses and in the process of deformation these defective places are converted into ultramicroscopic breaks of continuity with fully developed surfaces of separation. Thus even in its earliest stages the deformation of a solid body is at the same time a process of its gradual breakage.

There are numerous facts¹⁵⁴⁻¹⁵⁷ that show that

the stress is not the only parameter that determines the strength and brings about mechanical fracture. For a given stress the strength depends on the length of time the material under test is kept in the stressed

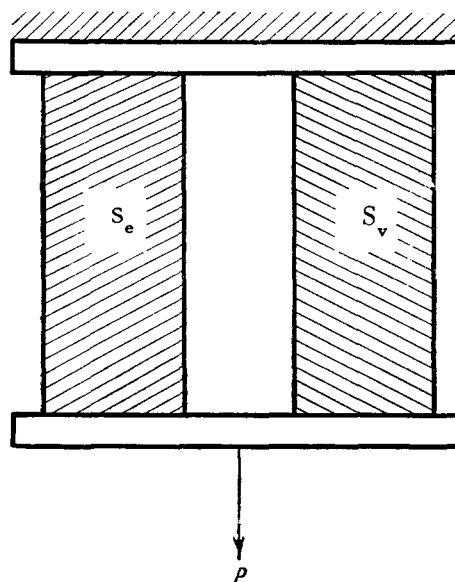


FIG. 26. Scheme of two-phase structure of brittle materials of the type of glass (Murgatroyd¹⁵⁹):

S_e -- elastic element; S_v -- viscous-plastic element

state, and on the number of changes of sign of the stress and the frequency of these changes. As a rule a specimen broken after deformation of brief duration appears considerably stronger. A deformation that has changes of sign or pulsations decidedly lowers the limiting strength.

The time dependence of the strength of solid materials has been widely studied. Orowan¹⁵⁸ proposed an explanation based on the assumption of the existence of Griffith cracks. In his opinion the lowering of the strength of glass with time is due to the adsorption on the surface of the glass of moisture from the air, which leads to a decrease of the surface tension. Orowan refers to papers by I. V. Obreimov and S. N. Zhurkov, who showed that the surface energy of some solids is several times as large in vacuum as it is in air.*

According to Orowan the ratio of the strengths of specimens for loads applied for short and long times is given by the equation

$$\frac{\sigma_{\text{short}}}{\sigma_{\text{long}}} = \alpha \sqrt{\frac{S_0}{S_{12}}} \quad (25)$$

*According to the data of I. V. Obreimov⁴⁰ the surface energy of mica is 19 times as large in vacuum as it is in air.

He regards fracture as a two-stage process. The first stage involves the slow growth of a Griffith crack, during which adsorbed material from the surface penetrates inside the crack and causes additional stress at its ends. When the total stress reaches the critical value, the disruption goes over into the second stage, which leads to rapid fracture.

On the basis of the hypothesis that brittle materials of the type of glass have a structure made up of quasi-viscous and elastic phases, Murgatroyd¹⁵⁹ postulated that the time dependence of strength is due to flow of the material in the quasi-viscous regions of the heterogeneous system. If such a system (Fig. 26) is subjected to tension, then in the course of time the stress in the elastic elements will increase on account of the relaxation of the stresses in the microscopic regions of the viscous-plastic type. When the con-

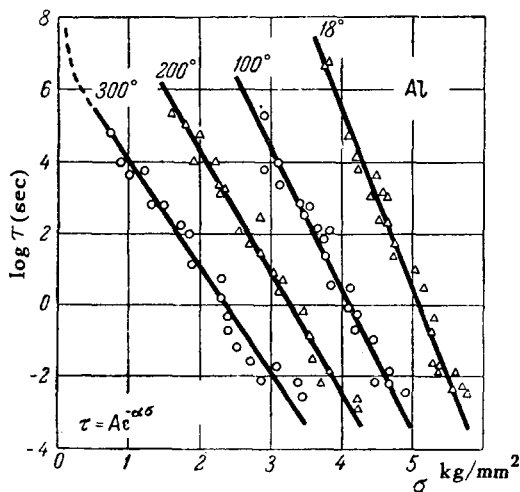


FIG. 27. Dependence of the endurance time τ on the stress σ for aluminum (99.99 percent) (Zhurkov and Sanfirova¹⁶²).

For a wide range of materials the endurance time τ under load obeys the general law

$$\tau = \tau_0 e^{\frac{U_0 - \gamma\sigma}{kT}} \quad (26)$$

where $U = U_0 - \gamma\sigma$ has the significance of an activation energy for the process leading to fracture, σ is the pulling stress, T is the absolute temperature, and τ_0 , U_0 , γ constant coefficients depending on the nature of the material.

The results obtained for the variation of the endurance time as a function of the stress σ (for $T = \text{const}$) and as a function of the temperature T (for $\sigma = \text{const}$) are shown in semilogarithmic coordinates in Figs. 27 and 28. It can be seen that the experimental data agree well with the straight-line relations.

centration of stress in the elastic elements reaches the limiting strength there is instantaneous fracture. The larger the applied load, the quicker the redistribution of the stresses from one phase to the other, and the shorter the lifetime of the specimen. These assumptions, however, are insufficient for the explanation of the observed dependence of the endurance time on the stress for various materials.

A detailed study of the temperature and time dependences of strength over a wide range of temperatures, stresses, and durations of the tests has been made by S. N. Zhurkov and his co-workers.^{112,160-162}

Zhurkov's experiments have shown that time dependence of the strength is an inherent property both of metals and alloys and of large-molecule amorphous substances, and is a general property of strength, not associated with any specific peculiarities of individual materials,

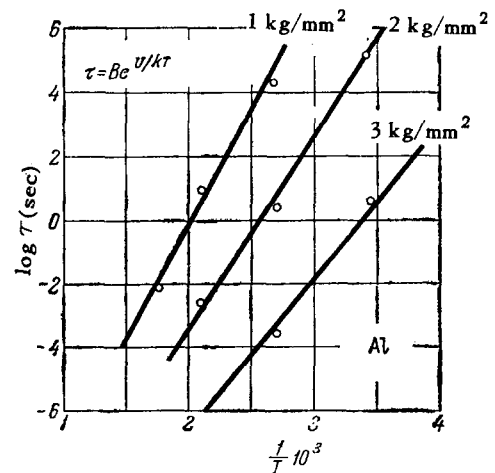


FIG. 28. Temperature dependence of the endurance time τ under constant load for aluminum (99.99 percent) (Zhurkov and Sanfirova¹⁶²).

The average speed of fracture, proportional to $1/\tau$, increases exponentially with increase of the stress and the temperature.

These experiments have made it possible to formulate a hypothesis of an activation mechanism for the fracture process, according to which the fracture of a body under load is regarded as a consequence of a process that develops in the material in the course of time. The duration of this process is determined by the temperature and the height of the activation barrier, $U = U_0 - \alpha\sigma$. The quantity U depends on the load and decreases with increase of the pulling stress (Fig. 29). It was found that the energy U_0 determined from Eq. (26) agreed with the value of the binding energy of the atoms in the crystal lattice of the pure

metal,* and the coefficient τ_0 of the exponential is constant for all metals and is of the order of magnitude of the period of the characteristic vibrations of the atoms in the lattice.

The endurance times of various materials can be determined on one hand by the time required for the formation of embryonic cracks, and on the other hand by the time for their growth. There are relatively few experimental papers on the kinetics of the origin and growth of fracture cracks, and the work reported is of a qualitative nature.^{113,114,154,163}

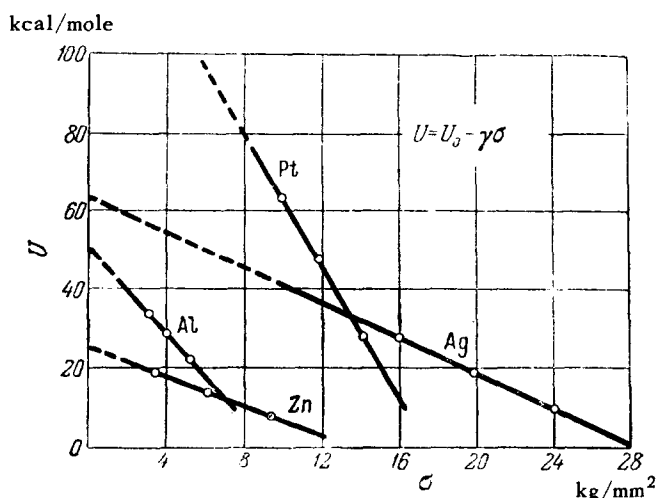


FIG. 29. Dependence of the activation energy U of the process leading to fracture on the tension σ for various metals (Zhurkov and Sanfirova¹⁶²).

A quantitative microscopic study by S. N. Zhurkov and E. E. Tomashevskii¹⁶¹ has shown that under conditions of uniaxial pulling with constant load the growth of microscopic cracks proceeds with acceleration and occupies the main part of the time until fracture. The rate of growth of the cracks rises with increase of the stress according to the law

$$v = v_0 e^{\frac{\beta\sigma_0}{1-\lambda/L}} \quad (27)$$

where λ is the length of a crack through the material, L is the width of the specimen, σ_0 is the mean applied stress, and v_0 and β are constants that depend on the nature of the material. The time until fracture calculated from the rate of growth of the cracks agrees well with the experimental data. In the opinion of these authors the dependence of the endurance time on the stress is determined by the kinetics of the growth of cracks.

B. Ya. Pines^{164,165} explains the time dependence of the strength by a slow "diffusion" growth of existing microscopic defects in the form of little embryonic

* For example, for zinc the binding energy is 27 kcal/mole and $U_0 = 25$ kcal-mole.

cracks until they reach the critical size required by the Griffith theory. He considers the molecular mechanism of the growth of embryonic cracks in connection with the presence of "vacancies" in the crystal lattice. The occurrence of large overstresses at the edges of a crack leads to a considerable decrease of the "activation energy" required for the replacement of the vacancies by atoms. This circumstance should decidedly accelerate self-diffusion and cause a sufficient "growing up" of the small embryonic cracks. The process of the preliminary growth of a crack is practically irreversible, since when the stress is removed the self-diffusion coefficient no longer has the higher value. The formula obtained by B. Ya. Pines for the endurance time of a body under tension agrees with the experimental results of S. N. Zhurkov.

A somewhat different vacancy mechanism of fracture has been proposed by I. Ya. Dekhtyar and K. A. Osipov.¹⁶⁶ They think of fracture at high temperatures as a continuously developing process of damage to the crystal structure in the "fracture zone." This zone, like the "site of fusion," is regarded as a certain number of disordered atoms surrounding a vacancy. Fracture begins when the critical concentration of holes in the dangerous zone reaches the value corresponding to the melting temperature.

10. FRACTURE INVOLVING CREEP

A particularly important question in the problem of the strength of metals at high temperatures is that of the nature of the fracture of metals when creep is involved. When there is creep the time dependence of the deformation is rather complicated. The second stage, however, goes with a constant speed. Processes of stationary quasi-viscous flow and of fracture involving creep are characterized by identical energy barriers and are closely related to each other.¹¹²

The character of the fracture is determined by the rate of creep, the temperature, and the form of the state of stress. Under conditions of simple pulling the question of greatest interest is that of intergrain disruption, since the main cause of lowering of the long-time strength is the formation and development of cracks along grain boundaries.

The microscopic mechanism of the origin and development of such cracks is considered from two points of view:

- 1) the initiation and development of cracks because of the concentration of stresses at the boundaries of grains owing to their relative displacement;¹⁶⁷
- 2) the appearance of cracks because vacancies accumulate at grain boundaries and unite into columns.¹⁶⁸⁻¹⁷⁰

It has been shown in a number of studies^{12,171,172} that in the deformation of a metal the grain boundaries

play the role of a viscous component, so that at sufficiently high temperatures and low rates of strain slip can occur along grain boundaries. According to a theory of Zener¹⁶⁷ the relaxation of the tangential stresses along grain boundaries that accompanies their displacement has the consequence that stresses become concentrated at places where three grains meet, and this leads to multilateral pulling and the formation of cracks.

More detailed physical descriptions of the mechanism of the origin and development of cracks during creep have been constructed in terms of the motion of vacancies. In an annealed metal there are not enough vacancies to form cracks. Vacancies get formed in the required numbers through the motion of dislocations in plastic deformation. The super-saturation with vacancies creates favorable conditions for their coagulation and precipitation on the surfaces of the embryonic microscopic cracks and pores already present in the metal. The most favorable conditions for the collection of vacancies occur near surfaces of separation: at the boundaries of grains and of blocks, slip bands, twinning planes, and so on. Intercrystalline disruption is observed at relatively small values of the stress and the rate of slip, when the accumulation of deformation occurs mainly on account of the displacement of grains. Separation of secondary phases from a solid solution, which most often occurs at grain boundaries, also assists the accumulation of vacancies at the boundaries and the formation of cracks.

The dislocation theory postulates that the avoidance of dispersed phases by moving dislocations leads to the intersection of dislocations and to intense vacancy production and the development of cracks at grain boundaries. In the case of a relatively large rate of creep (at high stresses) the disruption is localized along the slip planes, which are now the main form of plasticity and play the same role as grain boundaries with respect to the accumulation of vacancies and the formation of cracks.

Thus intragrain and intergrain disruption of metals during slip can be thought of as the result of the same processes of the formation of vacancies and their deposition at surfaces of separation, leading to the growth of cracks to the critical Griffith size.

The theory of modern alloys and of the development of materials that have great strength during creep is based on the idea of the necessity of producing obstacles to the motion of dislocations and suppressing the diffusive displacement of vacancies.

11. BRITTLENESS INDUCED BY COLD

It is well known that with certain plastic metals one finds that as the temperature is lowered there is a transition to a brittle state—brittleness induced by cold (cold shortness). The physical nature of this effect

still remains without an explanation.

In the search for an explanation of the transition from the viscous to the brittle state, resort is made to the results of A. F. Ioffe's study of rock salt.¹⁷³ Many workers have made repeated use of Ioffe's scheme to explain the effects of various factors on the cold-induced brittleness of iron and steel. According to this scheme the transition to the brittle state is due to the different temperature dependences of the brittle strength and the flow limit. The point of intersection of these two curves corresponds to the critical transition temperature that is ordinarily observed experimentally.

The critical temperature depends on the grain size of the structure, the chemical composition of the metal, the type of loading, and the test conditions. For example, with static loading of smooth specimens decrease of the grain size of ferrite to 20—30 microns makes iron highly plastic right down to liquid nitrogen temperature, whereas specimens with grain size 150—170 microns show brittle fracture at minus 196° C.¹⁷⁴ The effect of grain size shows itself mainly in the change of the brittle strength of iron. This increases sharply with decrease of the grain size of the ferrite.^{133,175} Changes of the rate of loading and particularly changes of the temperature affect the flow limit more strongly.

The experimental data indicate the presence of plastic deformation before brittle fracture, and that brittle fracture is connected with the action of tangential stresses. Therefore the curves of the temperature dependences of the flow limit and the brittle strength in A. F. Ioffe's scheme do not have to intersect.

The dislocation theory assumes that both the flow limit and the breaking stress are determined by the tangential stresses. For fracture to occur the following conditions must be fulfilled:

- 1) the tangential stresses in the slip plane of a grain must be large enough for the formation of accumulations of dislocations and the production of an elementary Griffith crack;
- 2) the stresses normal to the slip (or cleavage) plane that appear in an adjacent grain, caused by the action of an accumulation of dislocations at the grain boundary, must exceed the critical value at which growth of the crack occurs [cf. the Griffith formula, Eq. (2)].

In the dislocation model we can get an expression for the critical temperature for brittleness if we reduce the probability of brittle fracture to the probability of maintenance of the blocking of the dislocations.¹⁷⁶ The probability that liberation of the dislocations does not occur during a time t is given by

$$p = e^{-t/\delta}, \quad (28a)$$

where

$$\delta = \frac{1}{\nu} e^{U(\sigma)/kT}. \quad (28b)$$

Here $U(\sigma)$ is the activation energy necessary for the

freeing of a blocked dislocation, and depends on the stress σ acting on the dislocation; δ is the mean time that passes before the freeing of the dislocation (or the reciprocal of the probability per unit time for the freeing of the dislocation); ν is a constant with the dimensions of a frequency.

Assuming that for brittle fracture $p = 1$, we find from Eqs. (28a) and (28b) a formula for the critical temperature for brittleness:

$$T_{cr} = \frac{U(\sigma)}{k} \ln \nu t. \quad (29)$$

The experimental results confirm the relation described by Eq. (29) if we take p equal to the fraction of the specimens that undergo brittle fracture at a given temperature.¹⁷⁷

The dislocation theory enables us to get expressions for the dependence of the critical temperature for brittleness on the rate of strain ν and on the grain size d , in the form

$$\frac{1}{T_{cr}} = - \left(\frac{k}{U} \right) \ln \nu + C, \quad (30)$$

$$\frac{1}{T_{cr}} = - \frac{7}{2} \left(\frac{k}{U} \right) \ln d + C, \quad (31)$$

which also agree well with the experimental data.⁹²

At times it has been suggested that the tendency to cold-induced brittleness is determined by the type of crystal lattice: that metals and alloys with body-centered cubic and hexagonal lattices have this tendency, and those with face-centered cubic lattices lack it and retain rather high plasticity even under the most severe test conditions. But an increase of the plasticity observed for alkali metals (Li, Na) and tantalum,¹⁷⁸ especially at temperatures below minus 196° C, has cast doubt on the correctness of this connection of cold-induced brittleness with lattice type, since these metals have body-centered cubic lattices. The recently discovered phenomenon of the low-temperature polymorphism of metals^{98,179} has thrown light on the cause of the "anomalous" plastic properties of a number of metals at low temperatures.

Temperature polymorphism is ordinarily observed at rather high temperatures, under conditions in which the atoms of the crystal lattice are easily rearranged into a new and thermodynamically more favorable structure. At low temperatures also metals can possess stable structures that differ from those observed under normal conditions. In this case, however, the transition is practically suppressed because of the small diffusive mobility of the atoms, and even when kept for a long time at the low temperature the metal remains in an undercooled state with the structure that corresponds to higher temperatures.

It has been found that the alkali metals (and quite a number of other metals that do not have close-packed

structures) have a low-temperature modification--face-centered cubic--and make the transition to it under the influence of a small plastic deformation. The deformation facilitates the overcoming at low temperatures of the activation barrier to the rearrangement from one modification to the other. The considerable plasticity at low temperature is due both to the occurrence of the allotropic transformation during deformation and to the high plastic properties of the low-temperature modification that is formed.

The connection of the cold-induced brittleness of metals with the crystal-lattice type can be explained in terms of the different temperature dependences of the flow limit. The experiments show^{176,178} that the flow limit rises sharply as the temperature is lowered in the case of metals with body-centered cubic lattices, but not for those with face-centered cubic lattices (Fig. 30). There are also decided differences between the strain diagrams of metals with different crystal lattices. For example, the compression diagram of iron (Fig. 30) shows a fourfold increase of the flow limit without change of the hardening coefficient as the temperature is lowered from +20° to -183°. In the case of copper, on the other hand, the flow limit was practically unchanged, and the hardening coefficient increased at low temperatures (Fig. 30).

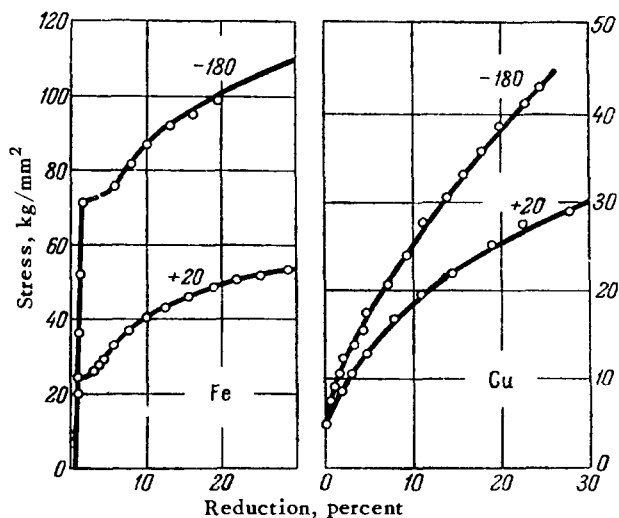


FIG. 30. Compression diagrams of iron and copper at room temperature and liquid-air temperature (Wellinger and Seifert¹⁸⁰).

Although there are a number of hypotheses,¹⁷⁸ the causes of the temperature dependence of the flow limit for cold-brittle metals in the low-temperature region are not yet known. A widely accepted hypothesis is that there is a causal connection between cold-induced brittleness of metals and their tendency to mechanical twinning. It is well known that twinning is a characteristic feature of the deformation of metals with body-centered cubic lattices. The transition from the plastic

to the brittle state is usually accompanied by a change of the mechanism of deformation--the replacement of slip by twinning. A number of experimental researches have shown that conditions that favor mechanical twinning at the same time favor brittle fracture. For example, higher rates of deformation and larger grain sizes lead to increase of the critical temperature for cold-induced brittleness, and also facilitate the formation of twinning layers.

The fact that cold-induced brittleness and twinning have the same temperature limit is commonly regarded as one of the main arguments in favor of the twinning theory. The twinning theory of cold-induced brittleness regards the formation of a brittle crack as the result of the effect of high local pulling stresses caused by the wedging apart of twinning layers.¹⁸¹ Studies in recent years^{174,182--187} have shown, however, that twinning in itself is not the cause of brittle fracture.

In reference 174 a study has been made of the effect of grain size on the conditions for the formation and development of twinning layers in technically pure iron (0.06 percent carbon). Under static pulling in liquid

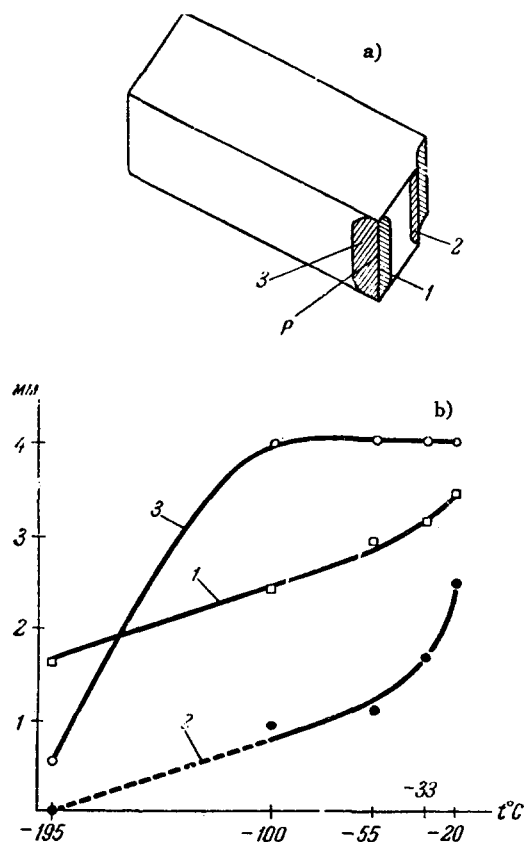


FIG. 31. Extent of the region of deformation by twinning at different temperatures in impulsive tests on iron (b), for the regions 1, 2, 3 indicated on the diagram of half a specimen (a) (Garber, Gindin, and Konstantinovskii¹⁷⁴).

Iron with homogeneous fine-grained structure (grain size 30 microns)

nitrogen specimens with structure homogeneous in grain size (150--170 microns in diameter) were fractured brittly without twinning. Specimens with sharply inhomogeneous structure (grain diameters from 20 to 600 microns) showed the presence of twinning after brittle fracture, but the crack ran through grains that did not contain twinning. With impulsive bending of small grain specimens (grain size 15 to 30 microns) the twinning was localized in the regions of plastic denting under the hammer (see regions 1 and 3 in Fig. 31), whereas in the place where the brittle cracking started (at the bottom of the preliminary notch, region 2 in Fig. 31) no twinning was observed at liquid-nitrogen temperature. At higher temperatures, when the specimens showed plastic bending, twinning layers appeared in this zone also. These experiments show that the occurrence and intensity of twinning are determined by the degree of concentration of stress in a grain, which depends on the size of the grain, the ratio of sizes of adjacent grains, and also the manner of application of the load and the temperature.

In experiments by N. N. Davidenkov and T. N. Chuchman¹⁸⁵ on impulsive bending of notched specimens of tantalum (at minus 269°) and of smooth specimens of Armco iron (at minus 196°) there was abundant twinning without brittle fracture. The authors draw the conclusion that even under the most severe conditions for cold-induced brittleness--low temperature, pulling stresses, preliminary cuts, and high rates of loading--the appearance of twinning does not necessarily bring with it brittle fracture.

Most studies of cold-induced brittleness have been made with technically pure metals. It is well known that admixtures of elements that form solid solutions by implanting (for example nitrogen, carbon, oxygen) help to increase the brittleness of metals with body-centered cubic lattices.¹⁸⁸ The effect of impurities is ascribed to the blocking of dislocations as in Cottrell's theory. Because of the small size of the pores in the body-centered cubic lattice, as compared with other types of lattice, implanting of atoms in it causes stronger distortions and leads to the appearance of tetragonal properties. The result is that by removal of impurities from brittle metals one can suppress the appearance of cold-induced brittleness. For example, Wain and his coworkers¹⁸⁸ have produced plastic chromium by removal of traces of nitrogen and carbon. Large-grained iron, produced by zonal casting¹⁸⁹ or by distillation in vacuum,^{186,187,190} also possesses plasticity at low temperatures, right down to $4.2^{\circ}K$.

Plastic fracture of iron of high purity is preceded by the formation and growth of twinning layers of thicknesses up to 20 microns, unusually thick for iron.¹⁸⁶ Technical iron of the same grain size, after brittle fracture at these temperatures, contains twinings of extraordinarily small thickness, not exceeding a micron.

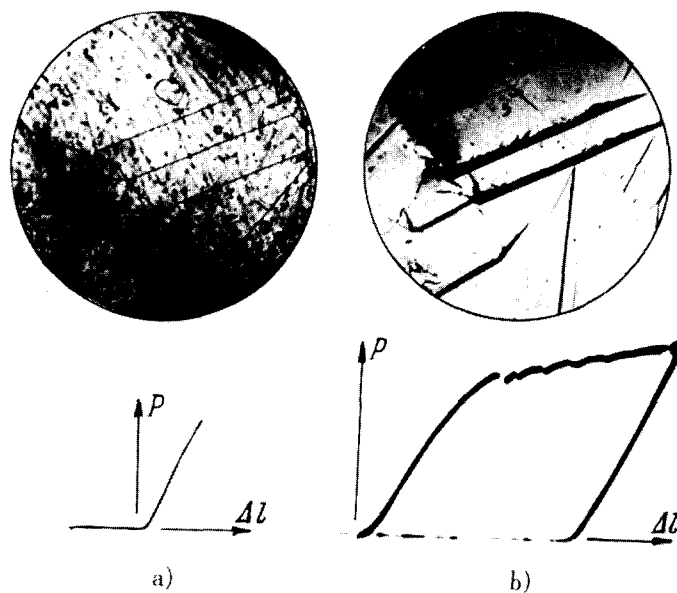


FIG. 32. Static pulling diagrams and microstructures of twinning layers for specimens of technical (a) and pure (b) iron (Garber, Gindin, and Konstantinovskii¹⁷⁴) (Garber, Gindin, and Starodubov¹⁸⁷).

Temperature during tests – 196° C. Grain size 150 to 170 microns. Magnification 40 diameters. (In the diagram the pulling of the pure iron was not carried to rupture.) Tensile strength of technical iron, 61 to 64 kg/mm²; tensile strength of pure iron, 40 to 41 kg/mm². The fracture of the technical iron was brittle, that of the pure iron, plastic (elongation up to 8 percent).

Figure 32 illustrates the radical difference between the static pulling diagrams and the twinning microstructures of specimens of large-grained technical and pure iron tested in liquid nitrogen. The simultaneous occurrence of twinning and cold-induced brittleness in technical iron can be explained by the harmful influence of impurities, which decidedly raise the twinning flow limit and strengthen the boundaries of the twinning layers. This must happen through the accumulation of the impurities in the boundary zones during mechanical twinning, by a mechanism like that of zonal purification. This must lead to limiting of the development of the layers and to the formation of cracks along planes of both basic and secondary cleavage.

A difference between this process and zonal purification is that at very low temperatures the activation of the impurities occurs on account of the energy of the mechanical work of deformation. Evidence in favor of the mechanism of removal of impurities from the metal during twinning and their accumulation at the moving boundary of the twin is found in the properties of single crystals of beryllium twinned throughout their entire

volume; in tests at 400° C such crystals showed increased plasticity and increased strength against subsequent basal slip.¹¹⁸

By stepwise low-temperature deformation of large-grained pure iron, with warming to room temperature between the steps, it has been possible to study the kinetics of the development of twinning and cracks. Twinning layers arose over the entire length of the specimen already at small fractional stretches, and their thickness increased greatly during the stretching process (Fig. 33). The formation of the first twinings was noted at stresses much smaller than the tensile strength.*

In the experiments with pure iron one could directly separate the temperature thresholds of cold-induced brittleness and of twinning and show that the production of the first microscopic cracks is not directly related

* In the tests in liquid nitrogen the tensile strength was 41 to 42 kg/mm²; twinning appeared already at stresses 17 to 18 kg/mm².

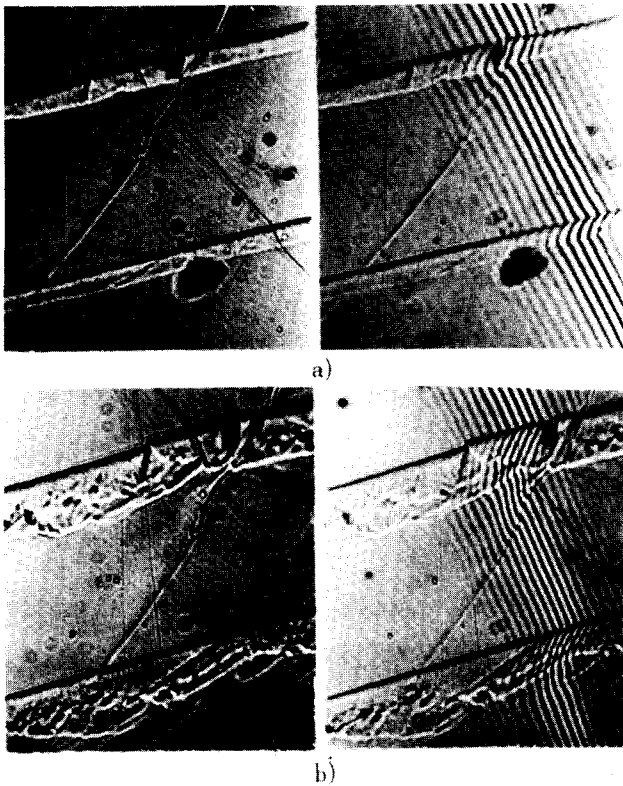


FIG. 33. Microstructure and interference pattern of the surface of a specimen of high-purity iron (99.99 percent) subjected to stepwise deformation at -196°C (Garber, Gindin, and Starodubov¹⁸⁷):

a) after 1 percent deformation; b) after intermediate warming (to 20°C) and additional deformation by 4 percent at -196°C . ($\times 440$)

to the appearance of twinning. The temperature for cold-induced brittleness of the iron in question lies below 4.2°K , whereas the threshold for twinning is above 77°K . At 4.2°K fracture occurs as a rule along cleavage planes; at 77° , with slip being more strongly developed, fracture occurs both along cleavage planes and along slip planes.

In many cases microscopic ruptures lead to such a redistribution of the local stresses at the vertex of the crack that there is production of wedge-shaped twinning layers with the broad base at the crack. In Fig. 34 one can see layers 2 that have appeared after the production of the crack 1 and are distributed along the slip bands.

Many observations show that fracture at low temperatures is stimulated by hardened slip bands to a greater extent than by twinning layers. This is in good agreement with the hypothesis that in the last stages of deformation there is pulling loose in the regions of slip bands, with subsequent partial reestablishment of contacts.¹¹⁵

Work done in recent years^{87,192} has revealed that metals with face-centered cubic lattices (Cu, Al) also twin at low temperatures, and in spite of this have considerable plasticity under pulling stress.

The absence of brittleness in twinned face-centered metals, and also the increase of plasticity in pure large-grained iron at low temperatures on account of twinning, are indications that mechanical twinning is not responsible for cold-induced brittleness. The tendency of a metal to such brittleness is determined by the type of crystal lattice and the impurities.

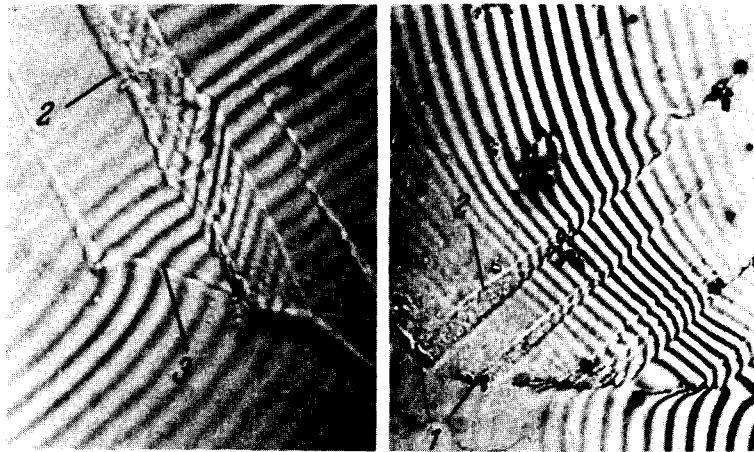


FIG. 34. Wedge shaped twinning layers in iron (99.99 percent) that have appeared at the vertex of a crack (Gindin and Starodubov¹⁸⁶):

Temperature during test, -196°C ; fractional elongation, 3.0 percent; 1—microscopic crack; 2—twinning layers; 3—slip band (the crack extends along a slip trace). ($\times 440$)

12. THE STRUCTURE OF HIGH-STRENGTH ALLOYS

Mainly as the result of x-ray studies by G. V. Kudryumov and his coworkers,¹⁹³⁻⁻¹⁹⁶ many details of the crystal structure of high-strength metals and alloys are now known.

Ordinarily a state with high strength arises as the result of plastic deformation, of the dissociation of supersaturated solid solutions under conditions of low mobility of the atoms, or also in tempered alloys annealed at low temperatures. All of these cases of hardening are characterized by the production of distortions of the crystal lattice and by the breaking up of the crystallites into blocks. In the case of alloys the hardened state is also associated with dispersed precipitation of the second phase, which leads to a fine-grained, submicroscopically inhomogeneous structure of the alloy. The mechanical strengths of metals and alloys can vary over wide ranges through changes of structure, and can reach large values as the result of the production of submicroscopic inhomogeneity of the crystal structure. The limiting strength of crystalline materials is determined, however, by the strength of the interatomic binding forces in the crystal lattice. The strength of the interatomic bonds in crystals and the small-scale structure of solid materials are the two closely interrelated factors that determine the resistance of a material to deformation and fracture. It is just these factors that fix the reserves available for raising the strengths of metals and alloys.

Under actual conditions the strength of the interatomic bonds is utilized only to a very slight extent. The degree of its utilization can be raised a great deal by producing a definite small-scale crystalline structure that leads to an increase of the "degree of simultaneity" with which the atoms take part in the opposition to the action of external forces.

One basic structural characteristic is the size of the blocks (regions of coherent scattering). In tempered or high-annealed alloys the blocks are relatively large ($\sim 10^{-4}$ cm) and the angles by which they are out of orientation are extremely small. But in alloys that are in a hardened condition (quenched, low-annealed, aged, or cold-worked) one finds that the blocks are much smaller and the disorientation angles larger. For example, in quenched steel containing 1 percent carbon the size of the blocks is about 2×10^{-6} cm.¹⁹³ Decrease of the block size begins at extremely small deformations. As the degree of deformation increases, the breaking up of the blocks falls off, and at a certain value of the strain, for example about 30 percent for iron, it practically ceases. Data given by various authors^{193,197,198} show that in strongly deformed iron or low-carbon alloy steel the blocks get reduced to a size of 1.3×10^{-5} to 5×10^{-6} cm.*

Experimental data on the dependence of the flow

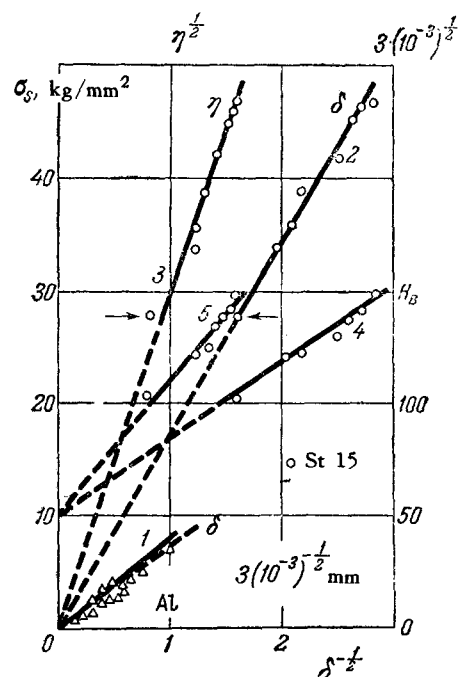


FIG. 35. Values of the flow limit σ (curves 1, 2, and 3) and the hardness H_B (4 and 5) plotted against the block size δ (1, 2, and 4) and the inhomogeneity $\eta = \Delta a/a$ of the distance between planes in the lattices of iron and aluminum (Rovinskii¹⁹⁹ and Rybakova²⁰⁰).

limit and the hardness of a metal on the size of the blocks are given in references 199 and 200. These results, obtained by Ball for aluminum¹⁹⁹ and by B. M. Rovinskii¹⁹⁹ and L. M. Rybakova for low-carbon steel,²⁰⁰ are shown in Fig. 35. There is a good linear dependence between the flow limit and the hardness on one side and the block size taken to the $-\frac{1}{2}$ power on the other.

Linear extrapolation takes the lines of $\sigma_s(\delta)$ through the origin, which shows that the flow limit of a metal goes to zero with increase of the size of the blocks.

When a metal that has been hardened is heated there is an increase of the block size; rapid growth of the blocks begins only above a certain temperature, for example in quenched steel at annealing temperatures above 450°C and in wrought iron above $350\text{--}550^\circ\text{C}$, depending on the composition.

Studies made recently in one of the laboratories of G. V. Kurdyumov have shown that the beginning of the growth of the blocks is accompanied by a marked decrease of the mechanical characteristics: the hardness, the flow limit, the strength (Fig. 36). For alloys Fe-Si

* Wood¹⁹⁷ and others have determined the stable size of the blocks broken up by plastic deformation for a number of metals: for copper, 0.7×10^{-5} cm; for aluminum, 10^{-4} cm; for molybdenum, 0.42×10^{-5} cm; etc.

and Fe-Ni and for nickel this work also gave results that indicate that distortions of the second kind do not cause hardening, but only accompany it. These distortions can be completely removed by annealing and produced again by suitable deformation without causing any change of the strength characteristics.*

The presence of distortions of the third kind is also not essential for high resistance to deformation. This has been shown clearly for specimens of pure iron quenched from the martensite state.²⁰¹ In this case no distortions of the third kind were found, but the hardening, block size, and distortions of the second kind were

at the same level as in the case of cold-worked iron.

An important part in determining resistance to deformation and fracture must be played by the amount of disorientation between blocks, and also by the total disorientation angle of a crystal. With increase of the degree of deformation the disorientation angle of the blocks increases, and in strongly work-hardened metals reaches the order of one degree. Large distortions are localized at the block boundaries. At present, however, we have no data on the contribution of the angular characteristics of the small-scale structure to the strength of metals and alloys.

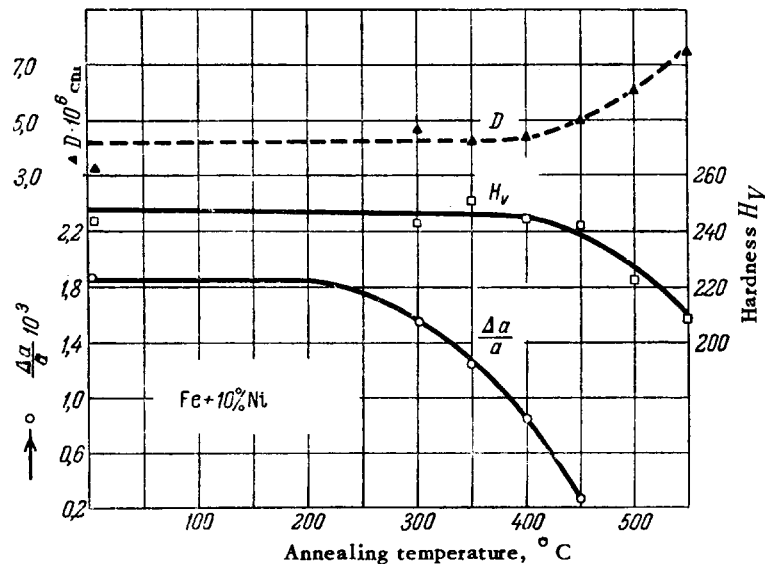


FIG. 36. Dependence of the hardness H_V , the inhomogeneity $\Delta a/a$ of the distance between planes (distortions of the second kind), and the block size D on the annealing temperature of a deformed iron-nickel alloy (Kurdyumov, Perkas, and Khondras²²³).

Generalizing from the results of many studies by G. V. Kurdyumov and his coworkers, B. M. Rovinskiĭ, and others,¹⁹³⁻⁻²⁰¹ we can draw the conclusion that the size and orientation of the blocks and the state of their boundaries determine the strength and other important mechanical properties of crystalline materials. Reduction of the block size is essential in any hardening process, whether thermal or mechanical, and must be regarded as one of the main crystal-structure factors in hardening. The level of improvement of the strength properties of materials that can be achieved is obviously limited by the value of the limiting size of the blocks obtained by plastic deformation or as the result of heat

* It is our opinion, however, that in some cases, for example in fatigue tests, distortions of the second kind can predetermine the character of plastic deformation and thus affect the hardening of an alloy.

treatment. It may be supposed that the structure of a crystal as composed of blocks fixes the value of the possible local overstresses, and the interatomic bonds fix the "theoretical strength" of a solid material.

The strength of the interatomic binding plays an especially important role in the behavior of metals and alloys at high temperatures, since it is on it that the position of the temperature range in which an alloy loses its strength depends. In fact, the mobility of the atoms in the lattice is determined by the ratio between the energy of the thermal vibrations and the height of the potential barriers, and the latter depends mainly on the strength of the interaction between atoms.²⁰²⁻⁻²⁰⁵

An analysis of the diffusive characteristics of alloys of various metals gives particularly clear evidence of the correspondence between high-temperature strength and the mobility of the atoms (Fig. 37).²⁰²

Large changes in the interatomic binding forces in

a crystal lattice can be produced by alloying; the strength of the binding forces may be either increased or decreased, depending on the nature of the alloying element.^{195,206,207} The concentration of the alloying component is also an important factor.^{208,209} For example, V. A. Il'ina and V. K. Kritskaya¹⁹⁵ found that manganese, niobium, molybdenum, and cobalt increase the forces of interaction between the atoms in the α -iron lattice, and vanadium decreases these forces (Fig. 38).

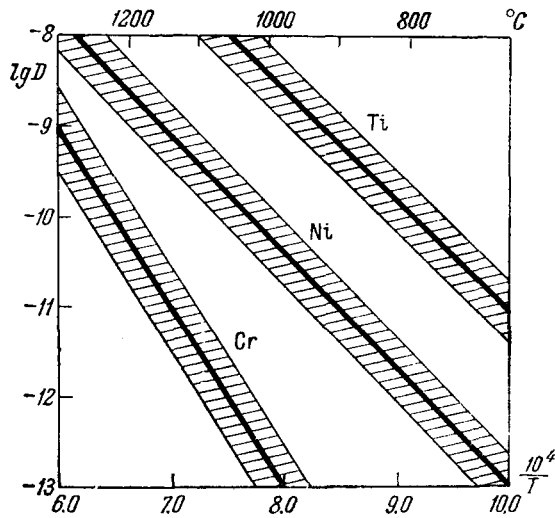


FIG. 37. Ranges of variation of self-diffusion coefficients as functions of temperature for alloys based on titanium, nickel, and chromium (Gruzin, Kurdyumov, Tyutyunik, and Entin²⁰²).

Besides changing the binding forces, alloying also serves the purpose of creating conditions for the production of high-strength states of alloys through dispersion hardening and the formation of a small-block structure owing to the precipitation of particles of a new phase.

Sometimes an increase of strength occurs under conditions in which the binding forces in the lattice are decreased (for example, martensite), but in these cases the increased strength is due to special structural states. An example of a great increase of strength (up to 350 kg/mm²) with unchanged binding forces is a patented steel wire produced by thermal and mechanical processing of a highly dispersed homogeneous laminar structure.

In pure metals (Fe, Mo, Ta, W)^{195,210} various changes of the fine-structure state by plastic deformation and heat treatment do not in themselves produce changes of the magnitude of the interatomic binding forces.

This conclusion cannot be regarded as final, however, since x-ray analysis provides measurements only of the volume average of the amount of static distortion of the crystal lattice.

V. I. Startsev and P. N. Aronova²⁰⁴ have studied the state of the crystal lattice in an isolated slip band of a crystal of rock salt. It was found that in the slip band the interatomic binding forces are much smaller than in the undeformed crystal, since the Debye temperatures* calculated from the x-ray data are almost twice those found in undeformed crystals. The lattice

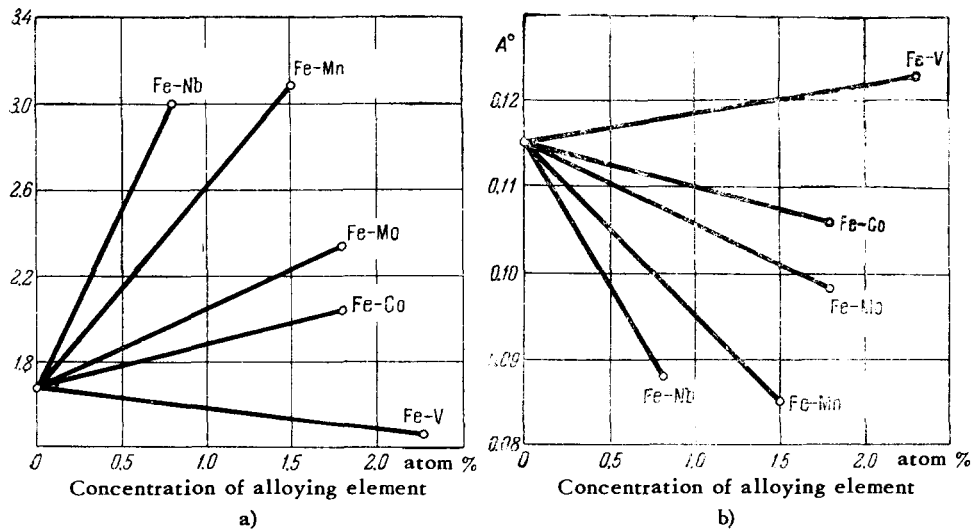


FIG. 38. Variation of quantities characteristic of the interatomic binding forces--a) $r\theta^2$ and b) $(u_{dyn}^2)^{1/2}$ --of alloys based on iron as functions of the nature and concentration of the alloying elements (Il'ina and Kritskaya¹⁹⁵).

in the slip bands is characterized by the appearance of large static displacements of atoms and is in a loosened-up state. This agrees with earlier observations by the Laue method, which showed strong distortions of the

* X-ray methods for studying the interatomic interaction forces in crystals evidently stand in need of more thorough theoretical analysis. This question is not discussed in the present review.

crystal lattice at boundaries between blocks. The total static and dynamic displacement of atoms in a slip band at room temperature (0.54 Å) is about three times the displacement of atoms in an undeformed crystal at the same temperature (0.18 Å).

Plastic deformation is not just localized in the slip bands; it also extends to regions at a rather large distance from them.* According to the data of reference 204 the characteristic temperature for an undeformed rock-salt crystal is 260 to 290°C; in a slip band it is 150 to 170°C, and in the undeformed region near the band it is 190 to 210°C. In crystals of solid solutions (Ni-Cr, Fe-Cr, Fe-W, Fe-Mn, etc.) the binding forces can be decidedly changed not only as the result of the alloying itself, but also in ways that depend on the type of heat treatment, and also by plastic deformation.^{206,208,194,210}

In a number of solid solutions²¹²⁻⁻²¹⁴ there are anomalies in the variation of some physical properties with the temperature (for example, the electric resistance, the moduli of elasticity, the hardness, and so on), although these alloys remain one-phase solid solutions over the entire temperature range. The most probable view is that both the change of the interatomic binding forces and the observed anomalies in the physical properties of solid solutions have a common nature and are due to the same process occurring in very small volumes of the crystal lattice. This process is of a diffusive character and is due to the occurrence under definite conditions of temperature of inhomogeneities of concentration in the one-phase solid solution. It evidently consists of a redistribution of the atoms in the lattice of the solution and of associated changes of the electronic interaction of the atoms. Therefore an increase of the binding forces must correspond to an increase of the degree of short-range ordering. This is confirmed by the results of references 2.5 and 2.6, in which there was a decrease of the degree of short-range ordering owing to plastic deformation.

An important role in hardening is played by the condition of the boundaries of grains and blocks. The regions of a crystal that are directly adjacent to grain or block boundaries are characterized by considerable distortions of the lattice. Although these boundary zones make up a small fraction of the total volume of the metal (~10 percent), they have a strong influence on many of its properties. The special state of the matter in the boundary layers is clearly manifested in measurements of the relaxation of stresses, of the dynamic elastic modulus, of creep, of internal friction, and of other properties. The changes of atomic mobilities

and diffusion rates in the zones near boundaries are of particular importance. For example, in zinc and silver self-diffusion occurs 10^4 to 10^6 times as fast along grain boundaries as it does through the grains themselves. As has been pointed out already, this leads to the occurrence of intergrain deformation and to ruptures along grain boundaries at high temperatures.

The importance of the grain boundaries is even greater owing to the fact that in alloys their chemical composition is usually quite different from the average composition throughout the grains, both on account of crystalline eliquation and in particular on account of internal adsorption.²¹⁷⁻⁻²¹⁹ The phenomenon of internal adsorption in solid materials consists of the appearance of nonuniformities of concentration of the dissolved components owing to structural nonuniformities of the solid solution. The effect of internal adsorption in solid solutions is influenced by the character and degree of the disorientation of the crystals at their contacts with each other, by the condition of the boundaries of the blocks of the mosaic and of the places where the subcrystalline structures are linked together that have arisen in the process of aging of the basic solution, and by other factors.²¹⁹

Intercrystallite internal adsorption goes on in the intercrystallite transition zones, whose thickness is only a few hundred times the interatomic distance.* Here a very great part is played by extremely small amounts of impurities, if they are surface-active. The phenomenon of internal adsorption and the special properties of layers adjacent to grain boundaries are of very wide occurrence in alloys and strongly affect their strength properties.

13. FATIGUE

The lowest resistances to fracture are observed with loads changing in sign when the changes are carried through a large number of cycles. Phenomenologically this effect is described as the engendering of a local rupture and subsequent gradual extension of the crack as the number of load cycles is increased. It is noted that in spite of low stresses, which are clearly below the flow limit in static tests, during fatigue tests there appear in certain regions of a specimen evidences of plastic deformation in the form of systems of slip bands. Fatigue cracks appear subsequently in just those regions where the slip traces are localized. Obviously a region of localization of slip bands is rich in embryonic cracks. The actual appearance of cracks can be ascribed to the effects of the process of plastic

* In its initial stage, plastic deformation of single crystals of beryllium¹¹⁹ is concentrated in regions adjoining the slip band; only in later stages does the deformation directly take in the volume of the slip band itself.

* Methods for detecting and studying internal absorption are based on the change of the ease of etching of the solid solution with change of the concentration of the dissolved element and on the use of radioactive isotopes.²¹⁸

deformation, which, as has been pointed out, is characterized by the breaking up and subsequent reestablishment of contacts. When there are reversals of sign of the deformation this feature of the process becomes more pronounced, so that packets of slips, and also cracks, are formed at lower stresses than in static deformation.

Direct proof of the occurrence of irreversible processes of the type of plastic deformation with hardening in microscopic regions during elastic vibrations has been obtained from experiments on the decrement of torsional vibrations of lead tubes, the gas pressure inside the tube being raised at the same time.²²⁰

From the results of experiments with single crystals of rock salt it has been established that under repeated plastic bending in alternate directions the hardening increases with the number of cycles and reaches a stable value after several hundred cycles.²²¹ Repeated reversals of twinning in a single twinning layer in calcite has also been found to lead to an increase of the hardening, which was terminated after 30 or 40 cycles by fracture along the secondary cleavage plane.²²²

Thus the appearance of cracks in fatigue fractures is a consequence of disruptions of structures in microscopic regions during plastic deformation, which is combined with hardening of the deformed regions of the specimen.

14. HARDENING

Since the strengths of plastic materials are mainly determined by their flow limits, methods for hardening, i. e., for raising the flow limit in plastic deformation, deserve special attention.

Recently R. I. Garber has shown that it is useful to consider two types of hardening:²²²

- 1) hardening that arises from the breaking up of blocks, as a result of which deformation is hindered;
- 2) hardening that arises from the diffusive redistribution of vacancies and impurities and their localization on slip bands or other separating boundaries produced during plastic deformation.

Mechanical breaking up of blocks occurs with stepwise application of loads and under conditions of nonequilibrium deformation, when the rate of increase of the stresses exceeds the rate of increase of the flow limit. In this case there is acceleration of the deformation, and the hardening is of an unstable nature and can lead to embrittlement.

Diffusive hardening can occur under smooth loading and rather small rates of increase of stress, at which there is approximate equality between the stress and the flow limit. Under these conditions the process of deformation proceeds in equilibrium, and increase of stress is accompanied by increase of the flow limit, the rate of deformation remaining constant. In diffusive

hardening the raising of the flow limit is combined with a considerable increase of the hardness and durability. From this point of view a particularly interesting phenomenon is that of stabilized creep, in which with constant load there is a prolonged deformation that proceeds with constant speed. The mechanism of this process can be described as the motion of a boundary layer with a speed equal to the rate of diffusive displacement of the vacancies that follow the shifting separating boundary; the rate of stabilized creep is then given by

$$\dot{\nu} = \nu \cdot d, \quad (32)$$

where d is the distance between planes and ν is the frequency of displacement of ions. Assuming that $\nu = \nu_0 \exp \{-Q/RT\}$, we get instead of Eq. (32)

$$Q = RT \ln \frac{\nu_0 d}{\dot{\nu}}. \quad (33)$$

Here ν_0 is the frequency of the characteristic vibrations of the given type of ions in the crystal lattice, and is $(1 \text{ to } 5) \times 10^{13} \text{ sec}^{-1}$.

If the condition (32) holds, i. e., if the temperature and the rate of deformation satisfy the relation (33) (with a definite value of Q), the deformation curves have a very smooth character and in general correspond to flow without appreciable hardening.* When, on the other hand, the relation (33) is not satisfied, hardening occurs, and in the case of deformation with twinning there are jumps of the value of the deformation, accompanied by a sharp drop of the resistance, corresponding to a transition from upper to lower flow limit.

Diffusive hardening is obtained with programmed deformation, which can be regarded as one of the methods of preliminary thermomechanical processing of metal parts.

A striking example of diffusive hardening is the increase of the flow limit of calcium when it is treated by programmed plastic deformation. The lower flow limit for a freshly-formed twinning layer is 40 g/mm^2 , whereas the flow limit obtained as the result of programmed hardening reaches values of 400 g/mm^2 .

CONCLUSION

Mechanical fracture of crystalline materials is a complex physical process, which has so far not received sufficient study, though already there is extensive experimental information and there have been a number of attempts at its interpretation.

* In reference 222 observations have been made on flow of calcite under compression at temperature 110°C and rate of deformation $\dot{\nu} = 0.4 \text{ } \mu/\text{sec}$ without hardening. Recently we have observed a similar case of flow in the stretching of a single crystal of beryllium with the (0001) plane at angle 45° with the axis of the specimen, at $t = 500^\circ \text{C}$ and $\dot{\nu} = 0.4 \text{ } \mu/\text{sec}$.

The theory of the ideal crystal lattice has shown that the strength of crystalline materials can be decidedly increased, and therefore the problem is not only of scientific interest but also of great economic and practical importance. From the available experimental material described in this review it follows that the low strengths of actual materials are due to defects in their structure, which lead to the concentration and nonuniform distribution of stresses. The problem of increasing the strengths of solid materials will evidently be solved through a more uniform distribution of the stresses, which can be secured by suitable thermal-mechanical processing together with proper choice of chemical composition.

The most important problems in the physics of strength of materials are those of the structural properties of solid crystalline substances, and especially the structure, shape, and size of the microscopic blocks, which are subject to wide variations both in the process of crystallization and in subsequent thermal-mechanical processing. The development of a physical theory of the strengths of real materials must start from considerations of this important structural factor.

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