TIME AND TEMPERATURE DEPENDENCE OF STRENGTH IN SOLIDS

V. I. Betekhtin and S. N. Zhurkov

UDC 539.4.019.1+539.4.019.3

There are two entirely different approaches to strength: the static (mechanical) one and the kinetic (thermal fluctuation) one.

Griffiths and Ioffe formulated the static approach long ago, and here failure is considered as a critical event (instantaneous loss of continuity), which occurs only when a certain limiting load is reached (the strength limit). No allowance is made for the effects of the thermal motion of the atoms on the strength.

In the kinetic approach, failure is considered as gradual accumulation of continuity defects (damage), with thermal fluctuations playing a decisive part. The process starts as soon as a load is applied and occurs at virtually any load. The limiting strength ceases to have a physical meaning, and the lifetime becomes the fundamental quantity characterizing the integral rate of failure for a given temperature and load, being the time from the instant of loading to that of breakage [1].

The basis of the kinetic approach is evidence that the strength is dependent on the temperature and on the time spent by the material in the stressed state. The present paper deals with this.

It has long been known that the strength is a function of time. For instance, in 1926 Welter [2] showed that the time to failure in copper, duralumin, or brass was dependent on the load. There was no doubt that the strength was time-dependent under specific test conditions (long-term strength at high temperatures). There was the evidence from slow failure in steels and on the weakening effects of the external medium, as

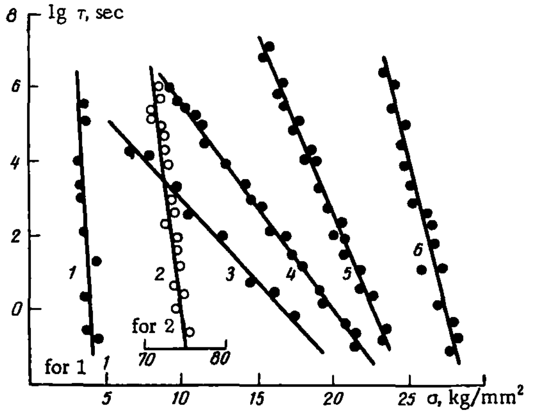

Fig. 1

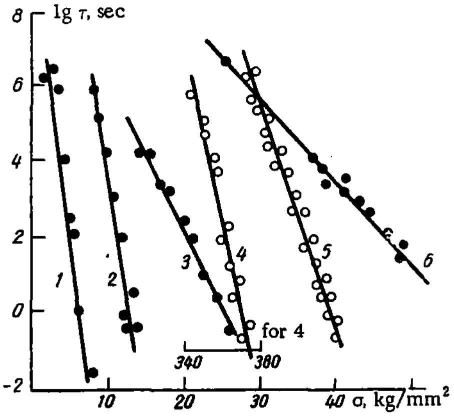

Fig. 2

Fig. 1. Dependence of τ on σ for simple bodies at 18°C and 1 atm: 1) NaCl; 2) Fe; 3) Ge (700°C); 4) Zn; 5) Al; (2700 atm); 6) Cu; (10^{-7} mm).

Fig. 2. Dependence of τ on σ for composite materials at 18°C: 1) lucite; 2) cement; 3) Al + 10% Zn (200°C); 4) Mo – Re alloy; 5) SAP-1 powder alloy; 6) Capron.

A. F. Ioffe Physicotechnical Institute, Academy of Sciences of the USSR. Translated from Problemy Prochnosti, No. 2, pp. 39-44, February, 1971. Original article submitted October 8, 1970.

well as much evidence that the deformation rate (essentially the time factor) affects the limiting characteristics of the strength.

However, virtually all the results were ascribed either to interaction of the material with the medium (effects of surfactants, corrosion cracking) or to structural changes in nonequilibrium or heterogeneous materials, or again to the effects of special test conditions (high temperature).

Then the effects of loading time on the strength was treated as the results of specific factors that reduce the strength in particular cases.

Systematic studies have been made on the relation of failure to loading time and to the atomic thermal motion in the Strength Physics Laboratory of the A. F. Ioffe Physicotechnical Institute. The scope of this paper allows us to examine these studies in some detail.

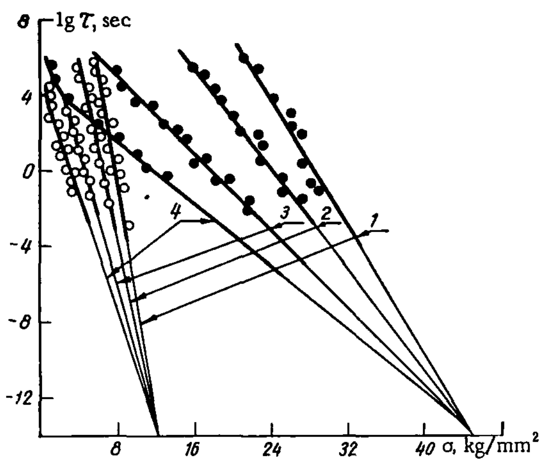

Fig. 3. Dependence of τ on σ for 99.9999% Al (left lines) and Al + 4% Cu (right lines) at temperatures (°C) of: 1) 18; 2) 100; 3) 200; 4) 300.

The effects of time on the strength were examined with uniaxial tension. The lifetime τ was measured as a function of stress σ and temperature T, with σ and T kept constant during stretching to failure for each individual specimen. A distinctive feature of the time-dependence studies was the wide range of times (10^{-3} to 10^{7} sec). Also, τ as a function of T was examined between −196°C and 1300°C.

The main tests were performed under ordinary atmospheric conditions, but some were done at 10^{-7} mm Hg [3, 4] and a few at hydrostatic pressures up to 10 kbar [5].

An important feature of the studies was that they were systematic, since it was required to elucidate the basic question whether a time dependence is always found, or whether it occurs only under specific test conditions or with materials having certain physicochemical properties.

We examined materials differing in structure and properties, which had metallic, covalent, and ionic bonds, including about 20 metals with various lattice types (Al, Fe, Ti, Zn, etc), which were given various preliminary mechanical and thermal treatments and (in certain instances) radiation treatment [6-9]. The metals varied in chemical stability (Cu, Pt) and in impurity content (1 to 10^{-4}%). Polycrystalline aggregates and about ten perfect single crystals (e.g., NaCl, Mo, Ge, Si) were used [4, 10-12]. Figure 1 shows typical examples of $\tau(\sigma)$ for various materials and conditions.

Single crystals and polycrystalline aggregates show a single type of behavior for τ as a function of σ, which for a constant test temperature T is

$$\tau = A\exp(-\alpha\sigma), \tag{1}$$

in which A and α are constants.

We examined some more complicated materials in order to establish whether this trend is general; about 20 polymers (Capron, Lavsan, Teflon), glasses, cement, wood [1*]*, and over 30 alloys [13-15]. Figure 2 shows τ as a function of σ for some of these.

The relationship is also exponential for these materials. In all, over 100 materials (including heterogeneous ones) show this relation for various external conditions, so there is a universal effect of time on strength for solids.

This feature is essentially incompatible with the view that there are limiting failure stresses; it confirms that the kinetic approach is correct. A decisive effect from time on the strength has been observed not only in tension tests (see [16, 17] in addition to the above) but also in torsion [18] and in fatigue tests [19].

---

*The asterisk to a literature citation indicates that it is not a primary source but a paper that gives the primary source.

TABLE 1

Material*	State	γ†, kcal-mm^2/mole-kg	σ‡, kg/mm^2	U_0, kcal/mole
NaCl (natural single crystal)	100	55	0.34	66 ± 3
	110	44	0.42	66 ± 3
NaCl (synthetic single crystal)	100	33	0.55	66 ± 3
	110	24	0.77	66 ± 3
	111	20	0.92	66 ± 3
NaCl (polycrystalline)	–	16	1.2	66 ± 3
Al (99.9999%)	300°C: tempering	10	3.5	52 ± 2
Al (99.96%)	550°C: tempering	7.8	4.5	52 ± 2
Al (99.96%)	400°C: tempering + 10^{22} n/cm^2 irradiation	5.2	6.8	52 ± 2
Al (99.96%)	300°C: tempering	3.5	10	52 ± 2
Al (99.3%)	rolled	2.35	15	52 ± 2
Al + 2% Mg	400°C: tempering	1.6	22	52 ± 2
Al + 4% Cu	310°C: annealed	1.2	29	52 ± 2
SAP-2 aluminum alloy	600°C: tempering	0.82	43	52 ± 2

---

* $\tau_0 = 10^{-13}$ sec for all materials.

† We can use [1*, 27] to estimate the overstress coefficient via the values for γ.

‡ For $\tau = 10$ sec and 500°C (for NaCl) and 18°C (for Al).

The time dependence as a function of temperature indicates the general relation of τ to σ and T, as Fig. 3 shows for pure aluminum and an aluminum alloy. A plot of log τ against σ gives a set of straight lines converging to a single point. Each line satisfies (1), and the slope is governed by T; the higher T, the less steep the line, i.e., the clearer the time dependence of the strength.

Figure 3 shows how the concept of a limiting strength arose; the plot of log τ against σ becomes steeper as T decreases and gives the impression that there is a threshold failure stress, which, if slightly exceeded, gives instantaneous failure.

The results give the following relation of τ to σ and T:

$$\tau = \tau_0 \exp\left(\frac{U_0 - \gamma\sigma}{kT}\right), \qquad (2)$$

where k is Boltzmann's constant and U_0, τ_0, and γ are constant coefficients.

Equation (2) is of kinetic type, which shows that failure is a gradual process in a strained body, not a critical event. Irreversibility is demonstrated by tests with interrupted loading, where the total τ equals the τ for specimens tested without interruption [1*]. There is now much direct evidence that failure develops against a background of submicroscopic and macroscopic cracks [20-24].

We now consider the coefficients in the equation for τ and the form of this.

It is found that τ_0 is of the same order (10^{-13} sec) for all solids, which coincides with a fundamental quantity, namely, the period of thermal vibration for atoms in a solid.

The energy barrier U_0 varies with the material, but it is a physical constant independent of the state for a given material. Table 1 gives results for aluminum and rocksalt.

These results show that U_0 is related to the short-range atomic order, which is determined by the atomic interaction; U_0 in fact agrees closely with the heat of sublimation [4-6], i.e., with a quantity related to the energy of atomic interaction.

The results show that failure arises from successive elementary acts of breakage in atomic bonds, where the decisive part is played not by the external force but by the thermal energy of the atoms, which accumulates at the bonds in a fluctuating fashion (thermal fluctuations). In fact, the equation for τ contains a Boltzmann factor, which coincides with Frenkel's expression for τ_{fl}, the time between two successive thermal fluctuations, which occur with the vibrational frequency τ_0 of the atoms:

$$\tau_{fl} = \tau_0 \exp(-E/kT), \tag{3}$$

where τ_0 is about 10^{-13} sec, as in (2).

This expression resembles the experimental equation (2), which shows that the thermal motion influences the failure not as a weak thermal background (oscillations of period 10^{-13} sec and energy kT) but as bursts of energy, the energy E of a burst being considered from (2) and (3) as needed to overcome the potential barrier $U = U_0 - \gamma\sigma$ set up by the interaction of adjacent atoms. The following arguments also show that thermal fluctuations are decisive.

The equation for τ has the form usual for a kinetic thermally activated process (compare Arrhenius's equation), and the decisive part is played by thermal fluctuations in overcoming obstacles in the short-range order (here, in breaking the bonds with adjacent atoms) [25].

We must now consider what is the role of the external force in failure in the kinetic approach. This force reduces the initial energy barrier U_0 by $\gamma\sigma$, facilitates thermal bond breakage, and gives direction to this process.

Factor γ has the following meaning in the kinetic approach: it represents the local overstress in the presence of a mean stress in the body. The increased loading in the atomic bonds should reduce U_0 more and so facilitate thermal bond breakage. The lower γ, the greater the real strength (Table 1), and γ for a given material is a function of its state, which is determined by thermal and mechanical treatments, presence of impurities, crystallographic orientation (for single crystals), and other factors.

Naturally, γ is structure-sensitive. For instance, in a polycrystalline metal it is determined by the dislocation structure of the block boundaries, which can be characterized via the angle between blocks [26-29], which increases substantially only at the start of loading, with very little subsequent change up to the failure point. The stabilized level is dependent only on the state of the material and defines γ (γ decreases as the angle increases) [26-28].† This relationship means that the local overstress level falls as the dislocation density increases at the stabilized block boundaries. This occurs because localization of plastic deformation (high local dislocation density) may not only facilitate failure (the various dislocation models for crack production) but also may retard it by relaxing the hazardous overstresses at the vertices of growing cracks. We may thus assume that production of block boundaries (regions of increased dislocation density) relaxes the overstresses involved in continuity failure.‡ This relaxation increases with the dislocation density at the block boundaries and causes γ to fall, which retards further development of the failures and so increases the strength.

The time dependence of failure sometimes reveals an anomaly in the slope of log τ against σ, or a knee on the curve [30, 31]. In such cases it is important to establish whether they arise from a change in the kinetic (fluctuation) failure mechanism or from a change in the conditions of operation of the mechanism. Mostly, structural studies show that the effect is due to structural instability in the material (unstable level of block disorientation, which causes γ to vary). Theoretically, it is possible for the thermal-fluctuation character of the failure to alter at 0°K, but a contribution from thermal fluctuation is unlikely near absolute zero. A detailed analysis from this viewpoint involves considering failure at high rates (by shock or explosion).

---

†Preliminary evidence indicates that this angle influences γ not only for metals but also for ionic and covalent crystals.

‡Formation of block boundaries ceases soon after loading (at the end of the first stage of creep) [28], and there are reasons for supposing that, at this point, the metal already has a high concentration of submicroscopic failures in continuity [32].

The kinetic approach indicates that most solids fail in the same way, and so there is little justification for division into viscous, brittle, high-temperature, and other forms of failure. Failure is based on a kinetic process of bond breakage (accumulation of vulnerability), with thermal fluctuations playing a major part.

This approach does not rule out specific features in the failure, which should be taken into account in strength calculations. These only serve to emphasize the common features.

The basic concepts of the kinetic approach have recently been confirmed on polymers, which are the most convenient for examination by physical methods. Infrared spectroscopy indicates that an external force loads atomic bonds and reduces the atomic interaction energy, thereby increasing the probability of fluctuation breakage while ESR reveals a high concentration of broken bonds even at the start of loading [1*]. Small-angle x-ray scattering shows that cracks of size 10^2-10^3 Å are produced [22], which leads ultimately to growth of macroscopic cracks [23].

LITERATURE CITED

1. S. N. Zhurkov, Vestnik AN SSSR, No. 11 (1957); Izv. AN SSSR, Ser. Neorg. Mat., 3 (1967).
2. G. Welter, Z. Metallkunde, 3 (1926).
3. S. N. Zhurkov, B. Ya. Levin, and É. E. Tomashevskii, Fiz. Tverd. Tela, 2, 2 (1960).
4. V. I. Betekhtin and A. N. Bakhtibaev, Fiz. Tverd. Tela, 12, 2 (1970).
5. A. I. Petrov and V. I. Betekhtin, Fiz. Tverd. Tela, 12, 9 (1970).
6. S. N. Zhurkov and T. P. Sanfirova, Dokl. AN SSSR, 101, 2 (1955).
7. S. N. Zhurkov and V. I. Betekhtin, Fiz. Met. Metalloved., 24, 5 (1967).
8. V. I. Betekhtin, in: Strength and Reliability of Metals and Alloys [in Russian], Chapter 2, Izd. LDNTP, Leningrad (1965).
9. V. M. Ilyasov, S. N. Votinov, and V. I. Betekhtin, Radiation Damage in Reactor Materials, Vol. 1, Vienna (1969).
10. S. N. Zhurkov, V. I. Betekhtin, and A. N. Bakhtibaev, Proceedings of the Second International Conference on Fracture, Brighton (1969).
11. S. N. Zhurkov, V. I. Betekhtin, and A. N. Bakhtibaev, Fiz. Tverd. Tela, 11, 3 (1969).
12. V. I. Betekhtin and A. N. Bakhtibaev, Dokl. AN SSSR, 188, 4 (1969).
13. V. I. Betekhtin, S. N. Zhurkov, and A. V. Savitskii, Fiz. Met. Metalloved., 10, 1 (1960).
14. S. N. Zhurkov, V. I. Betekhtin, and A. I. Slutsker, Fiz. Met. Metalloved., 17, 4 (1964).
15. S. N. Zhurkov and T. P. Sanfirova, Fiz. Tverd. Tela, 2, 6 (1960).
16. B. Ya. Pines and A. F. Sirenko, Dokl. AN SSSR, 134, 4 (1960).
17. B. M. Rovinskii and L. M. Rybakova, Fiz. Met. Metalloved., 9, 4 (1960).
18. V. A. Stepanov and I. E. Kurov, Fiz. Tverd. Tela, 4, 4 (1962).
19. V. R. Regel' and A. M. Leksovskii, Mekhanika Polimerov, 1 (1969).
20. S. N. Zhurkov and É. E. Tomashevskii, Zh. Tekh. Fiz., 27, 6 (1957).
21. I. A. Oding and Yu. P. Liberov, Izv. AN SSSR, Metallurgiya i Toplivo, 6, 75 (1962).
22. S. N. Zhurkov, V. S. Kuksenko, and A. I. Slutsker, Fiz. Tverd. Tela, 11, 11 (1969).
23. V. R. Regel' and A. M. Leksovskii, Mekhanika Polimerov, 2 (1970).
24. V. M. Finkel' and S. V. Serebryakov, Fiz. Met. Metalloved., 25, 3 (1968).
25. R. Z. Becker, Metallkunde, 2 (1925).
26. S. N. Zhurkov, V. I. Betekhtin, and A. I. Slutsker, Fiz. Met. Metalloved., 13, 5 (1962); 5, 5 (1963).
27. S. N. Zhurkov, V. I. Betekhtin, and A. I. Petrov, Fiz. Met. Metalloved., 23, 6 (1967).
28. V. I. Betekhtin, Strength and Mosaic Structure in Metals [in Russian], Izd. LDNTP, Leningrad (1968).
29. V. I. Betekhtin and M. M. Myshlyaev, Fiz. Met. Metalloved., 22, 1 (1966); 24, 3 (1967).
30. S. N. Zhurkov, V. I. Betekhtin, et al., Fiz. Met. Metalloved., 18 (1964).
31. S. N. Zhurkov, V. I. Betekhtin, and A. I. Petrov, Fiz. Met. Metalloved., 24, 1 (1967); 24, 2 (1967).
32. B. Ya. Levin, V. I. Betekhtin, et al., Fiz. Tverd. Tela, 12, 9 (1970).