

AT BRITTLE FRACTURE

G. M. Bartenev*

ABSTRACT

The process of fracture in solids is accompanied by the mechanical losses in the form of dissipated heat.

Three basic types of mechanical losses occur in the process of fracture in solids: 1) deformation losses, which are very highly pronounced at the tips of microcracks; 2) losses at the transformation of the elastic energy into the kinetic energy resulted from moving the walls of a growing crack apart at crack propagation; 3) losses at the breakdown of atomic bonds at the tips of growing cracks.

Consideration is given to the fundamental role of the mechanical losses in formulating the theory of solids fracturing even in the case when a model is used for an ideal brittle material.

Any process of fracture in solids is accompanied by the mechanical losses eventually taken into account by the dissipated heat δQ in the law of conservation of energy

$$\delta W = du + \delta Q \quad (1)$$

where δW is the change in the elastic energy of a specimen under stress, du the increase in the surface energy, and δQ the mechanical losses occurred in the process of fracture within a short interval of time dt . At $\delta Q=0$ the known Griffith equation is derived.

1. Types of Mechanical Losses at Fracture

There are several types of mechanical losses in processes of fracture in solids. Most known of these are the following: 1) deformation losses (elastic hysteresis, internal friction, plastic and viscous deformations, etc.) exhibited very distinctly in places of overstresses, at tips of microcracks and at boundaries with other flaws [1]; 2) partial transformation of the elastic energy into the kinetic energy resulted from moving the walls of a growing crack apart [2,3], or into the kinetic energy of "scattered splinters", and ultimately into heat. The third basically new mechanism of elastic energy dissipation at crack propagation is the dissipation of

* Department of Physics of Solids, Lenin State Teachers' Training University, Malaya Pyrogovskaya, 29, Moscow, G-435, U.S.S.R.

energy at the breakdown of atomic bonds at the tips of growing cracks.

In an ideal elastic solid body there are no deformation losses, except those of the 2nd and 3d types. In an ideal brittle body only irreversible (plastic) macro- and microdeformations, including local ones, are absent, but, unlike an ideal elastic solid body, it retains internal friction and is capable to dissipate the elastic energy. Consequently, in an ideal brittle body all types of losses are actually possible to occur, except losses associated with local or total residual deformations.

The paper discusses the problem of fracture in an ideal brittle body in the light of different types of mechanical losses which hitherto have not been given due consideration in the theoretical physics of strength. This, however, is quite understandable, as the investigators' attention was focused on such most important materials as metals and polymers in which the basic type of losses is solely that of deformation losses.

In literature mention is usually made of the first two types of losses observed in metals [1], and in solid polymers at impact [4] and the so-called controlled, slow fracture [5-8]. Of specific interest are the experiments with the controlled slow fracture which is very close to the so-called "equilibrium" process. These experiments have shown that in all cases most of the energy of fracture goes to form mechanical losses which are considerably greater than the surface free energy.

The energy of fracture is usually calculated per unit area of broken surface, hence it is written as $\delta W = \alpha^* ds$, where α^* is the characteristic fracture energy, ds the increase of newly formed surface. For an ideal brittle body and, especially, for real solid bodies, $\alpha^* > \alpha$, where α is the surface free energy. It is clear that α^* is not the constant of the material. This is quite obvious if we consider that the value of mechanical losses depends on temperature, velocity of fracture and sizes of a specimen. Shand [9] has calculated, from experimental data, a characteristic fracture energy for glasses at low and high velocities of fracture. It was found that at slow velocities of crack propagation α^* is close, but not equal to the surface free energy of the glass in a wet atmosphere $\alpha = 250-300$ erg/cm². With increasing the rate of crack growth, the mechanical losses will increase too, and at the velocity of 10 mm/sec. attain a high value ($\alpha^* = 5,000$ erg/cm²). When changing to the rapid phase of fracture, the mechanical losses will attain a maximum value, and the characteristic energy will be 7,500 erg/cm². When testing in vacuum, at low velocities of crack propagation $\alpha^* = 2,000$ erg/cm², which also is more than the surface free energy for glasses in vacuum.

Shand attribute the experimental results obtained only to the 1st type of mechanical losses. He believes that at low stresses and low velocities of crack propagation inelastic effects are not developing at crack tips, but at high stresses and high velocities they are essential and considerably influence the process of fracture. In this work [9], like in many others, other types of mechanical losses are not considered.

The necessity of taking into account the 2nd type of losses, i.e. the kinetic energy from moving the walls of growing cracks apart, especially at higher rates of fracture, was based by Mott [2] and by Bateson [3]. The 3d type of losses was discussed quite recently by the author and Mrs. I.

Rasumovskaya [10], using the molecular model [10,11] of a microcrack (Fig.1).

Bonds breakage occurs at the boundary of passage from the free surface to the "unbroken solid" (indicated by a dashed line). Particle 1-5 are still in the volume of the solid body, particles 6-10 are already in the free surface, and the act of breaking comes to the step 5-6. At the moment of bonds breaking the quasi-elastic force attains a maximum value (Fig.2) corresponding to the critical stress, the interparticle distance changes by jump, and the tip of the crack moves to a distance corresponding to that between two particles.

The dissipation of energy at the breakdown of bonds can be understood if we proceed from the following speculations [10,12]. If the quasi-static process of bonds stretching is possible up to the maximum μ point, then the breaking of bonds behind the μ point occurs spontaneously and at the rate of heat movement.

The 3d type of losses can be considered as that of mechanical losses in surface layers, however these losses are proportional to the number of broken bonds, but not to the velocity of crack propagation. In view of this fact the methods of controlled fracture, from Obreimoff's method of mica splitting onwards [13], provide not the surface free energy of the solid body in a pure form, but a value including the 3d type of mechanical losses per 1 cm² of the newly formed surface.

2. Criticism of Griffith's Theory.

The physical theory of strength by Griffith [14] initially was founded on the law of conservation of energy and considered brittle materials as solid media having flaws (microcracks). If the value of overstress at the tip of a most dangerous crack is equal to the theoretical strength σ_m , then, according to Griffith, a catastrophic (at velocities close to sound velocity) expansion of the crack occurs, and the specimen is splitted into parts. At that moment the average stress in the specimen corresponds to the critical stress, or the so-called maximum real strength of the specimen σ_k . At overstresses below the theoretical strength when $\sigma < \sigma_k$, the crack is not growing. According to Griffith, the critical stress is calculated from the following condition: the crack grows only when the decrease of the elastic energy in the specimen (due to the removal of stress around the growing crack) is equal to, or more than the increase of the surface potential energy when the formation of new surfaces occurs.

Griffith's ideas have been further modified in many later works in which the strength problem was discussed from the viewpoint of the elasticity theory of an absolutely elastic body. Most of the critical opinions concerning Griffith's theory were expressed by Griffith himself (on the limited applications of the elasticity theory, the reduction in elastic modulus with increasing stress, the effects of other cracks on the crack under consideration, etc.). However, the principal limitations of Griffith's theory, which were not shown by Griffith himself and his followers, are the following.

To say, the mechanism of fracture, according to Griffith, is, strictly speaking, related to the ultimate case, i.e. to the non-thermal process of fracture when the time dependence of the strength is expressed very poorly.

Only at low temperatures close to absolute zero the theory is physically justified to some degree, as the time dependence of the strength for solid materials is poorly expressed at very low temperatures, and the surface free energy practically coincides with the surface potential energy.

The basic limitation of Griffith's theory is associated with the fact that the mechanical losses at fracture are not taken into account. So, Griffith defines the critical stress from the following condition: the change in the elastic energy ΔW with crack propagation is equal to the increase of the surface potential energy dU . This condition can be considered almost identical only for the quasi-static process when the velocity of crack propagation is infinitely low. When the crack is growing at an ultimate velocity, the subsequent breakage of interatomic bonds and moving the walls of a growing crack apart will result in the dissipation of the elastic energy which will eventually transform into dissipated heat. When $\delta Q=0$, this means that the crack is not growing even if $\sigma = \sigma_k$ (according to Griffith).

All the works, from Griffith's onwards, are likely to identify two different terms - the theoretical strength and the critical overstress at the microcrack tip. This is not correct as the critical nature of fracture is observed only on the attainment of the maximum quasi-elastic force under the conditions of complex stress at the tip of the microcrack. This magnitude is called the critical overstress σ_k^* . Unlike the theoretical strength which is the constant of the material (at a given simple form of stressed state), the critical overstress can change depending on the size and shape of the crack and properties of the material.

The crack starts to grow at an infinite low velocity when the overstress σ^* at its tip is over σ_k^* ; however, the growth velocity is higher, the greater the overstress σ^* . This can be explained by the fact that a part of the elastic energy in the specimen is dissipated, and the degree of this dissipation is greater the higher the velocity of crack growth. Therefore, the crack growing at an ultimate has a value of overstress at its tip $\sigma^* > \sigma_k^*$, and at an infinitely slow crack propagation, $\sigma^* = \sigma_k^*$. This conclusion is well confirmed by the results obtained by Shand [15] who from experimental data and Neiber's equations for the coefficient of stress concentrations have calculated that, for glasses, the overstress at crack tips σ^* , at which fracture occurs at high velocities. This value for different glasses is 2-3 times greater than the theoretical strength calculated by Orovan's equation $\sigma_m = 0.1E$, where E is Young's modulus.

The quantitative record of all types of mechanical losses is one of the most important problems for the molecular theory of strength to solve.

REFERENCES

1. Proceedings of Intern. Confer. on the atomic mechanism of Fracture, Swampscott, USA, 1959.

2. N. Mott, *Engineering*, 165, 16, 1948.
3. S. Bateson, *Phys. Chem. Glasses* 1, 139, 1960.
4. W. Holzmüller, K. Altenburg, *Physik der Kunststoffe*, Berlin, 402, 1961.
5. F. Roesler, *Proc. Phys. Soc. B* 70, 201, 1957.
6. J. J. Benbow, F. Roesler, *Proc. Phys. Soc. B* 70, 201, 1957.
7. J. P. Berry, *J. Polymer. Sci.* 50, 107, 313, 1961; *J. Appl. Phys.* 34, 62, 1963.
8. N. Svensson, *Proc. Phys. Soc. B* 77, 876, 1961.
9. E. Shand, *J. Amer. Cer. Soc.* 44, 21, 1961.
10. G. M. Bartenev, I. V. Rasumovskaya, *Dokladi of Academy of Sciences, USSR*, 133, 341, 1961.
11. G. M. Bartenev, I. V. Rasumovskaya, P. A. Reh binder, *Kolloidal Journal, USSR*, 20, 655, 1958.
12. G. M. Bartenev, Yu. S. Zuyev, *The Strength and Fracture of High-Elastic Materials*, Moscow, 1964.
13. I. W. Obreimoff, *Proc. Roy. Soc. A* 127, 290, 1930.
14. A. Griffith, *Phil. Trans. Roy. Soc., London*, 221, 290, 1930.
15. E. Shand, *J. Amer. Ceram. Soc.*, 37, 52, 559, 1954; 42, 474, 1959.

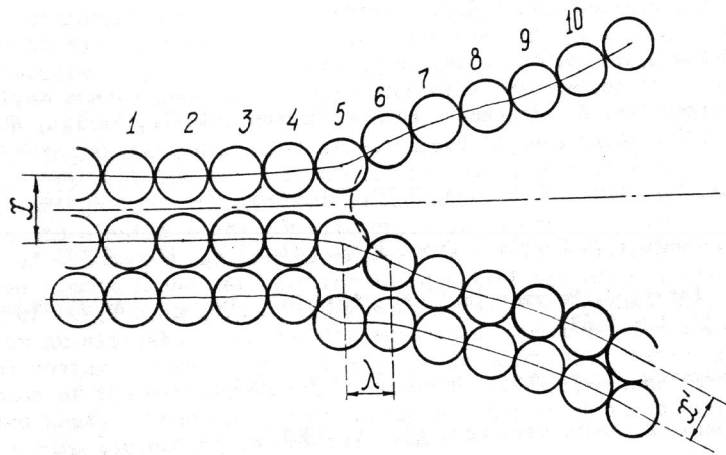


Fig.1. Molecular model of a crack in brittle materials.

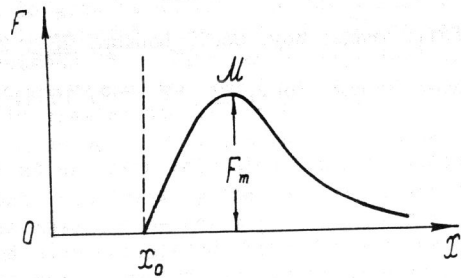


Fig.2. Dependence of force F acting between particles under strained conditions upon interparticle distance x ; x_0 is the interparticle distance in an unstressed state of the material.