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Resistance to crack propagation in ceramics subjected to thermal shock

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It is shown that the resistance to crack propagation in alumina, zirconia and porcelain, when subjected to thermal shock, is high when the initial strength is low and, vice versa when the initial strength is high. It is proposed that the high resistance to crack propagation results from discrete microcracking in the highly stressed zone ahead of the major crack, whereas the low resistance results from the unrestrained propagation of major cracks. Evidences of the crack patterns in alumina and zirconia substantiate this hypothesis.

1. Introduction

Traditionally, ceramics have been referred to as a class of materials which fail in a brittle manner when subjected to a thermal or mechanical loading. In practice, it means that, once a crack is initiated in ceramics, there is no resistance to propagation during fracture. This classical concept, first originated by Griffith, was found to be true in the case of glass, but proved inadequate in metals. The difference lies in the fact that the mechanism of energy dissipation is different in each case. In glass, the energy dissipation occurs primarily through the creation of surface energy. In metals, in addition to surface energy, considerable energy dissipation occurs through plastic deformation *prior to* and *after* the nucleation of cracks so that the propagation of cracks is slow. It is known that certain classes of ceramics, when subjected to a thermal shock, can also exhibit slow crack growth without undergoing plastic deformation. Consequently, for this class of materials, the crack initiation and propagation do not coincide as in glass and the fracture is non-catastrophic. This paper deals with slow crack growth in such ceramics.

2. Experimental data

Slow crack growth has been demonstrated recently for a certain class of alumina [1] and zirconia [2] when subjected to thermal shock. Hasselman [3] has defined such crack growth as quasi-static crack propagation, and attributed its occurrence to the lack of adequate elastic energy at fracture. The ratio of the rate of elastic energy (energy per unit crack length) released at fracture after a material is subjected to thermal shock to the same before thermal shock can be expressed as a measure of resistance to crack propagation, R, for the given amount of thermal shock. This elastic energy is equal to $\pi c^2 \sigma^2/E$ so that the rate is proportional to $c\sigma^2/E$, where c is the critical crack length, σ the stress at fracture and E the elastic modulus. Assuming that the elastic modulus before and after thermal shock remains the same, the resistance R is then proportional to $(\sigma_t/\sigma_i)^2$, where σ_t is the strength after thermal shock and σ_i is the strength before thermal shock. Fig. 1 illustrates the plot of R versus initial strength for Lucalox,* [1] a commercial alumina [4] and a commercial porcelain [5]. The critical quenching temperatures for thermal shock of these materials are 225 to 250°C, 180 to 220°C, and 600°C, respectively. In each case, the resistance to crack propagation increases sharply with decreasing initial strength of the material. As the strength increases, the resistance decreases exponentially, and the cracks become highly unstable, leading to catastrophic fractures. Between the data of porcelain and that of alumina, the initial strength varies by more than an order of magnitude. In spite of this, the nature of the curve is strikingly similar in each case. It is believed that similar curves can be generated for other materials as well. The curve for porcelain is of special interest. At a substantially low initial strength ($< 0.7 \times 10^8$ dyn

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*High density alumina, General Electric.

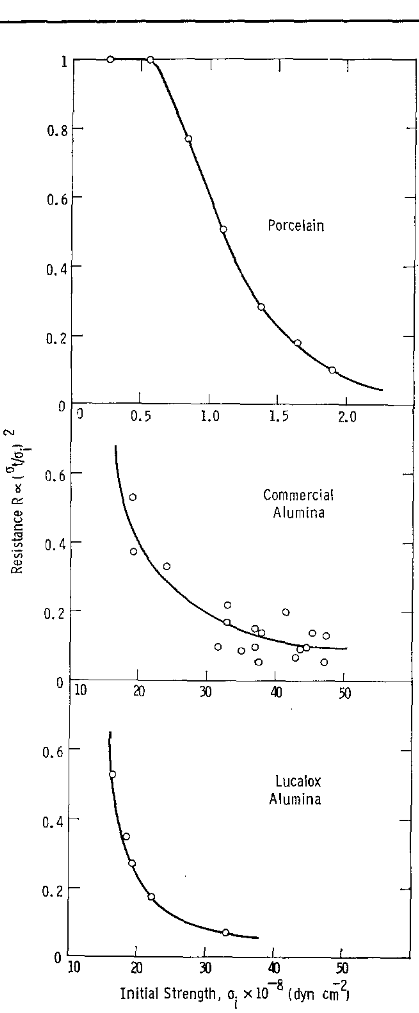

Figure 1 Resistance to crack propagation as a function of initial strength in thermally shocked Lucalox [1], commercial alumina [4] and porcelain [5].

cm^{-2}), the cracks are extremely stable and there is no degradation of strength on thermal shock; on the contrary, at a high initial strength ($> 2 \times 10^8$ dyn cm^{-2}), the cracks are highly unstable and the strength approaches to near zero value. The data for alumina are relatively few, but the results show a similar decrease in resistance with increasing initial strength. It is thus clear that alumina and porcelain behave like glass when the initial strength is high, but show macroscopic "plasticity" (high resistance to crack propagation) like metal when the initial strength is low.

3. Discussion

It is evident from Fig. 1 that the degree of energy dissipation on fracture varies with initial strength. It is high at low initial strength and low at high initial strength. Since plastic deformation can be neglected in alumina and porcelain, an alternative mechanism of energy dissipation must exist at low initial strength. It is proposed here that this energy dissipation occurs through microcracking in the highly stressed zone ahead of the major propagating crack. The stability of the major crack and hence the slow crack growth, is the direct result of the formation of these microcracks. The energy expended in creating these microcracks is obtained at the expense of the major crack whose propagation is thus restrained. Since most of these microcracks do not necessarily merge with the major crack, this energy may be considered as "lost". In the fracture of concrete, where a similar mechanism was first proposed by Glucklich [6], this "lost" energy was found to be about an order of magnitude higher than that of the formation of the major crack [7].

According to a semiquantitative treatment given by Glucklich and Cohen [8], the following energy equations apply during microcracking:

$$\text{energy demand } W_\mathrm{d} = 4\gamma c + 4\gamma \sum_{i=1}^{n} b_\mathrm{i}$$

$$\text{energy release } W_\mathrm{r} = \frac{\pi\sigma^2}{E} c^2 + \frac{\pi\sigma^2}{E} \sum_{i=1}^{n} b_\mathrm{i}^2 .$$

The first terms in the right hand side of both the equations are for an ideal Griffith material, and the second terms due to microcracks; γ is the surface energy, E the Young's modulus, σ the stress, c the half-length of major crack, b the half-length of discrete microcrack and n the number of microcracks. Glucklich and Cohen have qualitatively argued that as the fracture proceeds with microcracking, W_d undergoes a drastic change from a linear to an increasing power function of $\sum b_n$, whereas W_r remains merely a function of $\sum b_n^2$. The contribution of microcracks to energy demand thus exceeds that to energy release. Fig. 2 schematically illustrates [8] the difference in behaviour between a material which forms microcracks and one which does not form microcracks (ideal Griffith material). The important difference is in the shape of the energy demand curve W_d. This is a straight line in a Griffith material and an upward concave curve in material forming microcracks. The energy-release curve W_r is a second power para-

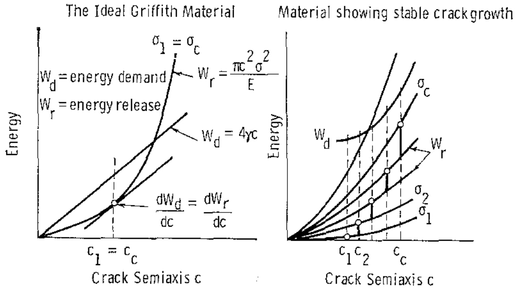

Figure 2 Schematic representation of the energy balance for a propagating crack in a flexible tensile loading system [6].

bola. Crack instability occurs when the slope of the two curves are equal.

It is seen that when the fracture proceeds with microcracks, there is always a lag of energy release behind the energy demand. The stress must be increased continuously from σ_1 to σ_c for the crack length to grow from c_1 to c_c, the critical crack length for fracture. At this point, the slopes are equal and the fracture occurs. The crack thus propagates in steps, alternately from instability to stability, and the catastrophic failure is prevented.

When the above concept is applied to the strength behaviour of alumina, zirconia and porcelain, it is seen that the crack propagation is similar to that of an ideal Griffith material when the initial strength is high, and that of microcracking behaviour when the initial strength is low. In the former case, all the stored elastic energy is used in an unrestrained propagation of few major cracks, whereas in the latter case, this energy is used in creating a multitude of microcracks in the highly stressed zone ahead of the major cracks. If this concept is correct, there will be a significant difference in the extent of crack propagation as the initial strength varies for a given material. At a low initial strength, there will be a network of small cracks (caused by the extension of innumerable microcracks), whereas at a high initial strength there will be a few large cracks across the ceramic body. This difference in behaviour is qualitatively illustrated in Figs. 3 and 4 which show the crack patterns in Lucalox [1] and zirconia [2], respectively, when subjected to thermal shock. In Lucalox, as the strength decreases from 32.7 × 10^8 dyn cm^{-2} (low resistance to crack propagation) to 16.5 × 10^8 dyn cm^{-2} (high resistance to crack propagation), the crack pattern is distinctly changed from runaway cracks to restrained microcracks. Similarly in Fig. 4, the crack pattern in partially stabilized zirconia shows numerous microcracks branching out from major cracks and thus preventing catastrophic fracture. It will be interesting to note the crack pattern in porcelain as a function of initial strength.

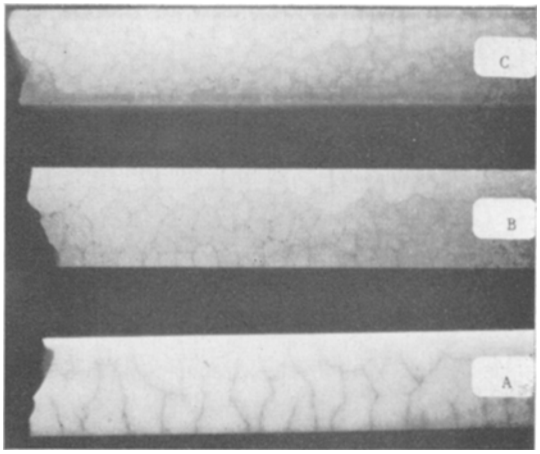

Figure 3 Crack propagation in Lucalox with varying initial strengths: (A) 32.7 × 10^8 dyn cm^{-2}, (B) 19.4 × 10^8 dyn cm^{-2}, and (C) 16.5 × 10^8 dyn cm^{-2} [1].

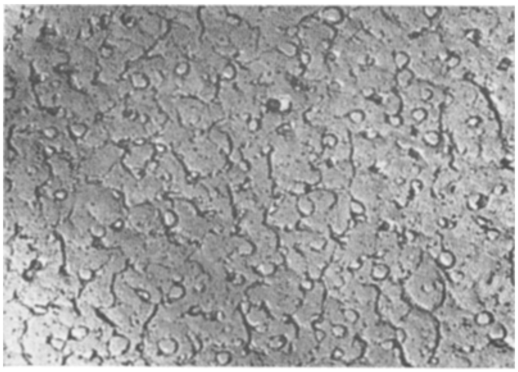

Figure 4 Crack propagation in partially stabilized zirconia [2].

A further evidence of the validity of the above concept is provided by Davidge and Tappin [4]. For the commercial alumina shown in Fig. 1, they estimated the amount of energy release versus energy demand for various aluminas studied. Except for three aluminas, the energy release on fracture was higher than the energy demand for all other aluminas. As expected, the relative strength degradation on thermal shock for these three aluminas were lower than those of other aluminas. These three aluminas are shown in the upper left hand side of the middle curve in Fig. 1. It is seen that the initial strengths of these three aluminas are lower than those of other aluminas and as expected, they show greater resistance to crack propagation than the other aluminas.

It is not clear as yet, however, why the microcracks are favoured in the above materials with low initial strength as opposed to those in high initial strength. This may arise from high anisotropy in alumina, when the grain size is large and the strength low, phase transformation in zirconia or simple heterogeneity in porcelain. The importance of microstructure in restraining the crack propagation upon thermal shock should thus be a major concern for future study. Its effect has not been clearly established yet in the literature.

In spite of this uncertainty, the curves such as those shown in Fig. 1, have some practical significance in designing with ceramics for a given thermal environment. It is important to note that the response to thermal shock for a given material is different for different initial strength. When the initial strength is high, there is a severe degradation in strength; when the strength is low, there is only a mild degradation in strength. In most practical applications, however, a balance is required between a preferred strength and a desirable thermal shock resistance. This balance in strength and thermal shock can be achieved by constructing curves similar to those shown in Fig. 1.

4. Summary

It has been shown that certain classes of ceramics such as alumina, zirconia and porcelain when subjected to thermal shock, show a high resistance to crack propagation (slow crack growth) when the initial strength is low and, a low resistance to crack propagation (rapid crack growth) when the initial strength is high. It is proposed that this difference stems from the different energy dissipating mechanisms of high and low strength materials, respectively. In the former case, all the stored elastic energy at fracture is used in an unrestrained propagation of the major crack, whereas in the latter case, this energy is used in creating a multitude of microcracks in the highly stressed zone ahead of the major crack. These microcracks are formed at the expense of the major crack whose propagation is thus arrested and the catastrophic failure prevented. Experimental evidences of crack patterns which result from thermal shocking of alumina and zirconia, are in agreement with the above hypothesis.

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