Initiation and Propagation of Fracture in Ceramics

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Summary

Ceramic materials fracture in a brittle mode at all but very high temperatures. Initiation may occur from surface irregularities, grain boundaries or pores, or from cracks resulting from manufacture, machining or other external agencies. The stress to fracture, σ_f, is then given by $K\left(\frac{2E\gamma_i}{C}\right)^{1/2}$, where γ_i is the effective surface energy for fracture initiation, C the surface flaw depth, (or half-depth for an interior flaw) and K depends on test configuration, flaw geometry and interaction between flaws. Where C is small, stress may be supported until it is sufficient to actuate microplastic flow, σ_Y, which then concentrates stress at grain boundaries and initiates cracking; then $\sigma_f = K'[\sigma_Y + K''\left(\frac{2E\gamma_i}{G}\right)^{1/2}]$ where G is the grain size. Internal stresses must be considered, along with uneven partition of stresses through variations in elastic moduli. Mechanisms have been inferred principally from measurements of γ_i, σ_f, and their dependence on temperature, grain size, pore content and impurity effects. Cooperative effects between flaws of subcritical size are especially likely in the surface region. The microscopic study of fracture origins is essential in order to identify the mechanisms and significat parameters for specific materials in particular temperature ranges.

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1. Introduction

Fracture phenomena in ceramic materials include all of the elements of fracture which apply to glasses, complicated by the anisotropy and sometimes by the microplasticity characteristic of single crystals. At very high temperatures, ductile yielding and viscous flow of one or more phases may participate. Finally, and most complicating of all, are the microstructural aspects. Even with complete information on the mechanical and thermal properties of the constituent crystalline phases and glasses, modified as they may be by such impurities as they contain in a ceramic, it still represents a tremendous challenge to deal with the interactions among the constituents under applied mechanical, chemical and thermal stresses.

Flaw growth to a size which is critical under the conditions of stress and environment proceeds, in a ceramic, within a system of variables so complex as to frustrate attempts at exact analysis. To begin with, the flaw shape, size and orientation must be understood. The local stresses effective at flaw boundaries are the resultants of applied stress components partitioned among the microstructural constituents in a pattern which depends on their various elastic constants and on their geometric configuration. Finally, development of the flaw into a critical crack must be expected to proceed unevenly in direction and discontinuously in time, advancing into weak or more highly stressed regions and waiting at other points. In this stage fracture is sensitive to environmental atmosphere as well as to the systematic and accidental features of microstructure in the immediate neighborhood.

Only when the crack size has far outstripped the grain size does it become nearly immune to local peculiarities of microstructure so that it can properly be treated in terms of average stress and average mechanical properties.

Progress towards understanding ceramic fracture has been systematically reviewed by Pugh [1], and briefer accounts have since been presented by a number of others.[2-4] The present paper will attempt to identify the principal areas of current interest and to call attention to certain considerations which appear to be neglected.

2. The Calculation of Strength

The fracture of ceramics has made extensive use of the principle pioneered by Griffith.[5] He had noted that the minimum applied tensile stress to fracture a glass sheet having a crack of defined geometry was inversely proportional to the square root of the crack depth. From experiments with artificially introduced cracks he evaluated the proportionality constant and then used the relationship to calculate the depth of the pre-existing flaw from which fracture had initiated in a specimen of glass loaded to failure with the same test geometry. If the term "Griffith crack" has any useful meaning (beyond the usual meaning of "crack" or "origin crack") it is the hypothetical crack of arbitrarily defined geometry having a dimension such as to permit fracture at the same loading stress as the measured strength of the specimen. It must be realized that both the shape and dimensions of the real crack will rarely be those of the Griffith crack; only postmortem examination can possibly show the real relationship between them. The actual flaw may be tilted or twisted with respect to the stress normal plane, it may be nonplanar, it may be deep and narrow, or it may be blunted; in all of these cases the corresponding Griffith crack will be smaller then the real flaw. (A proper analogy to the Griffith flaw size is the diameter of a particle as determined by sedimentation. The Stokes diameter of the particle is the diameter of a sphere of the same substance which would sink at the same velocity.)

Additional complications in the case of ceramic specimens have already been touched on. In the present contex it means that the microscopic details of the phases bounding the flaw and the localized stresses acting on them will certainly differ between the real flaw and the Griffith flaw. Success in interpreting flaw dimensions calculated in the Griffith tradition will depend on choosing an appropriate model flaw geometry and on supplying an energy term to fit the mode in which the origin flaw is extended under the conditions of test.

Orowan [6] suggested that the surface energy term γ_s, of Griffith might be enhanced by energy, γ_p, required by plastic deformation processes in the highly stressed region in front of the crack tip (Fig. 1)

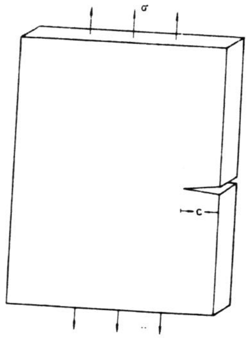

Fig. 1 Griffith-Orowan equation

$$\sigma_F = K \left[\frac{E(\gamma_s + \gamma_P)}{C}\right]^{1/2} \qquad (1)$$

σ_F = Fracture stress
γ_S = Surface energy
γ_P = Plastic work done
E = Young's modulus
K = Constant

Evans and Tappin[7] expanded on a development by Davidge and Evans[3] by using an experimentally determined "effective surface energy for fracture initiation", γ_i, in place of the problematic γ_s. They express the applied fracture stress in a form suitable for study of material parameters isolated from test geometry, i. e.,

$$\sigma_F = \frac{Z}{Y}\left(\frac{2E\gamma_i}{C}\right)^{1/2} \qquad (2)$$

where E is Young's modulus, and C is the depth of the surface flaw (or half-depth of an internal flaw). Y is the stress intensity coefficient defined by Srawley and Brown[8] and derived for various test geometries by Brown and Srawley[9]. Z is a flaw shape coefficient; its value is unity for a long shallow crack, is approximately 1.5 for a penny-shaped crack, 1.8 for a crack of equal width and depth and rises steeply thereafter. Certain microplastic effects may also be included in Z, inasmuch as the plastic zone relaxes stress at the crack tip and may be considered as adding to the effective crack depth[9]. The term γ_i includes the thermodynamic surface energy, γ_s, and effective energy contributions due to crack blunting and to creating cleavage steps and subsidiary cracks. Evans and Tappin feel that crack blunting is unlikely except at high temperatures and that generally the difference between γ_i and γ_s can be attributed largely to the work of plastic flow, in agreement with Orowan's expression.

The point may need further study. Guard and Romo[10], using a microbeam X-ray diffraction technique, observed plastically distorted grains in hot-pressed aluminum oxide having an average grain size of 20 um at distances as far as 60 μm below the fracture-exposed surface. Two zones were observed; the first, having a depth of about 10 μm, showed high distortion of the $\{11\bar{2}0\}$, $\{10\bar{1}\ 10\}$ and $\{01\bar{1}0\}$ planes, while the second (deeper) zone showed a lesser, nearly constant distortion on the $\{11\bar{2}0\}$ and $\{01\bar{1}0\}$. From these data the plastic work was calculated to be about 1.5 J/m^2, with little difference between values for specimens fractured at 20° and 1700° C. This compares with 1.1 J/m^2 calculated for the thermodynamic average surface free energy at room temperature[11]. Together these two energy terms are about one-tenth the work of fracture measured on similar aluminas[12], and substantially less than Wiederhorn's determination of 6.0 J/m^2 for the fracture surface energy on the weakest plane of single-crystal sapphire[13].

3. Fracture Energy

γ_i can be evaluated by any of the standard tests, such as the notched-beam or double-cantilever methods, reviewed recently by Coppola and Bradt[14], or by the newer double-torsion method[15]. The stress necessary to restart a previously introduced long deep crack is measured and γ_i is calculated from Equation 2, choosing values of Z and Y pertinent to the test. The resulting value can be used to calculate the size of the origin flaw in the same material tested to failure without precracking, Z and Y coefficients being reselected to fit the test geometry and the location and shape characteristic of the flaw which is assumed or observed to have generated the failure. The solution yields, in fact, a Griffith flaw dimension.

To what extent is the γ_i, determined in this way, suitable for discussion of propagation of a single fatal flaw? It must be clear that there are important differences between the situations at an arbitrarily established linear crack and at a single weakest-link site. In both cases various avenues of response to stress are likely to be involved along the perimeter of the flaw or crack, and different energies correspond to each:

1. Fracture through one or more crystalline phases.
2. Separation along twin or slip boundaries.
3. Separation of grains in a bicrystal relationship.
4. Separation of grains through impure boundary layers.
5. Microplastic deformation of crystalline phases.
6. Grain boundary sliding.
7. Ion migration.
8. Stress corrosion (including evaporation-condensation).
9. Viscous flow of liquid phase (at high temperature).
10. Elastic deformation, with a variety of elastic constants.

In the weakest-link flaw we are dealing with the exceptional case, i. e., an unusually large flaw size, an unusually facile avenue of propagation, an extraordinary peaking of local stresses through structural accident, or some combination of these. Even if accurate account is taken of the flaw size and shape, the fracture energy is likely to be considerably lower than that necessary to fracture along a continuous front.

At the continuous flaw front, continuity must be maintained in spite of initial nonlinearity both in the plane of fracture and departing from it, as well as tilt and twist misalignments of the leading elements of the fracture front. This must be expected to involve generation of subsidiary cracks, additional surface area because of roughness, and inefficient utilization of stresses through their resolution in following the easier avenues of advance. In cases where origin flaw size far exceeds the average grain diameter the specific energy for its extension should be closer to the γ_i determined on a large artificial crack und must approach it in the limit, especially if the origin flaw is the result of a single violent event, so that the process of selection from among many candidate flaws does not come into play. This point is essential to such questions as the interpretation of strength-grain size-porosity relationships.

In discussing growth of the critical flaw in selfbonded silicon carbide, McLaren, Tappin and Davidge[16] highlight the problem of γ_i by referring to an effective jump in γ_i as the growing flaw encounters a new grain configuration (Fig. 2).

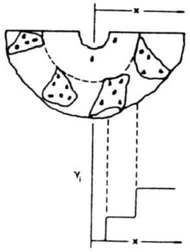

Fig. 2. Schematic representation of shifting γ_i values as growing flaw encounters new environments. (After McLaren et al[16]).

Thus the meaning of γ_i is interpreted not as a fundamental material property but as a function which varies with the shifting environment at the crack tip. (It would seem preferable to confine the symbol γ_i to the energy for fracture initiation as determined by a standard test; the energy for fracture through the constituent phases could be identified in a separate way.)

Work of fracture, i. e., the work done in generating new surface by a moving crack, is also frequently used in Griffith flaw determinations. It is subject to the same problems in application as the energy of restarting a crack. In addition it is not responsive to atmosphere in the same way as an origin crack because of its greater velocity.

Atmospheric effects, similar to those which are under intensive study in the glass area, are observed also in ceramics[17, 18] and sensitivity of strength measurement to loading rate is in part attributable to atmosphere, through effects such as stress corrosion. Atmosphere may also affect the ease of dislocation movement[19]. We[20] have just recently observed that cation migration to the region of the fracture tip, driven by steep stress gradient[21] and probably an associated gradient-induced electrical potential[22], plays a role in the static fatigue of glasses. The diffusing cations weaken the structure at the crack tip and permit faster crack advance in consequence. Figure 3 illustrates the effect. At low velocities, we believe that

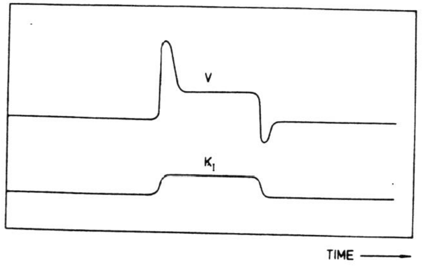

Fig. 3. Velocity of (slow) crack advance in response to increase and decrease in stress at the crack tip. (After Quackenbush and Frechette [20])

the crack tip region is more heavily enriched with respect to the cation. Upon sudden increase in applied stress the crack moves more rapidly into the enriched region but then slows as it passes through into the new, lower level of enrichment which corresponds to the new, faster dynamic equilibrium velocity. When the original stress level is abruptly restored, the velocity drops to a new low until the original cation enrichment level, and corresponding decrease in fracture energy, is re-established. The effect is to be expected in ceramics - and possibly in some crystals also. This suggests a further contribution to the energy of fracture, i.e. the work of ion transport.

4. The Material Parameters

Attempts to express the strength of a ceramic as a function of its grain size, G, and pore content, P, led to attempts, pioneered by Knudsen[23], along fully empirical lines. Knudsen's form was

$$\sigma_F = kG^{-a}e^{-bP} \qquad (3)$$

where k, a and b are empirical constants. A recent version[24] for the strength of a series of hotpressed aluminas is

$$\sigma_F = 142,500\ e^{-11.83P}\ G^{-0.60} + 3.33P \qquad (4)$$

Such empirical relations have been helpful in engineering practice but have contributed less towards understanding, since the "constants" vary markedly and unpredictably with changes in other material parameters. Distinction between closed pores and open must also be made, but it is not clear how to treat closed pores at grain boundary intersections[25]. Pores within grains have little effect on strength but do increase the measured work of fracture[24]. There are at least two ways in which intergranular pores may affect strength. They may contribute the originating flaw, and they reduce the elastic moduli through their reduction of the load-bearing area. The effect on Young's modulus has been expressed by Spriggs[26]

$$E = E_o e^{-bP} \qquad (5)$$

and by Hasselman[27]

$$E = E_o \left(1 - \frac{AP}{1-(A+1)P}\right) \qquad (6)$$

where E_o is the modulus at zero porosity; the constants A and b must be evaluated experimentally. The range of usefulness of both expressions is limited by the extent to which pore geometry remains fixed. They deal, moreover, with macroelasticity and not with the specific microscopic environment at the fracture origin.

The term e^{-bP} is often used to normalize measured strengths for examination of strength-grain size relationships[28], but its limitations should be remembered in that connection also.

The quantitative treatments have generally taken account of porosity by noting the total pore content[29]. I think this is too simple. The shape of the pore is a decisive factor. Sharp-edged pores act as stress raisers and are more effective as potential origins of fracture; they provide easy passage for the fracture front and so lower the work of fracture. Rounded pores, on the other hand, may actually raise the work of fracture above that of a nonporous material, as they are observed to do when they occur as inclusions within grains, by initiating cleavage steps.

The average pore size and the distribution of pore sizes also deserve attention. As a potential fracture origin, a single large pore can be decisive. Large pores lower the fracture energy more than the corresponding volume of fine pores. If we are to reach more than a rough rule-of-thumb treatment of porosity we must learn to deal with the dimensions and shapes of the most dangerous pores in a body in relation to strength, and with the shape mode and average size in questions of fracture energy.

The possibility that the origin flaw in dense ceramics is a separation along the boundary of a grain, suggests inserting the grain size, G, in a form of the Griffith equation

$$\sigma_F = KG^{-1/2} \qquad (7)$$

where K is a constant. Lack of fit to this expression when applied over wide ranges of grain size led Carniglia[30] to propose a two-stage model for Al_2O_3, MgO and BeO. (Fig. 4)

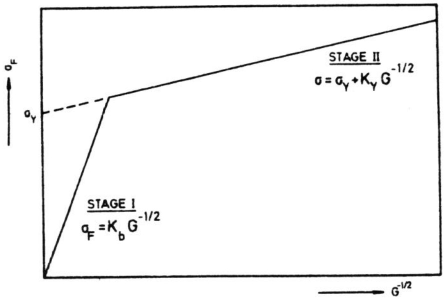

Fig. 4 Two-stage model for Strength vs Grain Size (After Carniglia[30]).

(8)

In Stage I, fracture is taken to originate at a grain-sized flaw, G, which grows catastrophically as soon as the stress level is sufficient to overcome the barrier provided by the complex of grains at its perimeter. The grain-sized flaw itself may have predated testing or it may have grown, in the grain boundary for example, at relatively low stress levels during the application of load. Stage II, adopted from a derivation for brittle metals by Petch[31], is reached when grain size is small enough to permit stress support to levels in excess of the critical limit for actuation of microplastic deformation. In addition to the energy for overcoming internal obstacles to microplastic yielding, i. e., the crystal property term σ_Y, the external barriers imposed by neighboring grains must be overcome; the necessary energy for this is represented by the grain-size-dependent term $K_Y G^{-1/2}$.

Mechanical twinning and dislocation mechanisms which, under sufficient stress, can provide microplastic flow in crystals of all but the exceptionally hard ceramic compounds are abundantly documented, and their interactions to produce cracking in bicrystals and polycrystalline bodies (Fig. 5) have been repeatedly demonstrated[32-35].

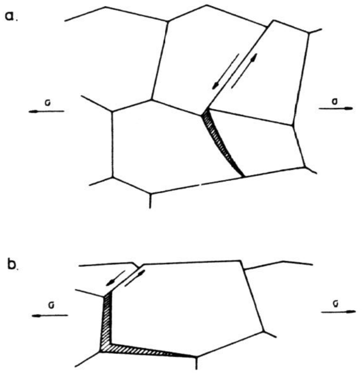

Fig. 5 Generation of cracks through microplastic mechanisms, a) by slip or twinning and b) by grain boundary sliding.

Ilschner[36] has explained the dual significance of microplastic yielding with respect to strength. On the one hand, it may form crack nuclei by piling up against obstacles as at grain boundaries; large grains allow correspondingly large pile-ups and so are more dangerous, under conditions of microplastic yielding, than are small grains. On the other hand microplasticity may relieve the stresses at the flaw tip and thus raise the stress to fracture. Its additional role in increasing the work of fracture has already been mentioned.

Rice[37] reviewed his own and published data for a large number of oxide, nitride, boride and carbide ceramics and was satisfied with their degree of fit to the Petch equation (Eq. 8) with a nonzero intercept of the σ_F versus $G^{-1/2}$ plot when allowance was made for porosity and for the effects of annealing, machining and impurity levels in the specimens. Differing methods of measuring grain size by various authors is given as one of many sources of the observed uncertainties in fit. The meaning of the intercept may be misinterpreted, according to Rice, if the stress to failure is not corrected for internally stored stress, σ_{int}, which arises from anisotropic contraction , $\Delta\alpha$, of the grains in cooling from the stress-relief temperature in manufacture. He calculates these stresses in a randomly oriented alumina to be

$$\sigma_{int} = \frac{\Delta\alpha}{2} E\,(T_{stress\ relief} - T_{test}) \qquad (9)$$

arguing, perhaps incorrectly, that the average stress will be half of the extreme value. (The average over all grain boundaries will be less than half the extreme for a hexagonal mineral such as alumina, while the stress at any prism face will be close to the extreme. The question is discussed from another standpoint by Wachtman and Dundurs[38].) The resulting internal stress values fail to account for differences in the Petch-plot intercept between randomly oriented hot-pressed alumina and fully oriented forged single crystals on the basis of any single stress-relief temperature, but the data (Fig. 6) were admitted to be somewhat scattered.

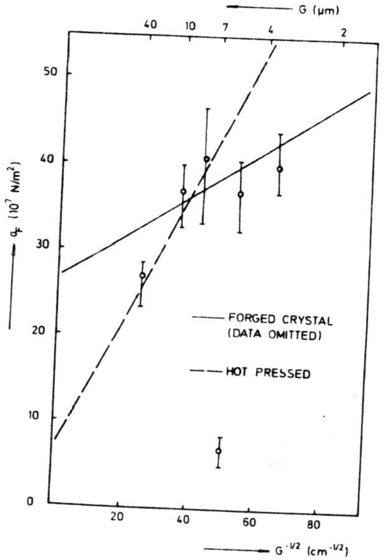

Fig. 6. Strength at -196°C of forged single-crystal and hot-pressed aluminas as a function of grain size (From Rice[37])

Nevertheless the existence of internal stored stresses in all but the isometric ceramics is quite real[39], and it would be astonishing if they did not sometimes play a part in fracture initiation and growth. Consideration of the grain boundary geometry of

dense oxide ceramics suggests that the predominant effect will be to tend to separate the most mismatched grains from one another. The principal consequence should be to provide grain-sized cracks which, on occasion, may adjoin one another or may link up with one another or with other flaws. But it does not seem reasonable to consider the internal stresses as simply adding to the applied stress.

The profound effect of impurities on the strength of oxide ceramics is well known. Not only cations but also anions may be deleterious and indeed anions may sometimes play the major role. Rice[19] reports for example that fractional percentages of residual carbonate and hydroxyl ions in hot pressed MgO cut strength in half; traces of certain cations were found to remain dissolved in the MgO crystallites and then raised strength, while in percentage amounts they segregated to grain boundaries and reduced strength.

Most work on grain size-strength relationships is untertaken at constant purity level in order to cancel out impurity effects. I think we must reconsider this strategy. In cases where the impurity is dissolved or evenly dispersed within the grains it may be justified. But in most instances impurities segregate in grain boundaries where they are accommodated in a layer whose area is inversely proportional to the square of the average grain diameter. Over the two- to sixty-micron range commonly studied, this represents a variation in boundary layer thickness of 900: 1 ! It would be amazing if this did not sometimes lead to a shift in fracture mode and other deviations from model behavior, particularly in rather pure materials. It would be interesting to study the strengths of a series in which the boundary layer thickness, rather than impurity content, were held constant.

The question of the grain size measurement itself appears to have suffered from extraordinary neglect. Knudsen[23], for example, presented particle size distribution data for his raw powders but only $\bar{d}_3$, the average diameter with respect to specific surface, for his fired specimens. Later

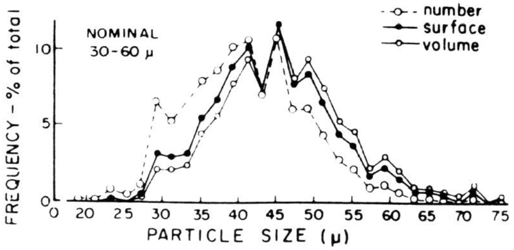

Fig. 7. Particle size distribution and $\bar{d}_3$ for Knudsen's ThO_2 raw powder[23].

authors have used the average length of intercept of a random line across a plane section or a fracture-exposed surface. Something of the $\bar{d}_3$ or linear intercept kind may be convenient for relating to the average work of fracture but it is not appropriate to predicting the weakest flaw in a strength measurement. Referring to Knudsen's data for ThO_2 in Fig. 7, his $\bar{d}_3$ is 45 μm, but it is apt to be larger grains, i. e., something between 70 and 75 μm, that will fix the size of the weakest flaw, while the energy to propagate will involve the smaller grains also.

A further difficulty lies in the fact that grain size is generally quite different at a fired surface from that in the interior, where measurements are usually made. This needs remembering in comparisons between as-fired and as-machined strengths.

5. Flaw Interactions

The possibility of cooperation between flaws to relieve stress, on the one hand, and to link up, forming larger flaws, on the other, has sometimes been suggested. (Fig. 8) We are indebted to Evans and Tappin[7] for criteria to identify the conditions where flaw linking will occur prior to catastrophic failure. They combine microscopic examination of actual fracture origins and potential origins with calculation of interactions between observed flaws with the aid of the Griffith equation normalized with respect to flaw geometry (Equation 2). Subsurface pore interactions are concluded to be far less significant than linking between a surface flaw and a subsurface pore. In the latter case, ligament failure, i. e., link-up, will occur when surface flaw size, C^*, is given by

$$C^* = \frac{2E\gamma_i}{\sigma^2 Y^2 K^2} \qquad (10)$$

where Y is the stress intensification coefficient for an edge-cracked

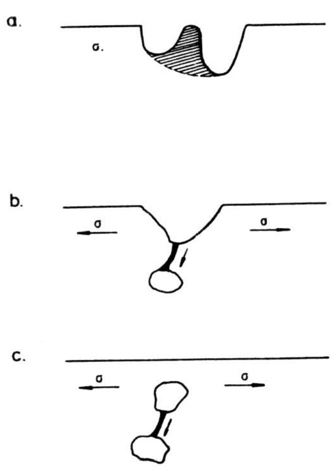

Fig. 8 Linking between flaws:
a) surface - surface
b) surface - subsurface
c) subsurface

geometry and K is the stress intensity factor for the ligament and involves the depth of the subsurface flaw. Stress to extend an isolated pore of radius a that just intersects the surface is given by

$$\sigma_b \approx \left(\frac{2E\gamma_i}{a}\right)^{1/2} \qquad (11)$$

and so linking will occur before catastrophe when

$$C^* > \frac{a}{K^2 Y^2} \qquad (12)$$

Figure 9 shows how the stress for linking a surface pore with a subsurface pore of the same diameter increases with separation distance. At small separation the link-up stress is low but the narrow flaw generated requires nearly as much energy to propagate to failure as does the surface pore alone. At greater separations the link-up stress increases and the propagation stress of the larger combined flaw lessens until a minimum in strength

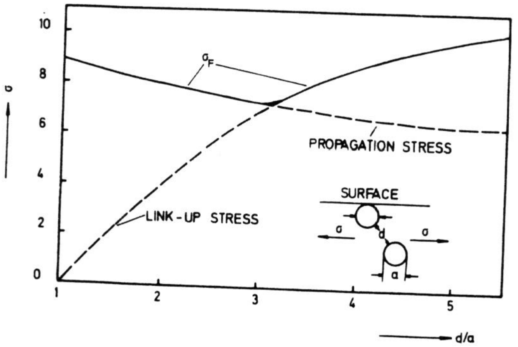

Fig. 9. Effect of pore separation on the applied stress needed for linking and on the stress for propagation of the flaw generated by linking. (After Evans and Tappin[7])

occurs when the separation is about three times the pore diameter. No linking occurs with further separation and there is less and less stress-raising influence from the presence of the subsurface pore, so that the stress to failure rises to that for an isolated surface pore.

Evans and Tappin[7] argue that no simple linear relationship can be expected to exist between fracture stress and $G^{-1/2}$ since 1) there is no necessary connection between grain size and flaw size, and 2) the energy for fracture initiation is itself a function of grain size.

6. Related Studies

Fracture is not always the ultimate catastrophe. In recent years has come the use of ceramics as light armor in which fracture of the ceramic is slow enough to frustrate penetration by a projectile. The sequence of initiation and growth of ballistically initiated fracture has been traced by Fréchette and Cline[40].

Scientific fracture studies have also taken cognizance of the practical experience that cracking of a refractory is not a catastrophe provided that the resulting cracks do not traverse the piece completely. In this connection Hasselman[41] distinguished between thermal shock fracture resistance and thermal shock damage resistance. He explains that the goal for thermal shock damage resistance should be the maximum strength after shock.

He therefore equated the total elastic stored energy at fracture with that expended to propagate the cracks half-way through the body, und calculated the optimum initial strength, S^*, of a sphere of diameter b as

$$S^* = 1/2 \left(\frac{35 \gamma}{2(1-\nu) b}\right)^{-1/2} \qquad (13)$$

where ν is the Poisson ratio. Porosity is helpful in spite of its effect in reducing thermal conductivity, by lowering the elastic moduli and therefore the stored energy.

A radical approach to the problem of thermal shock in rocket nozzle inserts was proposed by Robinson[42]. He precracked zirconium carbide inserts by water-quenching from high temperature and found the subsequent performance substantially improved. Evidently the intricate network of cracks introduced by the prequench[43] relieved shear forces at the surface during sudden heat in use, forces which otherwise produced shallow fragments, easily dislodged by the rocket blast stream.

It has long been known that the durablility of glass can be doubled by providing a surface layer under internal compressive stress. Before the flawed surface can be brought into dangerous tension by thermal or mechanical shock, this built-in compression must be overcome. Kirchner et al[44] have found it possible to double the strength of 96 °/o Al_2O_3 ceramics by quenching in oil from 1550°C. No improvement in strength was obtained from porous specimens which fractured from internal flaws[45]. Perhaps even more interesting is the observation that the quenched specimens were rendered much less subject to degradation of strength by abrasion[46]; abrasion damage was observed to penetrate to a lesser depth in agreement with this. (Fig. 10).

High-temperature fracture studies are not only appropriate to refractories, whose application involves high temperatures, but are also instructive in clarifying certain room-temperature fracture mechanisms. As an example of this, the data of Davidge and Tappin[43] on the strength of a siliceous 95 °/o aluminum oxide are instructive (Fig. 11). On rising temperature the strength decreases steadily until viscous flow of the glassy phase, beginning near 800°C, relieves stress at the crack tip and brings a sharp

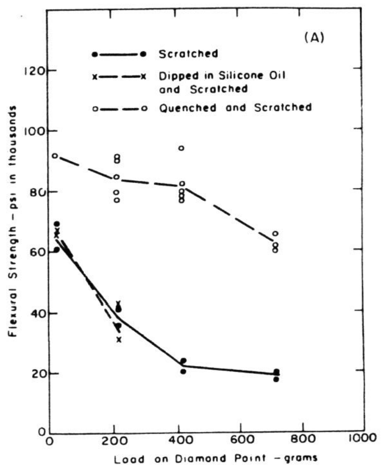

Fig. 10. Effect of abrasion with a loaded diamond point on the strength of 96% Al_2O_3 a) quenched and b) unquenched. (From Kirchner et al[46])

maximum in strength. Further increase in temperature brings a corrosive effect from the siliceous phase into play and the strength drops off sharply. They note that the flow properties of minor constituents can be important even if these are present in relatively small amounts. Similar conclusions were reached with respect to selfbonded silicon carbide[16], where a peak in strength

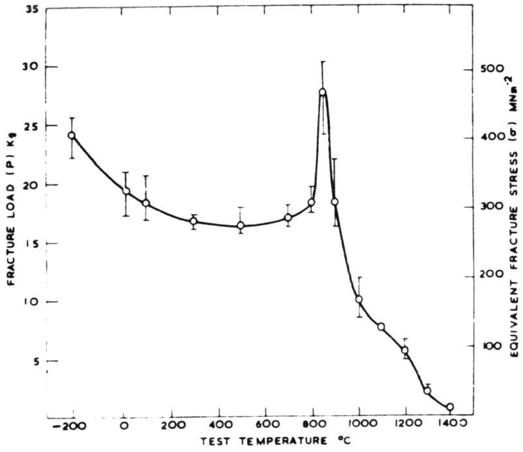

Fig. 11. Strength of a siliceous 95 % Al_2O_3 as a function of temperature (From Davidge and Tappin[43])

is observed close to the melting point of the metallic silicon phase. It seems natural to infer from this that microplastic mechanisms at room temperature may also work in this stress-relieving direction as well as sometimes serving to initiate cracking, as noted earlier.

7. Concluding Remarks

It is painful to have to stop so far short of having covered the subject assigned. Many aspects have had to be omitted and, worst of all, many people who have supplied key ideas to the understanding of fracture initiation and growth in ceramics have gone uncited. The papers listed here will, fortunately, provide reference to most of these.

It seemed more important, however, to emphasize a few salient points here rather than to try to miss nothing. Summarized, they are these:

1. The Griffith type of calculation, i.e., calculation of the dimension of the (hypothetical) Griffith flaw from measurements on large artificial flaws, is fraught with danger when applied to polycrystalline brittle materials such as ceramics, particularly when these are polyphase.
2. The present confusion concerning porosity and grain-size effects on strength is certain to continue until the shapes and size distributions are examined in the light of direct observations on actual origins, not as an occasional demonstration but throughout the ranges of the variables under consideration.
3. The treatment of impurity concentration as weight percent needs reconsideration in terms of such spatial distribution parameters as the thickness of grain boundary layers, particularly when a range of grain sizes is involved.
4. Ion migration, in the steep stress gradient and associated electrical potential gradient at the crack tip, is involved in questions of static fatigue of ceramics and can be expected to contribute to the work of fracture

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