AN ASSESSMENT OF THE STATE OF THE ART IN
PREDICTING THE FAILURE OF CERAMICS

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1. Introduction

The use of structural ceramics is increasing. As a consequence of the potential economy associated with their use, a great deal of research has been funded in an effort to develop sound, widely applicable models on which a comprehensive design methodology can be based. This report describes results of a survey of public literature on the prediction of failure in structural ceramics. The report is not a compendium of experimental results. Rather, it is an attempt to assess the adequacy of the models presented in the literature, and to summarize the degree to which existing design methodology based on these models has been established. In short, the report is an attempt to define the boundary between what is known and what is only suspected in predicting failure of structural ceramics.

Two kinds of failure are considered below: fracture upon exposure to an initial loading and fracture after an acceptably long service life. The former is governed by initiation of unstable crack propagation (fast fracture), while the latter is governed by fatigue mechanisms. We deal first with prediction of fast fracture, and then with predicting the length of service life.

2. Fast Fracture Prediction

It is generally assumed that fast fracture in a structural part is governed by some property of the material of which the part is made. This property, to which we shall refer as fast fracture resistance, depends on the chemical and thermal conditions to which the part is subjected. But for a given environment, one would expect fast fracture resistance to be manifested in the level of loading that can be tolerated without rupture. Furthermore, to the extent that a material is continuous, homogeneous, and isotropic, one would expect fast fracture resistance to be constant when the applied stress is constant. It is not surprising, then, that most fast fracture criteria proposed in the literature nominate the critical value of some parameter formed from elements of the stress tensor as the appropriate measure of fast fracture resistance. Thus, the usual criterion is that fast fracture occurs when the chosen stress state parameter attains its critical value.

2.1. Criteria for Linearly Elastic Material

Perhaps the earliest parameter assumed to govern fast fracture was maximum normal stress. In his famous statistical treatment of tensile strength data, Weibull (1939A, 1939B, 1951) assumed that fast fracture is initiated at a crack lying in a plane normal to the applied load. Such a crack would remain in its own plane as it propagated and would open only in Mode I. [For a description of the three basic modes of crack opening, see any general text on fracture mechanics, e.g. Kanninen (1985).] Given this assumption and uniaxial loading, the maximum normal stress is the logical candidate for the fast fracture resistance parameter. But the special configuration considered by Weibull is probably the only one for which one should expect the maximum normal stress criterion to apply. Nevertheless, it has been applied to other configurations. These include a very slender ellipse under biaxial loading [Griffith (1924); Paul (1968)], a sharp (plane) crack under biaxial loading [Erdogan (1963); Finnie (1973)], and an ellipsoidal cavity subjected to remotely applied combined loading consisting of normal and shear stresses [Paul (1976)].

The maximum normal stress criterion is not the only reasonable choice for a crack that opens purely in Mode I. Through linear elastic fracture mechanics (LEFM), normal stress at a crack tip can be related to the Mode I stress intensity factor (SIF) [Irwin (1957)], the Mode I strain energy release rate (SERR) [Griffith (1920); Paris (1965)], the J-integral [Rice (1968)], and the Mode I crack opening displacement (COD) [Wells (1963)]. Each of these has also been proposed as the parameter whose critical value represents a material's fast fracture resistance. But fast fracture criteria that account for only Mode I opening cannot explain the influence of hydrostatic stress on oxide ceramic tensile strength [Stucke (1975)] and on both tensile and shear strengths of NaCl [Kelly (1967)]. Nor can the Mode I criteria predict strength of SiC under various biaxial states of stress [Tappin (1978)]. Furthermore, the Mode I criteria are appropriate

only for a crack that remains in its original plane as it propagates. General stress states cause noncoplanar crack propagation.

An unusual alternative to the criteria cited above is the critical tensile strain criterion proposed by Shaw (1978). They found, however, that the critical value of tensile strain depends on mean normal stress, and so is not a material property. This is probably because a real crack grows when the total strain on interatomic bonds at the tip exceeds some critical value, so that a criterion based on only tensile strain is inappropriate. Considering only tensile strain is equivalent to assuming Mode I crack opening.

The various Mode I fast fracture criteria are inadequate because they fail to account for the fact that cracks subjected to general stress states open simultaneously in more than one mode (mixed-mode opening) and exhibit noncoplanar propagation. In the first attempts to account for these effects, the SERR criterion was applied to predict noncoplanar extension of a Griffith crack under biaxial loading, which induces simultaneous opening in Modes I and II [Swedlow (1965); Hussain (1974); Palaniswamy (1972, 1974, 1978); Nuismer (1975); Wang (1978)]. Palaniswamy (1978) contains a survey of mixed-mode criteria, an analysis of Mode III fracture, and a brief discussion of fracture under completely general stress. Recently, Ichikawa (1982) critically assessed relationships between the strain energy release rate and stress intensity factors, as predicted by various mixed-mode SERR criteria. This work includes derivation of a more consistent SERR criterion. Despite differences in the SERR criteria, their predictions of the stress required to propagate the crack and direction of crack propagation agree well with experimental data for some cases of biaxial stress.

Of all the attempts to account for mixed-mode crack opening and noncoplanar crack propagation, the most general are the strain energy density (SED) criteria. The original criterion [Sih (1974A, B, C; 1980)] is based on the total strain energy. A later version [Jayatilaka (1977A, B; 1978A, B, C; 1979)] is based on the strain energy of distortion. In general, the SED is a function of the local stress state and so varies continuously throughout an elastic body. As one moves from point to point around a crack, the SED somewhere attains a minimum value. The SED criteria state that when this minimum value rises to some critical value (the material's fast fracture resistance property), fracture occurs. The SED criteria were the first to deal with arbitrary, three-dimensional loading for which opening in Modes I, II, and III occurs simultaneously. For an extensive summary of the various crack geometries and loadings to which the total SED criterion has been applied, see Gdoutos (1984). For a summary of distortion SED applications, see Jayatilaka (1979).

In addition to the fast fracture criteria cited above, there are others in the literature. For a partial survey, see Swedlow (1976A). Many of these other criteria are based on the assumption of functional forms that have no physical basis, but can be fitted to some of the available experimental data. Examples of this type are the criteria proposed by Wang (1978), Stanley (1978A), and Alpa (1984).

Several comparisons of the more popular fast fracture criteria have been made. Petrovic (1976D, 1976E) compared the maximum normal stress, coplanar SERR, and noncoplanar SERR criteria for combined Mode I and Mode II opening. Batdorf (1977) compared the maximum normal stress and coplanar SERR criteria. Later, Batdorf (1978A) compared the maximum normal stress, noncoplanar SERR, and SED criteria for uniaxial and equibiaxial tension. Chen (1986) compared the maximum normal stress, noncoplanar SERR, and SED criteria for combined opening in Modes I and II, as well as in Modes I and III. It is worth noting that in the last two studies cited, the SED criterion was judged superior to the others examined. However, it should also be noted that (1) the differences between the various criteria are not necessarily significant for tensile loading, and (2) the ability of these criteria to predict fracture under compressive loading is questionable [Cotterell (1969); Swedlow (1976A)]. Since strengths in tension and compression are usually quite different, fast fracture criteria intended to be comprehensive contain separate parts for tension and compression [Jayatilaka (1978B, C; 1979); Stanley (1973)]. Discrepancies between theory and experiment for compressive loading may be associated with crack closing, for which opposite faces of the crack are in contact [Marshall (1984); Petrovic (1985)]. Criteria that account for sliding friction between opposite faces of a closed crack have been proposed [McClintock (1963); Swedlow (1976A)], but have not been firmly established [Hoek (1965)]. Comparison and discussion of various criteria when applied to biaxial stress states have been presented by Shetty and coworkers (1981, 1983, 1984A, 1984B).

The comparisons reported in the literature have only begun to resolve the issue as to which elastic fast fracture criterion is most generally applicable. Although agreement with experimental data is often good, the range of loading that has been considered is quite restricted. All that can be said with certainty is that for a given material placed in a given environment and subjected to a given test configuration, the relative merits of the existing criteria can be assessed. But should the material's microstructure, or its chemical or thermal environment, or the applied loads change, the criteria should be reassessed.

2.2. Criteria for Elastic-Plastic Material

Under certain conditions, most notably at elevated temperatures, even ceramics exhibit some plasticity. Since plasticity is a nonlinear phenomenon, its presence requires modification of the results of LEFM. One of the first attempts to account for plasticity at the tip of a crack in an otherwise elastic medium was presented by Dugdale (1960). Although useful for uniaxial stress, the Dugdale model is inappropriate when stresses are biaxial [Alpa (1979)]. A more general approach to elastic-plastic fracture mechanics is represented by the J-integral [Rice (1968)]. The J-integral fast fracture criterion states that fracture occurs when the J-integral attains a critical value that is the fast fracture resistance property referred to above. In the absence of plasticity, the J-integral criterion is equivalent to the Mode I SERR criterion [Kanninen (1985)]. But the J-integral extends the Mode I criteria into the elastic-plastic range. Many different applications of the J-integral in fracture mechanics can be found in Corten (1972), Kaufman (1973), Paris (1974), Swedlow (1976B), and Taplin (1978). The J-integral is limited, however, because it is defined only for biaxial stress states [Begley (1972)]. Hence, it cannot be applied for general loading. Furthermore, an underlying assumption of the J-integral criterion is that the strain energy density is singly valued. This assumption is inconsistent with unloading in the plastic range, and so invalidates the J-integral approach under these conditions [Thomson (1982)]. However, as pointed out in Begley (1972), the criterion is proposed to apply only to crack initiation, not crack propagation. It is not clear that unloading in the plastic range occurs during crack initiation. Hence, the J-integral criterion may be valid.

Recent efforts to model fast fracture in the elastic-plastic regime have focused on the "crack tip enclave" or "crack tip process zone" concept. A feature common to these models is that the crack tip is embedded in a small elastic zone that is surrounded by a plastic zone that, in turn, is surrounded by another elastic zone. Weertman (1978, 1980) applied the J-integral to a moving crack in an enclave model. Thomson (1978, 1982) focused on the dislocation density in the plastic region surrounding a Mode III crack. They replaced the J-integral with a contour integral that accounts for singular stresses at dislocations. Both of these enclave models suffer from lack of basic information concerning the boundaries between elastic and plastic zones. The two most important missing data are the locations of the boundaries and the stresses on those boundaries. This information expresses the influence of the surrounding dislocation field on the crack tip. Some insight into this influence may be found in the work of Majumdar (1981). But until much more work is done in this area, enclave models must be considered quite speculative.

2.3. Lattice Models

Our discussion has thus far been confined to concepts compatible with continuum mechanics. It has long been recognized that the process of fracture occurs at the scale of interatomic spacing, for which the continuum model is inappropriate. Barenblatt (1962) was one of the first to draw the attention of the engineering community to that small region within a few atomic radii of the crack tip where cohesive forces of lattice bonds must be overcome to achieve fracture. The dependence of these cohesive forces on lattice bond strain is highly nonlinear. Nonlinear models of these forces have been included in many studies of lattice deformations [Cribb (1967); Andersson (1970); Sinclair (1972, 1975); Smith (1975, 1978); Esterling (1976); Fuller (1978A)]. Reviews of the basic findings have been presented by Fuller (1978B) and Smith (1978). While many authors discuss the connection between lattice models and continuum models, no firm understanding of this connection exists. However, the fact that most, if not all, cracks in brittle materials are atomically sharp [Lawn (1980),(1983)] is persuasive evidence that atomic-scale models will be an important part of future fracture research.

2.4. Factors Influencing Fast Fracture

As indicated above, a plethora of fast fracture criteria are proposed and discussed in the literature. Nevertheless, no single criterion has been found to be clearly superior to the others. This is probably because each of the models on which the proposed criteria are based fails to account adequately for one or more of the factors described below.

2.4.1. Crack Geometry

Early work in fracture mechanics focused on simple crack geometries such as the Griffith crack, the ellipse, the ellipsoid, and the "penny-shaped" crack. It has been argued [Lawn (1975)] that both sharp and blunt cracks are essentially penny-like. But it has since been shown [Shetty (1983)] that spherical voids give different fracture strength than do planar cracks. Furthermore, crack propagation is a discontinuous noncoplanar, direction-changing process, and a crack front is not a smooth curve [Tappin (1978)]. It has also been shown [Adewoye (1983); Petrovic (1985)] that crack faces are irregular. It would seem, then that one of the reasons that no fast fracture criterion has been firmly established as satisfactory for a wide range of conditions is that none of the existing criteria incorporates a realistic crack geometry.

2.4.2. Microstructure

The relationship between microstructure and mechanical behavior of structural ceramics has been studied for many years. Summaries of early work are given in Stokes (1972) and Coble (1982). Early conclusions that rounded porosity raises toughness [Passmore (1965)], and that microstructure influences elastic modulus [Bailey (1970)] and strength [Dalglish (1975)], have

been followed by a steady stream of experiments in this area. It has been concluded that the microcleavage that initiates brittle fracture occurs within about two grain diameters of a macrocrack tip [Armstrong (1978)]. The relationship between microstructure and microcleavage has been studied directly [Cooke (1982)].

Virtually every feature of microstructure influences the failure of structural ceramics. Porosity has been shown to influence strength [Godfrey (1973), Shetty (1983)], toughness [Passmore (1965), Evans (1976)], fracture energy [Rice, R. W. (1978)], fatigue induced by thermal cycling [Ziegler (1982)], and critical flaw size [Hanney (1982)]. The effects of spherical and cylindrical cavities have also been studied [Evans (1979A, B)]. Inclusions have been shown to influence stress fields and crack initiation [Evans (1974A)]. Grain size affects strength, fracture toughness [Claussen (1975)] and fracture energy [Rice (1980, 1981B)]. The latter effect has been modeled by Rice (1981A) and Usami (1986).

Microcrack initiation and coalescence are highly dependent on microstructure [Cooke (1982)]. Morena (1983) has shown that microstructure affects the size of cracks formed by indentation damage.

As pointed out by Wiederhorn (1983), "...for many ceramics, the size of the crack relative to the size of the microstructure is dominant in establishing resistance to crack growth." Thus, it has been found that microstructure influences the deflection of a growing crack [Faber (1983A, B)].

Given the variety of microstructural features on which the details of fracture depend, one must conclude that the fast fracture criteria described above are rather simplistic attempts to capture the essence of an extremely complicated phenomenon.

2.4.3. Thermal and Chemical Environment

Because of thermal expansion mismatches between constituents and thermally induced changes in microstructure, thermal environment plays a role in fast fracture. For example, it has been established that the strength and fracture toughness of magnesia are affected by temperature. One of the microstructural changes associated with high temperature exposure of silicon nitride and other materials is crack healing. Annealing promotes healing, even in regions without residual stresses [Petrovic (1976A)]. Adewoye (1983) observed crack healing in the hot-stage of TEM. The importance of this effect is illustrated by the results of Marshall (1984), where it was found that annealing relaxes residual stress and so affects whether an initial crack propagates stably or unstably.

Presumably because of stress corrosion that occurs before loading, chemical environment has an influence on fast fracture. For instance, it has been shown that the presence of water decreases the strength of debased alumina [Davidge (1970)]. Of more importance for high temperature applications is the fact that oxidation has a strong influence on the strength of reaction-bonded silicon nitride (RBSN) [Jayatilaka (1975), Wills (1976), Godfrey (1978)].

For some materials, both chemical and thermal conditions are significant. Pabst (1983) has found this to be true for fracture toughness in some materials.

2.5. Statistics of Flaw Populations

The existence of a reasonable fast fracture criterion is, by itself, of little help in predicting fast fracture in a real part. Because real parts contain many flaws of various sizes, shapes and orientations, predicting failure from fast fracture cannot hope to succeed without taking into account the statistics of these variations. Ways of representing these variations, and various issues associated with these representations, are discussed below.

2.5.1. Functional Forms

Every structural part contains flaws. For each flaw, the value attained by the fast fracture resistance parameter (FFRP) depends on the level of the applied loads. Furthermore, for each flaw there is a critical level of loading for which the FFRP attains its critical value and fast fracture occurs. For a given level of loading, every part contains some number of "critical flaws," i.e., flaws whose critical level of loading is less than or equal to the given level. Thus, the density of critical flaws is a function of the level of loading. The greater the applied loads, the greater the critical flaw density. A central question in the statistics of fracture is the manner in which the critical flaw density depends on the level of applied stress. The function that describes this dependence is referred to as the critical flaw density distribution (CFDD).

Many different functional forms have been proposed for the CFDD. The first of these is that of Weibull (1949). Although it was originally proposed for heuristic reasons, Weibull's function was later recognized [Freudenthal (1953)] as the third asymptotic probability of smallest values [Gumbel (1958)]. It has been shown that fitting data to Weibull's three-parameter equation is not a unique process [Shih (1980)]. In the same study, it was found that differences between various fits to a single set of data are small in the range of the data, but large in extrapolated portions of curve.

McClintock (1974) noted the existence of three different asymptotic extreme value distributions, and argued that the CFDD should have the asymptotic distribution of the first kind. It was shown that if this were so, the distribution of breaking strengths in tensile tests

would be unlike any of the three asymptotic distributions.

Matthews (1976) showed that for certain simple stress states, the CFDD could be computed from fracture test data without assuming any functional form. This method was successfully applied to predicting strength distributions from three-point bend test data. Evans (1978C) extended the method of Matthews (1976) to biaxial configurations.

Jayalatika (1978A) proposed a CFDD function based on experimental data. The proposed function is different from those previously cited.

2.5.2. Issues

The superiority of any of the CFDD functions discussed above has yet to be established. This fact may be related to any or all of the issues cited below.

Most investigators assume that flaws are isolated and do not interact with each other. But it is known that in compression, cracks must coalesce before failure can occur [Brace (1963)]. Although stress fields for arrays of cracks have been studied to some extent [Nied (1983)], no attempt has been made to incorporate flaw interactions into existing design methodologies.

Most investigators assume that flaws are randomly oriented. But Batdorf (1978C) found that assuming flaws had a preferred orientation improved agreement between analysis and experiment. Furthermore, Shetty (1984B) cited fractographic evidence of Rice (1981C) that surface flaws generated in grinding of polycrystalline ceramics have two preferred orientations. This kind of information has yet to be used in design procedures.

Most investigators assume that there is only one population of flaws in a given part. But appropriate methods for dealing with multiple flaw populations are not the same as those for one population [Johnson (1983)].

2.6. Design Methodology (DM)

The primary task of any fast fracture DM is to synthesize the results of mechanical modeling (the fast fracture criterion) with the results of statistical modeling (the CFDD). The synthesis must be based on a logical scheme that does not violate any of the assumptions on which the models are based.

The earliest fast fracture DM to include a statistical model was that of Weibull (1939). The Weibull DM employs the maximum normal stress (S_{MAX}) fast fracture criterion and the Weibull CFDD. The scheme is to integrate the probability of failure per unit volume (or per unit area)

over the volume (or surface area) of the part in question to obtain the probability of failure of the entire part. The Weibull DM was originally conceived for a uniaxial stress state, and initially applied to tensile tests. Weill (1963) extended the Weibull DM to bending tests, in which the stress is uniaxial, but nonuniform.

The Weibull DM has been used for a biaxial stress state in which the maximum normal stress is a function of two space coordinates [Oh (1973); Vardar (1975)]. In these works, the Weibull surface integral was computed numerically.

Excellent agreement between analysis and experiment for biaxial stress with both principal stresses tensile has been obtained using the Weibull DM [Swank (1981)]. It should be noted not only that both principal stresses were tensile, but also that the specimens used were carefully chosen so as to be essentially identical in microstructure.

Finnie (1969) anticipated some of the problems associated with application of the Weibull DM when the stress state is compressive and/or triaxial.

Stanley (1973) recognized that in a multiaxial stress state, any one of the principal stresses could produce fracture. The idea is based on the fact that flaws are of different sizes. A large flaw perpendicular to a small principal stress could fail before a smaller flaw perpendicular to a larger principal stress. In other words, the "weakest link" is not necessarily a flaw aligned with the direction of the maximum principal stress. Hence, the presence of more than one nonzero principal stress should increase the risk of failure. (Because it considers only the largest principal stress, the Weibull DM does not predict an increased risk of rupture for multiaxial stress.) Stanley and coworkers combined the risks of failure due to all three principal stresses by invoking the principal of independent action (PIA), which is that the probability of failure associated with any one principal stress is not affected by the presence of another principal stress. This implies that for multiaxial stress, the probability of survival is simply the product of the probabilities of survival associated with each nonzero principal stress's acting alone.

Batdorf (1974B) argued that the PIA is not appropriate because all nonzero principal stresses contribute to the normal stress experienced by a flaw not aligned with any principal axis. Hence, if such a flaw were larger than any flaw perpendicular to a principal axis, it could fail first. The "Batdorf" DM, as it is commonly called, accounts for this effect. An important difference between it and that of Weibull is the presence of a stress-space integral inside Weibull's volume/surface integral. Although it originally used the maximum normal stress fast fracture criterion, the Batdorf DM has been modified to incorporate the SERR criterion [Batdorf (1978A, B, C)].

Jayatilaka (1977A, B) proposed a DM that, like that of Batdorf, accounts for interaction of principal stresses. Differences between the two are that Jayatilaka's 1) is only formulated for biaxial stress, 2) uses SED fast fracture criterion, and 3) replaces Batdorf's stress-space integral with one in crack-length space. The relationship between these two DM's has not been explored.

Problems common to all proposed DM's are failure to eliminate bad data in determining parameters [Neal (1983)] and extrapolating for untested strength ranges [Evans (1978A, B)].

Current DM has been incorporated in finite element computer programs. An example of this is discussed in Gyekenyesi (1985). The procedure used there is to apply the Batdorf DM to determine the probability of survival of each volume/surface element of the part. The probability of survival of the entire part is then computed on the assumption that failure of any one element precipitates failure of the entire part.

A brief summary of DM studies in the literature is presented in Tables 2-1. and 2-2.

<u>Investigator</u>	<u>Year</u>	<u>CFDD</u>	<u>DM</u>	<u>FFC</u>	<u>SS</u>
Alpa	1984	2	B	AH	B
Batdorf	1974A	PS	B	SMAX	-
Batdorf	1974B	PS	B	SMAX	B
Batdorf	1977	2	B	SMAX	B
Batdorf	1978A	2	B	SMAX, SERR	B
Batdorf	1978B,C	3	B	SMAX, SERR	B
Batdorf	1979	PS	B	SMAX	B
Davies	1973	2	W	SMAX	U
Evans	1978A,B	3	B	SERR	B
Evans	1978C	N	B	SMAX	B
Evans	1979A,B	2	B	SMAX	B
Finnie	1969	2	W	SMAX	B
Freudenthal	1953	2	W	SMAX	U
Giovan	1979	2,PS	W,PIA,B	SERR	B
Jayatilaka	1977A,B;				
	1978B,C	J	B	SED	B
Lamon	1983	2	B	SERR	B
Matsuo	1981	3	B	SERR	B
Matthews	1976	N	B	SERR	B
Rabie	1986	2	PIA	AH	B
Rufin	1984	2	B,PIA	SMAX	B
Shetty	1984B	2	B	SMAX, SERR	B
Stanley	1973;				
	1978B,C	2	PIA	SMAX	T
Stanley	1982	2	PIA	SMAX	T
Vardar	1975	2	B	SMAX	B
Weill	1963	3	W	SMAX	U

Table 2-1: Recent Studies of Fast Fracture Design Methodologies (Symbols are defined in Table 2-2.)

CFDD (Critical Flaw Density Distribution)

2	= two-parameter Weibull
3	= three-parameter Weibull
PS	= power series
J	= Jayatilka's
N	= none assumed

DM (Design Methodology)

W	= Weibull
PIA	= principle of independent action
B	= Batdorf

FFC (Fast Fracture Criteria)

AH	= ad hoc
SMAX	= maximum normal stress
SERR	= strain energy release rate
SED	= strain energy density

SS (Stress State)

U	= uniaxial
B	= biaxial
T	= triaxial

Table 2-2: Symbol Key for Table 2-1.

3. Life Prediction

Ideally, the method one uses to predict fast fracture should form the basis for predicting service life. The design methodology for the former should be extended to incorporate time-dependence. In practice, each attempt to accomplish this extension is found to have a narrow range of applicability. The primary reason for this is lack of understanding of the fatigue mechanisms operating during prolonged loading. These mechanisms are discussed next. Subsequently, existing design methodologies are assessed.

3.1. Static Fatigue Mechanisms

Any attempt to classify experimental studies dealing with mechanisms of static fatigue in structural ceramics is met with the problem of identifying the operative mechanism(s) in each of the experiments. In many cases, the investigator himself has been unable to do this. In that situation, the course usually adopted is to refer to the mechanism as either subcritical crack growth (SCG) or stress corrosion. Since SCG can be either purely mechanical or due to stress corrosion, or both, the labels chosen by investigators may not be precise. No attempt has been made herein to second guess the investigators in this regard.

3.1.1. Microcracking: Initiation and Coalescence

The importance of microcrack initiation and coalescence in fatigue has been recognized for some time. Hoek (1965) studied crack arrays and coalescence in compression because coalescence precedes failure in compression. Cooke (1982) found that microcracking precedes fracture under other conditions, as well. Hasselman (1983) concluded that over a wide range of temperature, microcracks form at pores and triple points and propagate along grain boundaries with viscous glassy phases. Microcracking around spherical inclusions has been studied by Green (1983). Very little is certain about the conditions that determine when microcracks are initiated, except that there is a minimum grain size below which initiation does not occur [Gesing (1983)].

Only a few attempts have been made to model microcrack initiation and/or coalescence. Lawn (1977) developed a model for crack formation in indentation tests. A more general approach was taken by Hoagland (1980) who simulated generation of microcracks ahead of a Griffith crack in an infinite sheet subjected to uniaxial tension normal to the crack. The study explored the influence of various microcrack generation criteria on the resulting damage zone. Although this last work was entirely numerical, it offers a framework within which realistic generation and coalescence models can be studied.

3.1.2. Stress Corrosion

Although stress corrosion has long been considered an important fatigue mechanism, the manner in which it changes crack geometry is still a matter of some debate. Charles (1958A, B, C) studied stress corrosion of soda-lime glass in water vapor, and concluded that crack lengthening occurred and led to failure. Doremus (1980) modeled stress corrosion of an initially elliptical crack and concluded that crack sharpening, rather than crack lengthening, is the mechanism of fatigue in glass. Michalske (1983) found that a stress corrosion limit exists in soda-lime glass. Below it, cracks blunt but do not extend. After blunting, raising the stress above the corrosion limit does not cause crack extension until after the crack resharpen.

As one might expect, stress corrosion is strongly dependent on the thermal and chemical environment. Gulden (1976) reported temperature and humidity dependence of stress corrosion in silicon nitride. Brooks (1978) found that for hot-pressed silicon nitride in a residual fuel oil bath, the sensitivity of stress corrosion to the sodium content of the oil varied greatly with temperature.

Because the rate of stress corrosion can be limited by the rate at which the corrosive substance can reach the corrosion site, the time-dependence of stress corrosion is of some interest. While studying the strength of Vickers-indented silicate glass rods immersed in acid solutions, Dabbs (1982) found a complicated time-dependence.

3.1.3. Crack Healing and Arrest

Prediction of fatigue life is complicated by the fact that damage does not always increase with time. Cracks can close and heal under certain conditions; and propagating cracks can be arrested upon entering regions of changed stress state or microstructure. Wills (1976) observed healing in RBSN from annealing, and Petrovic (1976B) observed the same phenomenon in SiC. An oxidation mechanism was postulated as the driving force in the latter study. Other instances of crack healing are referred to in Section 2.4.3, above.

An extensive survey of crack healing literature has been made [Case (1983)]. In this same study, Young's modulus measurements showed that healing depends on grain size and the presence of cavities. Davidge (1982) discusses this work. The thermodynamics of crack healing have also been considered [Rice, J. R. (1978)].

Except for atomistic models that allow for crack arrest via lattice trapping, little has been published on the subject of crack arrest. El Haddad (1979) observed that the crack growth threshold stress depends on flaw geometry, and concluded that sharp cracks can be arrested because the threshold stress rises as the crack grows.

3.1.4. Subcritical Crack Growth (SCG)

SCG has long been considered the most important mechanism in fatigue failure of ceramics. But, as will be apparent from the following discussion, SCG is not an easy process to model. The basic conclusions of every study of SCG include comments as to the mechanism governing SCG and/or the manner in which the crack velocity depends on the local SIF. This velocity-SIF relationship is commonly referred to as the crack growth law (CGL).

The mechanisms that have been suggested as governing SCG include grain boundary sliding, stress corrosion, dislocation movement, and crack coalescence, to name a few. Lange (1974) found that for hot-pressed silicon nitride (HPSN), SCG occurred by grain boundary sliding influenced by viscosity of the boundary phase. Evans (1975B) concluded that the mechanisms controlling microcrack growth and macrocrack growth are the same.

Evans (1980) found that inhomogeneities give rise to preferred sites for cavity nucleation. Nucleated cavities ultimately coalesce to make a macrocrack, which then grows to fracture. This same process was described by Porter (1981) and Hsueh (1981). In the latter study, it was also concluded that most polycrystalline ceramics should exhibit this processes.

The influence of stress corrosion on SCG has been cited by many investigators, e.g., Rowcliffe (1975), Bhaduri (1983). By comparing several materials, Byrne (1982) demonstrated a wide range of sensitivity to stress corrosion crack growth. It was pointed out in that work that studies aiming to relate stress to crack growth rate are irrelevant if stress corrosion is high.

Minford (1983) attributed SCG to grain boundary diffusion, and found that it does not occur below a measured static fatigue limit. Below that, flaw blunting, probably through viscous relaxation at the crack tip, occurs. The static fatigue limit, or threshold stress, was found by El Haddad (1979) to depend on flaw geometry (sharp vs. blunt). Using a lattice model, Thomson (1983) gave theoretical grounds for existence of the threshold stress and its dependence on flaw size.

The functional form usually assumed for the CGL is that the crack velocity is proportional to some power, n , of the local SIF. The power function form goes back at least as far as the work of Charles (1958C). Evans (1975C) applied the power function to SCG of silicon carbide in water, and found that it occurred only at high temperature. McHenry (1976) found results that agree with those of Evans (1975C).

The power function CGL curve is usually assumed to have three regions, not all of which will necessarily be exhibited by a given material. Using the three-region power function, Evans

(1975B) gave details of relating crack growth rate to stress rate in soda-lime glass.

There is ample evidence that the power function CGL is not fundamentally correct for ceramics. McHenry (1977), studying direct-bonded silicon carbide at high temperature, observed no stable crack growth. Rather, repeated, short-lived, irregular, unstable crack propagation and arrest were observed. Fracture was transgranular. Nevertheless, a classical three-region plot was given. Behavior in the middle region of the curve was different from that reported in Evans (1975C). On the other hand, hot-pressed silicon carbide (HPSC) at elevated temperature exhibited stable SCG and intergranular fracture. For bending of HPSN in a high temperature vacuum, Mendiratta (1978) found no simple relationship between SIF and crack velocity. It was concluded that crack front interaction with local heterogeneities produced highly variable crack velocity.

Other problems with the power function CGL have been documented. The SIF exponent, which should be a material property in each of the three regions, has been found to vary with flaw size [Freiman (1980)], thermal and chemical environment [Pabst (1983)], residual stresses [Chantikul (1978, 1981); Lawn (1979, 1981); Marshall (1979A, B)], and grain size [Gesing (1983)]. Gupta (1983) used Charles' growth law, with special attention to the dependence of its parameters on temperature and humidity. Doremus (1980) has argued that the mechanisms for microcrack and macrocrack growth are not the same, so that the power function CGL cannot be appropriate for both. Difficulties in applying the power function CGL in design are discussed in Section 3.3, below.

Some few alternatives to the power function CGL have been presented. Tradinik (1983) proposed a completely new growth law based on sound reasoning from first principles, but it has not yet been seriously studied. Whether this, or some other functional form will supplant the power function CGL in practice remains to be seen.

3.2. Creep and Plasticity

Although the fracture of structural ceramics is most often brittle fracture, the phenomena of creep and plasticity become evident for failure at elevated temperatures. The mechanisms of creep and plasticity in ceramics are essentially the same as those in metals, but the degree to which they occur is typically much less for ceramics than it is for metals.

Most recent work on creep in structural ceramics has focused on silicon nitride. Ashcroft (1973) identified creep in silicon nitride at high temperatures. Lange (1974) attributed high-temperature SCG in HPSN to creep caused by grain boundary sliding. The role of grain

boundary sliding has been recognized by many others. One of these is Din (1975) who concluded that the sliding was accompanied by some dislocation climb and/or glide. Another investigator who attributed silicon nitride creep to grain boundary sliding was Seltzer (1977), who found accompanying void formation. Arons (1979) concluded that HPSN shows strong viscoelastic strain recovery consistent with the theory that creep is due to grain boundary slide associated with viscous flow of the glassy boundary phase.

Wilshire (1982) found that unlike creep of MgO, CaO, and lime, creep of RBSN and HPSN is controlled by generation of microcracks at grain boundaries, which then allow grain boundary sliding. The process is very sensitive to microstructure, so that new experiments are needed every time the microstructure changes. Birch (1978) had previously reached many of the same conclusions as Wilshire (1982). The influence of microstructure was also found by Grathwohl (1978) for RBSN. There it was concluded that if the microstructure is not fine and uniform, it strongly influences the degree of internal oxidation at high temperature, which, in turn, strongly influences creep.

Creep of alumina has been examined by several people. In an early investigation, Walles (1970) interpreted results as reason for optimism regarding prediction of long-time creep. Hasselman (1983) calculated creep rates for many mechanisms, and found only slow crack growth to be fast enough to account for the observed creep rates. For polycrystalline alumina, Johnson (1984) associated creep with a process of crack nucleation, growth, and coalescence, accompanied by shear band formation. Dagleish (1984) and Blumenthal (1984) describe the same process as that discussed by Johnson (1984).

Birch (1975) concluded that creep of magnesia is by diffusion-controlled growth of the dislocation network.

Carter (1984) found that high temperature creep of reaction-bonded silicon carbide occurred by dislocation glide/climb, controlled by climb.

Many discussions of plastic deformation mechanisms in ceramics have been published, e.g., Marsh (1963), Burns (1966), Groves (1970), Evans (1970), Heuer (1975). These discussions do not go beyond the question of mechanisms to consider the macroscopic effects of plastic flow. As pointed out by Evans (1970) and others, the slip planes required for plastic deformation are not available in most ceramics. Consequently, the microcracking, void formation, and grain boundary sliding associated with creep in ceramics occurs before slip can occur [Heuer (1975)]. It is unlikely, then, that plastic deformation by slip processes will ever be a significant factor in the failure of ceramics.

3.3. Design Methodology (DM)

The goal of the DM is to compute the length of service life versus the history of loading, temperature, etc. To do this, one must have some information about the severity of the initial flaw population. While DM's differ in the manner in which information about the initial state is acquired, most use the power function CGL to track the evolution of strength. As pointed out by Wiederhorn (1976), there are three ways in which the initial conditions are commonly assessed; these are strength testing, proof testing, and NDE.

The usual method of tracking strength degradation is based on integration of the power function CGL to obtain time-to-failure versus strength and failure strength versus stress/strain rate. The details are discussed by many authors [Evans (1972); Davidge (1973); Evans (1974B, C)].

Davidge (1973) showed how to combine the strength tracking analysis with Weibull parameters from strength tests and SCG parameters obtained from constant strain/stress rate strength tests to create strength-probability-time (SPT) diagrams. It was pointed out that while the diagrams form the basis of a straightforward DM, they are valid only for the conditions under which the test data were gathered. McLaren (1975) developed SPT diagrams for alumina.

Wilkins (1974) advocated the use of homologous stress versus fatigue life plots.

Evans (1974B, C) showed how to combine the strength tracking analysis with the results of proof tests. The resulting DM has also been discussed and/or illustrated by Wiederhorn (1974), Evans (1975A,D), Jakus (1983) and Ritter (1976; 1978; 1983A, B). In this last work, the method was illustrated with SPT diagrams. Problems with using the proof test DM for the various regions of the three-region power function CGL are discussed by Evans (1975A) and Ritter (1983A).

Wiederhorn (1976) has performed an error analysis of the various DM's. In this work it was shown how to generate associated diagrams, compare, and choose an appropriate DM.

Jacobs (1976) has discussed the computation of 90% confidence bands on lifetime predictions from the proof stress DM. A Monte Carlo technique and law of propagation of errors were used. It was found that the bands extend over several orders of magnitude, but can be narrowed by raising the proof stress or lowering the allowable stress.

Lifetime prediction is very sensitive to the form of the CGL [Ritter (1983A), Wiederhorn (1984B)]. The DM's discussed above assume that a) SCG is the only operative fatigue mechanism

[Ritter (1983A, B)], and b) there is no change in the CGL for low velocities [Evans (1974B, C)]. Hence, the existing DM's probably have very narrow ranges of applicability. Extrapolation of results to strength ranges not tested (low crack growth velocity, long service life) is extremely dangerous [Bortz (1983), Evans (1974B, C, D), Ritter (1983A), Wiederhorn (1976).] A representative statement of the warnings encountered in these works is that from Evans (1974C):

"...predictions are based on extrapolation to crack velocities which are several orders of magnitude lower than the measured values; this extrapolation may be invalid, and the diagrams should, therefore, be regarded only as indications of performance until additional low velocity data are obtained."

If microcrack growth and macrocrack growth are not governed by the same equation, extrapolation to low velocities is not appropriate [Ritter (1977)]. Hence, before assuming crack velocity test data are applicable, one must demonstrate agreement between values of CGL parameters obtained from crack velocity tests and those obtained from strength tests. Ritter (1978) showed that a DM using the power function CGL successfully predicted time to failure for soda-lime glass. This is consistent with the fact that soda-lime glass seems to obey the same CGL for both low and moderate velocities.

Ritter (1983A) concluded that for some materials, CGL parameters must be determined in a test environment simulating service conditions because the parameters depend on the test used to obtain them. Pletka (1982) tested five materials by different techniques and concluded that for three of them, power function CGL parameters depend on the kind of test used to compute them. This behavior was attributed to the materials' coarse-grained microstructures.

Trantina (1979) compared the use of power and exponential functions for the CGL in life prediction of HPSN at high temperature. Both functions gave the same results for time-to-failure for constant stress rate and linear cyclic stress. But the exponential function did better at predicting strength degradation versus time. Lenoe (1983) pointed out several details of crack propagation that are inconsistent with the power function CGL.

Arguing from first principles, Tradinik (1983) derived a new CGL. The function was integrated to obtain a "strength function," which accounts for time-dependence of strength. It was shown how to combine the strength function with Weibull statistics to get time-dependent probability of failure. Pabst (1982) used the method of Tradinik (1983) and pointed out that the information required for an accurate "strength function" is not easy to get. It was concluded that accurate predictions of failure probability versus time may be "nearly impossible" to attain.

In addition to the alternative CGL suggested in Tradinik (1983), some other modifications of the prevailing DM have been proposed. Sines (1976) argued that when a crack moves through

changing geometry, fracture toughness changes. This implies that the usual fracture mechanics equations should be replaced with something like the R-curve approach. Jakus (1978) presented new statistical methods for analyzing fatigue data so as to reduce uncertainty in lifetime predictions. Lenoe (1983) argued that transient, nonlinear viscoelastic effects must be accounted for to get realistic lifetime estimates for silicon carbide and silicon nitride at elevated temperature. Fuller (1983) incorporated a residual stress term in the power function CGL and showed how it affects lifetime calculation. Ritter (1983B) showed how to include the effects of flaw healing and pit formation, provided suitable models for them are used.

Current thinking on this subject is not all positive. Wiederhorn (1984B) pointed out that many effects are not yet understood and argued for caution in applying existing DM's. Lenoe (1983) also said that much improvement is needed. Gee (1982) went so far as to say that not enough material data exists to establish a DM for structural ceramics. Papers on many aspects of assessing structural reliability of brittle materials can be found in Freiman (1984).

Several recent studies have offered suggestions regarding areas where better understanding is needed. Evans (1983) argued that the basic idea of existing DM's is simple in concept but involved in application. It was also said that we need better models of SCG in real microstructures and better flaw population characterization through improvements in NDE techniques. Bortz (1983) discussed DM problems. The need for fractography to verify the assumption that a single flaw population is present, and for better models of SCG, were articulated. A list of specific research areas recommended for support was also presented.

Several new suggestions for changes in the existing DM's have recently been put forth. Cook (1984) developed and used "master maps" to determine fracture toughness and lifetime parameters. Bogdanoff (1983) presented a phenomenological damage approach in which the strength degradation process is modeled as a nonstationary Markov chain. Strauss (1985) gave a new statistical approach in which period doubling theory (related to chaos theory of nonlinear systems) was applied to predicting microcrack formation, damage, and failure. Finally, Tighe (1983) and Wiederhorn (1983) developed "strength degradation maps." In this last work, the stochastic nature of the processes leading to failure was recognized. It was suggested that probability density functions be used to describe the evolution of strength with time. This idea is broad enough that it might eventually lead to a more widely applicable DM than presently exists.

4. Recommendations for Future Research

The general requirements for successful design for brittle fracture are a) mechanical models of fracture processes, b) statistical models of the microstructure and flaw populations of the material in question, and c) a logical scheme (design methodology) for combining the models. No matter how carefully design methodology is formulated, it can be no more accurate than the models it uses. Shortcomings in modeling imply inadequate design methodology. Thus, when existing design methodology fails to provide broadly useful results, one must first ask where the fault lies. The following discussion outlines shortcomings of existing mechanical and statistical models and raises questions pertinent to development of improved models.

4.1. Mechanical Modeling

Because brittle fractures occur at different rates and by different processes, the primary goal of mechanical models is to predict, for realistic situations, which processes occur and at what rates. An outline of the key issues associated with mechanical modeling of brittle fracture is given below.

For unstable macrocrack propagation (fast fracture), the first process to be predicted is fracture initiation. This is a stability problem and should be analyzed as such. An excellent introduction to this problem is given by Barenblatt (1962). He solved the problem for the simplest crack geometry and the simplest stress state. Apparently, his work has yet to be extended to realistic stress states and realistic crack geometries.

After initiation, fast fracture is generally noncoplanar, may be either transgranular or intergranular, may exhibit crack branching, and may even be arrested. Crack arrest is clearly the most important of these issues, but the others may play important roles in crack arrest. Although some of the experimental data on these phenomena have been carefully obtained, modeling of these processes has yet to deal with realistic crack geometry. Again, extension of the work of Barenblatt (1962) should relate directly to these phenomena.

Slow fracture can occur on either the macrocrack or microcrack scale. For the former, the process is essentially slow crack growth, such as that observed in, for example, double torsion tests. On the microcrack scale, however, several processes have been observed. These include initiation, growth, healing, and coalescence. This last phenomenon leads eventually to formation of a macrocrack, and so changes both the scale and the process of interest. While the first three of these processes are closely related to the problems of macrocrack propagation described above, no fundamental investigation of coalescence of cracks has been reported in the literature. Models of all these processes are sorely needed.

To predict occurrence and rates of slow fracture processes on the microcrack scale, a model must account for known mechanisms, such as grain boundary separation (cavitation), grain boundary sliding, and corrosion. Although these mechanisms are often referred to in published descriptions of experimental investigation of slow fracture, they cannot usually be identified separately. This situation will not change until models that define the conditions governing these mechanisms are developed.

Three of the most significant determinants of fracture processes, regardless of speed or scale, are stress state, microstructure, and crack geometry. Each of these inherently complicated factors is discussed below.

Because stress is essentially three-dimensional, the various processes involved in fracture depend not only on the level of stress, but also on the relative sizes of elements of the stress tensor. Changing the orientation of a flaw relative to the applied stress can change the stress levels required to initiate fracture processes and the rates at which the processes occur. Because the stress state is so important, no mechanical model of a fracture process is complete until it deals with arbitrary stress states. The influence of the stress tensor is closely related to the local microstructure and the details of flaw geometry.

Virtually all features of microstructure may have some influence on fracture processes. Probably the most important feature is grain geometry (both size and shape). Other important features include pores, inclusions, voids, triple points, and the intergranular phase. Fundamental investigations of the relationships between fracture processes and the various microstructural features have rarely been attempted. How do cracks interact with these features under complex stress states? Such questions need to be answered.

The boundary of a crack is a closed surface in three-dimensional space. In general, one dimension of the volume enclosed by the surface is very short relative to the other two. In other words, the crack is in some sense thin. The short dimension may even vanish, in which case the crack is said to be closed. In any case, the crack boundary can be divided into two faces lying opposite each other and meeting at a common edge. At this edge, or perimeter, the faces approach each other in such a way that the crack is atomically sharp. In general, the midsurface of the crack (the surface midway between the faces) is not planar. Still, the faces often lie roughly parallel to some plane over their entire extent. Even so, the faces themselves are not smooth. For microcracks, they probably follow grain boundaries; for macrocracks, they are composed of irregularly shaped bumps and depressions on the scale of grain size. Macrocracks may also exhibit "waviness" on a larger scale.

One consequence of their irregular geometry is that opposing crack faces interact in a complicated way when they are in contact under applied stress. For a crack that is closed or nearly so, stresses may be transmitted through interfacial contact at many points. If the applied stress should open the crack partially, the number of contact points, and so the transmitted stress, decreases. For some applied stresses, the crack faces are not in contact at all. Clearly, the relationship between applied stress and the degree to which stress is transmitted across the crack depends strongly on the details of the crack face geometry. Simplistic assumptions regarding crack geometry must be abandoned if the role of crack face interaction in fracture is to be understood. It is likely that the failure of existing models to account for crack behavior under arbitrary stress states is closely linked to the simplicity of the crack geometries investigated. A primary goal of future research should be to investigate fracture of cracks with more realistic geometries under arbitrary stress states.

Consider a rectangular array with a list of fracture phenomena in the top row and a list of microstructural features in the leftmost column. Let each each row-column innersection in the array represent a relationship between the phenomenon at the top of the column and the feature at the left of the row. One would hope that for a given crack geometry, a given stress state, and a given material, only a small number of these relationships is significant. A primary goal of future research should be identification and modeling of the most important of these relationships. Successful models will indicate which relationships are important under various conditions of stress state and crack geometry.

4.2. Statistical Modeling

Even if every relationship in the array referred to above were completely understood, modeling of fracture would not be complete. Real materials contain many instances of each microstructural feature and many flaws. To complete the modeling process, the distributions of flaws and microstructural features must be represented by statistical models. Important statistics of these distributions include density, size, shape, and orientation.

The primary difficulty in establishing statistical models of real materials is not a lack of reasonable mathematical functions for describing the distribution of flaws and microstructural features. Rather, the problem is the lack of data needed to verify the validity of the functions and their parameters. But since statistical models are an essential feature of an adequate design strategy, the data needed for verification must be acquired. A primary goal of future research should be assessment of the technology available for acquiring this data, and establishing the amount of data required for verification.

4.3. Design Methodology

Once mechanical and statistical models have been proposed, they can be used immediately to formulate a design methodology. But a useful intermediate step would be to evaluate their performance in some realistic situation. This is often attempted through experimental investigation, followed by discussion of the ways in which the results are or are not compatible with the proposed models. An alternate approach that has rarely been considered in the literature is direct simulation of a crack process (represented by a mechanical model) occurring in a field of microstructural features (represented by a statistical model). One example of this is the microcrack initiation study reported by Hoagland (1980). Such simulations are necessarily computer-based. But, on average, they are probably faster and cheaper than experimental programs. Furthermore, they are potentially rich sources of understanding of fracture processes. One key to their success is the use of models whose parameters can eventually be verified experimentally. This approach is so powerful that it should be used in the evaluation of all newly proposed mechanical and statistical models. A primary goal of future research should be to establish a computer-based fracture simulation that can incorporate a wide variety of mechanical and statistical models and crack geometries, as well as arbitrary stress states.

Design methodology must be logically consistent with the models it uses. It follows that any significant improvement in modeling requires that design methodology be reformulated to assure that consistency. Given the many unanswered, fundamental questions in both mechanical modeling of fracture and statistical modeling of materials, any effort to improve existing design methodology without improvements in modeling is probably of little significance.

5. Conclusions

The greatest weakness in existing design strategies for brittle fracture is in the narrow range of conditions for which the strategies are adequate. The primary reason for this weakness is the use of simplistic mechanical models of fracture processes and unverified statistical models of materials. To improve the design methodology, the models must first be improved. Specifically recommended research goals are:

- to develop models of cracks with realistic geometry under arbitrary stress states;
- to identify and model the most important relationships between fracture processes and microstructural features;
- to assess the technology available for acquiring statistical data on microstructure and flaw populations, and to establish the amount of data required for verification of statistical models;
- to establish a computer-based fracture simulation that can incorporate a wide variety of mechanical and statistical models and crack geometries, as well as arbitrary stress states.

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