Ambient and Elevated Temperature Fatigue Crack Growth in Ceramics and Ceramic Composites

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ABSTRACT

This paper examines the mechanisms of ambient and elevated temperature fatigue crack growth in ceramic materials under **cyclic compression** and **cyclic tension** loading conditions. The paper is divided into two parts. In Part I, we examine the origin and salient features of stable Mode I fatigue crack growth from stress concentrations in a wide range of ceramic materials subject to ***cyclic compression***. Results of experiments and analyses for ambient and elevated temperature failure are discussed for polycrystalline Al_2O_3, Si_3N_4, MgO-PSZ, Y_2O_3-TZP, as well as SiC whisker-reinforced Al_2O_3 and Si_3N_4. In Part II, we discuss the mechanisms of ***constant and variable amplitude tensile fatigue*** crack growth in ceramics and ceramic composites. New experimental results of crack growth rates and detailed transmission electron microscopy observations of crack-tip damage in an Al_2O_3-SiC composite subject to tension fatigue at 1400°C are presented. Mechanisms of fatigue crack closure in both compression and tension fatigue are briefly outlined. The fatigue characteristics of ceramic materials are compared and contrasted with those of metals in an attempt to develop an overall mechanistic perspective on cyclic load damage.

KEYWORDS

Fatigue, cyclic compression, cyclic tension, ceramics, mechanisms, crack closure, temperature effects, constant amplitude, variable amplitude.

INTRODUCTION

With recent advances in processing technology, the new generation of ceramic materials offer the potential for achieving a desirable combination of high temperature mechanical properties hitherto unobtainable in conventional metallic materials. The fracture toughness values for many advanced ceramics and ceramic composites appear promising; however, very little understanding exists about their resistance to fracture under *cyclic* loads, at both ambient and elevated temperatures, which are typical of potential service conditions. The aim of the present work to investigate the crack growth characteristics of ceramics under both cyclic compression and cyclic tension loading.

In this paper, we discuss on the mechanics and micromechanisms of fatigue crack growth in a wide range of ceramics and ceramic composites. A brief review of our recent work on compression fatigue in brittle solids is presented. This is followed by new results on tensile fatigue in a ceramic composite in 1400°C air environment. Results on the fatigue characteristics of ceramic materials under constant and variable amplitude fatigue in cyclic compression and cyclic tension are compared and contrasted with those of ductile solids in an attempt to develop a mechanistic understanding of damage mechanisms.

PART I. COMPRESSION FATIGUE

Background

It is generally acknowledged, based on the significant body of information available on fatigue of ductile solids, that crack growth under cyclic loads occurs by the continual blunting and resharpening of the advancing crack-tip. On the microscopic level, this process is aided by the to and fro motion of dislocations, i.e, reverse slip. It is often tacitly assumed that the absence of dislocation motion or plasticity inevitably promotes catastrophic brittle failure (as found in most ceramics at room temperature), with no inherent propensity for subcritical crack growth under cyclic loading conditions. It has been suggested that any apparent fatigue crack growth behavior observed in ceramic materials may be a direct consequence of stress corrosion cracking (e.g., Evans and Fuller, 1975). In the few studies where some effect of cyclic loading was observed (e.g., Guiu, 1978; Evans, 1980), no specific mechanisms of failure were identified.

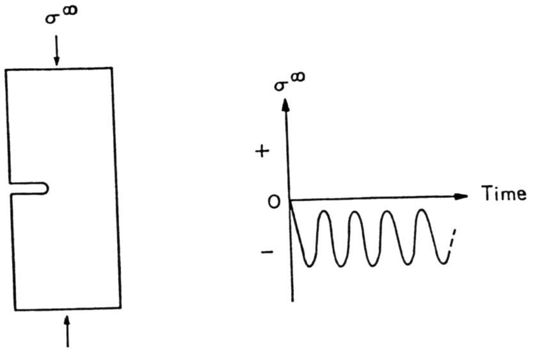

Fig. 1. Schematic of the compression fatigue test specimen set-up.

The first clear demonstration of an intrinsic mechanical fatigue effect at room temperature in a wide range of ceramics and ceramic composites was reported for crack growth in cyclic compression (Ewart and Suresh, 1986, 1987; Suresh and Brockenbrough, 1988; Suresh *et al.*, 1988). In this series of papers, it was established that the application of uniaxial cyclic compressive stresses, of maximum value substantially smaller than the unconstrained compressive strength, can lead to undesirable levels of stable Mode I fatigue crack growth. Figure 1 shows a schematic diagram of the compression fatigue specimen. It is observed that the fatigue cracks propagate, at a progressively decreasing velocity, along the plane of the notch in a direction macroscopically normal to the far-field compression axis, and then arrest completely. The rate of initial crack growth and the total distance of propagation prior to crack arrest are strongly influenced by the compression load amplitude, mean stress, specimen geometry, notch geometry, material microstructure and stress state. A significant feature of this crack growth phenomenon is that, despite vast differences in their microscopic deformation modes, the macroscopic crack growth behavior observed in ceramics and ceramic composites under far-field cyclic compression is similar to that documented for metallic materials (Hubbard, 1969; Reid *et al.*, 1979; Fleck *et al.*, 1985; Suresh, 1985).

Mechanisms of compression fatigue

Figures 2a-2d show examples of stable Mode I fatigue crack growth from a notch-tip under far-field uniaxial cyclic compression at room temperature in a polycrystalline Al_2O_3, a hot-pressed Si_3N_4, an Al_2O_3-33 vol.% SiC whisker composite, and an Y_2O_3-TZP transformation-toughened ceramic, respectively. The details of specimen geometry, loading conditions, and material microstructure pertaining to these examples can be found elsewhere (Ewart and Suresh, 1987; Suresh *et al.*, 1988). In nominally single phase systems, such as Al_2O_3, ***in situ*** examination of the notch-tip region during the very first compression cycle reveals predominantly intergranular failure (Figure 3). Grain boundary microcracks, which are nucleated during the loading portion of the compression cycle, can be clearly seen in this figure. The significant effect of grain boundaries in influencing notch-tip damage in brittle ceramics under cyclic compression can be further appreciated by noting that it has thus far not been feasible to induce stable fatigue failure in cyclic compression in α-alumina single crystals in any crystallographic orientation or in soda-lime glass (Ewart and Suresh, 1987).

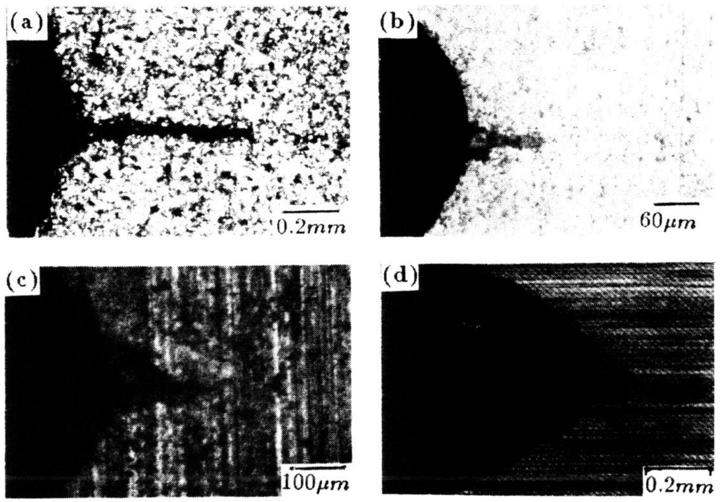

Fig. 2. Examples of Mode I fatigue crack growth under cyclic compression.
(a) Al_2O_3, (b) Si_3N_4, (c) Al_2O_3-SiC whisker composite, and (d) Y_2O_3-TZP.

Figure 4 is a schematic diagram of the constitutive behavior in cyclic compression, showing a nonlinear stress-strain response during compression loading and possible linear unloading paths with and without permanent strains. For single phase ceramics, the nonlinear behavior (line A in Figure 4) is induced by microcracking at the root of the notch. If all the microcracks which opened during compressive loading fully close upon complete removal of the far-field compressive stress, the unloading behavior would follow line B with no resultant permanent strains. On the other hand, if all the microcracks do not close because of frictional locking of asperities or the presence of debris particles, elastic unloading (similar to metals) would result. The corresponding unloading path, denoted by line C, would lead to the maximum amount of permanent strain. Most single phase ceramics exhibit an unloading behavior which falls within these two extreme, idealized cases, as modelled by line D. For the case of transformation-toughened ceramics (such as MgO-PSZ and Y_2O_3-TZP) at room temperature or for elevated temperature creep conditions of most ceramics, the nonlinear stress-strain response during compression loading (line A) is induced by martensitic transformation or power-law creep, respectively. The unloading path for either case may very closely follow line D. Ceramic composites, such as SiC whisker-reinforced Al_2O_3 or Si_3N_4 exhibit nonlinear behavior in compression as a consequence of microcracking and interfacial sliding. Here, the unloading path follows the trend exhibited by line D in Figure 4.

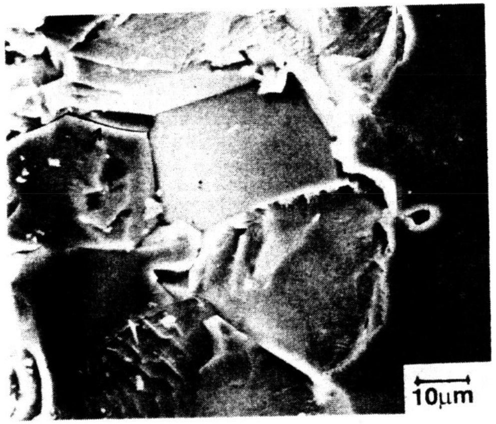

Fig. 3. G.B.microcracking at the notch-tip.

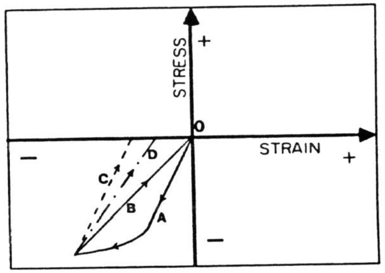

Fig. 4. Constitutive models for cyclic compression

Recently, detailed finite element simulations of residual stresses in notched plates under cyclic compression have been conducted for a wide range of constitutive behavior patterns and microscopic deformation modes, viz., microcracking, martensitic transformation, dislocation plasticity, and creep (Ewart and Suresh, 1987; Suresh and Brockenbrough, 1988; Suresh, 1988). In all these cases, it has been found that when permanent strains are retained at the notch-tip upon unloading from the far-field ***compressive*** stress (unloading paths C and D in Figure 4), large residual ***tensile*** stresses are generated at the notch-tip during the unloading portion of the fatigue cycle. As the maximum normal tensile stress easily exceeds the tensile strength of the solid in the vicinity of the stress concentration at the notch-tip, fracture occurs even after the very first compression cycle. Since the zone of residual tension is fully embedded in material elastically strained in compressed, a macroscopically similar, stable fatigue crack growth behavior ensues, irrespective of the microscopic deformation mode.

Development of crack closure

The fatigue cracks, initiated at the notch-tip under cyclic compression, propagate progressively slowly as a result of the monotonic increase in crack closure. Direct experimental measurements of crack closure in ductile solids in cyclic compression show that the fraction of the compression loading cycle during which the crack remains open decreases with an increase in crack length (Fleck *et al*, 1985; Suresh *et al.*, 1986; Pippan *et al.*, 1987). A similar trend is also observed for the case of ceramic materials. The effect of stress state on crack closure in cyclic compression, however, is found to be opposite to that in cyclic tension. While plane stress conditions promote a greater degree of (plasticity-induced) crack closure than plane strain in cyclic tension, the reverse is found to be true in cyclic compression. For ductile solids, the larger plastic zone size in plane stress leads to a larger residual tensile zone size at the notch-tip during cyclic compression, thereby promoting less closure (Suresh *et al.*, 1986).

In both brittle and ductile solids, application of compressive overloads during the very first cycle of compression fatigue leads to an increase in the size of the damage zone during compression loading and the residual tensile zone at the notch-tip upon unloading (Aswath *et al.*, 1988; Suresh, 1988). Consequently, crack closure stresses are significantly reduced and the total distance of growth prior to crack arrest is markedly increased. Similarly, large compressive loads can reinitiate a crack previously arrested due to crack closure in cyclic compression because of the flattening of fracture surface asperities and the crushing of debris particles (Aswath *et al.*, 1988). Table I, provided at the end of this paper, presents a summary of the mechanisms of fracture in ceramics under far-field cyclic compression.

Brittle solids, such as ceramics and ceramic composites, are up to ten times stronger in compression than in tension. This has led to the notion that loading of brittle materials in compression is much safer than tensile loading. Indeed, pre-stressing in compression is a common practice in a number of structural engineering applications. Our work demonstrates that undesirable levels of subcritical crack growth can occur in brittle solids, ***at compressive stress levels far below the compressive strength***, when stress concentrations are present and when cyclic compressive loads are encountered. Furthermore, a wide range of potential structural applications, for which the new generation of advanced ceramic materials are being developed, will involve varying compressive loads during service. The development of an effective fatigue design methodology for ceramics will inevitably require a detailed understanding of the mechanisms of crack growth under cyclic compressive loads.

PART II. TENSION FATIGUE

Background

The application of static or cyclic tensile stresses to very fine-grained, nominally single phase ceramics leads to catastrophic failure when the maximum applied stress intensity fractor reaches the fracture toughness. Recent studies, however, have shown that stable tensile fatigue crack growth can occur in ceramic materials which exhibit R-curve behavior under quasi-static loading conditions. In coarser grained alumina, which exhibits an R-curve in quasi-static tension, stable fatigue crack growth has been observed in cyclic tension and tension-compression loading (Reece *et al.*, 1988). The majority of studies on tensile fatigue at room temperature have focussed on documenting stable crack growth in different types of transformation-toughened MgO-PSZ (Swain and Zelizka, 1986; Bowman *et al.*, 1986; Dauskardt *et al.*, 1987) and in MgO-PSZ and Y_2O_3-TZP (Sylva and Suresh, 1987). The principal driving force for stable fatigue crack growth in these transforming ceramics is the creation of a transformation zone at the crack-tip due to stress-induced martensitic transformation, which leads to permanent volumetric and shear strains. Cyclic loading conditions also induce a hysteresis in the stress-strain response. Existing experimental evidence indicates that (i) at long crack lengths, crack growth under cyclic loading conditions are substantially faster than those under static loading (Swain *et al.*, 1986, Dauskardt *et al.*, 1987); (ii) fatigue crack growth rates are highly sensitive to the test environment, load ratio, and the extent of aging of the material (Dauskardt and Ritchie, 1988). For very small cracks, emanating from notches, crack propagation has been found to be highly discontinuous, with periodic crack arrest (Sylva and Suresh, 1987).

In this section, we present new experimental results on stable fatigue crack growth at 1400°C in an Al_2O_3-33 vol.% SiC whisker composite (commercially available as Grade WG-300, Greenleaf Corporation, Saegertown, PA). The process history and microstructural details can be found elsewhere (Han and Suresh, 1988). Single edge-notched specimens of this material (of width=10mm, length=50.8mm, and thickness=5mm) were fatigue pre-cracked in uniaxial cyclic compression to introduce a Mode I pre-crack at room temperature (similar to that in Figure 2c). Following pre-cracking, the specimens were subjected to cyclic tensile loads (in the four-point bend configuration) in 1400°C air environment at several different load ratios and cyclic frequencies. The crack length during the fatigue test was monitored through a view port in the furnace with the aid of a computer-controlled telescope; optical images of the specimen were monitored on a video screen (for details, see Han and Suresh, 1988). The behavior of the composite under cyclic loads is compared with the quasi-static crack growth response at 1400°C and with the failure characteristics of an unreinforced alumina matrix with a comparable grain size. Detailed transmission electron microscopy observations of the fatigue crack-tip region are discussed with a view to gaining an understanding of the mechanisms of cyclic load damage.

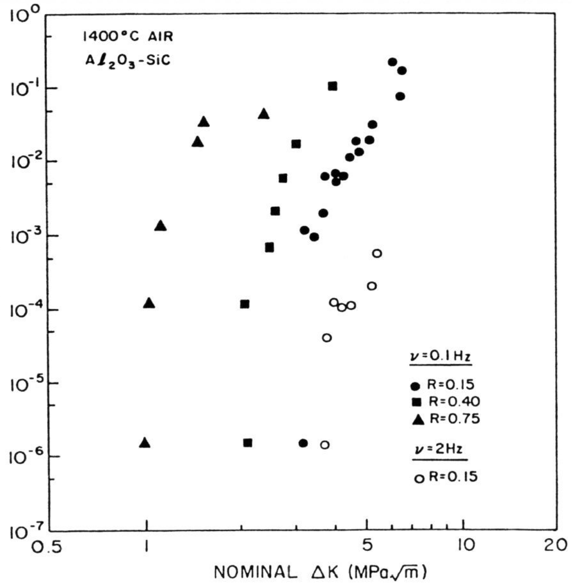

Fig. 5. Fatigue crack growth behavior of Al_2O_3-SiC whisker composite at 1400°C.

Tension fatigue results

Figure 5 shows the fatigue crack growth rate, da/dN, of the ceramic composite as a function of the nominal stress intensity factor range, ΔK, at load ratio values, R=0.15, 0.40 and 0.75 at frequencies (sinusoidal waveform) of ν=0.1 and 2 Hz. It is found that an increase in the load ratio leads to a reduction in the threshold stress intensity factor range and an increase in the overall fatigue crack growth rates. This behavior is similar to that observed in most metallic materials at both ambient and elevated temperatures. Furthermore, an increase in the cyclic frequency, from 0.1 Hz to 2 Hz, leads to an appreciable reduction in the rate of fatigue crack propagation over the entire range of growth rates. The variation of crack velocity, da/dt, with the applied stress intensity factor, K_I, is plotted in Figure 6 for the ceramic composite subjected to static load fracture at 1400°C (open square symbol). Superimposed in this plot are the crack velocities for the cyclic load tests ($da/dt = da/dN \times \nu$), taken from Figure 5, plotted as a function of the maximum stress intensity factor of the fatigue cycle, $K_{max} = K_I$. It is seen that the threshold stress intensity factor range values are up to about 33% higher and the near-threshold growth rates are up to three orders of magnitude slower under cyclic loads than the corresponding static crack growth case. (Here, the threshold is defined as the stress intensity value below which the crack velocity is small than 10^{-7} mm/sec.). Also shown in Figure 6, are the static and cyclic crack growth data for unreinforced alumina (at 1400°C) from the work of Evans and Fuller (1974) and Blumenthal and Evans (1975). It is interesting to note that the threshold stress intensity for high temperature crack growth in the Al_2O_3-SiC composite is up to three times greater than that in the unreinforced alumina with a comparable matrix grain size. Indeed, the fracture toughness of the unreinforced alumina at 1400°C is smaller than the threshold stress intensity factor for stable crack growth in the composite (shown in Figure 6). The fracture toughness of the composite at this temperature is 8.7-12.8 $MPa\sqrt{m}$ for loading rates of 0.235 to 2.35 kN/min.

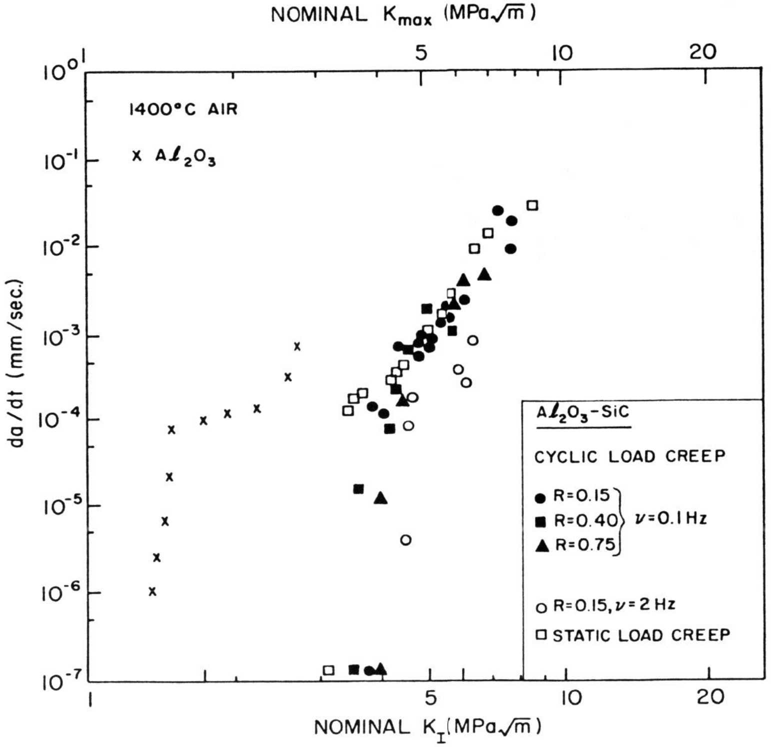

Fig. 6. Crack velocity vs. K_I plots from static and fatigue tests for the composite.

Figure 7a is an optical micrograph of a tensile fatigue crack in the composite at 1400°C at near-threshold stress intensity levels. Marked deviations of the crack path, manifested by deflections from the Mode I growth plane and crack bifurcation, are observed.

The principal mechanism of crack-tip damage in this material is cavitation and microcracking along grain boundaries and interfaces (Han and Suresh, 1988). An increase in the applied stress intensity factor range or the test temperature leads to a noticeable increase in the extent of crack-tip microcracking (Figure 7b).

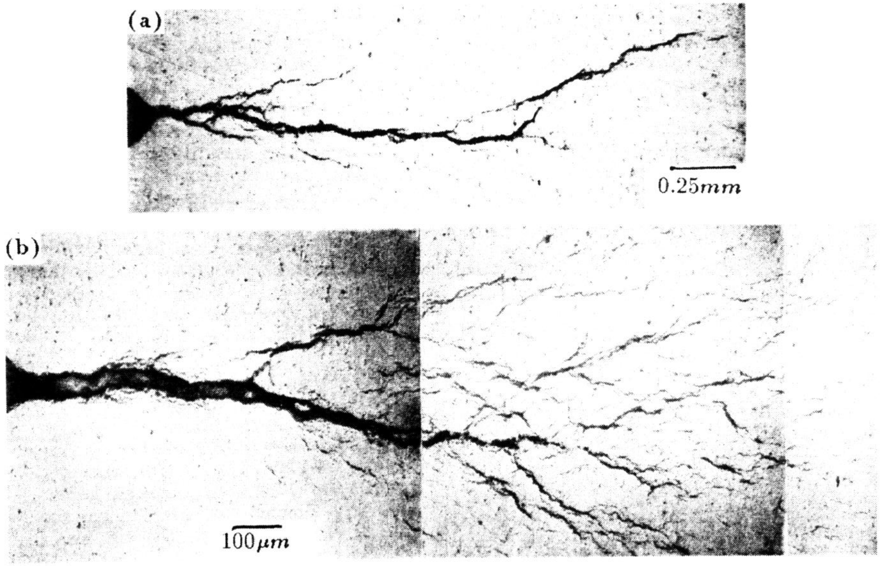

Fig. 7(a,b). Profiles of high temperature fatigue cracks in Al_2O_3-SiC composite.

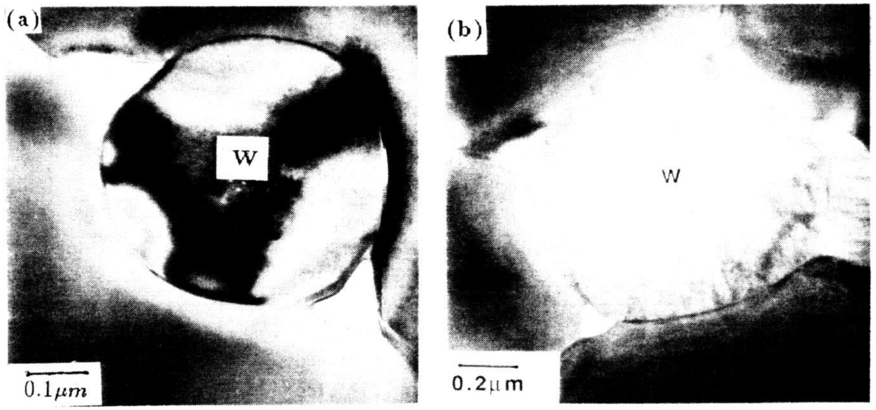

Fig. 8(a,b). TEM observations of interfacial cavitation at the crack-tip.

Figures 8a and 8b show transmission electron micrographs of foils taken from the crack tip in the ceramic composite which was subjected to a few tensile cycles at 1400°C at a low load ratio. These observations, in conjunction with the analyses of interfacial phases in this materials (Han and Suresh, 1988), reveal a significantly higher concentration of interfacial glassy phase at elevated temperature than at room temperature. It is well known that SiC oxidizes when exposed to air in certain temperature regimes; this oxidation leads to the formation of SiO_2 glassy phase as well as graphitic carbon. The transport of the oxidizing environment to the highly stressed region of the crack-tip promotes extensive cavitation. Since the glassy phase melts and undergoes viscous flow at the high temperature, extensive cavitation occurs at the interface between the matrix and the SiC whisker (denoted by "W" in Figures 8a and 8b). The growth and coalescence of these cavities during high temperature fatigue leads to the formation a diffuse microcrack zone ahead of the advancing crack-tip (Figure 7). Available results suggest that the shielding of the crack-tip by the diffuse microcrack zone leads to a reduction in the effective stress intensity for fatigue crack growth. Furthermore, the branching of the crack as well the closure between the faces of the main crack and of the branches enhance the beneficial effect of crack-tip shielding. (For a more quantitative description of these effects, see Han and Suresh, 1988). Consequently, the composite exhibits a far superior resistance to elevated temperature crack growth than the unreinforced matrix alumina, where significant cavitation (due to the oxidation of SiC) is not a relevant mechanism. Our results also show that the ceramic composite exhibits cyclic fatigue effects in that the growth rates are strongly influenced by the cyclic frequency and can be very different from the static load results, especially in the near-threshold region. Systematic studies of the effects of controlled environments, frequency, load ratio, and waveform on fatigue crack growth are underway in an attempt to separate out the mechanical and environmental contributions to high temperature fatigue.

In addition to the cavitation and microcracking damage mechanism, the ceramic composite exhibits a limited amount of dislocation plasticity, twinning, as well as subgrain formation within the alumina matrix at elevated temperature. Figure 9 is a dark field transmission electron micrograph showing the existence of dislocation within the alumina matrix and the formation of subgrain boundaries (pinned at the ends by SiC whiskers) within an alumina grain. However, the interfacial cavitation mechanism arising from the oxidation of SiC and the flow of glassy phase appears to the most dominant damage mechanism in the composite; matrix plasticity, albeit noticeable in a few grains, appears to be relatively less pronounced for the temperature and loading conditions investigated here.

Variable amplitude fatigue

As the alumina-SiC whisker composite exhibits stable fatigue crack growth over a wide range of ΔK, it is feasible to explore the effects of a tensile overload on transient crack propagation rates. Figure 10 is a schematic diagram of the effect of a 30-45% tensile overload applied to the specimen at baseline $\Delta K \approx 3.5$ MPa$\sqrt{m}$ during a low load ratio fatigue test. During the application of the overload there is a noticeable amount of crack extension (Figure 10b). However, for test temperatures in the range of $1400 - 1500°C$, delayed retardation of the fatigue crack is also observed (Figure 10c) because of (i) an increase in the size of the microcrack zone created by the overload and (ii) a consequent reduction in the near-tip driving force due to microcrack shielding. The number of microcracks per unit surface area is also found to increase by as much as a factor of two due to the application of the overload. Table I provides a summary of salient features and known mechanisms of tension fatigue crack growth in ceramics and ceramic composites under both constant amplitude and variable amplitude loading conditions.

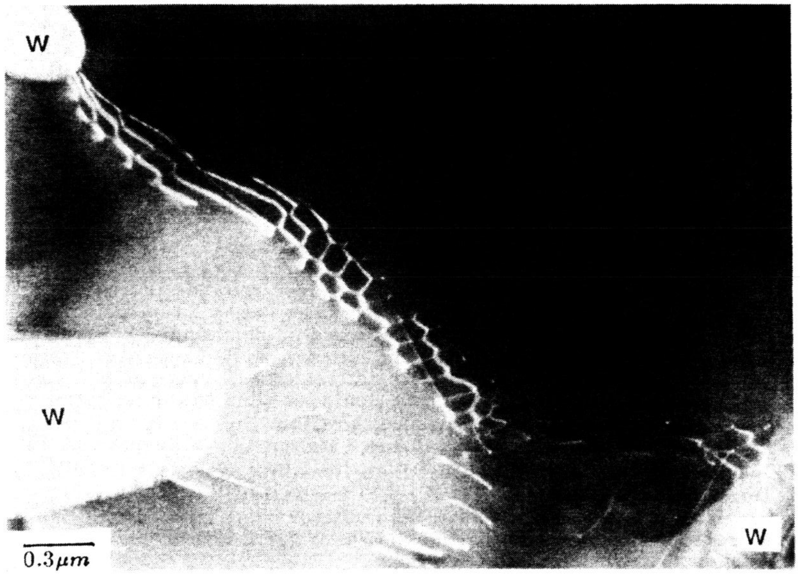

Fig. 9. Dark field TEM photograph of dislocations and subgrain formation

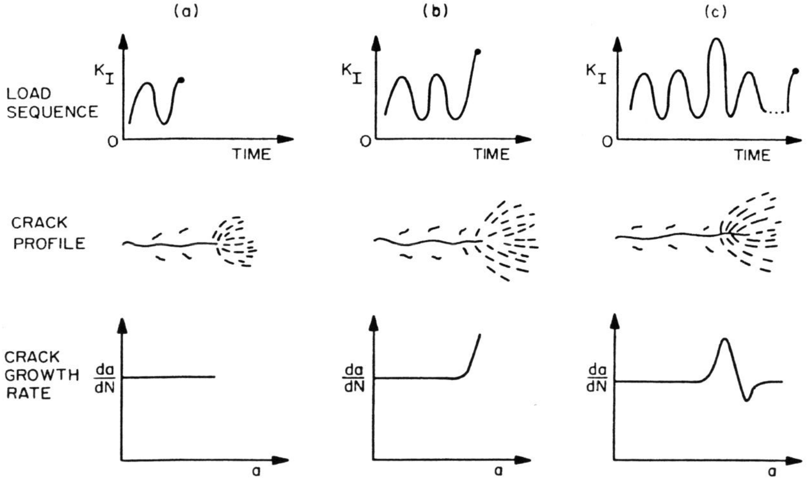

Fig. 10. Schematic of transient effects following overloads in Al_2O_3-SiC composite.

TABLE I. A SUMMARY OF FATIGUE EFFECTS IN CERAMIC MATERIALS

Materials	Factor/Variable	Mechanisms and Effects on Crack Growth Rates
Single phase ceramics	cyclic tension and tension-compression loading	A limited amount of subcritical crack growth in microstructures exhibiting an R-curve behavior; micromechanisms of any intrinsic mechanical fatigue effects mostly unknown. At high temp., lower failure life under cyclic loads than under static loads. A strong effect of specimen geometry and crack length on growth rates.
	cyclic compression loading of notched plates	stable Mode I crack growth due to residual *tensile* stresses induced at the notch-tip upon unloading from far-field *compressive* stress (when *permanent* deformation occurs in notch-tip damage zone.) Primary mechanism of damage is grain boundary microcracking.
Transforming ceramics (MgO-PSZ)	constant amplitude cyclic tension	Stable fatigue crack growth at ΔK values well below K_{Ic}. A reduction in threshold ΔK with an increase in load ratio, with overaging, and in the presence of a severe environment. Both volumetric and shear strains induced at the crack-tip due to transformation.
	constant amplitude cyclic compression	Mode I or mixed mode crack growth from notches depending on the constitutive response and extent of strain hardening in compression.
	tensile overloads	Temporary acceleration and delayed retardation following overloads, similar to metallic materials. Change in crack closure levels and in crack tip residual stress field due to overloads.
Single phase & multiphase ceramics	compressive overloads	Can markedly increase fatigue crack growth rates and crack growth distance ceramics in compression fatigue. Reduce extent of closure.

TABLE I. (continued...)

Materials	Factor/Variable	Mechanisms and Effects on Crack Growth Rates
Single phase & multiphase ceramics and ceramic composites	room temperature fatigue crack growth rates and crack closure levels	Influenced by: the "shielding" of the crack-tip due to martensitic transformation and/or micro-cracking; closure due to the bridging of crack faces by grains, debris particles, or reinforcements; closure due to transformed material in the *wake* of the advancing crack-tip.
Al_2O_3 & Al_2O_3-SiC composite	elevated temperature fatigue damage in constant amplitude cyclic tension	G.B. cavitation in Al_2O_3; interface cavitation and glass phase formation due to oxidation of SiC in the composite. A diffuse microcrack zone at the tip of the crack. Only limited extent of dislocation mobility, twinning, or subgrain formation at 1200-1500°C.
	elevated temperature fatigue damage in constant amplitude cyclic compression	An increase in temperature leads to an increase in both the notch-tip residual tensile stress and the size of the residual tensile zone. The total distance of crack growth is higher than at room temp. (at a fixed compression amplitude) due to creep and environmental interactions.
	elevated temperature crack growth rates in cyclic tension	Slower fatigue crack growth rates in the composite than in monolithic alumina. A reduction in fatigue threshold ..K with an increase in load ratio and a decrease in cyclic frequency.
	tensile overloads at elevated temperature	An increase in crack-tip micro-crack zone size. Crack acceleration due to the overload and a small amount of retardation due to micro-crack shielding.
	compressive overloads in notched plates at elevated temperature	A greater amount of acceleration in crack growth rates than at room temperature. Amplitude of the very first compression cycle markedly affects compression crack growth.

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