CRACK GROWTH BY CYCLIC FATIGUE IN YTTRIA-TETRAGONAL ZIRCONIA AND SILICON NITRIDE CERAMICS

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The cyclic fatigue for Y-TZP and silicon nitride is studied by means of indentation cracks. Crack growth rate as a function of the stress intensity factor plots are obtained. For both materials a "short-crack behavior" has been observed which is due to indentation residual stresses, since this behavior disappears when the residual stresses are considered in the equation to calculate the stress intensity factor. From fracture surface observations it is shown that crack wake contact has an important effect on cyclic fatigue.

INTRODUCTION

In the last years the existence of a cyclic fatigue effect in ceramic materials has been studied since materials subjected to fluctuating loads tent to fail earlier than what is expected merely by considering a static fatigue effect. The existence of cyclic fatigue in ceramic materials has been refuted in the past due to the lack of plasticity in these materials. It seems that the difference in the growth rate between static and cyclic fatigue is the difference in crack-wake contact between both modes of loading. The cyclic fatigue behavior of a broad range of materials has been reviewed by Suresh (1), including ceramics.

Since failure in ceramic materials is usually triggered by small flaws, it is important to study the fatigue and fracture behavior of these small cracks instead of hypothetical larger cracks which are not observed in ceramic structures. Steffen et al (2) have observed an inverse growth rate dependence on the applied stress which is known as a "short-crack behavior".

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One way of inducing small defects in ceramics is the indentation method, by the use of which a broad variety of crack lengths smaller than 1 mm can be generated and placed in the desired position on the specimen. This technique is used in the present work.

Indentation cracks have an important residual stress. This residual contribution to the crack growth rate causes the "short-crack behavior" observed in these cracks (3). In naturally occurring small cracks, this effect has been associated to restricted shielding in cracks with small wakes (2).

In the present study, indentation cracks in silicon nitride and Y-TZP will be subjected to fluctuating loads in order to characterize their growth rate in terms of the maximum applied stress intensity factor and (for the TZP) the corrected stress intensity factor considering the residual stress due to the indentation process. For this material the residual stresses have also been removed by annealing and the growth rate evaluated.

EXPERIMENTAL PROCEDURE

The zirconia tested is a 2.8 mole % Y-TZP with a grain size below 0.5 μm being nearly 100% tetragonal. It transforms in 29 % to the monoclinic symmetry in a static fracture test, as measured in the fracture surfaces. The material has a very little intergranular glassy phase. The silicon nitride is characterized by a microstructure where elongated grains (probably of β-Si_3N_4) are present.

The tests were performed in three and four point bending specimens of 5 x 25 x 3 mm thickness. The span was 20 mm in three point bending. In four point bending inner and outer span were 10 and 20 mm respectively.

The specimen surfaces were polished and one to four Vickers indents of 153 and 441 N were placed in several specimens. Care was taken in order to align the indents so that two opposite corner cracks were perpendicular to the long axis of the specimen. The indents were subjected to cyclic fatigue under tension-tension conditions with R = 0.1 and 10 Hz. All tests were performed at room temperature.

The residual stresses were removed in several Y-TZP specimens to determine the cyclic fatigue growth rate. These specimens were heated to 1100 °C for 45 min and cooled slowly, this treatment also reverts the tetragonal-monoclinic transformation hence removing crack shielding.

The total stress intensity factor in the presence of residual stresses is calculated by:

$$K_{max} = K_{max\ app} + \chi P / c^{3/2}.$$

The first term is calculated using the equations from (4) and is due to the nominal or applied stress. The second term corresponds to the residual stress contribution due to an indentation load "P" for a crack which has a length "c" measured from the center of the indent. $\chi = \delta\ (E/H)^{1/2}$ being "E" the modulus of elasticity, "H" the hardness and $\delta = 0.016$ for the TZP.

RESULTS AND DISCUSSION

In order to calculate the applied stress intensity factor it is necessary to know the aspect ratio of the semi-elliptical crack produced as a function of crack propagation. From measurements of the crack profiles in the fracture surfaces it was obtained that the aspect ratio, a/2c, was 0.4 constant throughout crack growth.

Fig 1 shows the crack growth rate as a function of the maximum applied stress intensity factor (Ka) for a material with indentation residual stresses developed by an indentation load of 153 N. In this figure the growth rate diminishes with Ka until a certain value above which it increases. This "short-crack behavior" is expected since initially, the residual stresses are very high and tent to decrease faster than the applied K increases for a given load as the crack growths.

In Fig 2 it has been taken into account the effect of the residual stresses using the equation previously proposed. It is clear from this figure that when the residual contribution is considered the "short-crack behavior" disappears. In Fig 3 the indentation load was 441 N; since the residual effect has been taken into account a short-crack behavior is not observed.

The growth rate as a function of the maximum applied stress intensity factor for the annealed TZP is plotted in Fig. 4. This figure is shifted to small values of K when compared to Figs 2 and 3 (about 0.5 and 1 MPa $m^{1/2}$ respectively) since the effect of crack shielding due to tetragonal-monoclinic transformation is absent for the initial cracks in the annealed condition, and increases with indentation load for the as indented specimens. Frictional contact between crack faces is enhanced for the case where residual stresses are absent since K_{min} is decreased; for this reason it is also expected higher growth rates in as indented (not annealed) specimens. Both effects (frictional contact and shielding) are also combined to increase the crack growth rate for small indentation loads which is observed when comparing Figs 2 and 3. It is important to state that the values of K are shifted between Figs. 2, 3 and 4 even though these figures have been plotted against the stress intensity factor and not the stress intensity factor range. This practice has brought different R data to a reduced scatter (3).

Fig 5 illustrates the different behavior observed in Y-TZP when the growth rate of static fatigue is compared to that of cyclic fatigue. It is clear that the cyclic fatigue effect is much more important than static fatigue, as crack propagation is observed for small stress intensity factors in the first case.

A particular problem observed in this material is that indentation cracks are much more radial than half-penny, no stress intensity factor was found for radial cracks. Despite this result, the fractographic study has shown that radial cracks tent to turn semi-elliptical at the earlier stages of propagation, then it can be considered for the as indented material that only the first propagation rates could be influenced by this effect since growth rates in the material with residual stresses are very high at the very beginning of growth (~ 10^{-7} m/cycle). This is not the case for the annealed material where the smallest growth rates are at the beginning of the crack growth propagation where the crack could still be radial.

An important result of this study refers to the difference in crack wake contact observed for cyclic and static fracture. In Fig 6 it is observed the fracture difference in Y-TZP for both modes of loading. The intergranular fracture observed corresponds to static fracture by overload, the crushed portion corresponds to cyclic fatigue, from this figure it is clear the effect of fracture surface asperities in this mode of fracture.

Silicon nitride has also be prone to cyclic fatigue. In Fig. 7 it has been plotted the crack growth rate in terms of the applied stress intensity factor for an indentation crack induced by a load of 153 N. As in the case of the TZP, miscount the indentation residual stress leads to a "short-crack behavior". From the fractographic results of the initial indentation cracks a constant relation $a/2c = 0.4$ was taken throughout propagation. In this material, the difference in fracture surfaces between cyclic and static fracture is the existence of small particles which emanate from the fracture surfaces. For the case of cyclic fatigue these small particles tent to coalesce (Fig 8) hence increasing the stress intensity factor at the crack tip by a fracture surface asperity effect.

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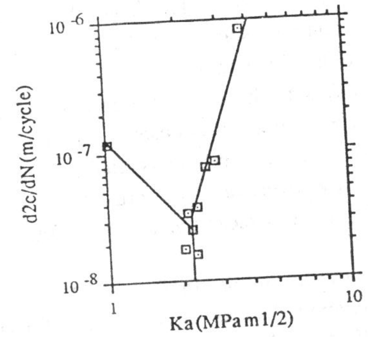

Fig 1. Crack growth rate as a function of the maximum nominal applied stress for an indentation load of 153 N.

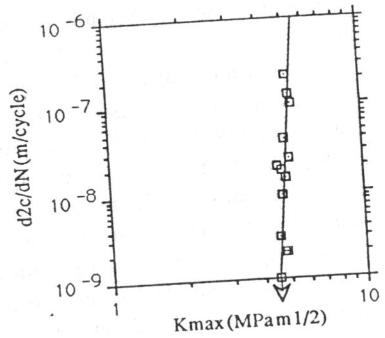

Fig 2. Crack growth rate as a function of the total maximum stress intensity factor for an indentation load of 153 N.

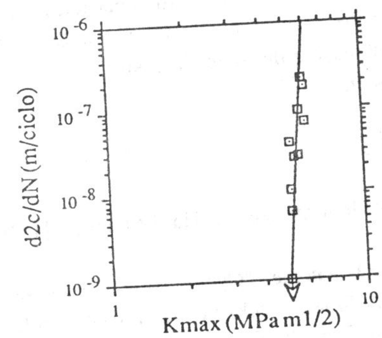

Fig 3. Crack growth rate as a function of the total maximum stress intensity factor for an indentation load of 441 N.

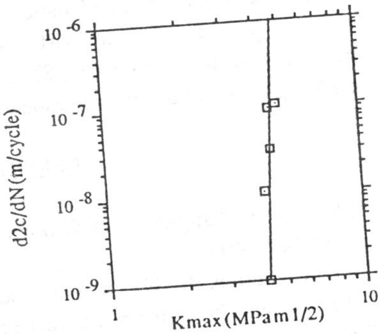

Fig 4. Crack growth rate as a function of the maximum stress intensity factor for an annealed specimen.

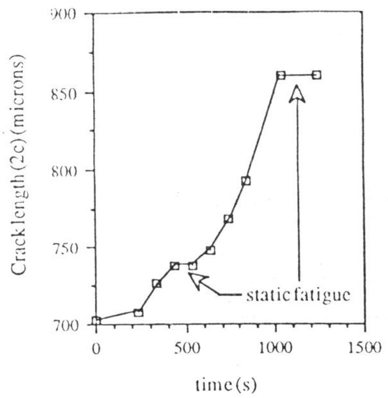

Fig 5. Crack lengths as a function of time for cyclic and static fatigue.

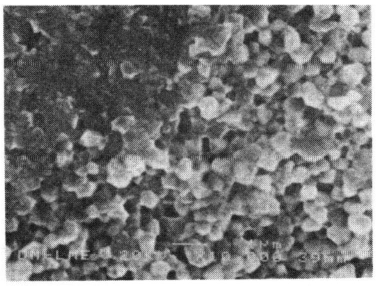

Fig 6. Fractographic differences between static (intergranular) and cyclic (crushed) fracture.

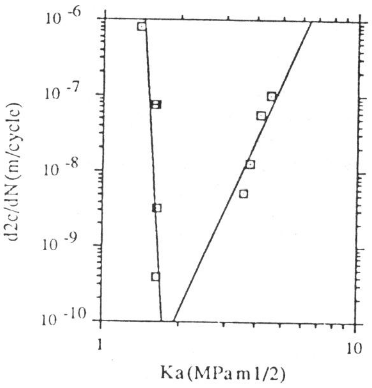

Fig 7. Crack growth rate as a function of the maximum nominal stress intensity factor for an indentation load of 153 N.

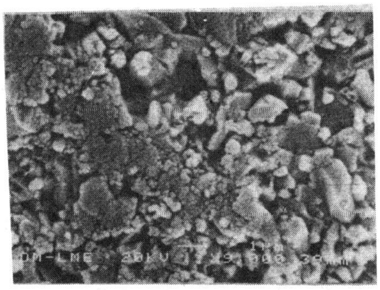

Fig 8. Cyclic fatigue fracture surface appearance.