FACTURE KINETICS OF CERAMIC MATERIALS IN CONTROLLED CRACK GROWTH TEST

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ABSTRACT

A method of fracture testing with controlled crack growth in ceramics is used for studying crack kinetics behavior under the given loading history. A computer-aided real-time data acquisition system enhances the informativity of a simple single specimen bend test in quantification of the crack growth parameters. Controlled fracture data are shown for alumina and for two grades of yttria-stabilized zirconia, all suggesting microstructure dependent fracture properties. Observed ambiguities in the crack growth diagrams for controlled testing are discussed to demonstrate the importance of the consideration of a crack growth history for the correct description of inequilibrium fracture behavior.

KEYWORDS
Fracture kinetics, R-curve, slow crack growth, alumina, zirconia

INTRODUCTION

Nonlinear fracture effects caused by the crack interaction with ceramic microstructure give rise to the obvious crack growth parameters dependence on the loading history. As a result in the lifetime prediction for ceramic structural elements the problems of undesired errors arise because of the uncertainty in the slow crack growth parameters. For the purpose of quantification and appropriate simulation of possible ambiguities in slow crack growth behavior the methodology of controlled fracture testing under predetermined loading history seems to be promising for better understanding of fracture kinetics.

The present paper addresses the question of controlled fracture testing with the emphasis on the crack kinetics monitoring, which is based on the improvement of displacement-controlled fracture testing in a stiff loading device. Some characteristic fracture kinetics data for alumina and tetragonal zirconia polycrystals are presented. These materials are expected to exhibit the microstructure dependent fracture properties, namely, the microcrack-induced rising crack growth resistance resulting from the thermal expansion anisotropy for the former, and the transformation toughening for the latter.
EXPERIMENTAL PROCEDURE

Standard single-edge notch bend (SSNB) test geometry has the monotonically rising K-calibration, which prevents control of the fracture process. The test geometries having descending part of the K-calibration curve (Müntz et al., 1983; Freiman, 1983; Pabst, 1975) was probably among the first who resulting in stable fractures of the specimen. Moreover, a stable crack growth under constant displacement rate is believed to give the receiving the K-drawings (Kleinlein and Haber, 1977; Pabst et al., 1983; Wilde et al., 1985); the following conclusions may be made. The use of the results obtained differentially on the machine rigidity, crack length, the indirect method of determination of the crack velocity using the only an average estimate of a real velocity, and also being hardly

Substantial improvement of the fracture data acquired during controlled fracture testing may be achieved by the use of data acquisition and processing (Benecke et al., 1984; Jung et al., 1986). The real-time monitoring of the crack growth. The application of the data acquisition and processing may enhance the informativity of the data and will help to avoid such traditional shortcomings of specimens, starting notch, or constant cross-head displacement rate is very approximate, thus, giving data on the rigidity of the testing machine.

The schematic showing the testing arrangement is given in Fig. 1 (Borovička, 1983). The test specimen is used, which bears the external load P, and is loaded to a simple three point SSNB geometry. In the test specimen, the crack length is measured by a high-speed displacement function D(t) resulting from the appropriate load control during the test. The load, deformation, and load points are monitored by the computer in real-time post-processing. The real-time load-displacement record represents the raw data of crack length; the cross-head displacement scheme for the test specimen in Fig. 1. It is possible to derive the compliance components of the test specimen in the PEL-unit as follows:

\[
K(t, A/W) = \frac{L(t)}{W^3} \left( C_{det} + C_{val} + C_{def} \right)
\]

where Y(A/W) is the calibration function of conventional three-point SSNB test (Scawley, 1976), L, B, and W are the specimen dimensions, A is the crack depth, Cdet is the testing machine compliance, Cval is the PEL-unit compliance, and Cdef is the total compliance in the specimen branch contained inside the PEL-unit. As shown in Fig. 2, calibration curve (1) has a convex shape with rising and descending parts (in contrast to the monotonically rising K-calibration). Eq. (1) is derived on the basis of linear elastic fracture mechanics, therefore, the limits of crack stability and instability are defined for a linearly elastic material. This follows from the consideration of crack length, the crack length for the stability of fracture may depend only on the compliance in the specimen branch Cdet. The specimen having starting notches close to the position of the maximum is the K-calibration should exhibit fully stable crack growth. Thus, the explicit introduction of the testing system rigidity in the K-calibration (1) allows to relate strictly the actual input and output of the fracture test, i.e., the time dependence of the stress intensity history on the cross-head displacement.

Calibration (1) may be directly employed in the stress intensity calculation for the predetermined U(t), but it is more accurate to use conventional calibration Y with respective measurements of Ps and Us. In linear elastic SSNB test the crack length \( A = \frac{W}{2} \) may be calculated from the compliance. Usually, in the compliance calculations of crack length several full or partial unloadings are used during the crack growth to define the magnitude of compliance change. By idealisation of the load-displacement diagram it is possible to simplify the calculation having only two reference crack lengths, the starting, and the final. The idealisation of the load-displacement diagram is based on the linearisation of the loading and unloading parts of record where no crack growth is assumed. The relevance of the idealisation and crack determination via compliance were estimated for each specimen by...
starts to move with low acceleration, which results in the gradual slope of respective extent of the V-K diagram. This extent appears to reflect the behavior observed in a conventional R-curve test. Crack retardation occurs under decreasing stress intensity with a more steep slope of the V-K diagram. If additional crack extension is induced by further increment of the cross-head displacement, the acceleration branch follows the former slope, while the retardation is shifted to the higher stress intensities with an equivalent increase in the crack growth exponent (see the data for AOT-21 specimen). The comparison of data in Fig. 3 shows that the crack growth exponent for the crack retardation is proportional to the preceding crack extension, and, therefore, is dependent on the loading history. The observed dependence is similar to the reported "memory" effect (Kahnans and Steinbruech, 1982), that was explained by microcracking related phenomena during the growth of a macrocrack.

Fig. 2. The influence of testing machine compliance on the crack stability. Comparison of the calculated and visually measured initial and final crack lengths. Since the crack velocity in SEMB test is not constant, it is not easy to measure actual crack length for comparison with the compliance specimen side or on its fracture surface. For all the materials studied of 0.01 m was always closer to linear calculation.

RESULTS AND DISCUSSION

Alumina and zirconia ceramics were investigated. The alumina ceramics of AOT-grade is a MgO-doped high-purity partially translucent ceramics of relatively coarse grain-size. To investigate possible transformation toughening effects on crack kinetics two grades of 2 mol.% of tetragonal zirconia polycrystals were tested, which may be related to the class microstructure typical of the materials containing the mixture of um is assumed to be tetragonal with imbedded coarse inclusions (5-10 um) contains additionally 5 mol.% of alumina. Mean grain size of zirconia inclusions as 2-5 um.

Alumina ceramics is expected to exhibit microstructure dependent fracture characteristics due to susceptibility to microcracking. In fact, strong ambiguities are obvious in the V-K diagram (Fig. 3). Initially the crack

Fig. 3. Crack velocity versus stress intensity diagram for AOT-alumina showing an ambiguity both in one-run test (11), and in two-run test (21) between successive runs.

Zirconia ceramics were tested under a single or several successive crack runs on each specimen. T2P2Y as well as T2P2Y5A showed similar fracture kinetics features (Figs 4 and 5), namely, the slow crack growth diagram reveals the plateau-like middle extent. Since such behavior is hardly dependent on the loading history, it is useful to consider the test in detail. The crack starts from the notch having a relatively high stress intensity and moves with acceleration to the maximal velocity in excess of 10^-3 m/s. It is not unusual for the transformation toughened ceramics and corresponds to the minor nonlinearity of the load line and almost flat R-curve with a very steep rising part (Burns and Swain, 1986). After
allowing the crack to retard (TZP2Y-20 or TZP2YSA-4) a considerable reduction in stress intensity follows, which corresponds to lower velocity changes, thus, the plateau like region appears in the diagram (n ~ 0.8). Finally, after a certain "threshold" value of stress intensity is achieved, the crack turns into the matrix region, forming left to right extent of the steep slope in the diagram but in the low stress intensity range. Thus, the obtained V-K diagram will be referred to herein as having high toughness steep, low toughness steep, and plateau extents. It is obvious that the diagram shows a well-defined similarity with the classic three-stage environmental slow crack growth diagram. However, all present experiments were conducted in the ambient laboratory environment, hence, the appearance of middle plateau extent, typical of the extremely low humidity is unlikely to result from the environmental effect. It is expected to be caused by microstructure related phenomena. Li and Pabst (1989) found a similar feature in the slow crack growth behavior for Mg-PSZ ceramics with a relatively low content of tetragonal phase. However, in double torsion tests they could probably record only the middle and right parts of the diagram shown in Fig. 4 or 5, because the D* measurements are carried out during the crack arrest stage. The abrupt change of the diagram slope was explained by the effective crack retardation, which takes place above the certain level of stress intensity. Another example of microstructure-induced effects reflected in the V-K diagram was observed by Troczinski and Nichols (1985) for a glass-metallic composite, where the horizontal extent in the low stress intensity region was attributed to the growth and interconnection of microcracks at the notch tip. In our experiments it is notable, that for the second and following crack runs (see TZP2Y-01, 02 in Fig. 4, or TZP2YSA-44 in Fig. 3), the starting stress and the initial value was substantially lower as compared to that of the first run and does not exceed the limits of the low toughness steep extent of the V-K diagram. This effect is believed to be similar to that observed by Li and Pabst, and may be explained as a consequence of stress induced phase transformation. During the initial loading of the notched specimen, the transformation zone develops, that induces the compression stresses giving rise to the starting value of K. After that, the following decrease in the stress intensity is not accompanied by the equivalent crack velocity reduction as is dictated by the high toughness steep extent of the diagram, and the crack rapidly overrides the compression zone. When the crack growth rate is repeated after full unloading, there is no more hindering influence of the starting compression zone, and the observed slow crack growth seems to occur without transformation toughening effects in the low region of the stress intensity variation. Transition from the low toughness steep extent of the V-K diagram to the plateau region (referred to a qualitative increase of crack retardation due to transformation; Li and Pabst, 1989) was never observed in any one of the subsequent runs, although the crack velocity did not exceed the upper boundary of the low toughness extent. However, the probability is retained to initiate the slow crack growth inside the low toughness region, after the specimen or crack growth is subjected to an overload. This in turn may be critical in any situation concerned with repeated loading, e.g., cyclic fatigue crack extension (Gratzschl and Li, 1991).
CONCLUSIONS

Controlled testing of alumina ceramics shows an ambiguous slow crack growth diagram in confirmation of the "memory" effect attributed to the irreversible microcracking around the crack tip. In the test with gradually in the crack acceleration exponents increases failure due to the transformation toughened zirconia of TZP2Y contradicts to those of microcracking in alumina. During crack resistance is measured, but during subsequent retardation the crack of the same crack show relatively lower toughness values. The observed reduction in stress intensity at the tip of the crack can lead to a decrease in the velocity, and characteristic plateau region appeared on the V-R diagram. The observed resistance to slow crack growth is indicative of possible weak may be susceptible to low stress intensity cracking after surviving an overload, despite the high initial crack growth resistance demonstrated fine-grained transformation toughened TZP-ceramics implies the implications of R-curve concept application in determination of fracture resistance.

REFERENCES