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Effect of Predamage on Crack Growth Behavior in a Particulate Composite Material

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The effects of predamage and strain rate on the crack growth behavior in a highly filled polymeric material were investigated. Centrally cracked predamaged sheet specimens were used to conduct crack propagation tests under two strain rates (0.05 and 5 min^{-1}) at room temperature. Experimental data were analyzed, and crack growth resistance curves as well as crack propagation curves, shown as the crack growth rate as a function of mode I stress intensity factor, were plotted. It is found that the crack growth behavior in predamaged specimen is highly dependent on the strain rate.

Nomenclature

a = half crack length, cm
$\dot{a}$ = crack growth rate, cm/s
C_1 = constant (see Eq. 1)
C_2 = constant (see Eq. 1)
K_I = mode I stress intensity factor, kPa $mm^{1/2}$

Introduction

IT is well known that defects exist in highly filled polymeric materials. These defects may be produced during the manufacturing process or in service. The defects or damage occur in the form of microcracks or microvoids in the binder or as dewetting between the binder and filler particles. Damage will not be confined to a specific location; rather it will diffuse into a relatively large area or zone. The damage evolution processes may take place by material tearing or by successive nucleation and coalescence of microvoids. The cumulative damage processes are time-dependent and are one of the main factors responsible for the time sensitivity of strength degradation as well as material fracture. Therefore, to evaluate both the integrity and reliability of polymeric material structures, it is necessary to determine damage effects on the material's constitutive and crack growth behavior.

There are a number of studies of crack growth behavior in highly filled polymeric materials available.[1–7] These studies used linear fracture mechanics theories to characterize crack growth behavior under mode I loading conditions. Experimental data were analyzed to obtain the mode I stress intensity factor K_I and the crack growth rate, $\dot{a}$. Statistical methods were used to determine the functional relationship between these two parameters. Results of the statistical analysis reveal that a power relationship exists between K_I and $\dot{a}$. The experimental finding supports the linear viscoelastic fracture mechanics theory developed by Knauss[8] and Schapery.[9] Although classical fracture mechanics principles are well established for single-phase materials, experimental data indicate that linear fracture mechanics theories have been applied to particulate composite materials with varying degrees of success.

The present study was undertaken to investigate the effects of prestrain, or predamage, on crack growth behavior in a highly filled polymeric material formed by embedding hard particles in the rubber. The specimens were 20.32 cm wide, 5.08 cm high, and 0.25 cm thick (Fig. 1). Prior to the crack propagation test, the specimen was stretched for 15% strain and then unloaded to 0% strain and removed from the test machine. While the specimen was under 0% load condition, a 2.54-cm crack was cut through the specimen's horizontal centerline with a razor blade (Fig. 1). Crack propagation tests were conducted with two different strain rates (0.05 and 5 min^{-1}) at room temperature. The recorded experimental data (crack length, load, and time) were used to calculate the crack growth rate $\dot{a}$ and the mode I stress intensity factor K_I as functions of time. Linear regressional analyses were conducted to determine the functional relationships between $\dot{a}$ and K_I. The results were analyzed to investigate the effect of strain rate on crack growth behavior in the material. In addition, the effect of predamage on crack growth behavior was also investigated, and the results are discussed.

Test and Data Reduction

In this study, prestrained sheet specimens were tested under constant strain rate at room temperature. A television camera was used to monitor crack growth, a tape recorder was used to record crack growth data for data reduction, and a strip chart recorder was used to record the load and time during the test. The raw data obtained from the tests were the half crack length a, the time t, and the load p corresponding to the measured crack length. These data were used to calculate the mode I stress intensity factor K_I and the crack growth rate $\dot{a}$. In calculating K_I for a given set of values of a and p, a nonlinear regression equation that relates the normalized stress intensity factor K_I/p to the half crack length a was used. The normalized stress intensity factors as a function of the two half crack lengths were calculated by the use of a three-dimensional elastic finite-element code. The normalized stress intensity factors are independent of Young's modulus, and therefore they can be used for elastic materials as well as viscoelastic materials. In calculating $\dot{a}$, the secant method was used. In the secant method, the crack growth rate was computed by calculating the slope of a straight line connecting two adjacent a-versus-t data points. The calculated average crack growth rate was assigned to a point midway between each pair of data points. It should be pointed out that, when analyzing the experimental data, the crack growth rates and the stress intensity factors at the left-side and right-side tips of the crack were calculated and analyzed separately. A detailed description of the data reduction method is shown in Ref. 7.

Results and Discussion

The material considered in this study is a lightly crosslinked polymer, highly filled with coarse solid particles. When subjected to external loads, it behaves like a viscoelastic material, and high nonhomogeneous local stress and strength fields can be produced in the material. Because of the high rigidity of the particles relative to the binder material, the magnitude of the local stress is significantly higher than that of the applied stress, especially when the particles are close to each other. Since the local stress and strength vary in a random fashion, the location and the degree of damage will also vary randomly in the material. The damage may be in the form of microcracks or microvoids in the binder or in the form of binder–particle separation, known as dewetting. When localized damage develops, the local stress will be redistributed. With time, additional

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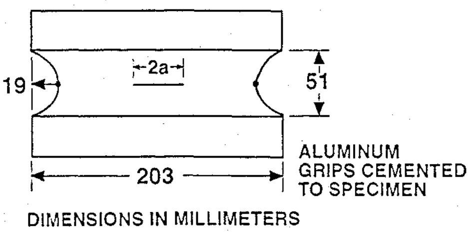


Fig. 1 Specimen geometry.

damage not confined to a specific location will develop and diffuse into a relatively large area or zone. Experimental data reveal that the damage state is closely related to the mechanical response and failure of the material. It is found that when a highly filled polymeric material is tested under constant strain rate, the initial linear portion of the stress–strain curve is associated with a stretching of undamaged material, with the filler particles bonded to the binder. As the external load is continuously increased, at a certain critical stress level, damage occurs. When the damage intensity exceeds a threshold value, the rigidity of the material is reduced. Usually this threshold damage state coincides with the transition from linear to nonlinear behavior. As the specimen is continuously stretched, the number of microvoids increases, and the formed microvoids start to grow and coalesce. This damage process is primarily related to the material's nonlinear response, and can be characterized by a bulk volume change during stretching.

Based on the preceding discussion on damage mechanisms in highly filled polymeric materials, it is expected that the material response and crack growth behavior will be closely related to a material's damage state. To obtain a fundamental understanding of the effect of predamage on the crack growth behavior, the damage characteristics near the crack tip need to be determined.

In the past, the Lockheed Research Laboratory's acoustic imaging system was used to measure the attenuation difference in a scanning mode. The recorded acoustic data were processed to create a visual indication of the energy absorbed in the material being inspected. A region of high ultrasonic absorption, i.e., a highly damaged area, will be shown as a dark area, and a region of unattenuated ultrasonic wave will produce a light or white area with 254 shades of gray in between. The acoustic image at a given strain level was plotted in the form of isointensity contours of the transmitted acoustic energy I_t to enhance the resolution of the damaged field.

The damage fields in a similar highly filled polymeric material were determined using acoustic imaging techniques.[10] Although the material investigated in Ref. 10 is different from the material in this study, the basic damage modes and mechanisms are similar. Therefore, the test results shown in Ref. 10 can be used to explain the effect of predamage in the damage fields near the crack tip in the material currently being investigated.

The acoustic images at 3% strain of a prestrained specimen, which was 7.62 cm wide, 6.35 cm high, and 0.508 cm thick, without crack and at 6% strain of virgin and prestrained specimens with crack are shown in Fig. 2. In Fig. 2, the number between two contour lines is the range of I_t between the prior and the next intensity levels. The small number indicates that the intensity of I_t is low or the damage intensity is high. Figure 2a shows the isointensity contour of I_t at 3% strain level after the specimen had been subjected to 9% strain and before a 2.54-cm crack was cut. This figure indicates that extensive damage was developed by prior strain and the damage intensity was uniform in the specimen.

The uniformly distributed damage intensity is an indication that the damage process is dominated by nucleation. In other words, the number of microvoids in the specimen increases with increasing strain level. This kind of damage nucleation and evolution process is a typical process that has been observed in prestrained specimens. When a crack is cut in the specimen, its presence will redistribute the stress in the specimen, especially near the crack tip region. The redistribution of stress will result in a modification of damage distribution. The higher stresses near the crack tip will induce higher damage intensity ahead of the crack tip as shown in Fig. 2b. Figure 2b shows the contours of I_t at 6% strain of a precracked virgin specimen. For comparison, the contours of I_t at 6% strain for the prestrained

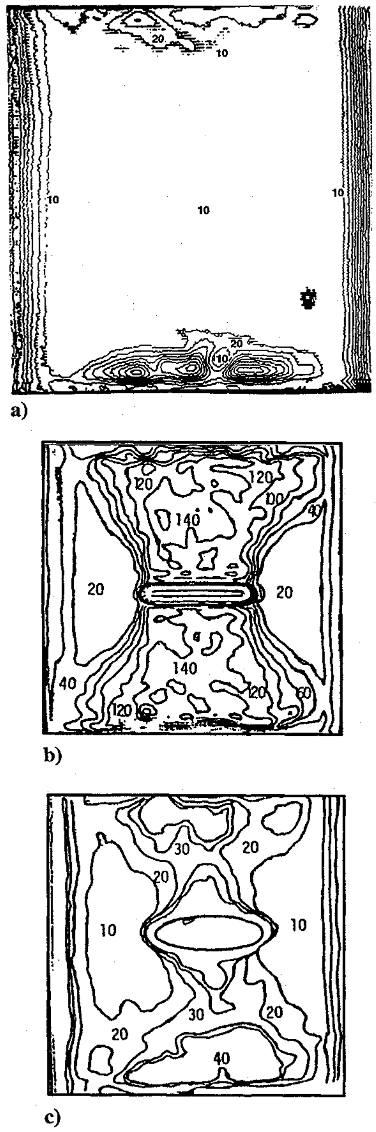


Fig. 2 Isointensity contour plot of acoustic image: a) 3% strain, before the crack was cut, b) 6% strain, after a 2.51-cm crack was cut, and c) 6% strain, virgin specimen, crack length = 2.54 cm.

specimen, Fig. 2a, after a 2.54-cm crack was cut is shown in Fig. 2c. Comparing Fig. 2b with Fig. 2c, it can be seen that predamage induces a larger damage zone with higher damage intensity ahead of the crack tip. The difference in the damage characteristics in the virgin and the predamaged specimens may induce different material response and crack growth behavior in the two specimens.

The preceding discussion is centered on the effect of predamage on the damage characteristics near the crack tip. In the following paragraphs, the effects of predamage and strain rate on material response and crack growth behavior are discussed.

Plots of applied load versus global displacement curves for virgin and predamaged cracked specimens tested at 0.05- and 5-min^{-1} strain rates are shown in Figs. 3 and 4, respectively. These two figures reveal that the effect of predamage on the load displacement curve depends on the strain rate. When strain rate is equal to 0.05 min^{-1}, predamage induces a significant reduction in Young's modulus and maximum load, whereas the displacement corresponding to the maximum load is relatively unaffected by predamage. Figure 3 also reveals that predamage increases the linear portion of the load displacement curve. However, when strain rate is equal to 5 min^{-1}, predamage has a relatively small effect on Young's modulus, and the load–displacement curve has a tendency to rejoin the virgin curve when the displacement is greater than 7.62 mm, or when the strain is greater than 15%, as shown in Fig. 4. A careful examination of the predamaged load–displacement curve in Fig. 4 reveals that the

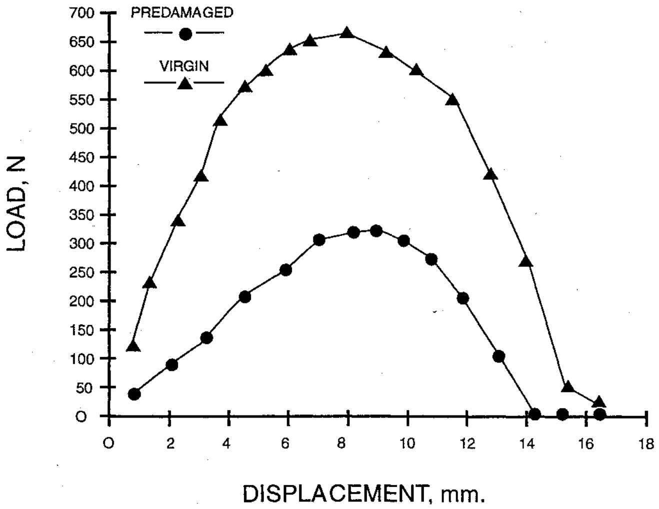


Fig. 3 Load–displacement curve of cracked specimen (strain rate = 0.05 min^{-1}).

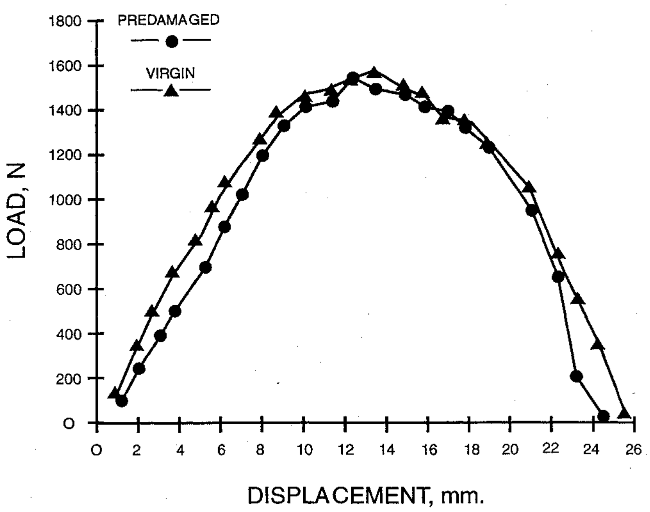
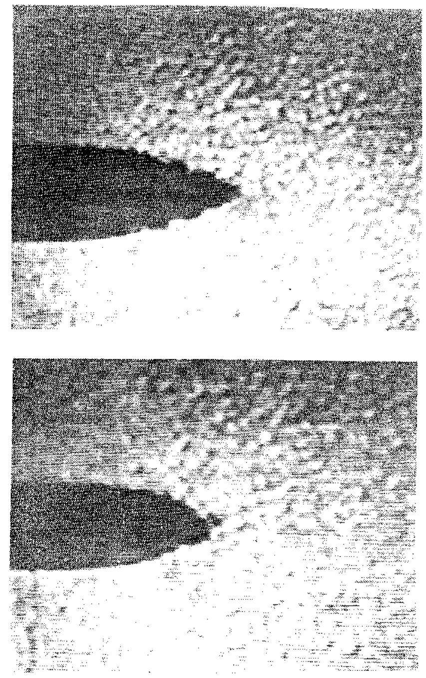


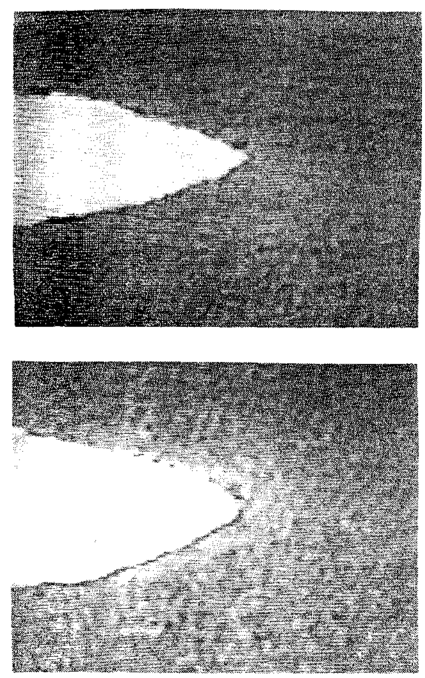
Fig. 4 Load–displacement curve of cracked specimen (strain rate = 5 min^{-1}).

curve shows a sigmoid shape characteristic of rubberlike materials (an upward sweep of the curve). The mechanism associated with the upward sweep of the load–displacement curve is closely related to the damage in the material. Owing to the nonhomogeneous nature of the material, the generated microdefects will be distributed more or less randomly in it. In addition, the microdefects are also initially randomly oriented with respect to the applied load. When this damaged material is stretched, the orientation of the microdefects is changed and the circuitous load path tends to straighten, resulting in a higher resistance to further deformation.[11] Consequently, a higher load is required to stretch the specimen further. It is interesting to note that the load–displacement curve for the low strain rate (0.05 min^{-1}) does not show the upward sweep. This difference is probably due to a combined effect of different degree of stiffness reduction associated with different crack length during crack growth and the viscoelastic nature of the material. Also, in Fig. 4, at 5-min^{-1} strain rate, there is no significant difference between the load–displacement curves of the virgin and the predamaged specimens. However, at 0.05 min^{-1}, predamage induces a significant effect on the load–displacement curve, as shown in Fig. 3. Since the crack growth behavior is controlled by the stress intensity factor, which is closely related to the applied load, it is expected that the strain rate will have a similar effect on the crack growth behavior in the virgin and the predamaged specimens. A detailed discussion on the effect of predamage and strain rate on crack growth behavior is contained in the following paragraphs.

Figures 5a and 5b are a typical set of photographs showing the crack surface profile during crack growth in the highly filled polymeric material. For the two strain rates considered in this study, crack tip blunting takes place both before and after crack growth. Owing to the heterogeneous nature of the material, the degree of blunting varies with the position of the advancing crack tip. This suggests that the local microstructure and damage state near the crack tip plays a significant role in the blunting phenomenon. Experimental data also revealed that, during crack growth, the crack tip was blunted and voids were developed ahead of the crack tip. The crack advanced by coalescing with these voids. When this occurred, the crack tip resharpened temporarily. Thus, the basic crack growth mechanisms consisted of a stop–growth–stop process, and the corresponding crack tip geometry consisted of blunt–sharp–blunt. This crack growth process is a highly nonlinear process, and it occurs in both the virgin and the predamaged specimens.



a)



b)

Fig. 5 Crack tip profiles: a) strain rate = 0.05 min^{-1} and b) strain rate = 5 min^{-1}.

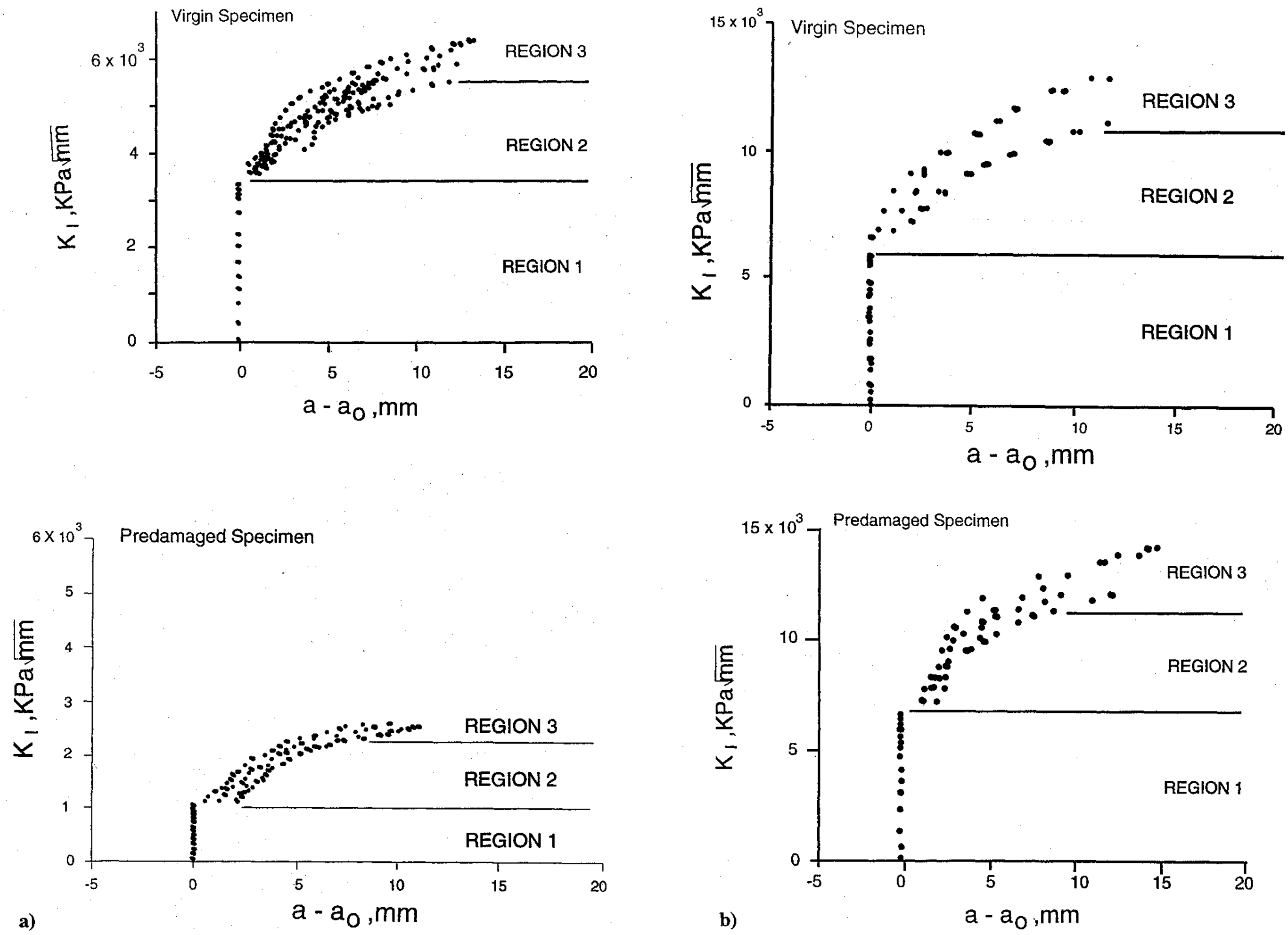


Fig. 6 Crack growth resistance curve: a) strain rate = 0.05 min^{-1} and b) strain rate = 5 min^{-1}.

The crack growth behavior in the highly filled polymeric material is investigated with the crack-growth-resistance (R-curve) approach. The concept of crack growth resistance was proposed for rate-insensitive materials by Krafft et al.[12] According to the R-curve concept, the increase of the crack extension force (the energy release rate G) is balanced by the increase of the crack growth resistance R, so that an equilibrium condition between G and R is maintained. The condition for the onset of unstable crack growth is reached when G is greater than R. It has been suggested by Krafft that, for a given thickness, the resistance to crack growth is primarily a function of the magnitude of the crack extension, and the R curve is a material property that is independent of initial crack length, specimen geometry, and loading condition. Although the R-curve concept was developed for rate-insensitive materials, preliminary experimental data indicate that it may be used to characterize the crack growth behavior in rate-sensitive materials such as highly filled polymeric materials and composite solid propellants.

The results of the data analyses were plotted as the mode I stress intensity factor K_{I} versus the crack extension $\Delta a = a - a_0$ where a_0 is the initial crack length (see Figs. 6a and 6b). For comparison, the R curves for virgin specimen data are also shown in Figs. 6a and 6b. There are three regions shown in Figs. 6a and 6b. The first region is defined as the crack tip blunting stage. Experimental data showed that, in the first region, the crack tip radius continually increased with increasing applied load. When the applied load reached a critical value for crack growth, the crack started to propagate, which defined the onset of the second region of the R curve. When determining the time corresponding to the onset of crack growth and the associated stress intensity factor, one must exercise caution because these factors are highly dependent on the microstructure and the damage state at the crack tip, especially when the specimen was damaged by prestraining. In the second region, or in the stable crack growth region, the basic crack growth mechanism was stop–growth–stop, and a considerable amount of stable crack growth took place. As the stable crack grew continually, the slope of the R curve decreased, which implied that the energy per unit extension required to continue crack propagation decreased. At a certain value of crack length, the transition from region 2 to region 3 was completed. A careful examination of the experimental data revealed that the transition region was located near the maximum load. This implied that region 3 was characterized by a continual decrease in the applied load with a continual increase in the crack length. Therefore, the onset of region 3 could be considered the onset of unstable crack growth. During the unstable crack growth stage, experimental data showed that the stop–growth–stop crack growth behavior also occurred in this stage. However, the time period in which the crack stopped became shorter as the crack length grew longer.

The effects of strain rate and predamage on the R curve are clearly indicated in Figs. 6a and 6b. For the predamaged specimens, the values of the critical K_{I} for the onset of crack growth are 1122 and 7006 kPa mm$^{1/2}$ for strain rates of 0.05 and 5 min^{-1}, respectively. Note that the values of K_{I} that correspond to unstable crack growth are 2528 and 11,955 kPa mm$^{1/2}$ for strain rates of 0.05 and 5 min^{-1}, respectively. These indicate that the values of K_{I} that correspond to the onset of crack growth and unstable crack growth increase with increasing strain rate.

Examining the R curves of the virgin and the predamaged specimens, we noticed that predamage has a significant effect on the R curve for the 0.05-min^{-1} strain rate, whereas it has a relatively small effect on the R curve for the 5-min^{-1} strain rate. In general, predamage induces not only lower values of K_{I} for onset and unstable crack growth, but also small slope in region 2 of the R curve. This implies that, for a given value of K_{I}, the increment of crack extension in predamaged specimen is larger than that in a virgin specimen. In

Table 1 **Summary of regression analysis**

Specimen	Strain rate, $\min^{-1}$	$\log C_1$	$\log C_2$
Virgin	0.05	−23.17	5.89
	5.00	−10.21	2.67
Predamaged	0.05	−14.65	4.07
	5.00	−13.32	3.54

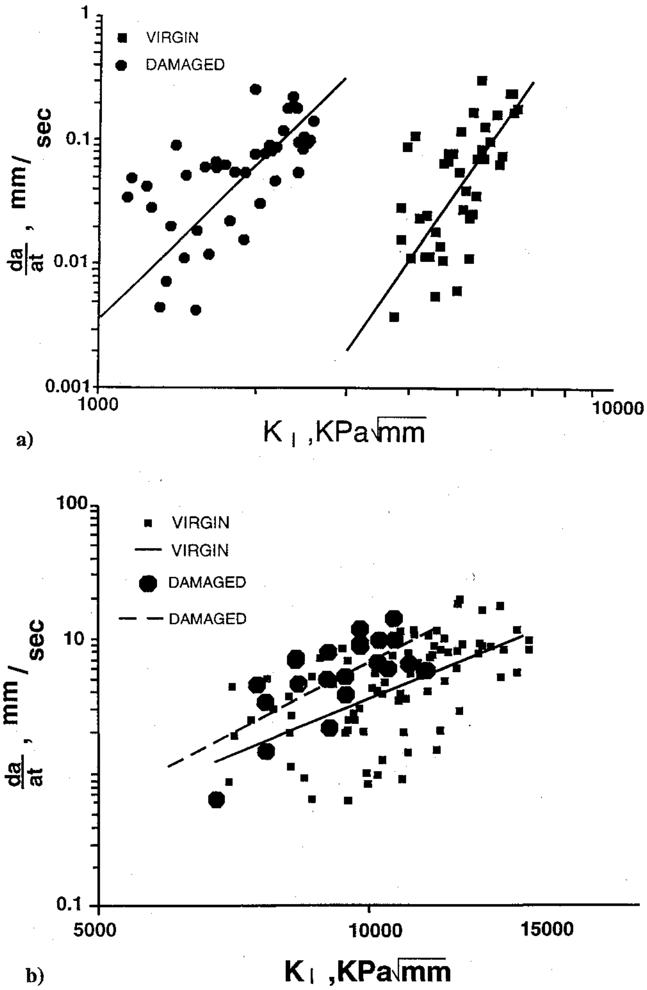


Fig. 7 Crack growth rate vs mode I stress intensity factor: a) strain rate = 0.05 $\min^{-1}$ and b) strain rate = 5 $\min^{-1}$.

other words, the resistance to crack growth in a predamaged specimen is lower than that in a virgin specimen.

Plots of $\dot{a}$ versus K_I for the virgin and the predamaged specimens tested at strain rates of 0.05 and 5 $\min^{-1}$ are shown in Figs. 7a and 7b. Based on these figures, in general, the crack growth data exhibit a rather large scatter, especially in the early stage of crack growth. The scatter seems to decrease as the stress intensity factor increases. On comparing the virgin specimen data with the predamaged data in Fig. 7, they show that the effect of predamage on crack growth depends on the strain rate. When strain rate is equal to 0.05 $\min^{-1}$, predamage has a significant effect on crack growth rate, whereas when the strain rate is equal to 5 $\min^{-1}$, the effect is relatively small.

The crack growth data, $\dot{a}$ and K_I, were analyzed using a linear regression analysis method, and the functional relationships between $\dot{a}$ and K_I were determined. In the analysis, the data in the early stages of crack growth were excluded. The results show that a power-law relationship exists between $\dot{a}$ and K_I, which is consistent with the theoretical results obtained by Knauss and Schapery. Mathematically, it can be written as

$$\dot{a} = C_1 K_I^{C_2} \tag{1}$$

in which C_1 and C_2 are constants. The values of C_1 and C_2 are shown in Table 1, and the regression lines are shown in Figs. 7a and 7b.

Conclusions

The principal conclusions that may be derived from the results of this study are as follows: 1) the crack growth mechanism consists of a stop–growth–stop phenomenon, which appear to be highly nonlinear; 2) a considerable amount of stable crack growth occurs before the onset of the instability; 3) the difference between the values of K_I corresponding to the onset of stable and unstable crack growth increases with increasing strain rate, 4) a power-law relationship exists between the crack growth rate and the mode I stress intensity factor; 5) at a strain rate of 0.05 $\min^{-1}$, predamage induces a significant reduction in the values of K_I corresponding to the onset of stable and unstable crack growth, whereas at a strain rate of 5 $\min^{-1}$ it has a small effect on these two K_I values, and (6) the effect of predamage on the crack growth behavior in the material is highly dependent on the strain rate.

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