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Fatigue Crack Retardation in Polycarbonate

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Synopsis

Results are presented from a recent study of the influence of tensile overloads on fatigue crack growth in polycarbonate. Fatigue cracks were grown under conditions of constant range in stress intensity factor in four-point bend specimens. The data presented here indicate that tensile overloads may significantly retard subsequent fatigue crack growth in polycarbonate. The period of delay in crack growth was shown to increase with the magnitude of the overload. Recovery of stable crack extension following the overload appeared to involve reinitiation of separate crack growth sites at the tip of the blunted crack tip, similar to the original crack initiation at sharp V-notches.

INTRODUCTION

Development of linear elastic fracture mechanics methods over the past two decades have provided the means to predict the influence of preexistent cracks on structural components. With the stress intensity factor, the linear elastic parameter relating crack length, remote load, and flaw geometry, the engineer can determine the critical crack size for a given structure and loading as well as predict the service life required for small cracks to grow to catastrophic dimensions.[1] This ability to predict crack growth lives by fracture mechanics methods rests largely on the pioneering work by Paris, Gomez, and Anderson[2] who demonstrated that the stress intensity factor is the controlling parameter for fatigue crack growth.

Although subsequent research by many investigators has confirmed that the stress intensity factor governs fatigue crack extension in a variety of materials, the influence of other external variables must also be considered. In a comprehensive review of polymer fatigue, for example, Manson and Hertzberg[3] summarize the literature which demonstrates that fracture mechanics techniques may be used to characterize fatigue crack growth in many polymers if one accounts for the effects of mean stress, temperature, cyclic frequency, environment, and specimen geometry (thickness). Data showing the dependence of fatigue crack growth rate on these external variables are reviewed along with suggestions for future research. Although load history is also known to have considerable importance on fatigue crack growth in metals (commonly referred to as fatigue crack retardation), this variable has apparently received little attention in polymer fatigue. Since the purpose of this paper is to demonstrate that prior loads can significantly influence subsequent fatigue crack growth in polycarbonate, the experience with metals will be briefly reviewed prior to discussing the present results.

FATIGUE CRACK RETARDATION

Fracture mechanics predictions of fatigue crack growth may lose accuracy if interactions between prior loads are not considered. It is well documented, for example, that peak tensile loads often retard subsequent fatigue crack extension in many structural metals.[4] These delay effects may be reduced, however, by subsequent compressive loading.[5] Since many structures see a wide variety of service loads, considerable research has been directed toward developing models to account for cyclic load history effects in order to obtain meaningful crack growth predictions.

Explanations for fatigue crack retardation have taken several approaches, including at least three possible mechanisms: fatigue crack closure, crack tip blunting, and residual stress fields. Although these models differ in form, it is generally agreed that the size of the crack tip plastic zone is a controlling factor for many load interaction effects.

The Elber closure concept,[6] for example, explains fatigue crack retardation in terms of residual compressive stresses left in the plastically deformed wake of the propagating crack. These compressive stresses act to physically hold the crack faces together, requiring a significant tensile load to separate the crack surfaces before the crack tip may extend. Elber suggests that overloads increase the residual compressive stresses, requiring a greater portion of subsequent tensile load cycles to overcome the residual forces. Since the effective portion of the load cycle for crack propagation is reduced, the range in stress intensity factor (ΔK) is lowered, resulting in slower crack growth rates.

Measurements of the change in crack opening load have been correlated with tensile overloads and subsequent delays in crack growth in aluminum specimens[6,7] as suggested by the closure model. In addition, interferometric measurements of crack surface displacements in poly(methyl methacrylate) (PMMA) specimens[8] confirmed the presence of fatigue crack closure in this polymer. Since the retardation effect was small for PMMA, however, it was not possible to correlate peak overloads with changes in the closure loads. Photoelastic measurements in polyester models[9] have also indicated closure in this polymer, although no attempt was made to relate closure with fatigue crack retardation.

The crack tip blunting mechanism[10] suggests that local plastic flow during the overload physically blunts the crack tip, reducing the stress concentration factor. A significant portion of subsequent cycling is then required to reinitiate a sharp crack tip in order to allow further crack propagation. Although this mechanism appears credible, crack tip blunting has not been presented in a form convenient for use with crack growth prediction techniques.

A third explanation for the retardation phenomena suggests that residual compressive stresses in the plastic zone reduce the effective tensile stress ahead of the crack tip. The overload creates a larger compressive stress field which lowers the effective tensile stresses, slowing the crack growth rates. Thus, this model places the effect of the residual stresses ahead of the crack tip, while crack closure assumes the residual stresses act behind the tip to close the crack faces.

Several models based on the effect of the plastic zone ahead of the crack tip[11-13] have provided, with limitations, useful crack growth estimates.

Although a considerable amount of work has been conducted in metals, studies of fatigue crack retardation in polymers have received much less attention. Pitoniak et al.[8] indicated that tensile overloads could perturb subsequent crack growth in poly(methyl methacrylate), although the amount of observed retardation was small in comparison with many metals. Since this polymer has a low fracture toughness (K_{IC} about 1000 psi-in.$^{1/2}$) resulting in a small crack tip plastic zone, the slight retardation effect is not surprising. One would expect fatigue crack growth in more ductile polymers, however, to be more influenced by load history. The purpose of this paper is to describe the results of some recent fatigue crack retardation studies in polycarbonate, a polymer with a much larger fracture toughness than PMMA (K_{IC} = 3300 psi-in.$^{1/2}$). The data presented here indicate that tensile overloads may, in fact, significantly delay subsequent fatigue crack extension in polycarbonate specimens.

EXPERIMENTAL PROCEDURES

Edge-notched bend specimens (1.375 in. wide by 7.5 in. long) were machined from a single sheet of 1-in.-thick polycarbonate. Care was taken to maintain the same directional orientation of all specimens to assure uniform mechanical properties. The machined ends were polished to transparency with metallographic lapping wheels to allow direct observation of the fatigue cracks. Next, to remove any residual machining stresses, each member was annealed at 280° ± 3°F for 24 hr and cooled slowly to room temperature. The 0.1-in.-deep V-notches were then sharpened with a scapel blade to provide a convenient source for initiating fatigue cracks.

The specimens were cycled in a four-point bend apparatus mounted in a closed-loop electrohydraulic test machine. The major span between load points was 6.0 in., while the minor span was 4.0 in. The cyclic frequency was maintained at 3 Hz, with the R ratio ($K_{\min}/K_{\max}$) kept near zero ($0 \leq R \leq 0.05$). No attempt was made to control the laboratory environment. Fatigue cracks were initiated at a ΔK of approximately 2000 psi-in.$^{1/2}$ Once a uniform crack front developed, the cyclic load was lowered and subsequent crack growth allowed to stabilize. In order to maintain conditions of constant range in stress intensity factor, the cyclic load was periodically reduced by predetermined amounts to correspond with crack growth increments of 0.01 in. In this manner, ΔK was kept constant within ±2% deviation.

Crack extension was measured from photographs of the crack plane taken through the transparent specimens with a 35-mm camera equipped with bellows. In order to provide a reference distance in the photos, a scale graduated in 0.005-in. increments was fixed to half of a polished polycarbonate specimen taped to the side of the test member. Thus, both the graduated scale and the crack plane were viewed at the same optical distance from the camera. By enlarging the photographic negatives approximately 20× with a filmstrip projector and measuring the projected image to the nearest 0.01 in., it was possible to resolve crack growth increments smaller than 0.001 in.

When the crack had grown a sufficient amount to assure stable growth (15,000 to 20,000 cycles), tensile overloads were applied. The overload procedure was accomplished by stopping the test, photographing the crack, applying the overload at the same loading rate as the baseline loading, and rephotographing the crack to record any extension due to the peak load. This entire process resulted in a delay of 30 to 60 sec, depending on the number of overloads. The cyclic loading was then adjusted to the original stress intensity range and the test continued. The experiment was usually terminated 15,000 to 20,000 cycles after fatigue crack growth had reinitiated and uniform growth was reestablished.

EXPERIMENTAL RESULTS

Materials Properties

Tensile tests conducted on specimens of standard ASTM configuration gave an elastic modulus of 324,000 psi, a 0.2% offset yield strength of 6000 psi, and a yield point of 9200 psi for the annealed polycarbonate. Identical results were obtained at loading rates of 0.01 and 0.1 in./min. Fracture toughness measurements from instrumented four-point bend specimens tested in compliance with ASTM recommended procedure[14] gave values of 3260, 3270, 3310, and 3400 psi-in.$^{1/2}$ (average value = 3310 psi-in.$^{1/2}$). Since the 1-in. specimen thickness exceeds the minimum plane strain thickness requirements ($2.5 \times (K_{IC}/\sigma_{YS})^2 = 0.32$ in., where $K_{IC} = 3310$ psi-in.$^{1/2}$ and $\sigma_{YS} = 9200$ psi), these values represent the plane strain fracture toughness K_{IC}. Johnson et al.[15] have reported similar room-temperature fracture toughness values for 0.25-in.-thick slow three-point bend specimens.

Fatigue crack growth rate data for the polycarbonate tested are shown in Figure 1. Points on this curve were obtained by two methods. The data represented by squares are from the constant ΔK baseline tests. Here, da/dN is the slope of the crack growth curve under constant ΔK conditions. The second method involved calculating the crack growth derivative by least-squares procedures[16] from constant-load tests as ΔK increased with crack extension to unstable fracture. As seen in Figure 1, data obtained by both methods are compatible. Fitting a straight line of the form

$$da/dN = C(\Delta K)^m \tag{1}$$

through these data gives the experimental constants $C = 6.7 \times 10^{-21}$ and $m = 4.87$, where the units of da/dN are in./cycle and ΔK is measured in psi-in.$^{1/2}$ Also included on Figure 1 is a band that encompasses fatigue crack growth data given in reference 3 for polycarbonate specimens tested at frequencies of 0.33 and 10 Hz. The present results agree well with these prior data.

Fatigue Crack Retardation

Fatigue cracks were grown under conditions of constant range in stress intensity factor at nominal ΔK levels of 750 and 900 psi-in.$^{1/2}$ Tensile overloads with λ values ($\lambda = K_{max}/\Delta K_{baseline}$) of 2.1, 3, and 3.75, were applied to the 750 psi-in.$^{1/2}$ baseline tests, while overload ratios of 2.1 and 3 were applied to the 900 psi-in.$^{1/2}$ baseline. The number of overloads alternated between 1 and 5, with

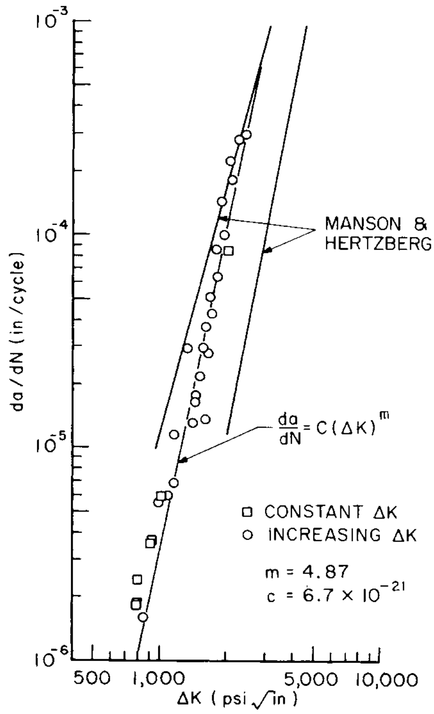

Fig. 1. Polycarbonate fatigue crack growth rate data.

the exception of one test in which 3500 overloads ($\lambda = 1.92$) were applied to a 750 psi-in.$^{1/2}$ baseline. These test conditions are summarized in Table I. Typical crack growth curves for some of the overload tests are presented in Figures 2 through 5. All of the retardation results are summarized in Table I. A schematic crack growth curve defining the parameters reported here is given in Figure 6.

Examining the crack growth curves of Figures 2 through 5, one notices that tensile overloads do, in fact, retard subsequent constant ΔK crack growth in polycarbonate. In order to isolate the influence of the experimental variables, the number of delay cycles (N_R) following the peak loads are plotted versus overload ratio squared (λ^2) for all of the tests in Figure 7. As seen by the following argument, λ^2 may be interpreted as an estimate of the increase in crack tip plastic zone size due to the overload. Theocaris and Gdoutos[17] have shown that for the small crack length-to-specimen thickness ratios encountered in the present tests, the Irwin plane strain plastic zone formula provides a reasonable estimate of the crack tip plastic zone size in polycarbonate. Thus,

$$r_y = \frac{1}{6\pi}\left(\frac{K}{\sigma_{YS}}\right)^2 \tag{2}$$

where r_y is the plastic zone, and k and σ_{YS} are the stress intensity factor and yield strength, as before. Since the overload stress intensity $K_{\max}$ is, by definition, equal to $\lambda \Delta K_{\text{baseline}}$, the ratio of the overload plastic zone to the baseline plastic zone reduces through eq. (2) to λ^2.

Examining Figure 7, one notes that the overload value λ (and, therefore, the

TABLE I
Summary of Retardation Experiments

Test no.	Baseline ΔK, psi-in.$^{1/2}$	No. of overloads	Overload ratio λ	a_p, in.	N_R, cycles	N_{ER}, cycles	da/dN_1, in./cycle	da/dN_2, in./cycle
310-1	750	1	2.1	.002	2,000	0	$.18 \times 10^{-5}$	$.18 \times 10^{-5}$
310-3	750	1	3	.003	4,000	2,500	$.19 \times 10^{-5}$	$.22 \times 10^{-5}$
500-2	800	1	3.75	.010	11,000	6,000	$.24 \times 10^{-5}$	$.24 \times 10^{-5}$
310-2	750	5	2.1	.003	2,500	0	$.18 \times 10^{-5}$	$.18 \times 10^{-5}$
310-4	790	5	2.85	.005	7,000	5,000	$.25 \times 10^{-5}$	$.18 \times 10^{-5}$
500-1	800	5	2.81	.004	6,500	5,000	$.28 \times 10^{-5}$	$.24 \times 10^{-5}$
500-3	800	5	3.75	.015	8,500	2,500	$.24 \times 10^{-5}$	$.25 \times 10^{-5}$
410	780	3,500	1.92	—	2,000	—	$.23 \times 10^{-5}$	$.24 \times 10^{-5}$
207-1	900	1	2.1	.002	2,000	700	$.35 \times 10^{-5}$	$.35 \times 10^{-5}$
207-3	900	1	3	.01	7,000	4,000	$.37 \times 10^{-5}$	$.38 \times 10^{-5}$
207-2	900	5	2.1	.004	2,700	900	$.35 \times 10^{-5}$	$.35 \times 10^{-5}$
207-4	900	5	3	.012	7,500	3,800	$.38 \times 10^{-5}$	$.35 \times 10^{-5}$

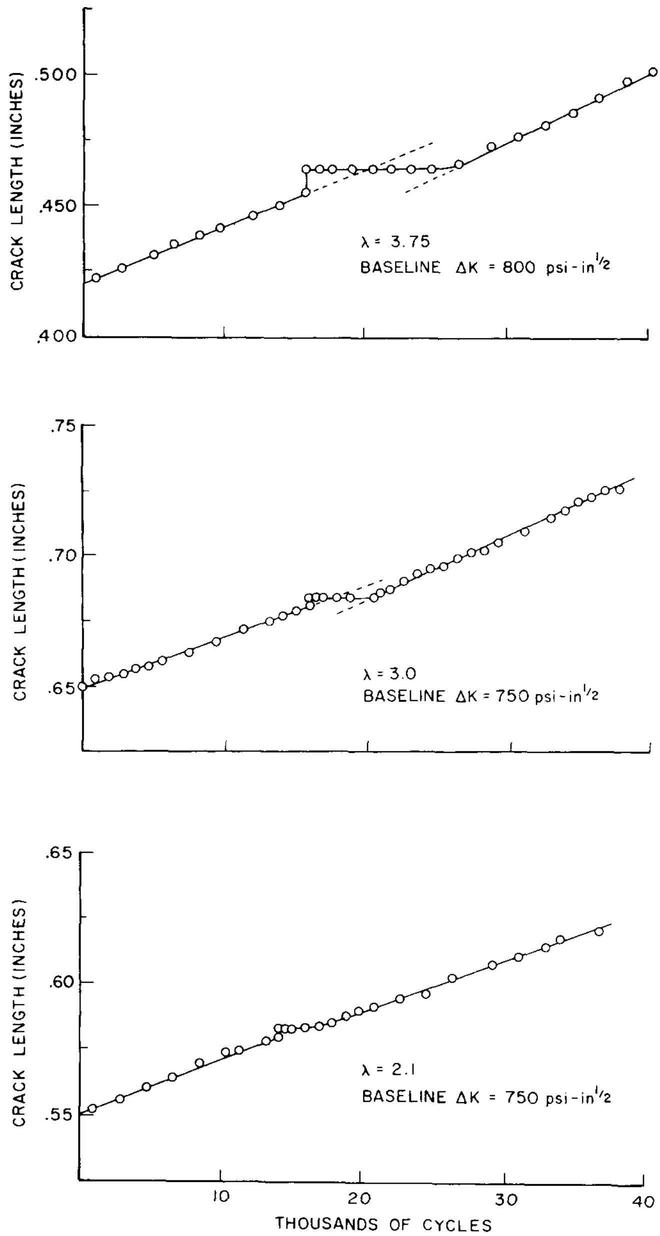

Fig. 2. Fatigue crack growth data for one overload applied to nominal baseline ΔK of 750 psi-in.$^{1/2}$

plastic zone ratio) has the major influence on retardation of the variables examined. Neither the number of overloads applied (1 versus 5) or the baseline ΔK level (750 versus 900 psi-in.$^{1/2}$) produced a major effect on N_R which could be separated from the experimental scatter. It must be pointed out, however, that only small variations in these latter variables were examined, and one should be cautious in extrapolating these conclusions to other test conditions or materials.

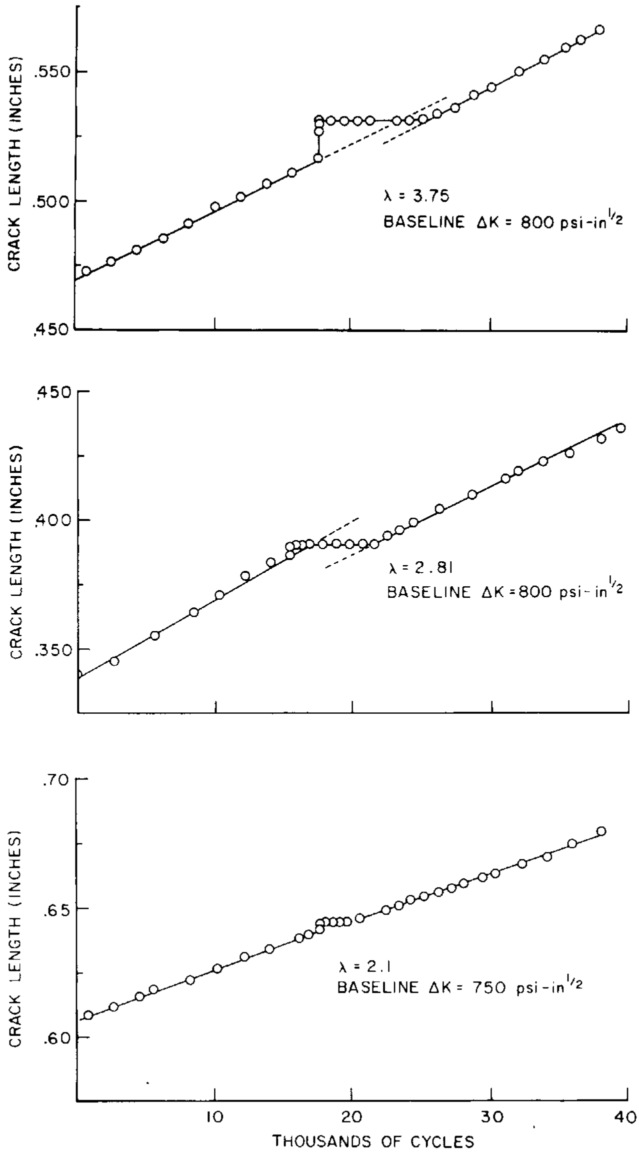

Fig. 3. Fatigue crack growth data for five overloads applied to nominal baseline ΔK of 750 psi-in.$^{1/2}$

It is also interesting to note from Figure 7 that N_R increased more or less linearly with λ^2. Although delay cycles also increase with overload ratio in metals, λ often reaches a value which causes infinite delay (crack arrest). Probst and Hillberry,[18] for example, found that if λ exceeded 2.23 in 2024-T3 aluminum alloy specimens cycled at a stress ratio $R = 0.3$, there was no measurable crack growth for over 10^6 cycles. Although any arrest overload value would be expected to differ with material, note that the largest delay period measured here was only 11,000 cycles for the rather large overload of $\lambda = 3.75$ (test 500-2). Thus, it would appear that the retardation phenomenon is not as significant in polycarbonate as in 2024-T3 aluminum, a conclusion which is not surprising if one remembers

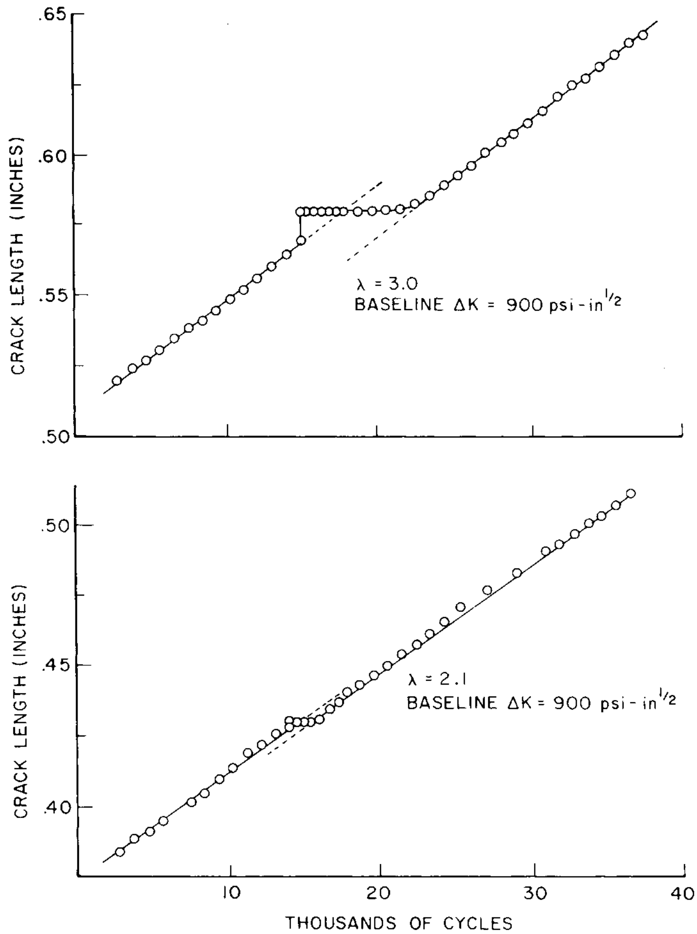

Fig. 4. Fatigue crack growth data for one overload applied to nominal baseline ΔK of 900 psi-in.$^{1/2}$

that the viscoelastic polymer would allow a much quicker relaxation of residual plasticity effects than the metal. This viscoelastic influence on crack tip behavior was also observed in reference 8 where residual crack surface displacements obtained by interferometric measurements on PMMA specimens decreased significantly when the specimen was allowed to rest at zero load.

Although there was no apparent change in the crack front on a macroscopic scale immediately following the overload, the observed retardation was evidently due to blunting of the crack tip. Figure 8a and b show schematically the notched specimen and the original fatigue crack initiation at several localized sites along the starter notch. (Recall that the transparent specimen material allowed direct visual observation of the entire crack plane prior to fracture.) Eventually, these small surface flaws united into a single front that propagated in a continuous manner as shown in Figure 8c. (Due to poor photograph quality, Figure 8c–e are shown schematically.) Following the peak load, however, this uniform extension ceased. Although there was no apparent change in crack tip appearance on a macroscopic level immediately after the overload (Fig. 8d), after

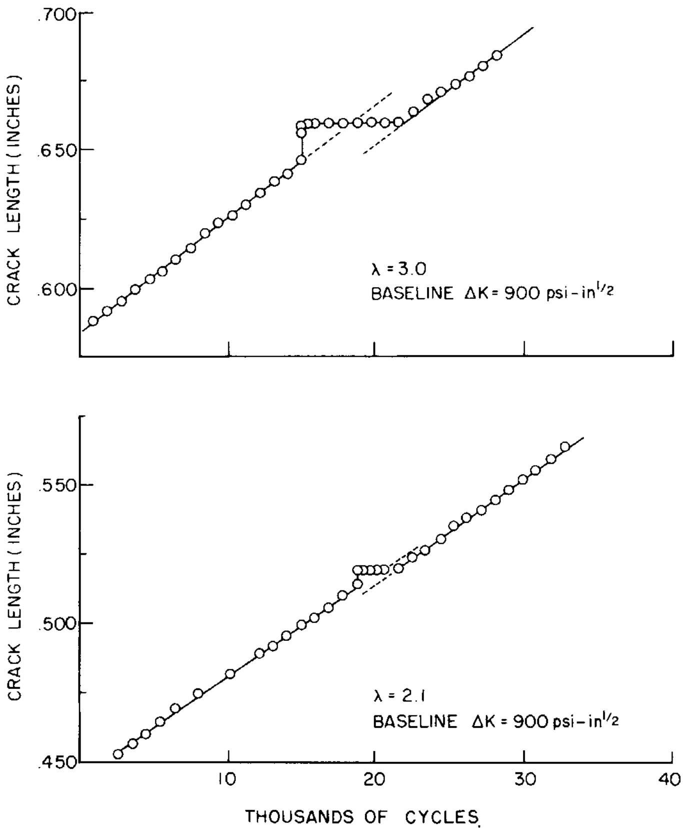

Fig. 5. Fatigue crack growth data for five overloads applied to nominal baseline ΔK of 900 psi-in.$^{1/2}$

a period of cycling the flaw began to extend independently at several localized points along the crack tip as in Figure 8e.

Thus, recovery of crack growth from the overload closely resembled the original fatigue crack initiation sites at the V-notch. Since the V-notch has a finite root radius compared with the "infinitely sharp" crack tip, reinitiation following the overloads suggests that the peak load blunted the crack tip. Once a stable crack front was reinitiated upon return to the baseline ΔK, the subsequent crack growth returned to the original propagation rate as summarized in Table I (compare da/dN_1 with da/dN_2).

CONCLUDING REMARKS

Tensile overloads were shown to significantly influence subsequent fatigue crack extension in polycarbonate specimens grown under conditions of constant range in stress intensity factor, although the retardation effect is not as large as for some metals. These results indicate that it is possible to extend the fatigue life of flawed polycarbonate specimens with tensile overloads. The number of delay cycles (N_R) following the peak loads were found to depend strongly on the magnitude of the overload ratio λ. For the conditions studied here, neither the

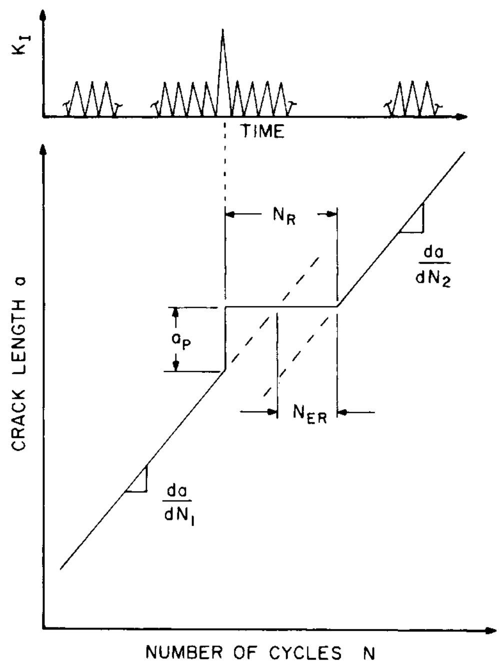

Fig. 6. Schematic of constant ΔK fatigue crack growth retardation due to tensile overloads.

baseline ΔK level, or the number of overloads appreciably affected N_R. Attempts to relate the total increase in specimen life N_{ER} (see Fig. 6) to these test variables were complicated by the additional crack extension a_p caused by the overloads. The data summarized in Table I generally indicate, however, that N_{ER} also increased with λ.

An attempt to apply the interferometric technique employed in reference 8 to this study was unsuccessful because of the roughness of the fatigue crack surfaces. The light rays transmitted through the transparent polycarbonate

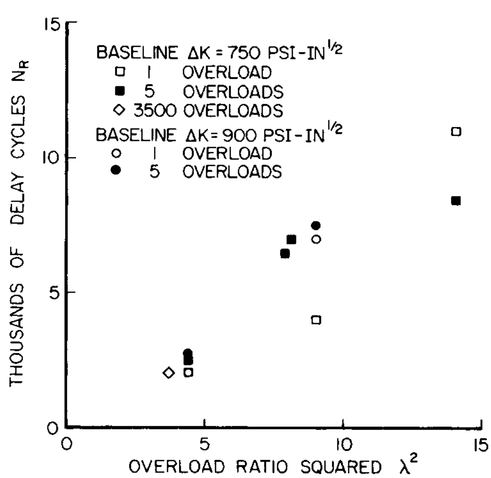

Fig. 7. Effect of tensile overload magnitude on fatigue crack retardation in polycarbonate.

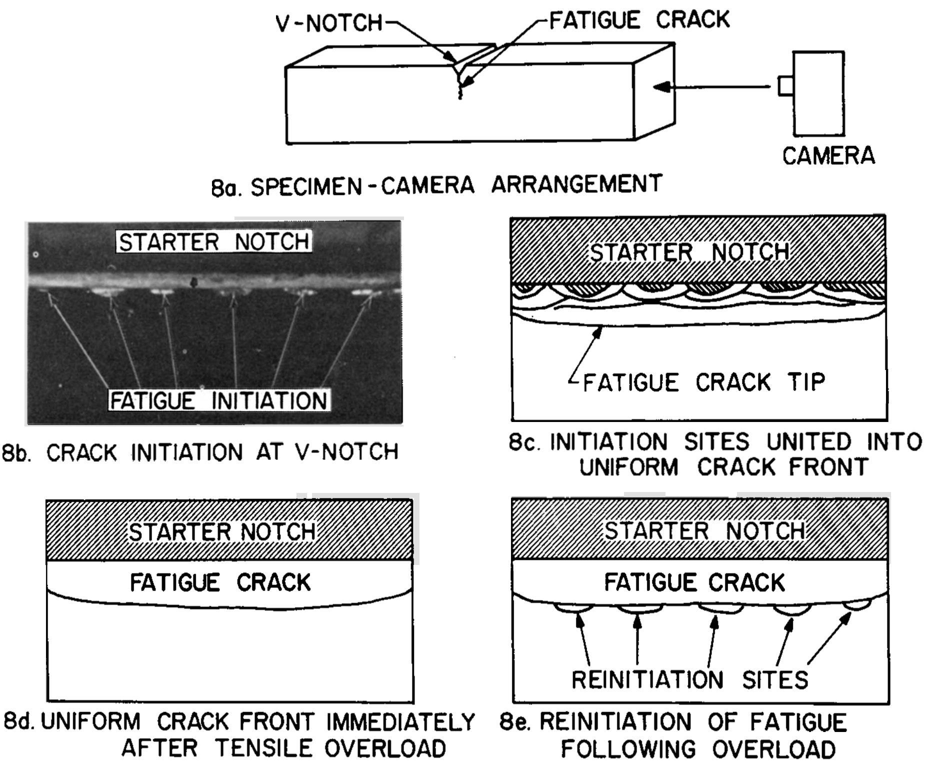

Fig. 8. Schematic view of fatigue crack initiation at original V-notch and reinitiation following tensile overload.

FIG. 8 SCHEMATIC VIEW OF FATIGUE CRACK INITIATION AT ORIGINAL V-NOTCH AND REINITIATION FOLLOWING TENSILE OVERLOAD

specimens were dispersed by the crack faces, preventing formation of the interference fringes. Although similar problems were encountered with crack roughness in the PMMA work reported in reference 8, it was possible to grow smooth cracks at low ΔK levels. Crack surfaces in the more ductile polycarbonate were unsuitable, however, even at the low $\Delta K = 750$ psi-in.$^{1/2}$ baseline levels.

Although further efforts to use other experimental methods for measuring closure loads in metal specimens[6–7,19–21] were not employed here, it would appear that the closure mechanism does not completely describe fatigue crack retardation in polycarbonate. Clearly, the reinitiation of separate crack growth sites along the crack tip following the overload suggests that crack blunting might be a more appropriate model for fatigue crack retardation in polycarbonate. More rigorous examination of this point remains, however, for future work.

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