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Fatigue Crack Propagation in Polymeric Materials

R. W. HERTZBERG, H. NORDBERG*
Department of Metallurgy and Materials Science

J. A. MANSON
Department of Chemistry, Lehigh University, Bethlehem, Pa., USA

In order to gain a better understanding of matrix-controlled fatigue failure processes in non-metallic materials a series of fatigue tests were performed on several different polymer materials representing different classes of mechanical response. Fatigue crack propagation rates between 5×10^{-6} in. cycle^{-1} (127 nm cycle^{-1}) and 4×10^{-4} in. cycle^{-1} (10 300 nm cycle^{-1}) were measured in nylon, polycarbonate, ABS resin, low-density polyethylene and polymethyl methacrylate. A strong correlation was found between the fatigue crack propagation rate and the stress intensity factor range prevailing at the advancing crack tip. Whereas metals exhibit comparable fatigue growth rates for a given stress intensity range when normalised with respect to their static elastic modulus, the polymer materials exhibited a 1300-fold difference in crack growth rate for a given normalised stress intensity range. This observation dramatically illustrates the importance of understanding molecular motion and energy dissipation processes in polymer materials as related to their chemistry and architecture. The relative behaviour of the different polymer materials could be generally correlated with their reported damping characteristics.

1. Introduction

In order to design composite materials intelligently, a better understanding of failure mechanisms is clearly needed. Although the stresses encountered frequently are cyclical in nature, relatively little information has been generated concerning the fatigue response of composites. In addition, the interaction between stress and environment has not been clearly defined.

Most of the fatigue data for composites in existence have appeared in the form of stress vs. cyclic life plots as shown in the recent work by Hofer *et al* [1]. While these data do help in the evaluation of relative performance of materials, they do not permit a detailed analysis of the fracture process. Specifically, it is not possible to separate the initiation and propagation stages of the stress vs. cyclic life curves. Since most engineering components have flaws present before testing, it is important to isolate the propagation stage of fracture and define the fracture mechanisms and their dependence on the stress environment.

To better understand the rôle of matrix material in the fatigue failure processes in non-metallic composites, an exploratory study of fatigue crack propagation in nylon 66, low-density polyethylene (LDPE), acrylonitrile-butadiene-styrene resin (ABS), polycarbonate (PC), polymethyl methacrylate (PMMA) and epoxy resin ERL 2256 with ZZL 0820 hardener was performed.

2. Experimental Procedures

2.1. Materials Selection

A series of fatigue tests were performed on several polymer materials representing different classes of mechanical response. In this manner it was hoped that a better understanding of the fatigue response of various polymer structures and morphologies could be obtained and that

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*Currently at the Swedish Institute for Metal Research, Stockholm, Sweden.

TABLE I Types and sources of materials used

Material*	Specifications	Source
1. Epoxy resin	ERL 2256 resin, ZZL 0820 hardener	Air Force Materials Laboratory
2. Acrylonitrile-butadiene-styrene (ABS) resin	standard "house" grade	Westlake Plastics Co., Lenni, Pa.
3. Nylon 66	standard "house" grade	Westlake Plastics Co., Lenni, Pa.
4. Low-density polyethylene (LDPE)	standard "house" grade	Westlake Plastics Co., Lenni, Pa.
5. Polycarbonate (PC)	standard "house" grade	Plastics Mart, Allentown, Pa.
6. Polymethyl methacrylate† (PMMA)	standard "house" grade	Plastics Mart, Allentown, Pa.

*Nominal thickness, $\frac{3}{16}$ in. (0.48×10^{-2} m) in each case.
†Additional test with 0.24 in. (0.62×10^{-2} m) plate.

TABLE II Tensile properties of polymers for crack growth studies

Polymer, specimen		ϵ, strain rate (sec^{-1} $\times$ 10^4)	E (psi $\times$ 10^{-5})		σ, tensile strength (psi)	
Polyethylene	- A	~5.5	0.26	(0.3-0.6)*	1460	(1200-3500)
	- B	~11	0.33		1200	
Nylon 66	- C	~5.5	4.3	(2-4)	11 900	(9000-12 000)
	- D	~5.5	4.0		12 000	
ABS	- E	~5.5	2.8	(3-4)	6070	(6000-8000)
	- F	~5.5	2.7		5800	
PMMA		—	—	(4-5)	—	(6000-9000)
Polycarbonate		—	—	(3-4)	—	(9000-10 000)
Epoxy - RC-1		—	4.3†		15 000†	

*Data in parentheses represent typical ranges quoted in Modern Plastics Encyclopedia, 1968, and Materials Selector, 1968.
†Quoted values - supplier.

such information might suggest approaches to the development of tougher epoxy resins. Several groups of polymers were selected in order to assess the relative importance of elastic modulus and inelastic energy dissipation in determining fracture toughness (tables I and II). It was assumed, in other words, that plane strain fracture toughness, K_{Ic}, is a function of both modulus, E, and the rate of energy dissipation. Amorphous and crystalline polymers, and one two-phase polyblend, were included.

Polycarbonate (PC) and polymethyl methacrylate (PMMA) were chosen as a pair of essentially amorphous polymers which have nearly equal moduli of elasticity, but which vary significantly with respect to well-established fracture energy and damping characteristics [2-4]. For example, the brittle temperatures of PMMA and PC are reported to be +45° C and −200° C, respectively [5, p. 53]. As a further indication of the relative difference in fracture energy between the two resins, PMMA and PC exhibit Izod impact energies of about 0.04 and 12 ft lb in.$^{-1}$ of notch (2.16 and 647 mN m^{-1}), respectively [6]. While the authors are not aware of K_{Ic} values for PC, Watts and Burns [7] reported K_{Ic} for PMMA to be 1.45 ksi in.$^{1/2}$ (1.59 MN m$^{-3/2}$), reflecting low toughness. The fundamental reason for such differences may be attributed to the fact that, as shown by damping curves as a function of temperature, PMMA exhibits several discrete narrow mechanical transition regions, while PC has a rather broad and diffuse spectrum [2, 3]. The spectrum of PC extends well below ambient temperatures and indicates the existence of an appreciable degree of main-chain segmental motion at room temperature.

The damping behaviour of the epoxy resin ERL 2256 is believed to be similar to that of other epoxy resins which typically exhibit a relatively low degree of energy absorption [8] and limited segmental motion, and which resemble, in this sense, PMMA. Typical values for the

Izod impact energy of epoxy resins are low – about 0.2 ft lb in.$^{-1}$ (10.8 mN m^{-1}) of notch [6].

To elucidate the rôle of elastic modulus, a pair of tough polymers was chosen for evaluation: nylon 66 and low-density polyethylene (LDPE). Each of the materials chosen exhibits rather broad and complex relaxation behaviour, but differ markedly from the other in the magnitude of E [9, 10, p. 779]. Thus, nylon 66 ($E = 4 \times 10^5$ psi) (2.76 -N m^{-2}) has an Izod impact energy value of about 1 to 2 ft lb in.$^{-1}$ (54 to 108 mN m^{-1}) of notch, while LDPE ($E = 2.5 \times 10^4$ psi) (1.73 -N m^{-2}) does not break under the conditions of the Izod test [6].

Finally, to evaluate the rôle of a molecularly dispersed elastomeric component in an otherwise brittle matrix, a high-impact-strength ABS resin was selected. This resin serves as an example of a common approach to toughening brittle plastics: the use of graft polymerisation or mechanical blending to provide a highly deformable second phase dispersed in a rigid matrix [11, p 81; 12, 13]. While polystyrene by itself would be expected to be approximately as brittle as PMMA, the ABS resin has an Izod impact energy value of about 7 ft lb in.$^{-1}$ (378 mN m^{-1}) of notch. It also has a broad mechanical relaxation spectrum [6, p. 121].

2.2. Test Procedure

Single-edge-notch, pin-loaded specimens, 3 in. wide (7.6×10^{-2} m), 12 in. long (30.5×10^{-2} m) and $\frac{3}{16}$ to $\frac{1}{4}$ in. (0.48 to 0.63×10^{-2} m) thick were tested in tension-tension loading. The PC fatigue test conducted at 0.33 c/s was performed in an Instron machine while all other tests, covering a frequency range of 1 to 10 c/s, were performed in an MTS electro-hydraulic closed loop testing machine. From the raw data of crack length versus number of fatigue cycles, the macroscopic growth rate, da/dn, was related to changes in the crack tip stress intensity field. For the specimen configuration used in this investigation,

$$\Delta K = \Delta\sigma \sqrt{\pi a}\, \mathrm{f}\,(a/b)$$

where ΔK = stress intensity factor range; $\Delta\sigma$ = gross stress range; a = crack length; f(a/b) = correction factor for particular specimen geometry in terms of crack to half panel width ratio.

3. Experimental Results and Discussion

3.1. Linear Polymers

Fatigue crack propagation rates between 5×10^{-6} in. cycle^{-1} (127 nm cycle^{-1}) and 4×10^{-4} in. cycle^{-1} (10 300 nm cycle^{-1}) were measured in the nylon, polycarbonate, ABS resin, low-density polyethylene and the PMMA polymers. A strong correlation was found between the fatigue crack propagation rate and the stress intensity factor range prevailing at the advancing tip (figs. 1-6). Whereas this correlation has been shown in the case of metals [14] it was surprising to note the applicability of this approach in the case of these polymers possessing widely varying mechanical response with respect to their non-linear deformation behaviour.

The relative behaviour of the different polymer materials could be correlated generally with their damping characteristics. For example, nylon 66, which exhibits a rather broad energy absorption spectrum below room temperature and a fairly low glass transition temperature, showed the highest resistance to crack propagation (fig. 6). The existence of loss peaks below room temperature implies the occurrence of segmental motions, which may permit localised viscoelastic flow or creep. In addition, the mechanical loss *per se* implies a dissipation of energy in the motion itself. Low-density polyethylene – also having a broad energy absorption spectrum and, in addition, superior toughness under impact loading – was, on the other hand, much less resistant to crack propagation,

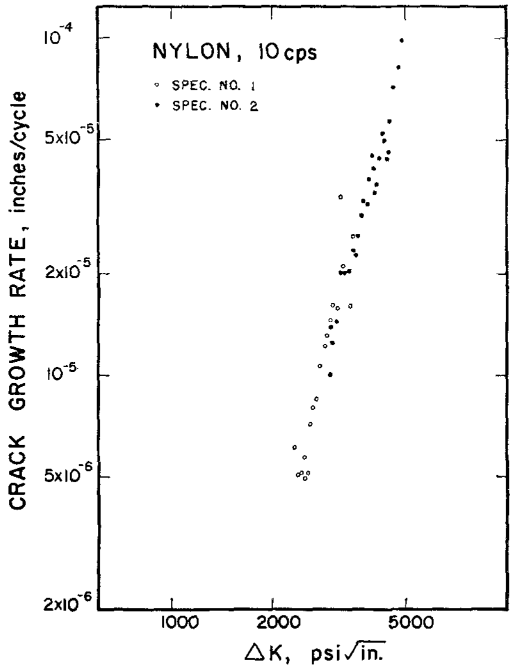

Figure 1 Fatigue crack growth characteristics of nylon 66.

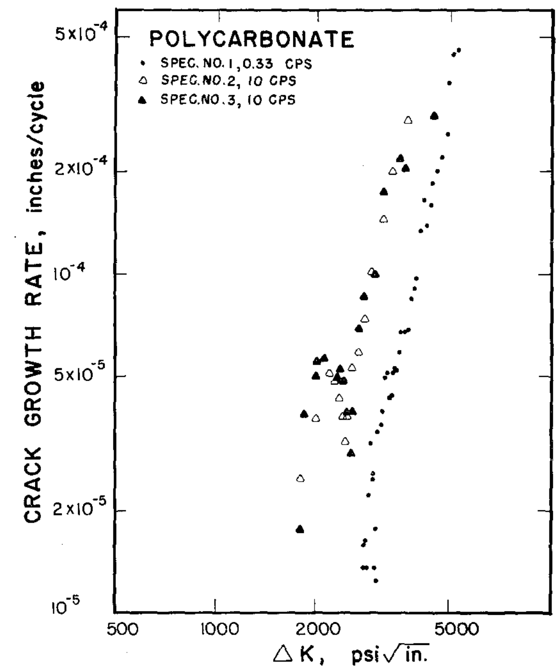

Figure 2 Fatigue crack growth characteristics of polycarbonate revealing effect of test frequency.

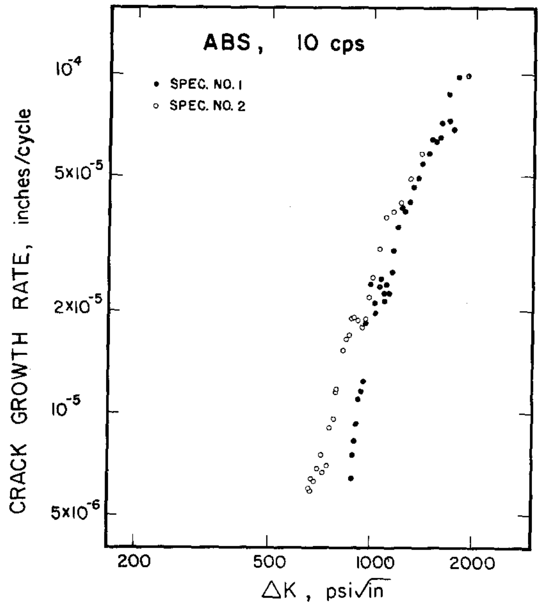

Figure 3 Fatigue crack growth characteristics of ABS resin.

apparently due to a much lower elastic modulus.

The suspected rôle of damping characteristics in conferring toughness at a given level of modulus was confirmed by the results for PC and PMMA. Although the value of E was in fact somewhat higher for PMMA, the resistance shown to crack propagation was very much less than was the case for PC. This behaviour was consistent with both the brittle temperature and

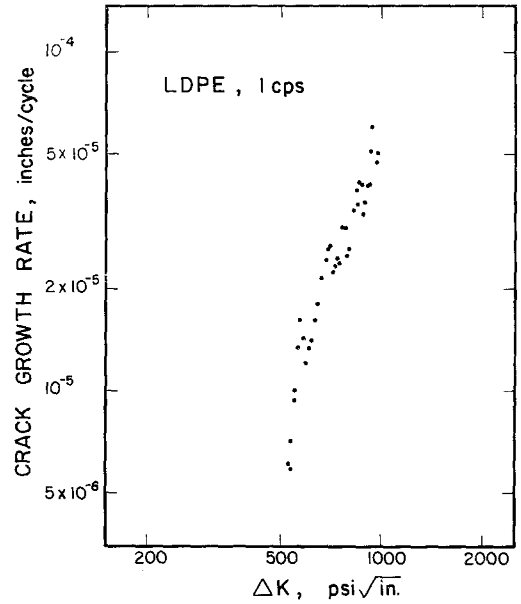

Figure 4 Fatigue crack growth characteristics of low-density polyethylene.

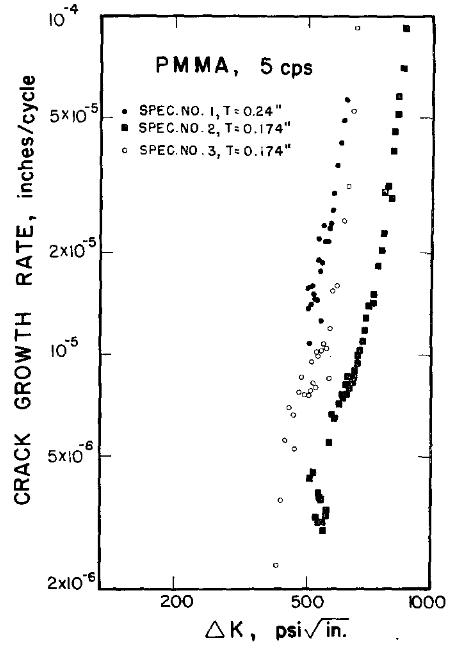

Figure 5 Fatigue crack growth characteristics of PMMA.

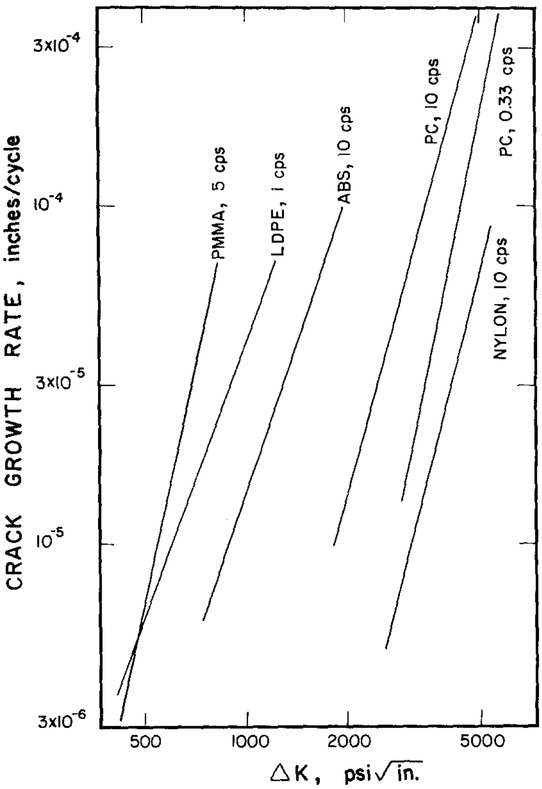

Figure 6 Comparative fatigue crack growth rate behaviour of several polymer materials. (PMMA: polymethyl methacrylate; LDPE: low-density polyethylene; ABS: acrylonitrile-butadiene-styrene resin; PC: polycarbonate resin).

impact behaviour. It is noted that the crack propagation characteristics and the K_{Ic} values at fracture (approximately 1450 psi in.$^{1/2}$ (1.59 MN m$^{-3/2}$) of PMMA are in agreement with the published results of Watts and Burns [7].

Since the K_{Ic} level in polystyrene, 1600 psi in.$^{1/2}$ (1.75 MN m$^{-3/2}$) [16], is comparable to the PMMA value, 1450 psi in.$^{1/2}$, found in this investigation and by others [13] it may be possible to compare the fatigue results of the toughened ABS resin with PMMA behaviour. Accordingly, it is suggested that the improvement of toughness by grafting or polyblend approach gave rise to lower fatigue crack growth rates in the ABS resin as compared to PMMA.

Thus, great differences in crack propagation rates were observed for polymeric materials differing only moderately in tensile elastic modulus at room temperature. While different metals exhibit comparable fatigue growth rates for a given stress intensity range when normalised with respect to static elastic modulus [15], the polymer materials exhibited a 1300-fold difference in crack growth rate for a given normalised stress intensity range. This observation reflects differences in the damping behaviour and viscoelastic response of the polymers examined. For example, the improvement in fatigue response of the PC polymer with decreasing cyclic rate may be associated with greater viscoelastic energy absorption at lower strain rates (figs. 2, 6). Therefore, it appears more appropriate to adjust the applied stress intensity factor for comparison of fatigue performance with a more meaningful measure of the elastic modulus, taking into account both time-dependent and time-independent strains [5, 18].

3.2. Epoxy Resin ERL 2256

Gripping problems were encountered initially (for example, frequent fracture at the pin-loading holes) until the specimens were reinforced with backing plates. Attempts were made to determine the fatigue crack propagation characteristics of the epoxy resin at room temperature. No measurable crack extension was observed before catastrophic failure. An attempt was made to conduct a fatigue test at 90° C in order to introduce some additional mechanisms for energy dissipation which could lead to plastic deformation. In this instance negligible stable crack extension was observed. Microscopic observation (fig. 7) did reveal an extremely small region of stable crack extension as observed by the presence of fatigue striations on the fracture surface. This region extended less than 0.02 in. (0.51 × 10^{-3} m) from the original crack tip. The range of spacing

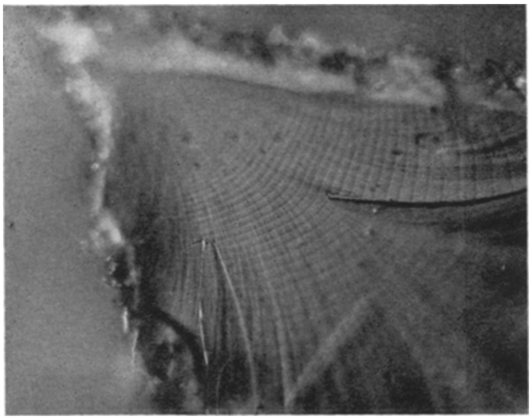

Figure 7 Local region of stable crack extension in epoxy resin revealing fatigue striations (×320).

of these striations, a measure of stable crack extension rate, for the estimated stress intensity level at the starting crack tip, revealed the epoxy resin to possess the worst fatigue performance. This observation is indicative of the brittle behaviour of the epoxy resin ERL 2256 which exhibited an estimated fracture toughness in the range of 1000 psi in.$^{1/2}$ (1.1 MN m$^{-3/2}$).

4. Conclusions

While these materials exhibited various degrees of non-linear behaviour, it was possible to correlate the crack propagation characteristics for each of the polymers examined by means of the stress intensity factor range. It appears that a correlation between the different polymer materials requires some normalising factor accounting for time-dependent and time-independent strains.

The fatigue results with the series of polymers emphasised that improved resistance to crack propagation requires substantial internal energy dissipation mechanisms. These may arise from secondary transitions or flow due to mobility of chain segments, from the presence of dispersed crystalline regions, or from a second elastomeric phase dispersed in a glassy matrix. If such mechanisms exist, high elastic moduli need not be sacrificed.

The epoxy resin ERL 2256 was found to be extremely brittle and possessing limited capacity for stable crack extension. From the fracture toughness standpoint, this material would appear to be unsatisfactory as a matrix material in a reinforced composite.

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References

1. K. E. HOFER JR and E. M. OLSEN, Technical Report AD 652.415, April 1967.
2. R. F. BOYER, *Polymer Eng. and Sci.* **8** (1968) 161.
3. J. HEIJBOER, *J. Polymer Sci.* **5** (1967) 3755.
4. S. G. TURLEY, *Polymer Preprints*, American Chemical Society **8** (1968) 1524.
5. E. H. ANDREWS, "Fracture in Polymers" (American Elsevier Publishing Company, New York, 1968).
6. Modern Plastics Encyclopedia (McGraw-Hill Inc., New York, 1968).
7. N. H. WATTS and D. J. BURNS, *Polymer Eng. and Sci.* **1** (1967) 90.
8. T. K. KWEI, *J. Polymer Sci.* **4** (1966) 943.
9. A. E. WOODWARD, J. A. SAUER, C. W. DEELEY, and D. E. KLINE, *J. Colloid Sci.* **12** (1957) 363.
10. M. P. GROENEWEGE *et al*, in "Crystalline Olefin Polymers", edited by M. Raff and A. Doak, "High Polymers", vol. XX (Interscience Publishers, New York, 1965).
11. A. V. TOBOLSKY, "Properties and Structure of Polymers" (Interscience Publishers, New York, 1960).
12. F. R. MCGARRY and A. M. WILLNER, MIT Research Report R68-8, March 1968.
13. J. A. BRYDSON, "Plastic Materials" (D. van Nostrand Company Inc., Princeton, NJ, 1966).
14. P. C. PARIS and F. ERDOGAN, ASME Paper 62-WA-234, 1963.
15. S. PEARSON, *Nature* **211** (1966) 1077.
16. Y. KATZ, P. L. KEY, and E. R. PARKER, *Trans. ASME* Ser. D **90** (1968) 622.
17. "Fracture Processes in Polymeric Solids", edited by B. Rosen (John Wiley and Sons Inc., New York, 1964).
18. F. R. EIRICH, *J. Appl. Polymer Sci.* **4** (1961) 1.

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