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Strain Rate Dependence of Failure Processes in Polycarbonate and Nylon

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Synopsis

A study has been made of two types of failure, namely, monotonic fractures using Charpy-type specimens and fatigue crack propagation using rectangular plates containing an initial central notch. The work was conducted on an amorphous polymer (polycarbonate) and a semicrystalline polymer (nylon N 6.6). Monotonic tests were performed in an Instron testing machine between 0.002 and 20 in./min, and in a Charpy testing machine between 2000 and 11800 in./min. The cyclic tests (under constant K conditions) were carried out at frequencies that ranged from 0.1 to 20 Hz. A model for the relationship between the cyclic rate of crack growth and appropriate LEFM parameters, previously described, has now been converted into cyclic strain energy transformations. In computing the strain energy, the value of the time-dependent modulus of the material was used; and under cyclic loading conditions the glassy (short time) value was employed. The authors have proposed that the modulus measurements obtained at very low temperatures, where the viscous response of the material is highly restricted, will approximate the glassy value, E_g, found by conducting relaxation measurement tests at different temperatures down to $-197°C$. Within the range of tests conducted, the fracture toughness values of both PC and N 6.6 apparently decrease with increase in loading rate. Fatigue crack growth in both materials is influenced by loading frequency and cyclic waveform. These variations may be related to the magnitude of the viscous energy loss and hence to the available energy for crack extension per cycle.

INTRODUCTION

An analysis of the failure strengths of polymeric materials may be conveniently expressed in terms of the time and temperature dependences of their mechanical properties. Irrespective of the nature of the external load, the response of viscoelastic solids will be conditioned by the time and temperature histories of the loading process imposed upon them.

This behavior is indicative of the complexity of the nature of such problems as the determination of the combined effects of, say, the relaxation moduli, the mechanical hysteresis energies, and the characteristics of the boundary loadings combined with a formulation of the time and temperature effects on each of these parameters. However, in recent years, as the use of polymeric solids has increased, the failure of such materials under static and cyclic loading conditions has been the subject of numerous investigations.

Monotonic fracture and fatigue studies have been carried out employing the concepts of linear elastic fracture mechanics (LEFM).[1,2] Successful applications of LEFM concepts to a number of thermoplastics have been demonstrated, although such applications are limited to a range of load, time and strain in each material, where approximate linear elastic behavior may be assumed.[3] Consequently, failures containing large amounts of plastic deformation are not discussed.

Linear elastic fracture mechanics has provided the designer with quantitative relationships between various parameters influencing the failure of bodies containing flaws. A parameter which represents the combined influence of applied stress, body geometry, and the current crack length is introduced into the field of failure analysis. This parameter, termed the stress intensity factor, K, has been tabulated for a wide range of stress fields in flawed bodies.[4] Achievement of a critical level of the stress intensity factor corresponds to catastrophic failure conditions.

In the present paper, two types of failure studies will be discussed in detail: (1) monotonic fracture studies using standard Charpy-type specimens, and (2) fatigue crack propagation studies using large rectangular plates containing an initial center notch. The stress intensity factors for these two geometries have been given previously.[6,7]

It was decided to study the behavior of an amorphous polymer (polycarbonate, PC) and a semicrystalline polymer (nylon N 6.6) under the above two loading conditions. Monotonic fractures at various strain rates and cyclic failure patterns under different loading frequencies were investigated. A model for the relationship between the cyclic rate of crack growth and appropriate parameters associated with LEFM principles, which has previously been described,[8,9] is now presented in terms of cyclic strain energy transformations.

For the computation of strain energy, the value of the time-dependent modulus of the material, $E(t)$, was used, but under cyclic loading conditions (and also under the conditions of fast dynamic fracture), it was assumed that the use of the glassy (short time) value, E_g, would be sufficiently accurate.

A convenient procedure for the determination of the E_g modulus for thermoplastic materials is described, and the results of such measurements on a number of thermoplastics (including PC and N 6.6) are presented.

MATERIALS, SPECIMEN DESIGN, AND TEST PROCEDURES

Materials used were polycarbonate of bisphenol A (Makrolon, Bayer) and nylon 6.6 (ICI Maranyl) which had been produced in the form of extruded sheets. Methods of preparation of the three-point-bend specimens (dimensions $10 \times 10 \times 55$ mm) used in the static fracture tests and the center-notched plates (dimensions $7 \times 12 \times {}^1/_4$ in.) used in the fatigue studies have been fully described in refs. 6 and 8, respectively. Descriptions of the associated equipment used may also be found in the above papers.

The dimensions of the rectangular-plate fatigue specimens were selected such that crack growth over a length of about 3 in. could be monitored. These tests, in tension, were carried out under the conditions of constant maximum levels of the stress intensity factor, the conditions being achieved by a gradual reduction in the level of applied load as the crack grew in length.

In the static fracture tests, a 10,000-lb-capacity Instron tensile testing machine was used for the low load rates and a 120-kg-capacity instrumented Charpy impact machine for the high load rates.

In the modulus measurements studies, unnotched bars of similar form to those used in the Charpy tests were employed, and the tests were performed at cross-head speeds of 0.2 in./min at cryogenic temperature (−197°C, obtained in a liquid nitrogen bath). Other modulus measurements, at higher temperatures, in the range between +20°C to −197°C, were carried out in baths filled with various mixtures of liquid nitrogen and petroleum ether. Test readings were recorded after the specimen had reached a steady temperature.

STATIC FRACTURE TESTS

The critical value of the stress intensity factor at the point of instability (catastrophic fracture) is termed the fracture toughness of the material, and in plane strain conditions it is designated by K_{1c}. This parameter is a measure of the energy required for crack extension in the material, and it is to be expected that, in rate-dependent materials (such as polymers), K_{1c} will change as a function of variations in the applied loading rate, environment, and temperature. It is anticipated that at lower temperatures and higher loading rates, the material will behave in an increasingly less ductile manner.

In order to analyze the behavior of PC and N 6.6 under such conditions, a range of tests was carried out to study the variation of fracture toughness value with change in loading rate, other conditions being constant.

Load Rate Effects

The influence of loading rate on K_{1c} was studied in three-point bend tests at room temperature (21°C) in air, over a range of cross-head speeds between 0.002 and 11800 in./min. The tests were performed in accordance with the ASTM standard,[7] and the fracture toughness values were calculated using the same procedure as described in ref. 6, i.e., by calculating the strain energy release rate and then converting it to the stress intensity factor term. The results of these tests have been presented in Figures 1 and 2, which clearly indicate the variations in K_{1c} as the loading rate is increased. The variations at relatively high cross-head speeds, however, do not seem to be as large as those suggested in ref. 10 for PC. The discrepancy may lie in basic differences in the materials tested and in testing techniques including the nature of the initial notch in the specimens.

Slow bend fractures in nylon 6.6 were in the form of very ductile failures; under these conditions, the LEFM concepts are not applicable and the quoted K_{1c} values may be described only as "apparent" toughnesses; the three tests carried out at this rate exhibited a large deviation in results (Fig. 2). However, as the loading rate was increased, a gradually decreasing ductility was observed, and the measurements of K_{1c} became more consistent.

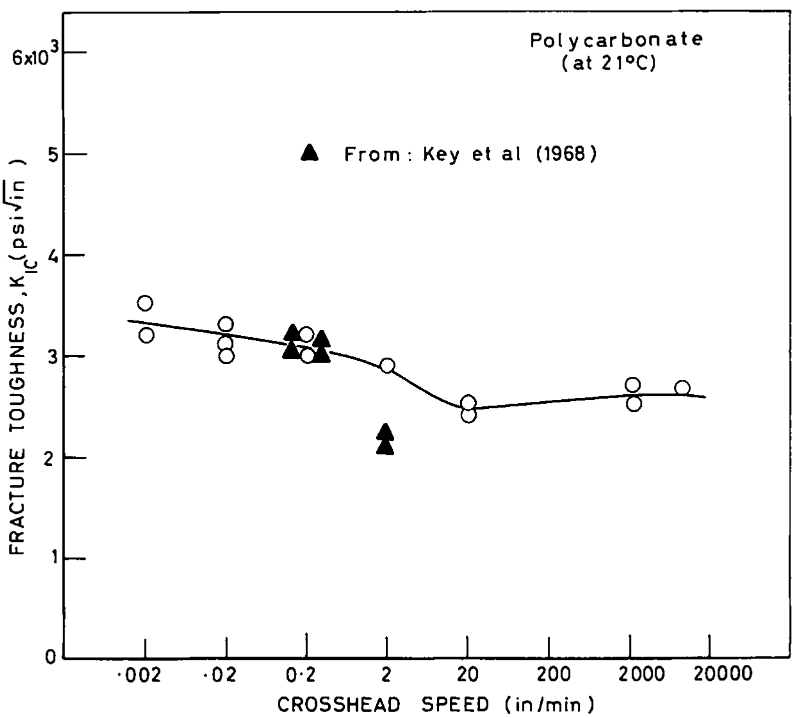

Fig. 1. Variation of fracture toughness of polycarbonate with cross-head speed at 21°C.

Fracture surface studies have also confirmed such a gradual transition in behavior pattern; see, for example, ref. 11 for studies on PC. Quasi-brittle materials usually exhibit a stable crack growth period (termed the slow growth) preceding the final unstable fracture. The length of this slow growth zone can be correlated with the material ductility, and if $K_{1\,init}$ is calculated at the point of initiation of slow growth rather than at the catastrophic failure point, a value different from and lower than K_{1c} will be obtained; in the presence of an increasing slow growth region, a decreasing value for $K_{1\,init}$ will result. Similar behavior may be observed with increasing test temperatures. Thus, as the degree of ductility is reduced through variations in testing conditions, smaller slow growth regions will be obtained, and the fracture toughness values calculated at the beginning and at the end of this region will become closer to each other.

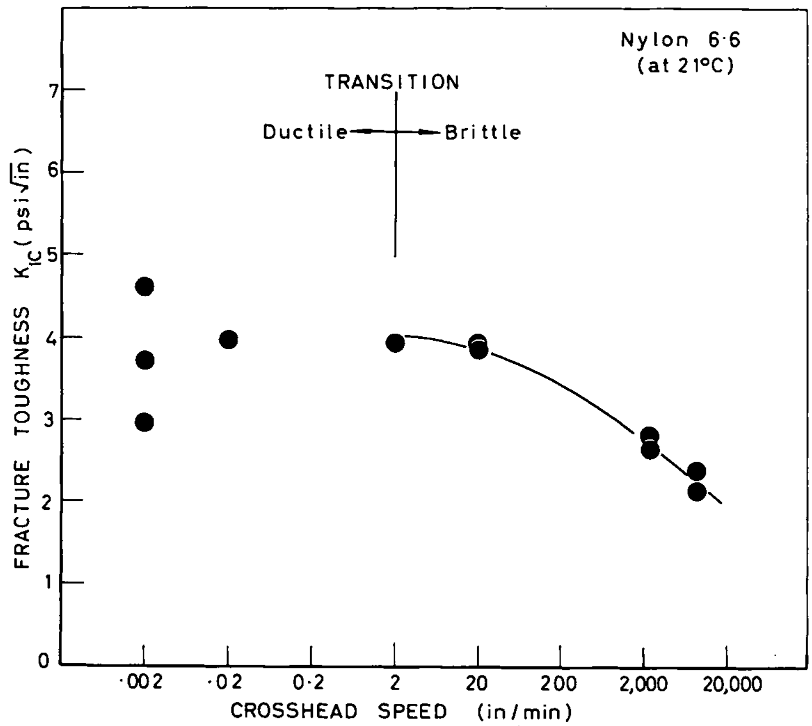

Fig. 2. Variation in fracture toughness of nylon 6.6 with cross-head speed at 21°C.

CYCLIC CRACK PROPAGATION

A comprehensive program of experimentation, results of which have in part been reported previously,[8,12,13] has shown that a crack propagation model of the following form may successfully be used in predicting the cyclic rate of crack growth:

$$\dot{a}_N = \beta\lambda^n \tag{3}$$

where $\lambda = K_{\max}^2 - K_{\min}^2$, and β and n are numerical parameters—to be determined empirically—dependent upon frequency of loading, material properties and other testing conditions.

The form of this crack growth model, enables the crack propagation rate to be expressed in terms of G, the strain energy release rate, in a load cycle.[3] The K^2 term is related to G by the following.[14]

$$G = \frac{K^2}{E}\zeta \tag{4}$$

where

$$\zeta = \begin{cases} 1 - \nu^2 & \text{plane strain} \\ 1 & \text{plane stress} \end{cases}$$

Equation (3) can thus be converted into the following form:

$$\dot{a}_N = \beta \left(\frac{E}{\zeta}\right)^n [G_{\max} - G_{\min}]^n \tag{5}$$

or, briefly,

$$\dot{a}_N = MA^n \tag{6}$$

where $A = G_{\max} - G_{\min}$.

In ref. 3, a review of instances of application of an energy-based model for crack propagation as well as its specific advantages have been presented.

Effect of Variation in Frequency on Cyclic Damage

Evidence indicating the strong frequency dependence of cyclic crack growth rate in polymeric materials has been produced.[8,9,13] It has been found that in general, under constant stress level conditions, an increase in frequency leads to a decrease in the cyclic rate of growth (Fig. 3). How-

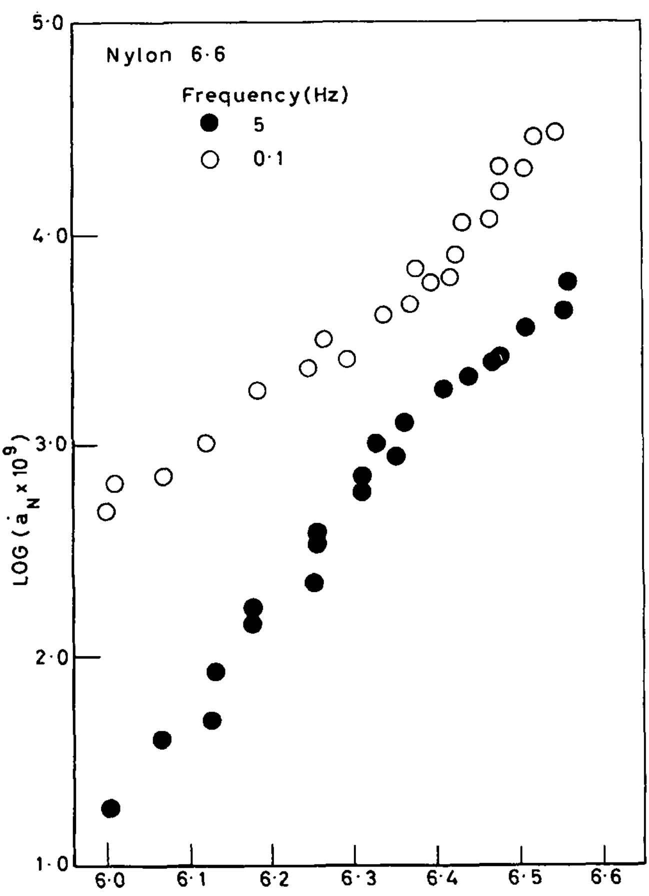

Fig. 3. Effect of frequency variation on cyclic crack growth rate in polycarbonate at 21°C.

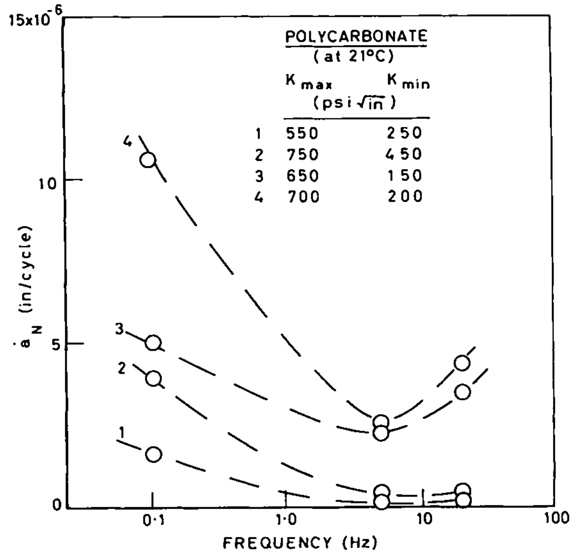

Fig. 4. Dependence of cyclic crack propagation rate in N 6.6 on parameter λ at 21°C.

ever, in terms of time to failure, a rise in frequency will result in a shorter total life.[13]

In the case of N 6.6, data from tests at three frequency levels, namely, 0.1, 5, and 20 Hz, have clearly indicated such a trend.[9] However, in the case of PC (Fig. 4), a deviation from this pattern of behavior is observed above a certain frequency level.[8] Other studies[2,11] have shown that it is possible to obtain an increase in the cyclic rate of crack propagation with increasing frequency. The following discussion may explain, at least in part, the above-mentioned frequency effects:

The Influence Attributed to the External Loading

During the fast cycling process (sharper sinusoidal wave form), the material is subjected to high loads for a shorter period of time per cycle. The crack growth rate is proportional to the size of the plastically deformed zone at the crack tip, as indicated by eq. (3).[3,8] The plastic zone may be considered as the crazed material at the crack tip through which the crack propagates. There is evidence of the time and strain dependence of craze growth in PMMA[15] and PC,[16] namely, that if the same strain is applied for a shorter period of time, the size of the crazed zone developed will be smaller. Observation of the fracture surfaces produced under various load rates also confirms such a hypothesis: as the loading rate is increased, the fracture surfaces become more smooth, indicating a more brittle type of behavior. In the discussion of variations, with load rate, of the energy required to create a unit fracture surface, however, one must also refer to the data relating the fracture toughness value to the crack speed, $\dot{a}_t$, in the material. Such data on PMMA, for example, shows a rise in K_{1c}, indicating a higher fracture toughness as $\dot{a}_t$ increases up to a certain level corresponding to crack speeds of a few inches per second. Subsequently,

K_{1c} remains constant, but the situation becomes more complicated as $\dot{a}_t$ continues to rise to very high values (1000 ft/sec), and K_{1c} then begins to rise again.[17] Clearly, if the crack growth rate in fatigue falls within the first section of this behavior pattern, one would expect an increasing resistance to crack growth as the loading rate (or frequency) is increased.

Data relating crack speed and fracture toughness are not, unfortunately, available yet for most materials although some have recently been produced on PC[18]; and hence detailed discussion of such results, at present, has to be limited. The fracture toughness measurements discussed above clearly indicate the gradual change in the specific fracture energy of PC and N 6.6. However, such data must be extended to a wider range of loading rates before a general discussion covering behavior of stable and accelerating cracks is possible.

The Influence of Intrinsic Viscoelastic Properties of the Material

From consideration of analytical solutions of viscoelastic fracture processes,[19] it is evident that the role of the viscoelastic absorbed energy is important in the availability of propagation energy required for crack extension. This loss energy, $\dot{W}_v$, is a function of E'' (the loss component of the complex modulus of a polymeric material, $E^* = E' + iE''$); $\dot{W}_v$ is related to E'' and E' in the following form[8]:

$$\dot{W}_v = \frac{\pi E''}{E''^2 + E'^2} [2S_m + S_a]^2 \text{ per cycle}$$

S_m and S_a being the mean level and amplitude of the applied stress cycle, respectively.

The loss modulus E'' is itself a function of frequency. At low frequencies, where large recoveries may be made in delayed elasticity, as well as at high frequencies when there is not sufficient time for molecular motions required for stress relaxation to take place, E'' has a small value. In the intermediate range, E'' has a peak at a level of frequency termed the resonance frequency. Clearly, depending on which side of the E'' peak a frequency range is chosen, the loss modulus may increase or decrease with increasing frequency levels. The specific trend of variations in E'' will thus influence $\dot{W}_v$, and consequently the magnitude of the energy available for crack propagation will be affected. Quantitative substantiation of the above is not possible at present as comprehensive relaxation data for PC and nylon 6.6 at room temperature are not yet available.

MEASUREMENT OF GLASSY MODULUS E_g

The value of E in eq. (4) is that pertaining to a linear elastic material. Hence, in an application of this equation to polymers, the value of E must correspond to the instantaneous modulus of the material which will be an appropriate choice for all dynamic loading conditions. It is assumed that the value of this modulus will approximate the short time (glassy value) of $E(t)$ which is defined as E_g in relaxation spectra. The value of E_g for most thermoplastics is not available in the literature, as the normal procedures of relaxation studies would require measurements of load and strain

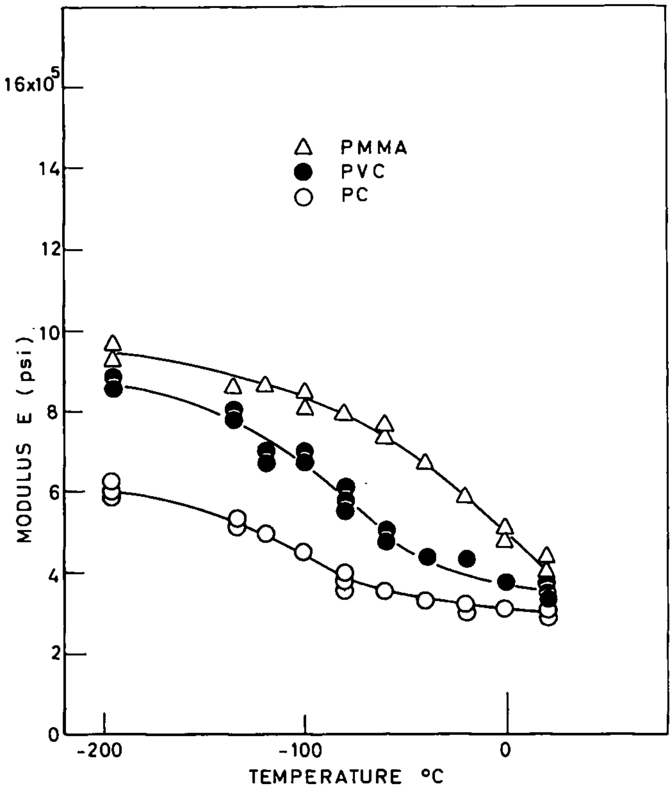

Fig. 5. Variation in time-dependent modulus of PMMA, PVC, and PC with temperature.

at exceedingly short times and hence are often impractical. In the present program of studies, it was proposed to make relaxation measurement tests in gradually decreasing environmental temperatures until the achievement of cryogenic temperature (i.e., from +21°C to −197°C). Data relating the value of the modulus to temperature, measured after approximately 1 min, at a constant cross-head speed of 0.2 in./min, for PC and N 6.6 are given in Figures 5 and 6. In these figures, similar data obtained on poly(vinyl chloride) (PVC), polyacetal (PA), and poly(methyl methacrylate) (PMMA) are also included so that a general comparison of the behavior of these different materials may be made. The figures show the general trends of behavior in the modulus/temperature relationships, and one may assume that an approximate plateau region has been achieved in the data on PVC, PMMA, and PC. However, for PA and, more prominently, for N 6.6 the plateau region is less pronounced; indeed, the modulus of N 6.6 seems to be still rising at a temperature of −197°C. The specific values of the moduli at −197°C and at 20°C are shown in Table I.

The room temperature data in the table were compared with relaxation data given in ref. 20, and a reasonably good degree of correspondence was observed. Also, for N 6.6, the value of the modulus obtained at −197°C was compared to data provided in ref. 21, and excellent correlation was obtained (Fig. 6).

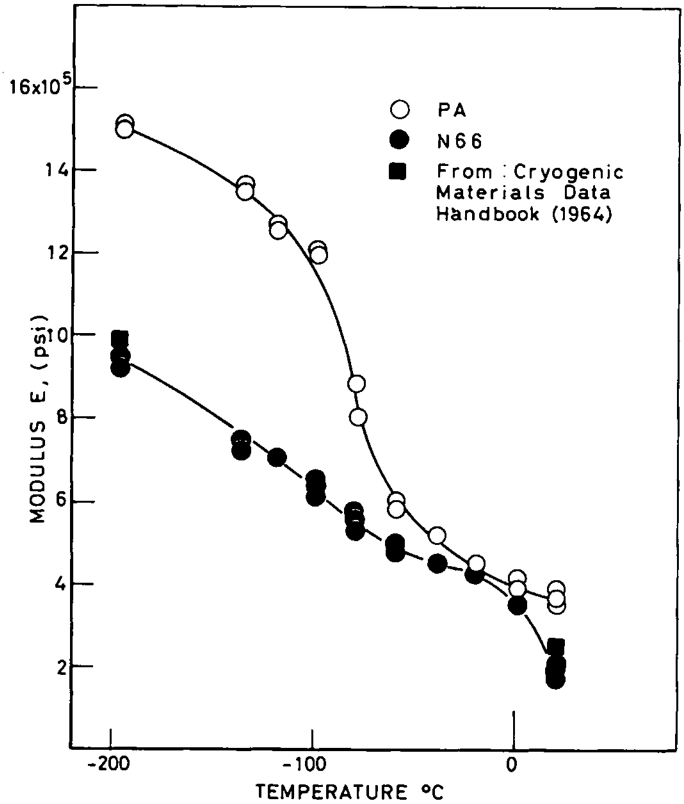

Fig. 6. Variation in time-dependent modulus of PA and N 6.6 with temperature.

DISCUSSION

Application of energy-based approaches to failure prediction and to analysis in engineering materials has been immensely fruitful. Studies of ductile-brittle fractures of metals, originally initiated by the work of Griffith,[22] fracture in elastic and viscoelastic polymers,[19,23] tear and fatigue of rubbers,[24] and fatigue of polyethylene[25] have indicated the possibility of development of a unified approach to the analysis of failure processes in general. Equation (3), which has been based on empirical data from

TABLE I
Moduli at 20°C and −197°C

Material	Modulus, psi	
	20°C	−197°C
PVC	364,000	800,000
PA	371,000	1,510,000
PC	310,000	622,000
N 6.6	180,000	936,000
PMMA	419,000	942,000

fatigue crack growth studies, is in the most suitable form to be converted into an energy-based criterion, eq. (6). Possibility of the application of this growth model to fatigue failure in metals has been explored with some degree of success in ref. 3. It is within the context of such an energy-based criterion for failure that one may be able to incorporate with facility the effects of various parameters associated with loading conditions. For instance, the influence of loading rate (or frequency) and temperature can be easily predicted, in the light of the empirical data and the discussions presented in previous sections, from the analysis of the effects of such external parameters on various energy components involved in a failure process, as formulated in ref. 19.

The specific role of strain rate in influencing the amount of energy available for crack extension may be discussed in terms of the magnitude of the hysteresis energy loss. In the monotonic fracture tests, this phenomenon is observed in the form of a reduction in total energy required for crack extension as the loading rate is increased, and thus the material is increasingly more restricted in exercising its viscous response. In cyclic loading, the effect of an increase in frequency manifests itself in terms of a reduction in the cyclic rate of crack growth which may, again, be partly explained in terms of the variation in viscous energy components.

From the above discussion the following conclusions may be drawn with regard to the effect of loading rate on the behavior of PC and N 6.6 in failure:

The fracture toughness values of both PC and N 6.6 decrease with an increase in loading rate, at least within the presented range of experimentation. Fatigue crack growth rate in both materials is influenced by variations in loading frequency and cyclic waveform, as these parameters have a considerable effect on the magnitude of the available energy for crack extension. The rate of growth can be directly related to this energy, eq. (6).

The trend in the variation of the cyclic rate of growth, $\dot{a}_N$, with frequency was found to be consistent throughout the range of tests on PMMA reported in ref. 13, i.e., a gradual decrease in $\dot{a}_N$ as frequency is raised. In the case of PC, however, this trend is not continuous throughout the whole range of test frequencies: $\dot{a}_N$ initially decreases with an increase in frequency, but subsequently increases again as the frequency is raised further (Fig. 3). It is suggested that this variation in behavior is possibly related to a change in frequency/E'' and crack speed/K_{1c} dependences.

Notation

A	$G_{max} - G_{min}$
$\dot{a}_N$	cyclic rate of crack propagation da/dN
$\dot{a}_t$	crack speed da/dt
$E(t)$	time-dependent modulus of polymers
E_g	glass modulus
E^*	complex modulus
E'	real part of E^* (storage modulus)

E''	imaginary part of E^* (loss modulus)
G	strain energy release rate
G_{max}	maximum level of G
G_{min}	minimum level of G
K_1	stress intensity factor in mode 1 crack opening
K_{max}	maximum level of K_1
K_{min}	minimum level of K_1
K_{1c}	critical value of K_1; corresponds to unstable fracture conditions
$K_{1\ init}$	stress intensity factor calculated at the point of initiation of slow (stable) growth
S_a	amplitude of applied stress cycle
S_m	mean level of applied stress cycle
M	numerical constant in crack growth equation
n	exponent in crack growth equation
$\dot{W}_v$	viscous energy absorbed per cycle of loading
β	numerical parameter in crack growth equation
λ	$K_{max}^2 - K_{min}^2$

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