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A model was previously proposed by Hertzberg, Skibo, and Manson to account for the sensitivity of fatigue crack growth rates to frequency. The model was based on the hypothesis that the mechanical β transition controls the fatigue deformation and on the use of time-temperature relationships to determine the match between the β transition frequency and temperature and the frequency and temperature of the test. An alternate model was recently proposed by Williams, who interpreted the frequency sensitivity in terms of

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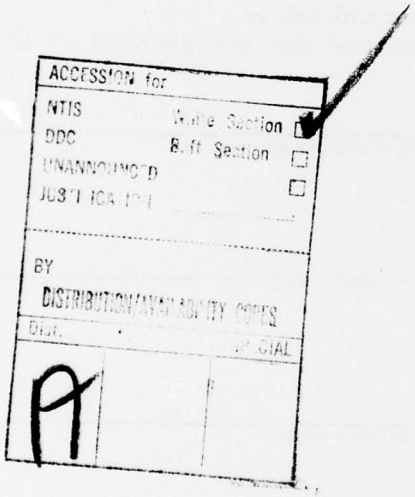
the sensitivity of fracture toughness (and Young's modulus) to frequency. In this communication, results of new experiments with nylon 66 are presented and discussed. Since Young's modulus was independent of test frequency for nylon 66, and for several other polymers, it was concluded that the authors' earlier hypothesis was not invalidated.



DISCUSSION OF "A MODEL OF FATIGUE CRACK GROWTH IN POLYMERS"

by

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J. K. Donald***

Williams has proposed an interesting model to describe fatigue crack propagation (FCP) in polymeric solids and to account for a number of experimental observations.[1] The purpose of this communication is to (1) examine the basic assumptions underlying the model, (2) compare recent data with values predicted from the model and (3) present alternate explanations for polymer fatigue behavior.

The first assumption is that upon unloading and reloading a craze at the crack tip, some of the craze ligaments become damaged, thereby reducing the craze stress σ_c. From this a two-stage craze zone is envisioned in which the newly formed craze material at the craze tip experiences a stress σ_c while the remaining part of the craze sustains a lower stress, $\alpha\sigma_c$. Use of this assumption leads to values of σ_c and $\alpha\sigma_c$ for several polymers in the ranges (325-720) MPa and (29-2,016) MPa, respectively (see Table II in References 1 and 2). In contrast, use of the Dugdale plastic strip formulation leads to values typically in the range of 40-80 MPa.[2] Since experimental values for crazing stresses are

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comparable for each of a number of polymers, it seems unlikely that the two-stage model can be generally valid, at least as currently stated. Instead we suggest that the bulk of the craze experiences a uniform stress, σ_c,[3] similar to that postulated in the Dugdale plastic strip model.

We certainly agree with Williams that cyclic-stress-induced weakening will take place in some of the fibrils that span the craze. However, we postulate that the load across the craze will be redistributed among the remaining unbroken craze fibrils.[3] These fibrils are then envisioned to stretch further, thereby leading to enhanced orientation hardening. With further cycling, additional fibrils are expected to break and the remaining ligaments would correspondingly become more highly oriented. We suggest further that a steady-state balance is struck between these two competitive processes--weakening through fibril fracture, and strengthening due to orientation hardening of the remaining fibrils--with consequent development of a constant stress, σ_c, across the craze. Quantitatively, this stress level should correspond to the product of load-bearing fibril strength σ_f and fibril volume fraction v_f so that $\sigma_c \approx \sigma_f v_f$. By way of confirmation, we find that computations involving fracture band widths, based on the Dugdale model, permit one to infer uniform craze stresses in several polymeric solids that, as mentioned above, are in good agreement with values reported in the literature by others.[2]

Proceeding further, we believe that the weight of evidence does not support Williams' explanation for the sensitivity of FCP rates to test frequency. He argues that polymer crack growth rates may be given by

$$\frac{da}{dn} = A \left(\frac{\Delta K}{K_c}\right)^n \qquad (1)$$

where $\frac{da}{dn}$ = fatigue crack growth rate

A,n = material property

ΔK = stress intensity factor range

K_c = fracture toughness

Indeed relationships of this form has been proposed by Wnuk[4] and supported by extensive experimental findings by several groups.[5-7] Using Equation 1, Williams then proposes that the sensitivity of FCP rate to frequency is controlled by the strain-rate sensitivity of K_c, the latter being given by

$$K_c^2 = E \cdot \sigma_{ys} \cdot COD \qquad (2)$$

where E = elastic modulus

σ_{ys} = yield strength

COD = crack opening displacement.

Earlier Williams[8] assumed that the yield strain ϵ_y could be estimated from Hooke's Law as

$$\epsilon_y = \frac{\sigma_{ys}}{E} \qquad (3)$$

so that $K_c = E\sqrt{COD \cdot \epsilon_y}$. The use of Eq. 3 in this situation seems questionable, especially since the modulus values used were defined at a strain of 3½ percent.

In any case, using yield strength and secant modulus data, along with Hooke's Law, Williams concluded that the yield strain was insensitive to strain rate and that the frequency sensitivity of K_c was due only to strain-rate-induced changes in E.

We disagree with this analysis in principle and on the basis of lack of correlation with both existing data and new test results reported below. First, by defining a <u>secant</u> modulus at a strain of 3½ percent and assuming

a true elastic limit at a much lower strain level, one would expect the secant modulus to be strongly sensitive to the yield strength (see Figure 1). It would not be surprising then to find the strong frequency sensitivity of E that was reported by Williams.[8] On the other hand, moduli of typical glassy polymers are stated to be relatively insensitive to strain-rate,[9] and even semi-crystalline polymers show relatively small time-dependent changes below T_g.[10] If da/dN is to be changed by an order of magnitude (as is the case with some polymers[11-13]) then E would have to change by a factor of 1.33 even if we assume the high value of 8 for the exponent n in Eq. 1 and assume that the frequency sensitivity of K_c is due only to strain-rate-induced changes in E.

Second, in order to examine directly the frequency dependence of E, we recently obtained compliance measurements from standard compact-tension samples, using the same geometry used to generate our FCP test results. These measurements were obtained under cyclic loading conditions at test frequencies ranging from 0.1 to 100 Hz. With the aid of data processing from an on-line PDP-8e computer, 20 to 100 individual data points (depending on test frequency) corresponding to specimen load P and associated crack opening displacement v were identified for each loading cycle. These values were used to establish a best-fit slope of the ΔP-Δv line. Between 2 and 40 such slopes were then used to define a final average slope. For a given crack length to specimen width ratio a/W and specimen thickness B, the modulus of elasticity of each sample could then be computed from the known compliance calibration relationship for the test specimen. It is of particular note that for most of the materials tested, the computed value of elastic modulus did not change to any significant degree (Table I). Since the body of the compact tension sample is predominantly elastic and

experiences very small strains, the results from these compliance measurements should reveal the material's *elastic* modulus rather than the *secant* modulus which Williams reported at a strain level of 3.5 percent. Because the FCP process and the associated crack tip stress intensity conditions are controlled by the elastic volume surrounding the small crack tip zone, the values of E reported here are considered to be more meaningful in assessing the FCP frequency dependence on $E(\dot{\epsilon})$. The reported E values are in some cases higher than those normally reported, based on conventional stress-strain data, but in general agreement with values reported based on dynamic mechanical data.[14] Higher values would not be surprising since the specimen strains are very low.

The results confirm the relative insensitivity to frequency anticipated for diverse polymers. Note in poly(vinyl chloride) (PVC), polystyrene (PS), and poly(phenylene oxide) (PPO), that the measured values of E changed by only about one percent for each of several decade changes in cyclic test frequency. Compare this lack of modulus-frequency sensitivity with the previously documented strong FCP frequency sensitivity for these materials.[11-13] Clearly, frequency-induced changes in E cannot as a general rule account for the large frequency sensitivity factors (FSF) reported. Instead, we maintain our view that FCP frequency sensitivity is largely controlled by a resonance condition between test machine frequency and the frequency of molecular segmental motions associated with the β damping peak.[11-13]

For the case of commercially prepared PMMA, we report an 8-10 percent change in E per decade change in test frequency which is considerably smaller than that reported by Williams, based on the 3.5 percent secant modulus.[6] This highlights the difference in E based upon different strain level reference points. When the $E(\dot{\epsilon})$ values are included in Eq. 1 along

with the material parameter n, the computed change in FCP rate per decade change in frequency is less than that actually measured. In fact, a similar $E(\dot{\epsilon})$ sensitivity was found in laboratory-cast PMMA and, yet no sensitivity of FCP to frequency was found in this material (Table I). Finally, tests were conducted on dry nylon 66 to establish both the FSF and $E(\dot{\epsilon})$. We found no change in E as a function of test frequency and no sensitivity of FCP to frequency (see Figure 2), as was also the case for nylon 66 containing an unknown amount of moisture.[11] The reason for the difference in frequency sensitivity of FCP rates between our results and those of El-Hakeem[15] (see reference 1, Figure 9) for dry nylon 66 is not clear at this time.

We certainly agree that the value of E to be used is the value at the test frequency selected. However while the secant modulus may show a strong strain rate sensitivity (presumably related to the strong strain-rate sensitivity of σ_{ys}), we conclude that the secant modulus is not the appropriate parameter for evaluating modulus effects on fatigue crack propagation behavior. Rather, we conclude that the modulus defined at small elastic strains is a more meaningful parameter for this purpose. However, the fact that $E(\dot{\epsilon})$ does not explain the FCP dependence on test frequency, supports our previously reported hypothesis that β-peak related segmental motions hold the key to the FCP frequency sensitivity in polymeric solids.[11-13]

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MATERIAL	0.1 Hz	1 Hz	10 Hz	100 Hz	n	$^{*}(E_{10Hz}/E_{1Hz})^{n}$	FSF**
Nylon 66 Dry	4160 MPa	4190 MPa	4190 MPa	4140 MPa	6.4	1	1
PVC ($\bar{M}_w$=1.4 x 10^5)	4590	4560	4600	4670	4.2	1.05	2.3
PVC ($\bar{M}_w$=2.3 x 10^5) + 6% DOP)	4290	4340	4410	4470	5.0	1.08	-
NORYL	3410	3490	3550	3580	4.9	1.09	2
ABS	3100	3170	3210	3210	3.7	1.05	1
PS	3930	3990	4120	4200	2.8	1.09	2.2
PMMA (laboratory cast $\bar{M}_w$=1.9 x 10^5)	5210	5740	6320	6980	9.3	2.44	1
PMMA (commercial $\bar{M}_w$=1.6 x 10^6)	3590	3960	4320		8.0	2.02	2.6
PC	3250	3270	3320		4.2	1.06	1

* Ratio of modulii at frequencies of 1 and 10 Hz

** M. D. Skibo, R. W. Hertzberg and J. A. Manson, Fracture 1977, Vol. 3, ICF 4, Waterloo, Canada, June 1977.

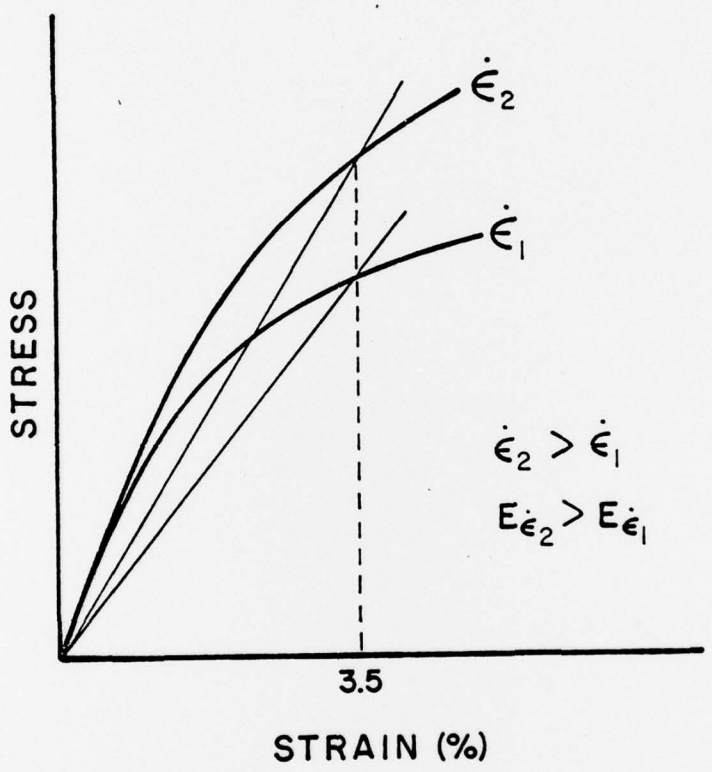
$\dot{\epsilon}_2$
$\dot{\epsilon}_1$
STRESS
$\dot{\epsilon}_2 > \dot{\epsilon}_1$
$E\dot{\epsilon}_2 > E\dot{\epsilon}_1$
3.5
STRAIN (%)

Figure 1

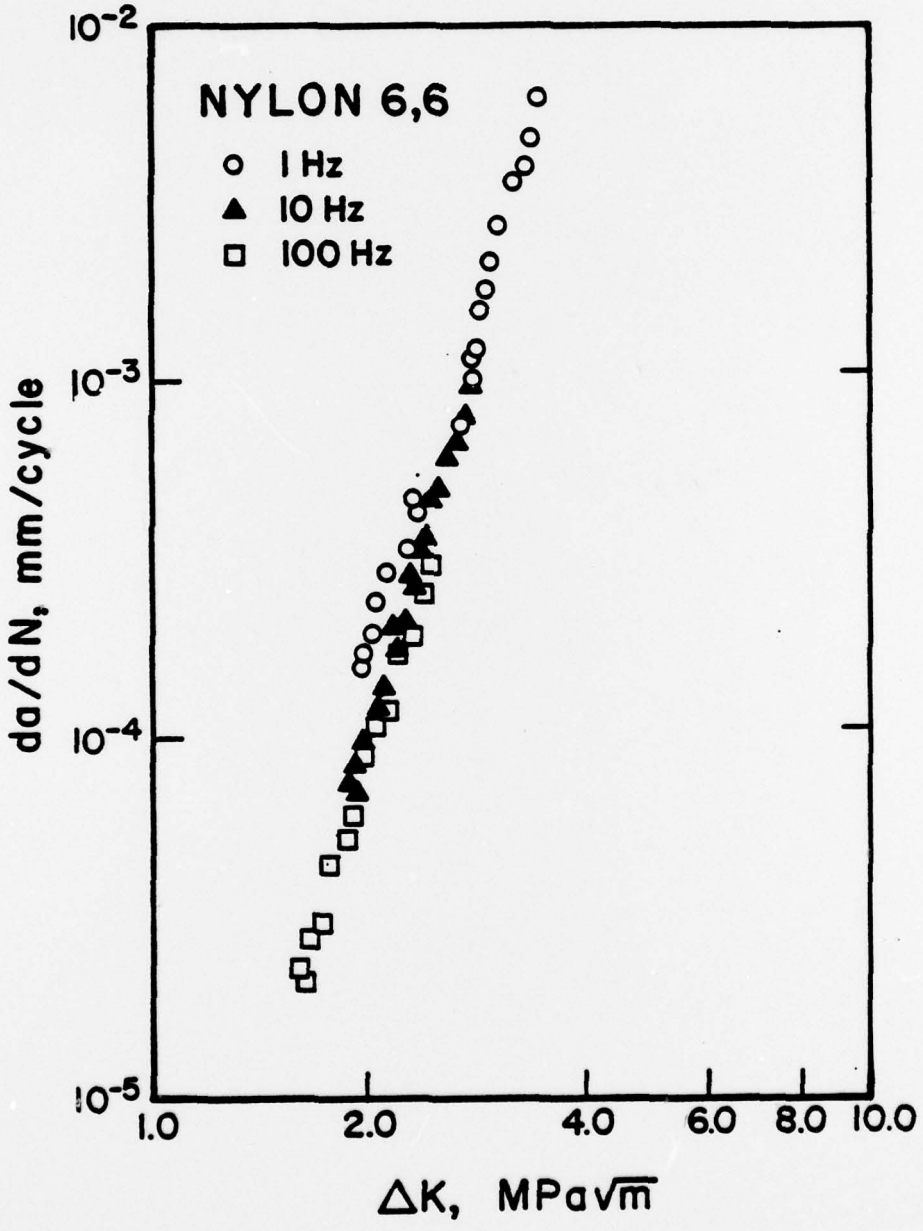
NYLON 6,6
1 Hz
10 Hz
100 Hz
10^{-2}
10^{-3}
10^{-4}
10^{-5}
da/dN, mm/cycle
1.0
2.0
4.0
6.0
8.0
10.0
ΔK, MPa√m

Figure 2

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