Comparative crack layer analysis of fatigue and creep crack propagation in high density polyethylene

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(Received 20 March 1989; revised 26 June 1989; accepted 1 July 1989)

This paper presents a comparative analysis of crack propagation in high density polyethylene (HDPE) under fatigue and creep loading conditions. This analysis is focused on the issue of mechanistic similarity between creep and fatigue failure in this material. Demonstration of such similarity is a crucial step in establishing the validity of fatigue as an accelerated laboratory test for long-term field failure under creep conditions. The Crack Layer approach is utilized as the analytical tool of the present investigation. It is demonstrated that, within the limits of the current analysis, mechanistic similarity with creep failure is preserved when fatigue is used as an accelerating agent for HDPE crack propagation.

(Keywords: fatigue; creep; crack propagation; crack layer; accelerated testing; specific enthalpy of damage)

INTRODUCTION

In two earlier papers fatigue and creep crack propagation in high density polyethylene (HDPE) have been analysed individually[1,2]. The Crack Layer theory[3] has been shown to describe fracture behaviour with a great deal of success in each case. At this stage a concurrent examination of HDPE fracture under these very different types of loading would seem in order.

Such a comparison would be instructive from an engineering as well as an academic viewpoint. From an academic perspective a favourable comparison between fatigue and creep fracture mechanisms would be another step toward a unified approach to polymer fracture. Furthermore, if it could be shown that the Crack Layer methodology produces the same Υ^* values for both creep and fatigue then rationalizations for HDPE crack resistance might begin to be considered on the molecular and morphological bases.

From an engineering standpoint, fatigue is very desirable as an accelerating agent for fracture testing. Accelerated test development has been the focus of a great deal of interest for many years. The validity of an accelerated test is contingent on its ability to produce failure which is mechanistically identical to that observed in the field. The mechanism of failure is the central issue in the Crack Layer approach. Hence, this approach is a potentially valuable tool in analysis and development of accelerated test procedures. Crack propagation in HDPE represents the first opportunity to apply the Crack Layer formalism in this manner. If mechanistic similarity could be established between fatigue and creep crack propagation then a lifetime of service under creep conditions in the field could potentially be simulated in a few days of laboratory testing. The practical implications of this are rather obvious.

All discussions in this paper will refer, unless otherwise indicated, to fatigue specimens tested under a mean stress of 7.94 MPa and creep specimens tested under a constant stress of 7.94 MPa. Preceding papers[1,2] should be consulted for details of sample preparation, experimental procedure and analysis.

DISCUSSION

Fracture propagation kinetics

In the earlier papers it was shown that both creep and fatigue HDPE samples exhibit 'brittle' crack tip and damage zone characteristics in the early stages of crack propagation. As the crack continues to grow, crack tip and damage zone features become, in both cases, progressively more 'ductile' in appearance.

Shown in *Figure 1* are the crack propagation rates for representative creep and fatigue specimens plotted as a function of crack length. It is quite evident that the behaviour is qualitatively identical in both cases. Three kinetic regions are identifiable under both creep and fatigue: initial acceleration, followed by a deceleration and then reacceleration to ultimate failure (in each case ultimate failure occurs via large scale yielding of the unbroken ligament). It is also noteworthy that the transition from deceleration to reacceleration occurs at essentially the same crack length in each case.

While the qualitative kinetic behaviour appears to be virtually identical under creep and fatigue, there are obvious quantitative differences. The magnitude of the deceleration is greater in the fatigue sample and, much

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0032-3861/90/030435-05

more significantly, the fatigue process appears to have a more pronounced effect on the 'ductile' crack propagation regime. It is certainly not surprising to find that crack propagation rates are significantly higher under fatigue conditions. In fact, total sample lifetime under fatigue is less than half of that under creep. What is interesting however, is the apparently preferential fatigue acceleration of the 'ductile' kinetic regime. Traditionally, one would expect creep loading to favour ductile behaviour, since it would seem to enhance polymer tendency to flow. Under fatigue the material gets at least a limited opportunity for recovery.

Micromechanisms

Figure 2 contains scanning electron micrographs of the damage zone in creep (*Figure 2a*) and in fatigue (*Figure 2b*). The similarities are again quite obvious. In both cases the basic unit of damage is the fibrillated void. There is however, an equally obvious difference in the geometry of these fibrillated voids. Voids formed under creep conditions appear to be somewhat serpentine in appearance, whereas their counterparts from the fatigue sample are rather straight. This is illustrated further in *Figure 3*, where the higher magnification micrographs of individual voids are presented.

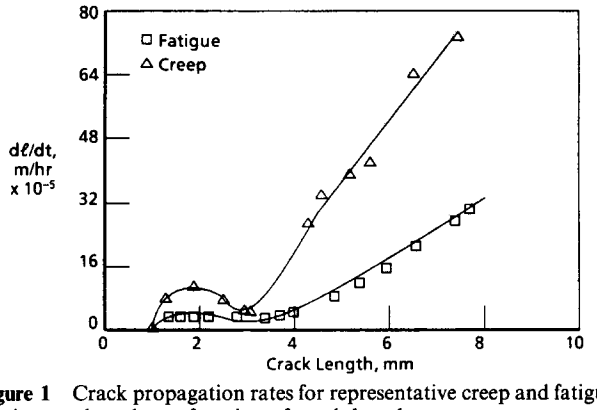

Figure 1 Crack propagation rates for representative creep and fatigue specimens plotted as a function of crack length

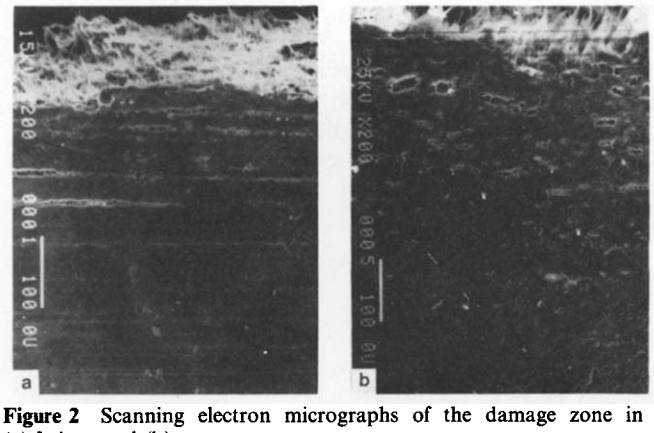

Figure 2 Scanning electron micrographs of the damage zone in (a) fatigue and (b) creep

It is difficult to provide an unequivocal rationale for this geometrical difference. One explanation which does come to mind involves the relative amounts of time spent on damage creation under creep and fatigue conditions. As mentioned earlier the total lifetime of the creep sample is approximately twice that of the fatigue sample (where the mean stress under fatigue is the same as the creep stress). The process of damage creation and evolution is intimately intertwined with sample failure, thus it is safe to say that fibrillated void creation under creep consumed (or was allotted) more time than under fatigue. Given more time it is plausible that these voids were able to follow more closely the lines of greatest weakness in the material. Perhaps void geometry under creep is the reflection of interspherulitic boundaries or the boundaries of some larger morphological assemblies.

As already mentioned, formation of fibrillated voids is not the only damage mechanism associated with HDPE fracture. Under both creep and fatigue pronounced thinning was observed within the active zone. Thinning profiles for representative creep and fatigue specimens are presented as a function of crack length in *Figure 4*. Also shown is the thinning profile obtained under fatigue at a mean stress of 3.75 MPa. Again, qualitatively the thinning profile obtained under creep agrees well with that from the fatigue sample. Thinning is present throughout the entire lifetime of the crack, but becomes more rapid as the 'ductile' crack propagation regime is entered ($l > 3.5$ mm).

It is quite apparent, on the other hand, that the *extent* of thinning at a given crack length within the 'ductile' regime is greater for the sample under fatigue. While somewhat unexpected, this observation may provide some insight into the comparative kinetic behaviour discussed earlier. It can be concluded from *Figure 4* that fatigue enhances the yielding damage mechanism. Yielding was earlier asserted to be the dominant damage mechanism in the 'ductile' crack propagation region under both creep and fatigue. Therefore, the previously

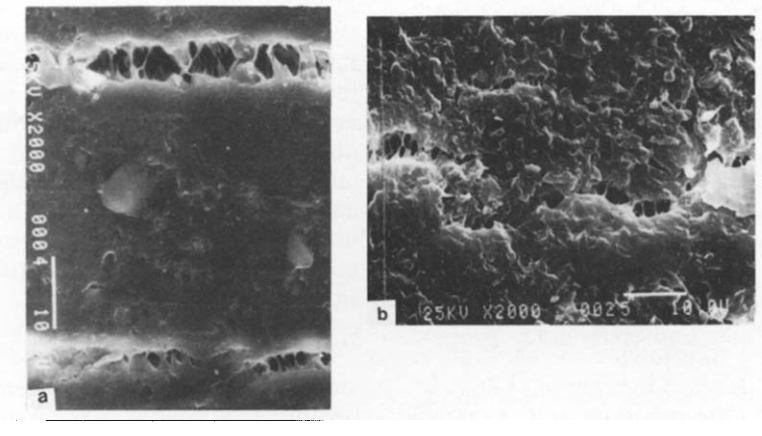

Figure 3 Scanning electron micrographs of voids in (a) fatigue and (b) creep

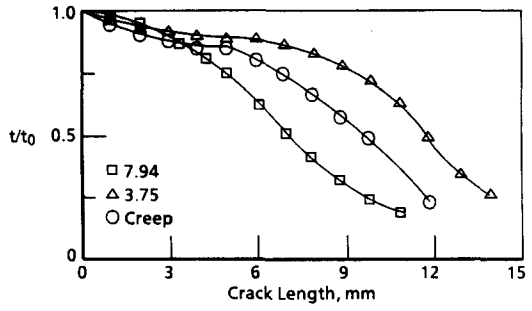

Figure 4 Thinning profiles for creep and fatigue specimens as a function of crack length

curious tendency of fatigue to accelerate 'ductile' crack propagation preferentially now appears to be sensible.

It is now reasonably clear that fatigue accelerates 'ductile' crack propagation in HDPE by accelerating the dominant damage mechanism in that kinetic region. The reason for this accelerated yielding under fatigue is not quite so apparent. A possible explanation may be advanced on the basis of the thinning profile obtained under fatigue with a mean stress of 3.75 MPa. It appears that the yielding mechanism is very sensitive to applied stress. At 3.75 MPa the transition to accelerated yielding occurs at a much longer crack length than at 7.94 MPa. This, of course, is as expected since the tendency of lower stresses to promote brittle rather than ductile fracture is well known. Hence, it is not unreasonable to propose that accelerated yielding under fatigue is caused by excursions to higher stress during each cycle. Though the *mean* stress of the fatigue test is exactly the same as the constant creep stress, during each fatigue cycle the sample experiences higher stresses for some fraction of the time. This exposure to higher stress is apparently sufficient to accelerate the stress-sensitive yielding process to the observed extent.

In addition to the effects of exposure to higher stress, the role of fatigue heating should be mentioned. By its very nature, fatigue loading results in heat accumulation within the sample. This could, in part, contribute to the 'ductile' acceleration.

Resistance moment and active zone evolution

Figure 5 contains the total resistance moment, R_t, plotted as a function of crack length for creep and fatigue loading conditions. The behaviour is virtually identical both qualitatively and quantitatively in each case, with the slight deviations likely due to inherent limitations of the measurements rather than phenomenological differences in the fracture process. Such agreement is a strong suggestion of the fundamental unity of HDPE fracture whether under creep or fatigue.

Even more striking evidence is presented in *Figures 6* and 7. Here, active zone expansion and distortion are plotted as functions of crack length for creep and fatigue specimens. These elementary active zone movements lie at the very heart of the fracture process and may, in some sense, be regarded as its phenomenological fingerprint. The apparent identity of these fluxes for both loading conditions strongly implies mechanistic similarity of HDPE failure under creep and fatigue. In addition, it suggests that the energy release rate J_1 can no longer be used as the sole similarity criterion in fracture processes where significant damage is present.

Generally, the energy release rate J_1, is used in lieu of test specific parameters such as crack length, applied stress or stress intensity factor, to characterize fracture. Similarity of the fracture process for a given material is established on the basis of the J_1 value[4,5]. The data in *Figures 6* and 7 seem to indicate that active zone evolution, expansion in particular, is the controlling agent in HDPE fracture. Thus, the M integral[3,6], which physically corresponds to the potential energy release rate due to active zone expansion, should be considered along with the J integral as a similarity criterion.

The authors would be remiss not to point out that much more detailed analysis would be necessary before the statement above could be transferred from the realm of intriguing conjecture to one of undeniable fact. The most crucial step in this analysis would be *in depth* microscopic characterization of damage density and its evolution with crack length. Active zone fluxes plotted in *Figures 6* and 7 are measures of global active zone geometry. Until it can be shown quantitatively that damage density and its behaviour are identical under both creep and fatigue, modification of fracture similarity criteria to include the M integral will have to remain a tantalizing possibility.

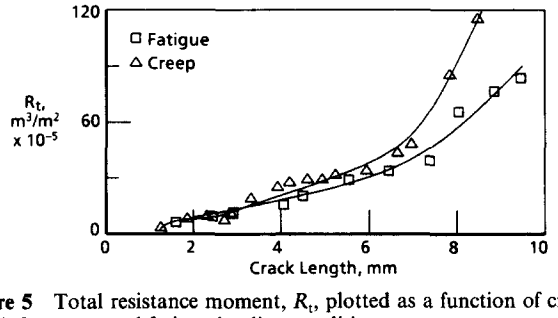

Figure 5 Total resistance moment, R_t, plotted as a function of crack length for creep and fatigue loading conditions

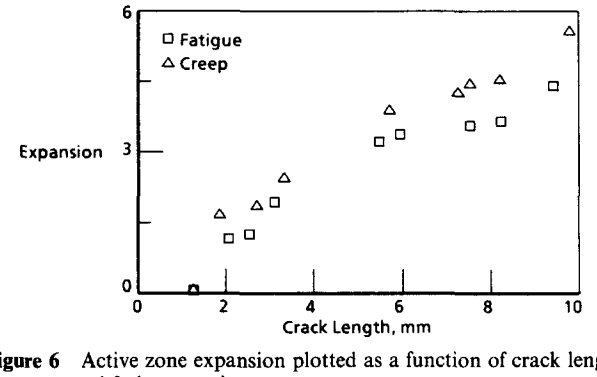

Figure 6 Active zone expansion plotted as a function of crack length for creep and fatigue specimens

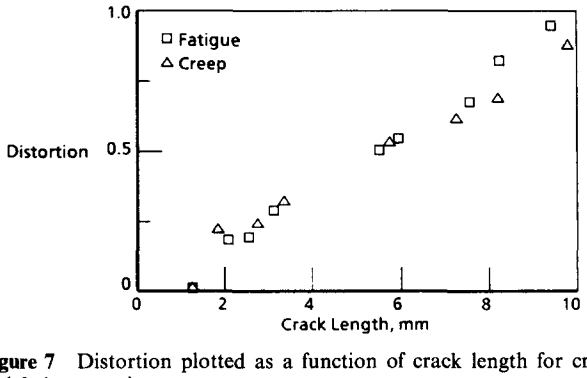

Figure 7 Distortion plotted as a function of crack length for creep and fatigue specimens

Energy dissipation rate

The rate of energy dissipation on damage creation within the active zone, $\dot{D}$, was defined as

$$\dot{D} = \beta \dot{W}_i$$

where $\dot{W}_i$ is the rate of irreversible work done on the active zone. In the fatigue case this quantity was evaluated from the hysteresis loops of the load–displacement curves recorded throughout the experiment and had units of energy per cycle. Under creep, the rate of irreversible work was computed as the product of the load and the rate of displacement due to irreversible processes[2] and had units of energy per hour. The results of these computations are presented in *Figures 8* and *9*. The evolution of the rate of irreversible work is qualitatively very similar for both creep and fatigue. In each case W_i increases slowly within the 'brittle' regime and much more rapidly as the highly dissipative 'ductile' damage mechanism becomes more dominant. It appears that the rate of irreversible work increases slightly more rapidly in the 'brittle' regime of the fatigue sample, but the difference is sufficiently small that it is difficult to determine whether it is mechanistically significant.

A direct quantitative comparison between the rates of irreversible work is not meaningful. Apart from the fact that they were evaluated differently, they are also more sensitive to the loading conditions than any other parameter. Clearly the amounts and rates of work done and of heat dissipated are going to be greater in the fatigue process. Yet it is not clear how great the difference should be and how to treat it quantitatively. Hence, at least at this point, the reader will have to be satisfied solely with the qualitative agreement.

Υ^* and β evaluation

The specific enthalpy of damage Υ^* and the characteristic dissipation coefficient β evaluated for creep and fatigue HDPE fracture are shown below. As already discussed, the 'brittle' and 'ductile' regimes crack propagation were analysed separately:

Creep. 'brittle': $\Upsilon^* = 1.3\,\text{cal g}^{-1}$, $\beta = 8.8 \times 10^{-3}$;
'ductile': $\Upsilon^* = 1.0\,\text{cal g}^{-1}$; $\beta = 1.4 \times 10^{-3}$.
Fatigue. 'brittle': $\Upsilon^* = 2.2\,\text{cal g}^{-1}$; $\beta = 0.004 \times 10^{-3}$;
'ductile': $\Upsilon^* = 1.7\,\text{cal g}^{-1}$, $\beta = 0.23 \times 10^{-3}$.

The agreement in Υ^* values is quite acceptable within the limits of experimental error and assumptions made along the computational trail. The disparity of β values is a consequence of the difference in loading conditions in the same fashion as the rate of irreversible work. Physically, β is the characteristic dissipation coefficient which serves as the proportionality constant between the rate of energy dissipation on damage creation in the active zone and the rate of irreversible work done on the active zone. It can be visualized as a measure of the efficiency with which work done on the material is converted into damage rather than dissipated as heat. It is evident from these results that during the fatigue process a much smaller portion of the total energy input into the system is utilized for damage formation. This is actually quite intuitive, since, as was pointed out in the previous section, fatigue is expected to result in higher heat dissipation.

In fairness, it should also be mentioned that β is subject to considerable error as it is evaluated from the slope of a least squares line and should be judged by its order of magnitude at best.

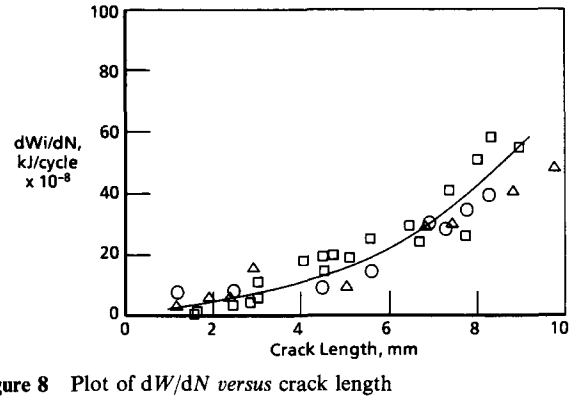

Figure 8 Plot of dW/dN *versus* crack length

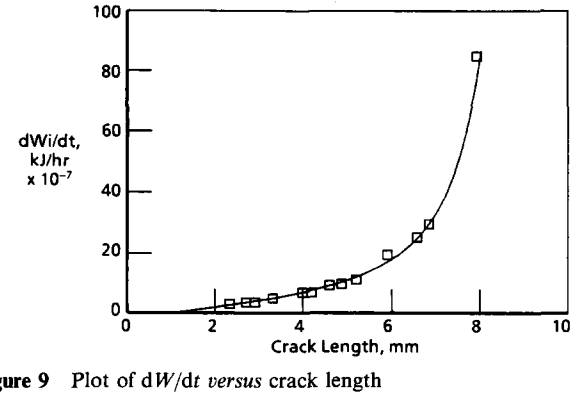

Figure 9 Plot of dW/dt *versus* crack length

Meaning and proper utilization of Υ^*

From the first paragraphs of this and the preceding[1,2] papers much attention has been paid to the specific enthalpy of damage Υ^*. It is this parameter which sets the Crack Layer theory apart from all others by enabling it to account for microstructural variations among materials. The actual use of Υ^* however, has yet to be discussed adequately in this or any preceding publications.

By its very definition, Υ^* is an enthalpy, a *thermodynamic* parameter which corresponds to the energy difference between virgin and damaged material. As such, it is most certainly *not* a *kinetic* parameter which could be associated directly with the rate of fracture. In other words, it is not the activation energy for the fracture process. Polymer fracture is a complex, multistep process and to find a single parameter exercising direct control over the rate of crack propagation one would need to identify and analyse the rate-limiting step of this process. To illustrate the point, consider polymers A and B and let the same damage mechanism be available to both. Furthermore, let A be more resistant to fracture. That is, let lower crack propagation rates be observed in A under comparable loading conditions. Crack Layer analysis may very well produce a lower Υ^* value for polymer A. This would not be at all contradictory. The energy difference between virgin and damaged states in A may be less than that in B, but the energy barrier for the interconversion of these states may be higher. The enthalpy term does not concern itself with time, it merely indicates the relative stability of the states. It is the activation energy, the energy barrier which must be overcome along the pathway from one thermodynamic state to the other, which determines the rate at which the more stable state will be achieved. Therefore, one

cannot use Υ^* values of different polymers as the sole measure of their resistance to crack propagation.

How, then, should Υ^* be utilized and how does the Crack Layer theory deal with fracture kinetics? The latter part of the question shall be addressed first. The kinetic parameter in the Crack Layer theory is the resistance moment. It is this quantity which measures the ability and tendency of a given material to resist crack propagation. It is this quantity, which in combination with Υ^*, can allow bona fide material lifetime predictions. If the constitutive equations for resistance moment evolution are known for polymers A and B and the specific enthalpies of damage for the respective mechanisms are available, then unambiguous predictions of fracture resistance for these materials can be made.

Constitutive equations for resistance moment evolution will be, no doubt, forthcoming at some future time. The present utility of Υ^* is still potentially significant, however. The specific enthalpy of damage is more than just an academic curiosity. It can be used today to help design more fracture-resistant materials. For a given material under specified loading conditions the resistance moment evolution and Υ^* can be evaluated experimentally. Once the damage mechanism has been ascertained, the crack propagation resistance of the material can be improved by lowering Υ^*. This can be achieved through either chemical or physical alteration of the material. As long as the mechanism of damage remains unchanged, a decrease in Υ^* will result in increased fracture resistance.

CONCLUSIONS

Within the discussed limits of the present analysis the following conclusions may be drawn: (1) mechanistic similarity of the HDPE fracture process under creep and fatigue has been established. Consequently, fatigue may be used as a means of accelerated testing for this material; (2) the Crack Layer methodology can be applied successfully, consistently and relatively simply to both creep and fatigue crack propagation in HDPE. In each case the analysis yields a specific enthalpy of damage on the order of $1\text{–}2\,\mathrm{cal\,g^{-1}}$. At this stage, the value is sufficiently reliable to justify the pursuit of molecular and morphological rationalizations of its magnitude.

ACKNOWLEDGEMENT

The authors are grateful to the Gas Research Institute for its support of this work under grant no. 5083-260-0940.

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