Fatigue of viscoelastic polymers: 1. Crack-growth characteristics

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Fatigue crack propagation studies conducted in reversed loading have revealed similar behaviour to that obtained in tensile-only cyclic loading. With low-density polyethylene a two-stage growth characteristic is obtained and the improvement in fatigue life-time that is produced by the introduction of a compressive component into the loading cycle appears to be associated with a shift in the stage I—stage II transition to coincide with a different point in the crack development. Tests conducted at different temperatures within the range 261—308K show the same general behaviour, but the stage I—stage II transition is temperature sensitive. Studies on a second low-density polyethylene, a high-density polyethylene and a plasticized poly(vinyl chloride) have been included for comparison and it is shown that significant differences exist in the fatigue behaviour of different polymers as well as much common ground.

INTRODUCTION

The growing importance of engineering polymers has stimulated considerable interest in their fatigue behaviour. Much of the work reported in the literature has been conducted on glassy polymers, usually poly(methyl methacrylate), polystyrene or polycarbonate[1,2]; considerable success has been achieved by the analysis of crack-growth data from these materials using the stress intensity factor approach[1,3–5]. Ductile polymers do not lend themselves so readily to this kind of analysis, but Andrews has shown that some of the ideas developed in the study of fracture in rubbers can be adopted for this class of materials[6]. The cyclic crack growth rate can be expressed in the form

$$\frac{dc}{dn} = B\mathscr{T}^n \qquad (1)$$

where $\mathscr{T}$ is a fracture mechanics parameter and B and n are material constants; B is found to be sensitive to the test conditions, particularly the temperature. Andrews and Walker found[6] that in low density polyethylene (LDPE), there were two distinct regions of crack propagation, labelled stages I and III, in which equation (1) is obeyed, separated by a less well-defined region, (stage II), in which the crack growth rate remained more or less constant. The crack growth characteristics presented by Andrews and Walker were derived from many tests, and each test provides a large number of data points since $\mathscr{T}$, which is a function of c, continually changes throughout a test. Good superposition of data from different tests was found to be possible for stages I and III, but the transition from stage I to stage II was found to depend systematically on the value of W_0, the strain energy density at maximum extension, which was varied from test to test. In addition to their study of the crack growth characteristics, Andrews and Walker were able to identify with considerable confidence that the intrinsic flaws from which fatigue crack growth occurred in unnotched spherulitic low-density polyethylene were the interspherulitic boundaries[6].

The purpose of the present study was to extend the work of Andrews and Walker, primarily to investigate the effect of a compressive component in the load programme. The previous study[6] was conducted in tension only but many service components suffer loadings in both senses. The machine on which the previous studies were conducted was extensively modified to enable operation of closely monitored and controlled tests under reversed loading conditions[7], and full advantage of the facilities offered by this machine were taken in the study reported here. Tests were conducted not just with symmetrical reversed loading, but with a series of different values for ϵ_c/ϵ_t, (ϵ_c = strain amplitude in compression; ϵ_t = strain amplitude in tension). The effect of temperature was also investigated.

Some comparative studies were performed in tension only on a non-spherulitic low-density polyethylene, (LDPE B), a high-density polyethylene (HDPE) and a plasticized poly(vinyl chloride), (PPVC).

EXPERIMENTAL

Materials and test-piece fabrication

LDPE was chosen for study for several reasons. Its molecular and crystal structures are fairly simple and unlikely to introduce into its fracture behaviour peculiarities which would frustrate comparison with other ductile polymers and the development of a general understanding of the fatigue characteristics of this class of material. There have been numerous studies of the structure, morphology, mechanical properties and deformation mechanisms of LDPE and although there remain many unanswered questions the

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Table 1

Designation	Source	Specifications				Form	Method of fabrication
		$\bar{M}_w$	$\bar{M}_n$	Density (kg/m^3)	*MFI*		
LDPE A	ICI: WJG 11			918	2.0	Granules	Compression-moulded into sheet, stamped into dumbbells and re-moulded in close-fitting moulds
LDPE B	G H Bloore Ltd			922		Extruded sheet 0.96 mm thick	Cut into parallel strips
HDPE 1	RAPRA	$\sim 10^5$	$\sim 10^4$	960	2.5	Granules	Compression-moulded into sheet 1.6 mm thick. Cut into parallel strips
PPVC	Stanley Smith & Co.	$\sim 10^5$	$\sim 4.5 \times 10^4$			Extruded sheet 1.08 mm thick	Cut into parallel strips

foundation of knowledge from which to work is substantially greater than for most other ductile polymers. LDPE A was selected since this was one of those used previously by Andrews and Walker[6]; one of the results of the earlier work was that there are marked differences in crack propagation behaviour between different grades of LDPE. Designations and sources of the polymers used are listed in *Table 1,* together with some relevant specifications.

Test-pieces were cut into strips from sheet or, in the case of LDPE A, compression moulded into parallel-sided dumbbells by a routine described elsewhere[7,8]. This involved a final remelting and slow-cooling stage under light pressure designed to give a good surface finish and an absence of internal moulding stresses or voiding. Specimens cut from sheet were 10 or 15 mm wide with a gauge length of 25 mm, and oriented parallel to the extrusion direction where applicable. Moulded specimens[7] were approximately 3 mm wide, 3 to 3.5 mm thick, and had a gauge length of 10 mm.

Specimens were aged at room temperature for at least 6 weeks prior to testing; this precaution is thought to be both necessary and adequate for this purpose[6,9].

Fatigue testing

Cyclic loading experiments were conducted at 0.25 Hz on a machine designed and built for this programme[7]. A special feature of the machine is that dynamic creep, accumulated during cyclic loading, can be sensed, measured and compensated for without interrupting the test. Strain-controlled testing is normally preferred since interrupted static creep failure could occur under stress-controlled conditions[2]. Even if this latter mode does not lead to catastrophic failure, modification to crack growth behaviour associated with cyclic creep cań occur, especially at low frequencies[10,11]. In conventional strain-controlled fatigue tests the movable crosshead is cycled between fixed limits. Consequently, if any cyclic creep occurs, producing a change in the unstressed dimensions of the test-piece, then the strain must also change. If the specimen suffers a net creep extension then the stress cycle will be displaced in the negative (compressive) sense. This effect is quite independent of and additional to any cyclic softening that occurs[12]. In the test arrangement employed in the current work the creep is sensed and compensated for so that the ratio of the amplitude of strain in tension to that in compression remains constant[7]. In an investigation into the influence of this ratio upon fracture behaviour such control is, of course, essential.

The specimen is enclosed in a Perspex cabinet which permits inspection of the specimen and measurement of the crack length using a travelling microscope. A controlled temperature is maintained inside the enclosure by means of an open circuit air flow which is heated or cooled as required[7,8].

METHOD OF ANALYSIS

Basic equations

In order to compare the experimental data with equation (1) a value for $\mathscr{T}$ must be derived. In the case of an edge crack in a semi-infinite sheet under a uniform applied tensile stress acting perpendicular to the crack axis[6,13].

$$\mathscr{T} = kcW_0 \tag{2}$$

where c is the crack length, W_0 is the input energy density at points remote from the crack and k is a function which varies slowly with applied strain. This is identical to the form derived for elastomers, for which k takes values between π at small strains and unity at large strains, and W_0 is the strain energy density at points remote from the crack[11,14]. The analysis of Andrews and Walker is based upon equation (2), leading, in combination with equation (1) to:

$$\frac{dc}{dN} = B(kcW_0)^n \tag{3}$$

This has been taken as the starting point in the current work. The values of W_0 and k have been determined by the methods described below for all of the materials employed and at each of the test temperatures used. Results are presented on log–log plots for comparison with equation (3). The original formulation was for a semi-infinite lamella and its application to finite-width specimens was also considered by Andrews and Walker[6]. No significant error is involved as long as the ratio of crack length to the specimen width, b, is less than 0.2, and they based their analysis on data recorded while this condition was fulfilled. To confine the data of the present study in similar fashion would have been unacceptably restrictive because the specimen geometry was changed to avoid buckling in compression, and a substantial number of data points correspond to values for c/b siderably higher than 0.2. A modification to the value of k should then strictly be applied, but this is unlikely to be significant except in the latter part of most of the experiments.

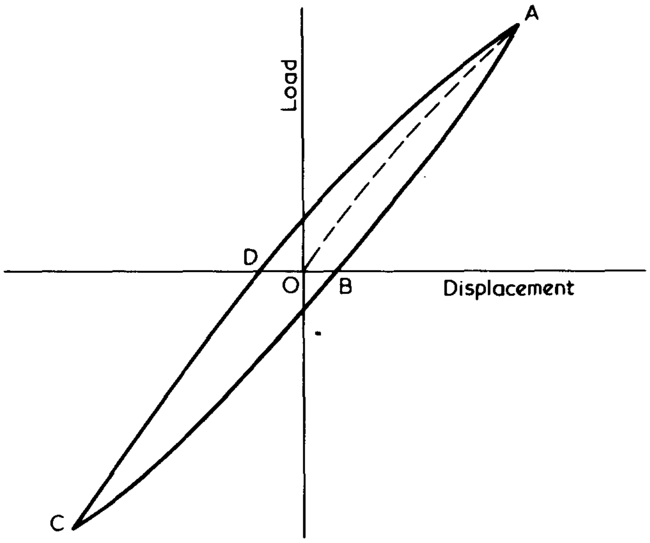

Figure 1 Stable load–deformation hysteresis loop for a polymer after several cycles of reversed loading (schematic). OA shows the initial loading

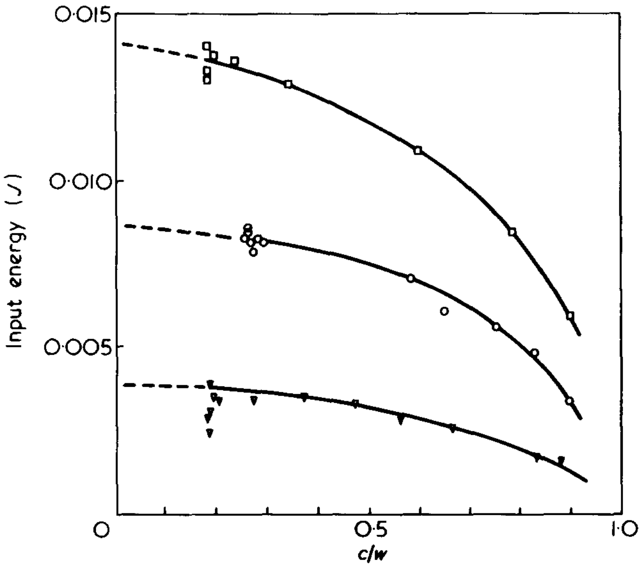

Figure 2 Change in input energy with crack length for fatigue cycling at constant strain. ϵ_t = 0.031 (□); 0.024 (○); 0.016 (▼)

Measurement of W_0

W_0 is defined as the work done on unit volume of material during monotonic loading in the absence of a crack; in reversed loading deformation, W_0 derives from the tensile part of the cycle and is given by the area under DA, *Figure 1.* Values of the total input energy measured during fatigue crack growth experiments on LDPE A at three strain amplitudes are shown plotted against crack length in *Figure 2.* On extrapolation to zero crack length, the intercept with the input energy axis can be used as an estimate of the value of W_0. An alternative way of obtaining W_0 was also employed; unnotched specimens were cycled until a steady-state hysteresis loop was attained, from which W_0 was estimated directly. The two methods were in good agreement as can be seen in *Figure 3.*

Measurement of k

Three independent methods of estimating k were used.

(i) Monotonic extension of notched specimens. For a specimen containing an edge crack the total energy change in the system due to propagation of the crack in simple extension is given by:

$$\frac{-\partial \mathscr{E}}{\partial A} = kcW_0$$

where A is the fracture area. For a specimen of thickness h this becomes:

$$\frac{-\partial \mathscr{E}}{\partial c} = 2hkcW_0 \tag{4}$$

A series of specimens with different notch lengths were tested to produce a set of load–deformation curves, shown schematically in *Figure 4.* The value of total energy loss, $\Delta\mathscr{E}(c_1)$, associated with a crack length c_1 is therefore given

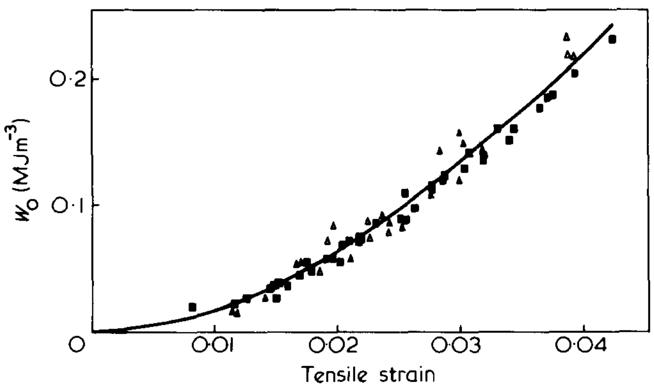

Figure 3 Dependence of W_0 on tensile strain for LDPE A at 293 K. △, Notched specimens; ■, unnotched specimens

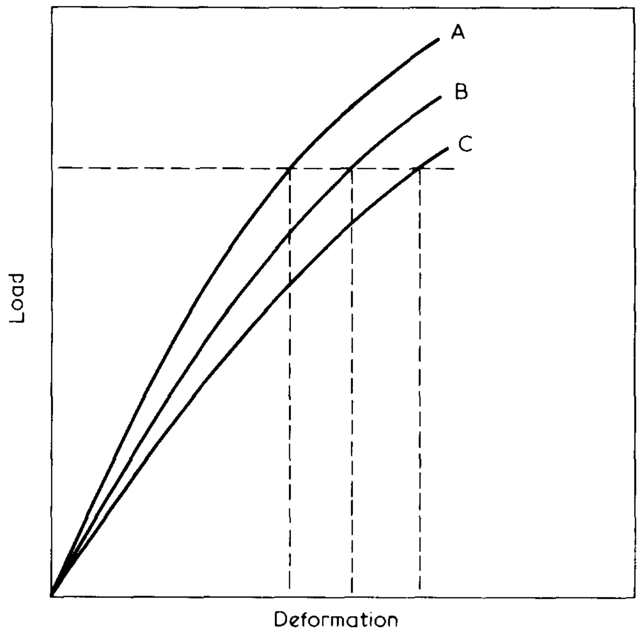

Figure 4 Schematic diagram of load-deformation curves for tensile specimens containing edge cracks of different lengths. A, c = 0; B, $c = c_1$; C, $c = c_1 + \Delta c$

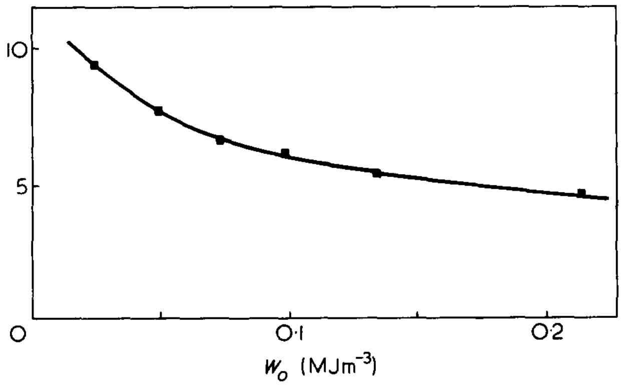

Figure 5 Dependence of k on W_0 obtained from monotonic extension tests

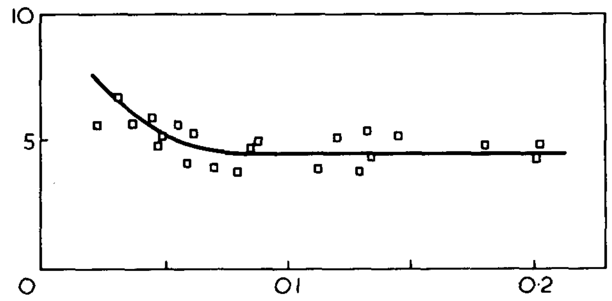

Figure 6 Dependence of k on W_0 obtained from the load–deformation curves generated during cyclic loading tests

by the area OAB and plots of $\Delta\mathscr{E}(c)$ *versus* c at constant W_0 were thus generated. The gradient, $\partial\mathscr{E}/\partial c$, was then measured at several values of c and plots of

$$\frac{-1}{2hW_0}\left(\frac{\partial\mathscr{E}}{\partial c}\right) \textit{ versus } c$$

constructed for different values of W_0. These plots were linear, as required by the theory, and gave values of k as shown in *Figure 5*.

(ii) Use of load–deformation curves generated during cyclic loading tests. On plotting input energy *versus* c^2 from *Figure 2* a straight line is obtained with a negative slope, $-s$, i.e.

$$\frac{-\mathrm{d}\mathscr{E}}{\mathrm{d}(c^2)} = \frac{-\mathrm{d}\mathscr{E}}{2c\mathrm{d}c} = -s \qquad (5)$$

On comparing equations (4) and (5) we can write:

$$k = \frac{s}{hW_0}$$

and values of k derived by this method are shown in *Figure 6*.

(iii) Crack growth rate superposition method. If the fatigue crack growth rate is plotted against cW_0 for several tests performed at different W_0 levels a series of parallel straight lines are obtained, as observed by Andrews and Walker. The method adopted by Andrews and Walker was to relate the shift parallel to the cW_0 axis required to achieve superposition to variations in the value of k. This method is comparative and does not yield an absolute value for k, but when applied to the data obtained in the current study was found to show a fairly similar W_0 dependence to that found by the other methods (*Figure 7*).

In the previous paper[6] the asymptotic value of k at large values of W_0 was chosen to be unity for convenience, but in the work presented here the value is taken to be 4.5 in agreement with the results obtained using methods (i) and (ii) above. As a consequence any k-dependent data found in ref 6 must be adjusted to take into account the absolute value of k before attempting comparison with the results of the present paper.

RESULTS

Crack-growth characteristics of LDPE A

Crack growth data for LDPE A tested at room temperature at three different values of ϵ_T, the total strain amplitude, $(=\epsilon_t + |\epsilon_c|)$, but with the same value of ϵ_t in each case is shown in *Figure 8*. Only stage I data is plotted, and remarkably good superposition is obtained. On comparison of these results with those obtained by Walker using the same material in tension only it is found that the gradient is smaller than that obtained in the earlier study (*Table 2*). If the asymptotic value for k of 4.5 is applied to Walker's results, replacing unity in the original computations, the results of both studies display a fair degree of overlap (*Figure 9*), particularly when it is noted to what extent the characteristics were found to be displaced from one material to another, (see *Figure 10* of ref 6). The other result evident on inspection of *Figure 9* is that stage I behaviour extends to much higher $\mathrm{d}c/\mathrm{d}N$, (and kcW_0).

Stage I behaviour terminates fairly abruptly and the crack growth characteristic becomes horizontal, ('stage II') with $\mathrm{d}c/\mathrm{d}N$ remaining constant while kcW_0 continues to increase, (*Figures 10* and *12*). This is the same result as that obtained in tensile-only tests by Andrews and Walker[6]. All specimens failed before the development of any stage III, though this may be a consequence of the specimen geometry.

Transition from stage I to stage II

If results obtained using the same ϵ_T but a different ϵ_t are compared the value of $\mathrm{d}c/\mathrm{d}N$, at which the transition from stage I to stage II occurs rises sharply with ϵ_t (or kW_0) (*Figure 10*). On recording the stage I–stage II transition points for all tests, irrespective of their ϵ_T values, on a $(\mathrm{d}c/\mathrm{d}N)_{\mathrm{I\text{-}II}}$ versus kW_0 plot this trend is maintained, *Figure 11*. The positive gradient is in direct conflict with the result obtained by Walker[6]. The scatter is fairly large, but is pro-

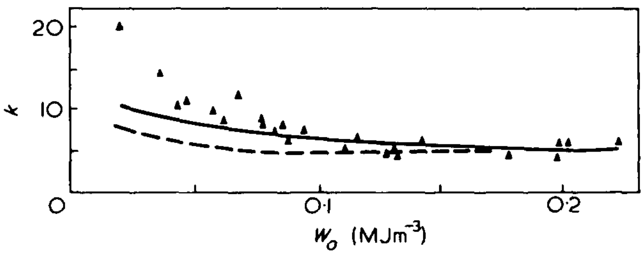

Figure 7 Dependence of k on W_0 from the superposition method (data points). The results of the other techniques are shown for comparison: the solid line is transposed from *Figure 5* and the broken line from *Figure 6*

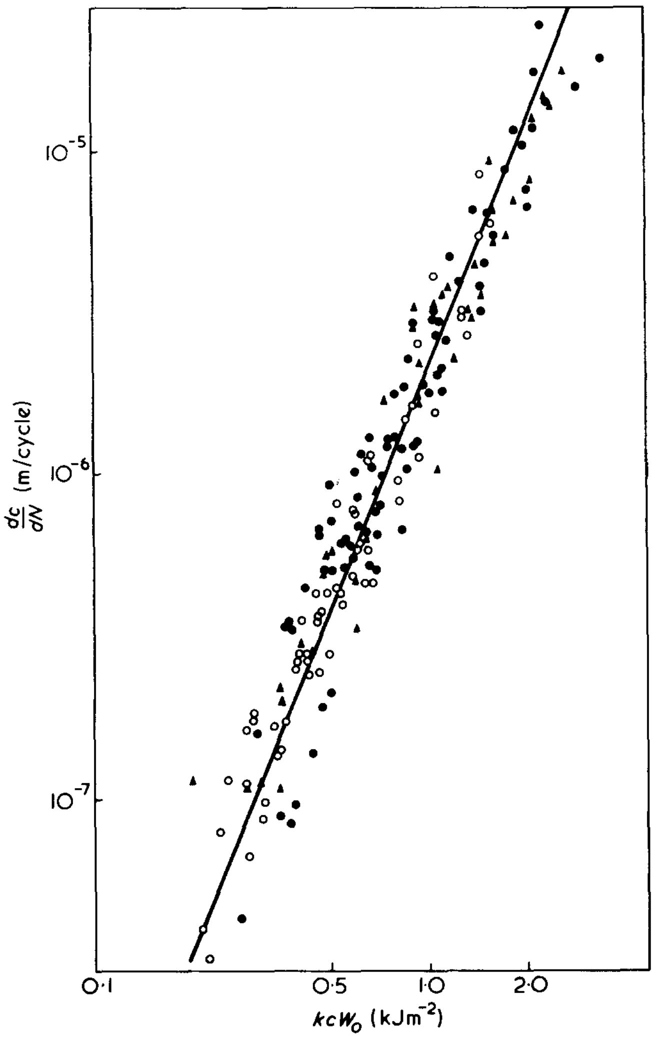

Figure 8 Stage I of the crack growth characteristic of LDPE A at 293K. ϵ_T = 0.066 (●); 0.046 (▲); 0.034 (○)

Table 2

Material	n		
	Stage I	Stage II	Stage III
LDPE A	2.5	0	–
LDPE B	3.0	0	2.5
HDPE	3.0	0.75	–
PPVC	3.5	1.0	–

bably accounted for by a secondary effect associated with the compressive component of the deformation cycle. This is illustrated in *Figure 12* where it is seen that for specimens tested at the same ϵ_t, (and therefore kW_0), the value of $(dc/dN)_{I-II}$ tends to decrease slightly as ϵ_c increases.

Effect of temperature

Tests conducted at five different temperatures were found to give the same slope for stage I of the log(dc/dN) versus log(kcW_0) plot, but the lines were displaced relative to one another in a systematic manner, meaning that the constant, B, in equation (1) must be temperature-dependent, in common with the findings of other investigations in which this point has been examined. The location of the stage I–stage II transition is also temperature-dependent, as can be seen in *Figure 11*.

Crack-growth data for LDPE B

Three-stage crack growth, in which $(dc/dN)_{I-II}$ again increased with kW_0, was displayed by these specimens (*Figures 13* and *14*). Note that this series of experiments was conducted in tension only, and that the test-piece thickness was much less than that used with LDPE A.

Crack-growth data for HDPE and PPVC

In both of these materials specimen failure occurred too soon after the onset of stage II to permit the detection of any stage III. The values for n for both stage I and stage II of both materials are shown in *Table 2*. The stage I–stage II transition was found to occur at a particular value of $\mathscr{T}$ in HDPE, while for PPVC the value of $(dc/dN)_{I-II}$ was found to increase with W_0 (*Figure 15*), as was observed with LDPE A and B.

DISCUSSION

The major objective of this work was to investigate the effect of introducing a compressive component into the deformation programme on fatigue crack propagation in low-density polyethylene. As a starting point the crack-growth characteristics were analysed with reference to the kcW_0 parameter derived from the tensile part of the programme only. Remarkably good superposition of crack-growth data from tests conducted with very different $\epsilon_t/\epsilon c$ ratios were obtained, so vindicating this procedure. It can therefore be concluded that the compressive deformation has at most a secondary effect with low-density poly-ethylene, (within the limits tested in the study described here).

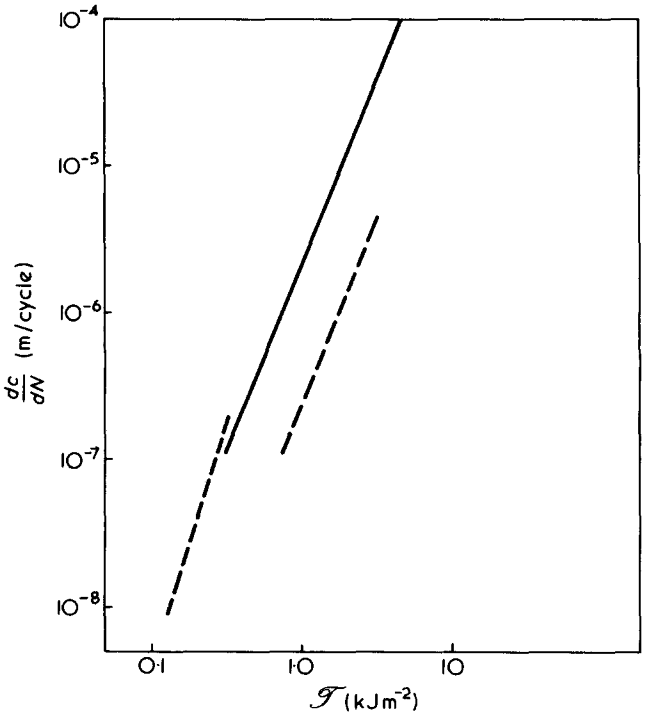

Figure 9 Comparison of fatigue crack growth characteristics obtained in the present work under reversed loading conditions with the results of Andrews and Walker obtained using tensile-only testing. Solid line: stage I results obtained in the present work. Broken lines: Andrews and Walker stage I (lower left) and stage III (upper right). [k = 4.5 (k = 1 originally[6])]

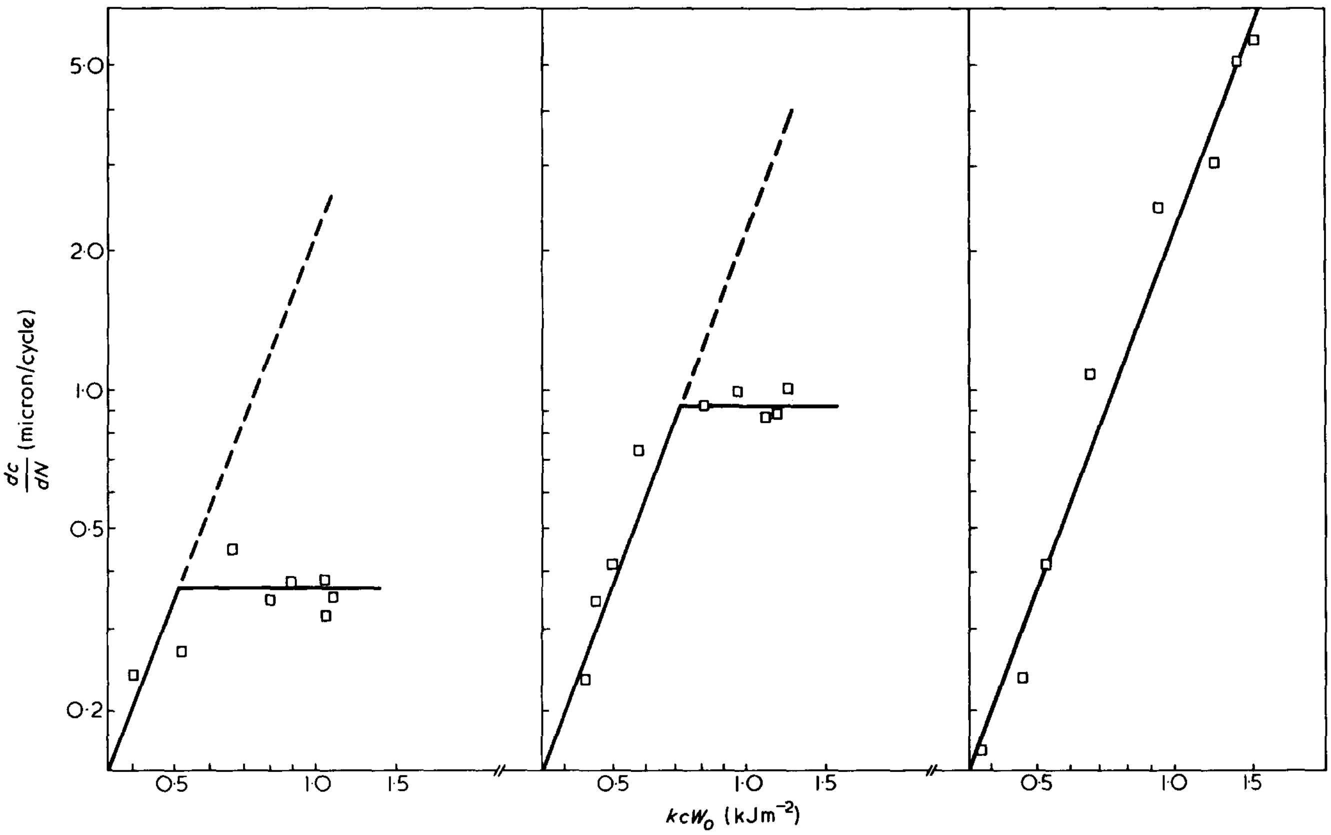

Figure 10 Crack growth characteristics for LDPE A at 293K. ϵ_T is constant but ϵ_t is changed to give (a) kW_0 = 0.43; (b) 0.53; (c) 0.60 MJ/m^3

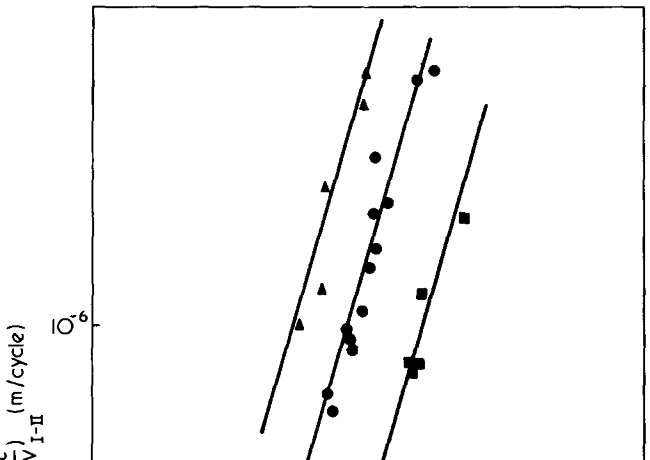

Figure 11 Dependence of the value of the cyclic crack growth rate at the transition from stage I to stage II crack growth, $(dc/dN)_{I-II}$, on kW_0 for LDPE A at three different temperatures: 285K (▲); 293K (●); 308K (■)

Another way of presenting data from these tests is to be found in a previous paper[7]. A reduced fatigue life was defined as 'that contribution to the total fatigue life accrued during propagation of the crack from 0.8 to 2.0 mm'; in this way discrepancies attached to irreproducibility of the razor notch depth and to the influence of the specimen boundary opposite to the notched face are removed. *Figure 7* of ref 7 shows that the fatigue life so defined is very sensitive to ϵ_t but relatively insensitive to ϵ_c, tending to increase very slowly with increasing $|\epsilon_c|$ (at fixed ϵ_t). At first sight this might appear to indicate that the log(dc/dN) *versus* log(kcW_0) plots are less sensitive to changes in ϵ_c, but the two results can be reconciled when the point at which the stage I–stage II transition takes place is taken into consideration. In stage II the crack-growth rate is more or less constant, while if the stage I–stage II transition were to be postponed and the stage I characteristic followed further, the rate would continue to get progressively higher as the crack length (and hence kcW_0) increases. Thus the sooner is the transition from stage I to stage II, the lower is the overall rate of crack growth. Even if a third stage of crack-growth were to be quickly promoted the overall rate would remain lower than if stage I were to be maintained throughout, since the stage III part of the characteristic always lies beneath the stage I portion.

On returning to the data upon which ref 7 is based it is found that the stage I–stage II transition frequently corresponds to a crack length between 0.8 and 2.0 mm and therefore affects the reduced fatigue life. An example is to be found in *Figure 10* of the present paper; the stage I–stage II transition occurred at a crack length of approximately 1.2 mm for the test at kW_0 = 0.43 MJm^{-3}, at $c \sim$ 1.3–1.4 mm for kW_0 = 0.53 MJm^{-3}, while the transition could not be located

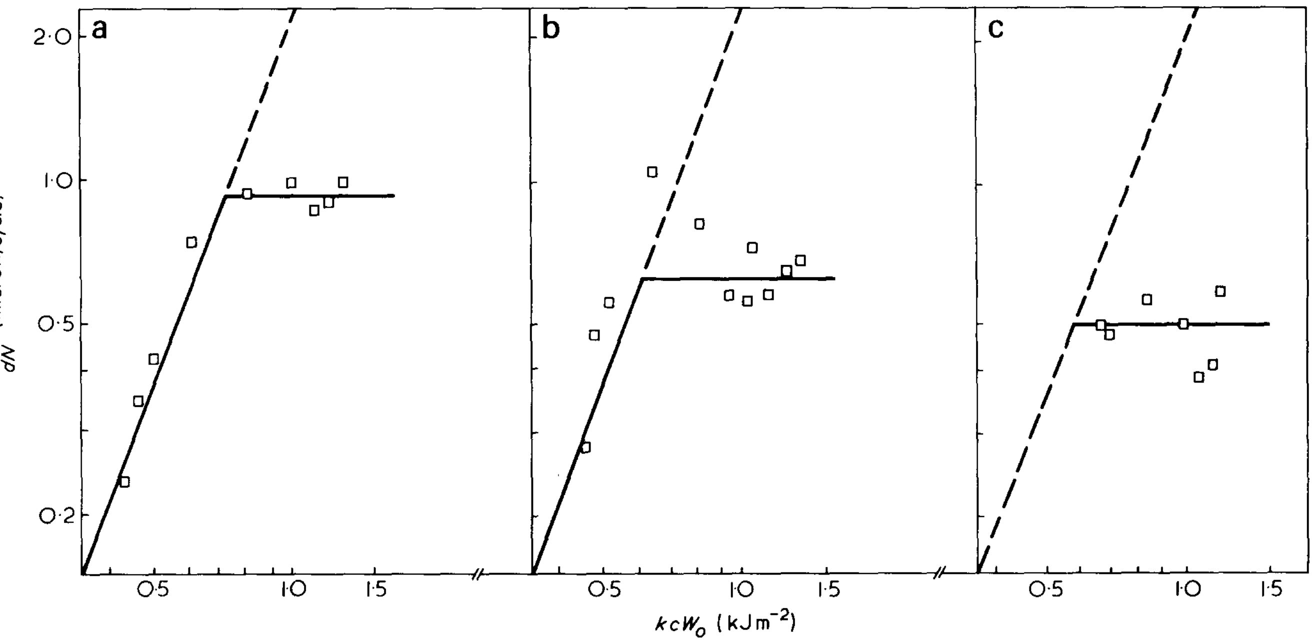

Figure 12 Crack growth characteristics for LDPE A at 293K showing the effect of changing ϵ_c. ϵ_t is held constant at 0.019 and (a) ϵ_c = 0.014; (b) 0.026. (c) 0.037

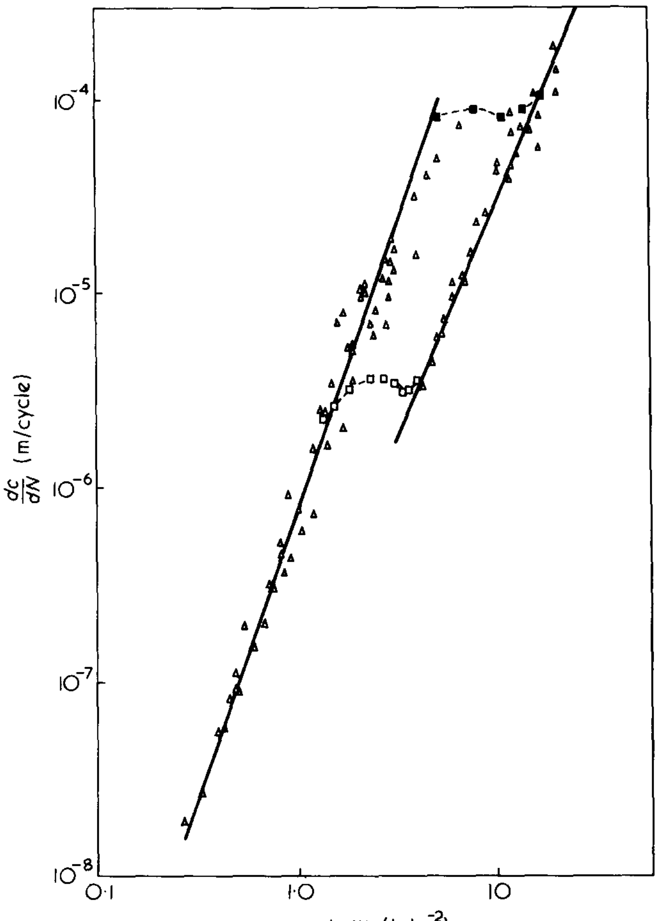

Figure 13 Three-stage crack growth characteristics obtained with LDPE B at 293K. The two stage II branches correspond to kW_0 = 2.41 (■) and 0.53 MJ/m^3 (□), respectively

with confidence in the test at kW_0 = 0.60 MJm^{-3}, stage I growth being maintained well past the point at which the crack reached 2.0 mm and continuing almost to the point of failure. These tests were all performed at the same values of ϵ_T so that a low value for kW_0 corresponds to a high value of $|\epsilon_c|$, and the early transition to stage II therefore accounts for the apparent fatigue life benefit provided by the compressive loading that was noted in ref 7. The same effect is displayed in *Figure 12* above, where the stage I–stage II transition occurs progressively earlier as $|\epsilon_c|$ increases for a set of tests at constant ϵ_t. Again the position at which the transition occurs corresponds to a crack length within the region in question ($0.8 < c < 2.0$ mm), and the different proportions of stage I and stage II growth explain the variation in the reduced fatigue life.

From the results presented in *Figure 11* it appears that there is a strong correlation between the stage I–stage II transition and the tensile component of the deformation programme. While the amplitude, ϵ_t, is likely to be important in itself it should also be recognized that for a series of tests conducted all at the same frequency, as with the programme described here, the strain rate will vary from test to test. Although the strain rate in tension is dependent upon the exact values of both ϵ_t and ϵ_c it will generally increase as ϵ_t increases and hence as kW_0 increases.

If the stage I–stage II transition were a strain-rate effect then it would be expected that any response promoted by increasing the strain-rate (and hence kW_0), could be produced instead by decreasing the temperature. In *Figure 11* the positive slope of each of the lines indicates that $(dc/dN)_{I-II}$ increases with kW_0, while by following any line drawn at constant kW_0 it can be seen that an increase in $(dc/dN)_{I-II}$ can alternatively be achieved by decreasing the temperature, so supporting the strain-rate effect hypothesis. A small strain rate therefore favours earlier promotion of ductile behaviour (stage II) than a higher strain rate, as is consistent with normal experience

In considering the influence of the compressive component of the deformation cycle it must first be noted that for a particular value of ϵ_t the time spent in tension per cycle decreases as $|\epsilon_c|$ increases, while the strain rate also increases as $|\epsilon_c|$ increases. Both effects might be expected to favour brittle behaviour and thus to cause postponement of the stage I–stage II transition. This is contrary to observation

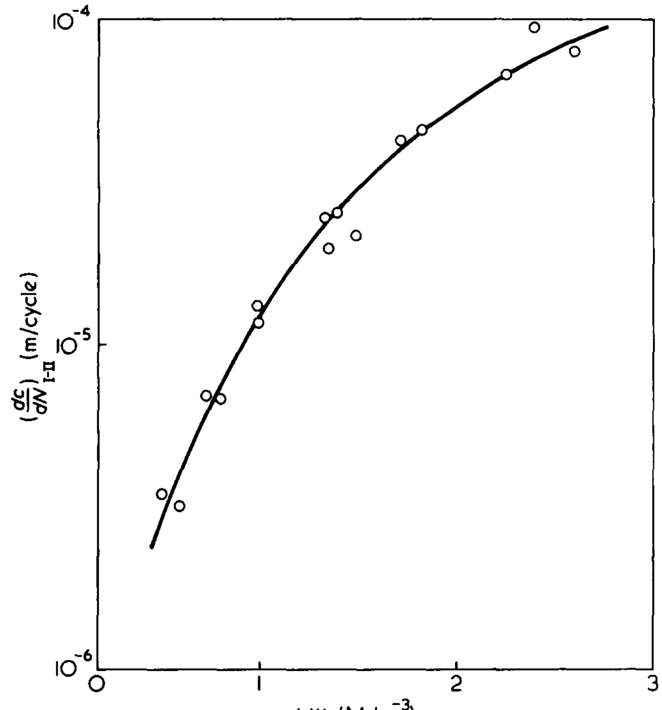

Figure 14 Dependence of $(dc/dN)_{I-II}$ on kW_0 for LDPE B at 293K

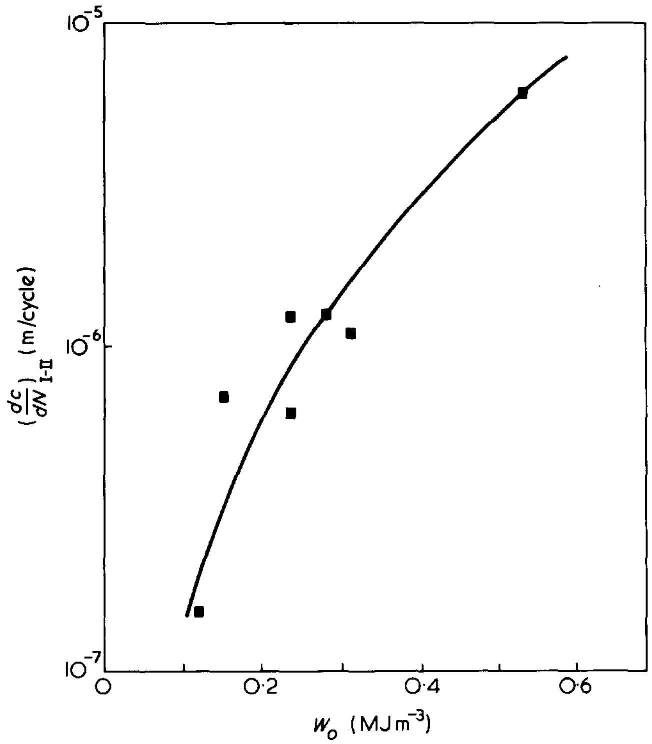

Figure 15 Dependence of $(dc/dN)_{I-II}$ on W_0 for PPVC at 293K

and although the trend in the opposite direction is not dominant it does seem to suggest that some structural change must occur during the compressive deformation that conditions the material for the transition.

The tests performed on LDPE B in tension provided crack growth characteristics similar to those for LDPE A and other low-density polyethylenes investigated by Andrews and Walker[6]. The transition from stage I to stage II followed the same trend as that observed in the present work with LDPE A, with $(dc/dN)_{I-II}$ increasing with increasing kW_0. This is the opposite of the result obtained by Andrews and Walker with a different grade of low-density polyethylene. At this point in time we have no explanation for this apparent conflict between the earlier work and the current work but we do not regard it of sufficient importance to reverse our general conclusions.

With both HDPE and PPVC the crack growth rate was found to rise in stage II with increasing kcW_0, and approximated to equation (1) with values of n much less than those found to apply to stage I (see *Table 2*). These tests were also conducted in tension only. It was again found that $(dc/dN)_{I-II}$ increased with kcW_0 for PPVC, but with HDPE, within the limits of testing applied here, the transition appeared to occur at approximately the same value of kcW_0, (and hence of dc/dN), in all cases.

A discussion of the microstructural aspects of fracture will be presented in a further paper[15] dealing with a fractographic study conducted in parallel with the work described above, using optical and scanning electron microscopy.

CONCLUSIONS

(i) In low-density polyethylene the addition of a compressive component to the deformation programme makes little difference to the general behaviour of fatigue crack propagation, a growth characteristic being obtained very similar to that found in tension. The absence of stage III may be a consequence of the test-piece geometry, but as it is the least important part of the failure process it has not as yet received much attention.

(ii) The major effect produced by altering the intensity of the compressive component appears to be a variation in the point at which the transition from stage I to stage II occurs.

(iii) The temperature dependence of the stage I–stage II transition is consistent with it being (tensile) strain-rate dependent, though the exact nature of the effect caused by the compressive component remains to be established.

(iv) Detailed differences exist in the crack-growth characteristics of different materials. The non-appearance of a 'stage III' in HDPE and PPVC may be due to the completion of failure before a stage III can develop, but another possibility is that the transition from stage I to stage III occurs in these materials without an intermediate 'stage II'. The fact that the second stage showed a non-zero (positive), value for the index n is consistent with this idea. Clearly more work is needed in this area.

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