FATIGUE DAMAGE ACCUMULATION PROCESSES IN ENGINEERING METALS AND PLASTICS

R. W. Hertzberg

Department of Metallurgy & Materials Engineering, Lehigh University, Bethlehem, PA 18015, USA

ABSTRACT

Fatigue fracture processes in engineering plastics are examined with particular attention given to the kinetics of discontinuous growth band formation. The width of these fracture bands (broken crazes) corresponds to the plastic zone size as computed from the Dugdale plastic strip model. Recent studies have shown that the early stages of craze damage mainly involve craze lengthening. Later stages of damage are believed to involve craze thickening by fibril stretching rather than drawing fresh material into the craze from the surrounding bulk. A different type of discontinuous growth band--one containing a pair of shear bands--has been reported and the presence of the shear bands related to the ratio of applied stress to yield strength levels.

KEYWORDS

fatigue, plastics, electron fractography, striations, discontinuous growth bands

INTRODUCTION

Increasing attention has been given to the evaluation of fatigue fracture processes in engineering metal alloys and structural plastics. Important studies have been conducted to determine the role of many important parameters such as: various stress and/or strain variables; pre-existing defects of different sizes and shapes; environmental factors such as test temperature, cyclic frequency, wave form and environment; and a multitude of material variables. From the time period roughly corresponding to the first International Conference on Fracture, a new approach to fatigue analysis--the study of fatigue crack propagation (FCP) processes--has received careful and extensive attention. Important relationships have been identified such as the Paris relationship which relates FCP rates to the prevailing stress intensity factor conditions at the tip of a growing crack (Paris, 1964; Paris and Erdogan, 1963).

$$da/dN = A\Delta K^{m} \quad (1)$$

where da/dN = fatigue crack growth rate
A,m = material and environmental test variables
ΔK = stress intensity factor range

With such information, the remaining cyclic life of a flawed component could then be estimated by integration of Eq. 1, thereby leading to the development of damage tolerant design procedures. More recently, studies have shown that the power-law relationship of Eq. 1 represents an oversimplification of true material behavior. Of particular note, it has been shown that crack growth rates in metals and engineering plastics become vanishingly small at very low ΔK levels (Fig. 1). These "threshold" ΔK values (ΔK_{th}) are analogous to the endurance limits of materials as determined from cyclic loading experiments on unnotched samples.

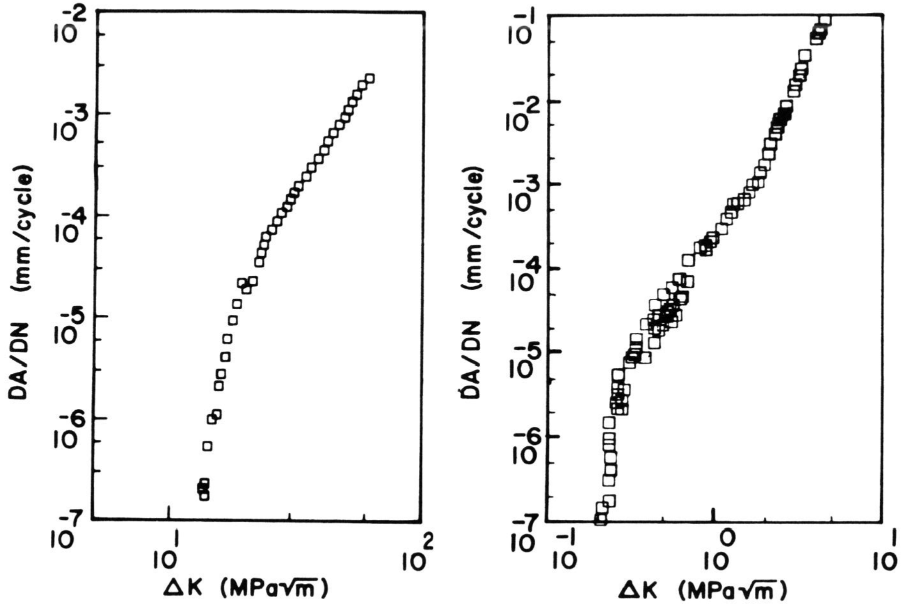

Fig. 1. Fatigue crack growth rates versus ΔK revealing log-log linear region at higher ΔK levels and threshold condition at low ΔK levels. a) HIP'd Astroloy (R=0.5); b) ABS plastic (R=0.1).

One research area that has engaged the present author for many years is the study of fatigue fracture surface micromorphology in metals and plastics. It is the objective of this paper to review some earlier studies and discuss recent findings that add to our understanding of polymer fatigue fracture. Specific attention will be given to the discontinuous cracking process that has been observed in several engineering plastics.

FATIGUE FRACTURE PROCESSES

In recent years, increased attention has been given to the analysis of fatigue processes in engineering plastics (Hertzberg and Manson, 1980). This has resulted from the use of lighter weight materials such as plastics and their composites in structural components to reduce energy costs in automobiles, boats and planes, etc. Phenomenological studies have shown that the crack growth rates in numerous engineering plastics and their composites (both particulate-and fiber-filled) can be correlated with ΔK (recall Eq. 1). Furthermore, both metals and polymers reveal parallel sensitivities to the influence of mechanical, environmental and material variables on FCP rates at given ΔK levels. Surely, conflicting data trends between metals and plastics are to be expected based on the vastly different damping characteristics of polymeric solids from those in metals. For example, metal alloys generally reveal little or no test frequency sensitivity when experiments are conducted in an inert atmosphere but exhibit slower fatigue crack growth rates with increasing test frequency when the sample is placed in an aggressive environment. In polymeric solids, it has been shown that in the absence of an aggressive specie, crack growth rates in different materials can increase, decrease or remain the same; the latter mentioned complex response has been attributed to the degree and extent of self-heating of the polymer during cyclic loading (Hertzberg, Manson and Skibo, 1978a,b; Skibo, Hertzberg and Manson, 1979; Hahn and co-workers, 1982). When such heating is localized, so as to presumably blunt the crack tip, growth rates decrease with increasing frequency; growth rates have been shown to increase with increasing frequency when the specimen experiences widespread heating that increases specimen compliance and associated strain per loading cycle.

Parallel studies of metal and plastic fatigue have also focused on characterization of the resulting fracture surface micromorphology (Hertzberg, 1983). In metals, several micromechanisms have been identified in FCC, BCC and HCP metal alloys at various ΔK levels. At high ΔK levels, for example, the fracture surface reveals extensive evidence of microvoids along with widely spaced fatigue striations. At intermediate ΔK levels, the fracture surface reveals a greater density of striated regions with each striae representing the extent of crack advance as a result of a single loading excursion. Bates and Clark (1969) proposed an empirical relationship of the form

$$\text{striation spacing} \approx 6\left(\frac{\Delta K}{E}\right)^2 \quad (2)$$

that normalizes the results for many metal alloys of varying stiffness (E). Equation 2 is used extensively in failure analysis studies and in basic materials research studies and, as such, represents the most quantitative and most useful fractographic relationship available in fracture analysis. It should be noted, however, that fatigue striation spacings do not usually correspond to the average crack growth rate at any given ΔK level (Fig. 2). This arises from the presence of other fracture micromechanisms that contribute to the overall macroscopic growth rate. The second power dependence of the striation spacing on ΔK has suggested that the growth increment (striation width) of the crack resulting from a given loading cycle may be related to some fraction of the crack tip opening displacement given by Goodier and Field (1963)

$$2V = \frac{K^2}{\sigma_{ys}E} \quad (3)$$

where $2V$ = crack opening displacement
E = elastic modulus
σ_{ys} = yield strength
K = prevailing stress intensity factor

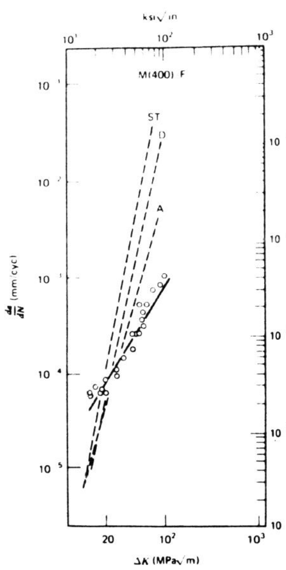

Fig. 2. Macroscopic and microscopic (o) crack growth rates in hot rolled steel (Heiser and Hertzberg, 1971).

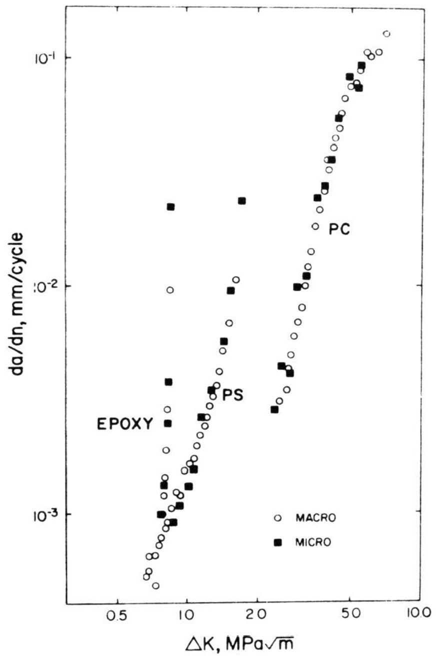

Fig. 3. Macroscopic and microscopic crack growth rates in epoxy, polystyrene (PS) and polycarbonate (PC) (Hertzberg, Skibo and Manson, 1979).

Proceeding further, more recent studies have shown that the striation mechanism in metals ceases to be observed at progressively lower ΔK levels (Hertzberg and Mills, 1976). In some instances, a region of intergranular failure is observed over some intermediate ΔK regime (Cooke and co-workers, 1975). In general, there is a tendency toward the development of a highly faceted fracture surface appearance as ΔK approaches ΔK_{th} that is strongly reminiscent of a cleavage-like appearance. Such markings have been seen in iron, aluminum, nickel, and titanium based alloys and reflect fracture processes that are strongly dependent on metallurgical factors. Yoder, Cooley and Crooker (1977), for example, identified such faceted growth with ΔK conditions corresponding to the development of a crack tip cyclic plastic zone that is equal to or smaller than the microstructure component (e.g., grain boundary, pearlite colony or martensite lath boundary) that controls overall fatigue cracking behavior.

The fatigue fracture surface micromorphology of engineering plastics is equally complex. At relatively high ΔK levels, many polymers reveal

fatigue striation spacings that are in excellent agreement with macroscopically determined crack growth rates (e.g., see Fig. 3) (Hertzberg, Skibo and Manson, 1979; Hertzberg and Manson, 1980). This excellent correlation between macroscopic and microscopic growth rates reflects the fact that 100% of the fracture surface in this ΔK regime is striated; hence, only one micromechanism is operative. Contrast this with the findings in metal systems. Furthermore, since one finds a close match between striation spacings and the associated macroscopic growth rates in polymers (Fig. 3), the ΔK dependence of polymer striation width is not constant but varies greatly with material. Obviously, striation width-crack opening displacement correlations (Eqs. 2 and 3) are suspect for the case of polymer fatigue (Hertzberg and Manson, 1980). A similar conclusion was drawn by Döll, Könczöl and Schinker (1983), based on striation spacing and crack opening displacement measurements.

DISCONTINUOUS GROWTH BANDS IN ENGINEERING PLASTICS

At lower ΔK levels in several polymeric solids (Hertzberg, Skibo and Manson, 1979; Hertzberg and Manson, 1980) and at all ΔK levels investigated thus far in poly(vinyl chloride) (PVC) (Hertzberg and Manson, 1973), a second set of fracture bands were observed but which did not correspond to the prevailing macroscopic growth rate in this ΔK regime. Instead, these fracture bands reflected discrete crack advance increments that appeared after several hundred to several thousand cycles of total crack arrest (Fig. 4a). Elinck, Bauwens and Homés (1971) were the first to recognize

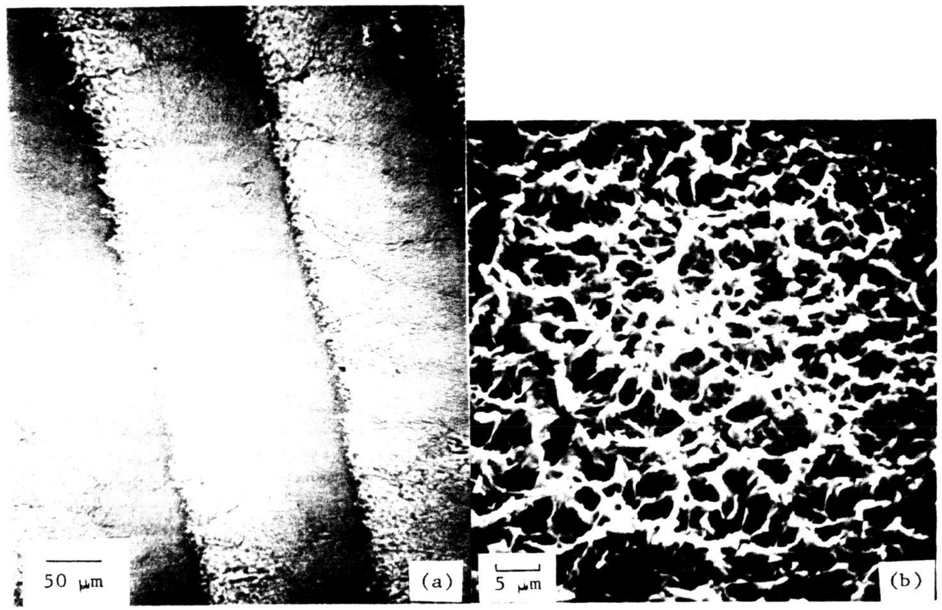

Fig. 4. Discontinuous crack growth in PVC. a) Arrow indicates crack direction. Note void gradient; b) Void structure in DGB. Note ruptured fibrils.

the discontinuous nature of crack growth associated with these fracture bands and demonstrated that the band width corresponded to the size of the Dugdale (1960) plastic strip zone as given in Eq. 4.

$$\text{Plastic zone length} = \frac{\pi}{8}\,\frac{K^2}{\sigma_{ys}{}^2} \tag{4}$$

Subsequent studies (Hertzberg, Skibo and Manson, 1978, 1979) revealed that these bands were produced by continuous growth of a craze zone ahead of the advancing crack front, followed by sudden breakdown of the craze. Earlier attempts to monitor the kinetics of craze growth led to the conclusion that the length of the craze increased rapidly at first to about 60 to 80% of its final length within the first 10% of the band's cyclic lifetime (Hertzberg, Skibo and Manson, 1979). During the remaining lifetime of the band, the length increased to that predicted from Eq. 4. Estimates of the band's cyclic stability were obtained by dividing the band width by the prevailing macroscopic crack growth rate. Figure 5 shows clearly that the stable lifetime of these bands is measured in hundreds and thousands of loading cycles and not the single loading cycle associated with the formation of a fatigue striation.

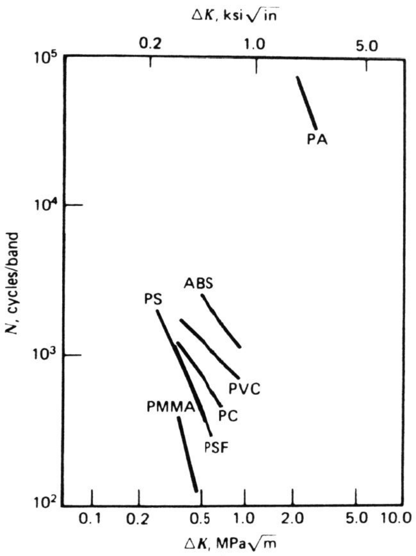

Fig. 5. Cyclic stability of discontinuous growth bands in several polymers as a function of ΔK (Hertzberg, Skibo and Manson, 1978).

Though no craze thickness measurements were attempted with these earlier studies, it was speculated that the craze thickened progressively with continued cycling to some limiting value (perhaps the critical crack opening displacement at the prevailing stress intensity level as computed from Eq. 3). As such, it was suggested that the growth of the craze length to the

Dugdale dimension ℓ_D constitutes a necessary but not sufficient condition for fracture of these fracture bands. Instead, the key to the sudden breakdown of the band was believed to correspond to a cyclic strain-induced stretching of the craze fibrils to their ultimate length, t_{max}, which was less than COD_{max} for the given ΔK level. If $t_{max} > COD_{max}$, then continuous crack growth should occur and no discontinuous growth bands (DGB) would be expected. The latter condition was found to exist under low test frequency conditions (Skibo, Hertzberg and Manson, 1976) and when the material possessed either a high molecular weight or high molecular weight tail in a lower average molecular weight polymer (Skibo and co-workers, 1977; Janiszewski, Hertzberg and Manson, 1981). The model for discontinuous growth band formation describes craze thickening as occurring by fibril stretching which contributes to orientation hardening and fibril strengthening. At some point, the individual chains in the fibrils are envisioned to disentangle and/or fracture, thereby leading to a gradual reduction in their volume fraction within the craze. It was further envisioned that the reduction in fibril volume fraction is counterbalanced by an increase in the average strength of the remaining fibrils due to orientation hardening. The net effect was believed to be the establishment of a relatively constant stress across the craze zone (consistent with the Dugdale formulation) based on a relatively constant product of fibril strength and remaining fibril volume fraction, $\sigma_f\,v_f$. After fracture of n fibrils and the associated stretching (and strengthening) of the remaining fibrils due to the necessary load transfer, a critical condition would be reached when the (n+1)th fibril were to break; the remaining fibrils would no longer be able to stretch (i.e., $t_{max} < COD_{max}$) and the entire craze would suddenly tear apart. The period during which the craze was both growing in length and thickness prior to its sudden breakdown could then be described as a metastable equilibrium condition representing a balance between strain-induced fibril strengthening and the weakening process of void formation and growth.

A closer examination of these DG bands reveals a clearly defined void gradient with large voids located near the crack tip and small voids found near the craze tip. This pattern suggests that the craze breakdown process occurs by a void coalescence mechanism with the void size distribution reflecting the internal structure of the craze just prior to crack extension. Note that this void size gradient parallels the craze opening displacement distribution across the craze. It should be recognized that these microvoids represent small regions of empty space that are surrounded by broken fibrils (Fig. 4b). Schinker, Konczöl and Döll (1982) also reported seeing fracture patterns in PVC that included broken fibrils.

RECENT DGB FINDINGS

Recent studies have contributed much to further our understanding of the discontinuous growth band formation and breakdown process. Rimnac, Hertzberg and Manson (1983, 1984) reexamined the kinetics of DG band extension in plasticized samples of PVC and the associated formation of fracture bands up to 230 μm in length. Figure 6 shows the band length plotted as a function of the number of loading cycles for samples subjected to different constant ΔK levels and possessing different amounts of plasticizer content and different molecular weights. These craze length measurements were obtained with a traveling optical microscope being trained on the surface of the specimen (along the line of sight A) such that craze appeared in profile view (Fig. 7). It is clear that the final length and stability of the bands varies markedly with different ΔK levels, plasticizer content and molecular weight. In each case, however, the craze

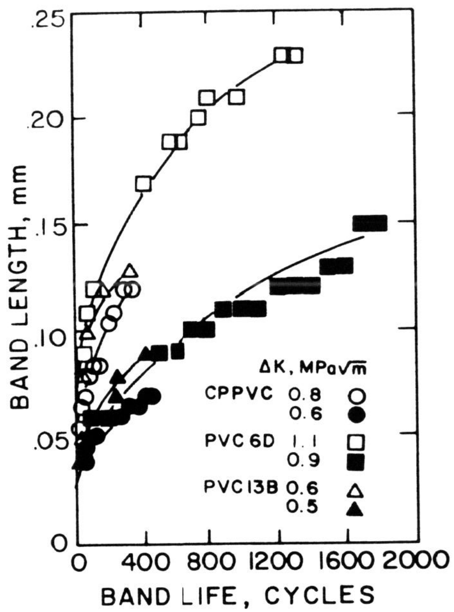

Fig. 6. Discontinuous band length versus band life for PVC samples with different molecular weight and ΔK levels (Rimnac, Hertzberg and Manson 1984).

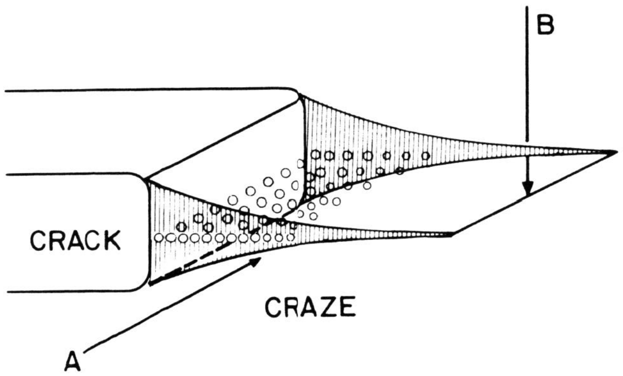

Fig. 7. Directions of viewing craze growth. A--profile view; B--normal view of interference fringes (Rimnac, Hertzberg and Manson, 1983).

growth rate $d\ell/dN$ is seen to increase rapidly at first but then quickly decelerates to much lower levels. One may then describe band growth by a relationship of the form

$$\ell = A\,N^{m} \tag{5}$$

where ℓ is the band length, N is the number of cycles, A is a constant dependent on material and test variables and m is the slope (0.3) of the line described by log (ℓ) vs. log (N). When these data are normalized by the final craze length ℓ_D (defined by the Dugdale dimension) and the cyclic stability N* (recall Fig. 5), respectively, the data fall along a single line (Fig. 8) described by

$$(\ell/\ell_D) = B(N/N^*)^{0.3} \tag{6}$$

This result points out that regardless of the PVC material parameters or stress intensity level, the kinetics of craze development and final breakdown during cyclic loading are the same. Furthermore, if Eq. 5 is differentiated, the craze growth rate $d\ell/dN$ may be described by

$$d\ell/dN = C/N^{0.7} \tag{7}$$

which parallels the craze strain vs time relationship during monotonic loading of impact-modified PVC (Bucknall, 1977) and is consistent with Andrade-type transient creep (Andrade, 1910).

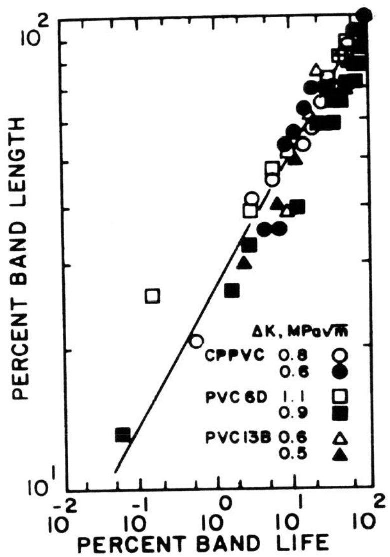

Fig. 8. Percent band length versus percent band life for three PVC resins (Rimnac, Hertzberg and Manson, 1984).

It is interesting to consider an alternative relationship between the dimensionless damage parameter ℓ/ℓ_D and the fractional life parameter N/N*. Fong (1982) postulated that the rate of change of a damage parameter Δ with respect to some fractional life parameter x varies linearly with the damage parameter such that

$$\Delta = \frac{e^{kx}-1}{e^{k}-1} \qquad (8)$$

where k is a constant to fit the experimental data. (With respect to the parameters used in Eq. 6, $\Delta = \ell/\ell_D$ and x = N/N*). Some selected values of k are seen on the linear plot depicted in Fig. 9. When the data from Fig. 8 are replotted on linear scales, the data for the different samples tend to follow the form of Eq. 8 with a fitting constant k that falls roughly between -2 and -4 (Fig. 10).

For the case where $k \approx -4$, Wong (1982) noted that Eq. 8 would be very sensitive to the incubation and initiation of damage corresponding to the early stages of fatigue damage. From the data given in Fig. 10, we may conclude that the early stages of cyclic-induced craze damage are dominated by the craze lengthening process. Since the craze lengthening damage parameter is not as sensitive a parameter at large fractional lives (say, $x \geq 0.8$), the terminal stages of craze breakdown should be correlated with a damage parameter that changes more rapidly as x approaches 1.0 (i.e., $k > 0$). It is tempting to speculate whether a different dimensionless damage parameter, based on the ratio between the instantaneous craze thickness and critical craze opening displacement, might be that more sensitive parameter which characterizes final craze breakdown. This author reached a similar conclusion in an earlier publication (Hertzberg, Skibo and Manson, 1979).

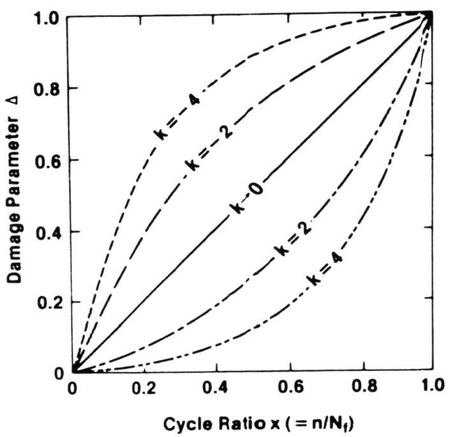

Fig. 9. Schematic representation of dimensionless damage parameter versus cyclic damage ratio (Fong, 1982).

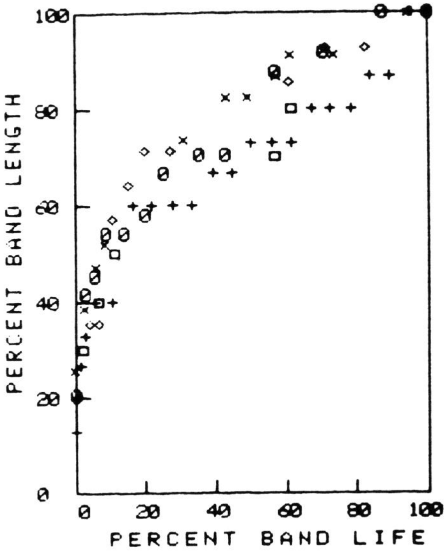

Fig. 10. Data from Fig. 8 replotted on linear scale.

Further studies are indicated to determine whether the rate of fatigue damage accumulation within the growing craze is of an exponential or power law form. Regardless of the form of the relationship, however, the correlation of a dimensionless damage parameter with the fractional life parameter appears to hold much promise in characterizing fatigue damage in engineering plastics as well as in metal alloys and composites.

As noted earlier, the fracture surface micromorphology of the discontinuous growth band reveals microvoids (and surrounding broken fibrils) which decrease in size toward the craze tip. To characterize the progress of craze breakdown at craze lifetimes less than N*, Rimnac, Hertzberg and Manson (1983, 1984) examined the fracture surfaces of DG bands that had been broken open after 25%, 50% and 75% of N*. In each instance, the length of the broken band (Y) was equal to the length of the uninterrupted band X (i.e., ℓ_D) in apparent disagreement with the results shown in Fig. 10 (see Fig. 11). It is also noted that the micromorphology of the interrupted fracture bands are markedly different from the normal bands that preceded them. Instead of a simple void gradient pattern, these bands reveal an initial region of decreasing voids followed by a zone of craze patch. The latter pattern has been identified by Beahan, Bevis and Hull (1975) as representative of rapid fracture along the craze-matrix interface of a pre-existent craze. As such, these fractographic features suggest that the crack progresses initially along the craze mid-rib and through a network of microvoids before jumping alternately from one craze-matrix interface to the other. That is to say, the craze apparently develops to its full length (defined by Eq. 4) before 25% of its life has been consumed and that the crack path switches from the mid-rib location to the craze-matrix interface. It is interesting to note that the mechanism transition point corresponds in each case to the craze length dimension measured along the line of sight A (data shown in Fig. 8).

To reconcile these contradictory findings, additional craze length measurements were made, this time along the line of sight B (Fig. 7). As Kambour (1968) pointed out some time ago, interference fringes are found when a craze zone is viewed in a direction perpendicular to the craze plane (line of sight B (Fig. 7). Figure 12 shows the fringe patterns from DG bands that were interrupted after 25% and 50% of their anticipated life, respectively. While the ℓ/ℓ_D ratios for these two fractional lifetimes were 0.65 and 0.9, respectively, the observed fringe patterns showed that in each case, the craze had grown to its maximum length ℓ_D. Therefore, the data generated along line of sight A (Fig. 7) do not represent a measure of the total craze length as was originally thought (Hertzberg, Skibo and Manson, 1979); rather this measurement describes some level of damage within the craze that is discernible along line of sight A. The fracture appearance noted in Fig. 11, therefore, is consistent with both interference fringe measurements that identify the total craze length and profile craze length measurements (line of sight A) that define the transition point from the void coalescence to patch fracture mechanisms.

Additional optical interference measurements pertaining to discontinuous growth band development and breakdown have been reported by Döll, Konczöl and Schinker (1983); Schinker, Konczöl and Döll (1982), Döll (1983), Konczöl, Schinker and Döll (1983) and Lang and co-workers (1984). These experiments have provided useful information pertaining to the cyclic lengthening and thickening processes within the craze zone. Of particular note, Konzcöl, Schinker and Döll (1983) concluded that the craze thickens with continued cycling by fibril stretching and associated orientation

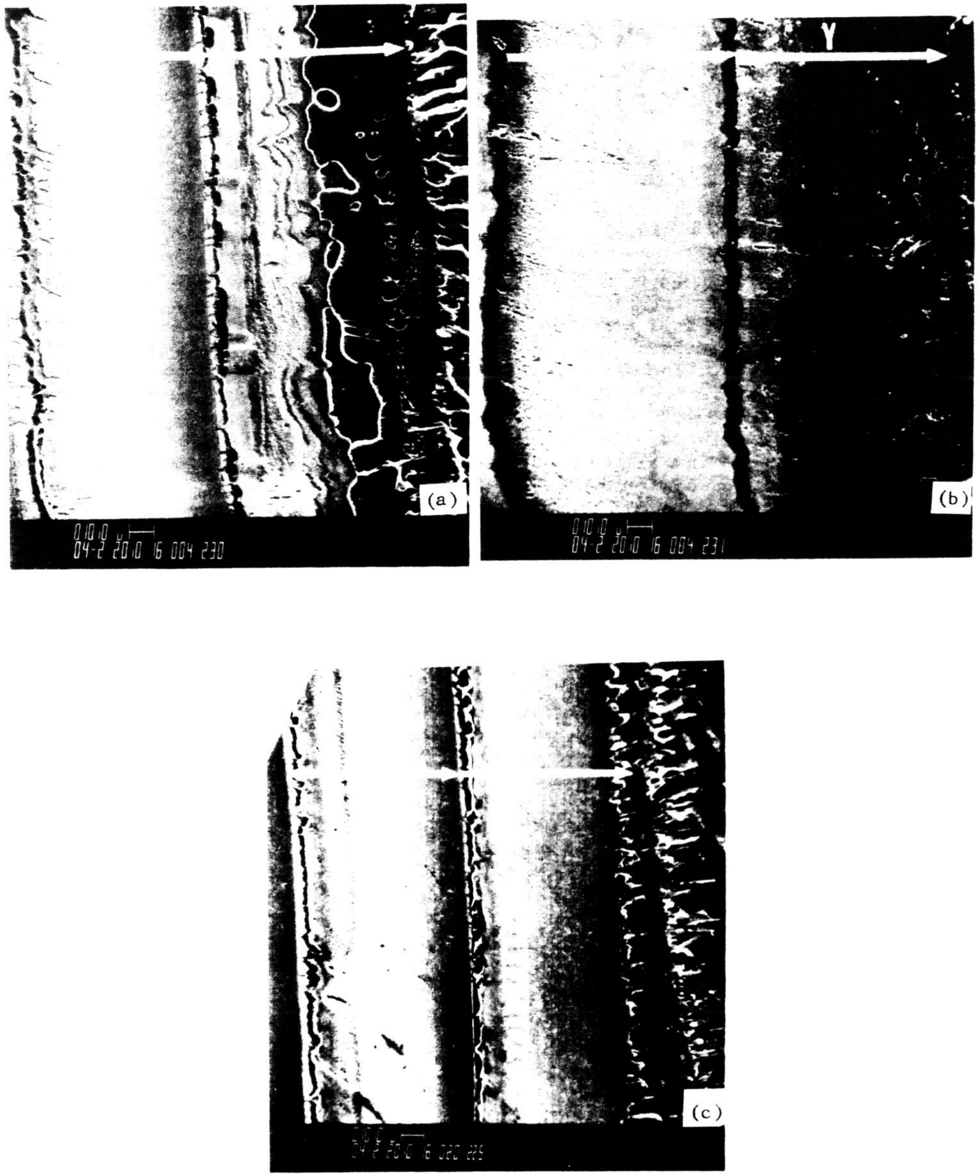

Fig. 11. Morphology of interrupted DG bands in PVC after a) 25%, b) 50%, c) 75% life (Rimnac, Hertzberg and Manson, 1984).

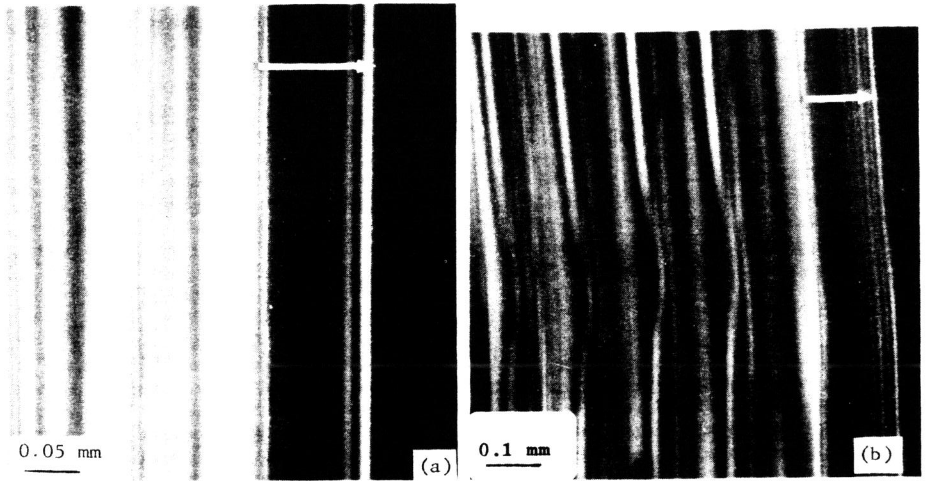

Fig. 12. Interference fringe patterns for DG bands interrupted after a) 25% and b) 50% life (Rimnac, Hertzberg and Manson, 1984).

hardening. This finding is consistent with the DGB model described earlier; however, Lauterwasser and Kramer (1979) and Verhaulpen-Heymens (1979) found that craze thickening in polycarbonate and polystyrene under quasi-static loading conditions took place by drawing fresh material into the craze from the surrounding bulk. Apparently, surface drawing of new material into the craze takes place under static loading conditions and should also occur at low cyclic frequencies. Since the fibrils are not stretched significantly but rather are lengthened by the addition of new material, fibril length can exceed COD_{max} before failure; DBG formation would then be precluded (Fig. 13a). On the other hand, when fibril stretching occurs under rapid cyclic loading conditions, DGB formation would occur since the fibrils are strained to failure at lengths less than COD_{max} (Fig. 13b). The development of discontinuous growth bands therefore, reflects a competition between two craze thickening processes: fibril stretching and surface drawing of new material into the body of the craze.

A point of controversy remains pertaining to DGB formation and breakdown processes. Test results involving interference fringe and fracture surface measurements are interpreted by this author to show that the entire craze breaks down at the point of instability and that a completely new craze begins to form with continued load cycling. By contrast, Schinker, Konczöl and Döll (1982) and Takemori and Matsumoto (1981) interpreted their results to show that the crack jumped through approximately 2/3 of the craze. It is not clear why these results differ from the observations described above; further studies are indicated.

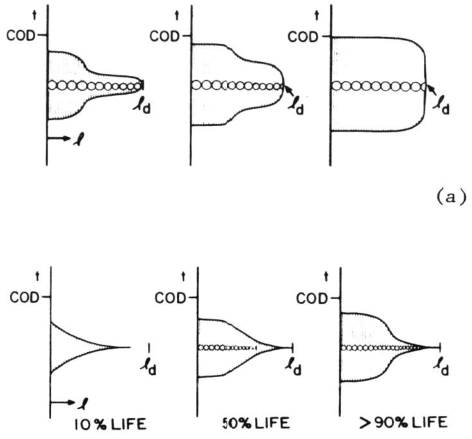

Fig. 13. Schematic representation of a) Non-DGB formation and b) DG band formation where $t_{max} < COD_{max}$ (Rimnac, Hertzberg and Manson, 1983a).

EPSILON-DGB MORPHOLOGY

Finally, a different type of DGB has been reported by Matsumoto and Takemori (1982) and Mills and Walker (1980). Discontinuous growth bands were developed in polycarbonate and polysulfone which formed in association with a pair of shear bands above and below the crack plane. The combination of the two shear bands and the associated craze band was referred to as an ϵ-DG band since these deformation markings resembled the Greek letter epsilon. By contrast, no such ϵ-DGB markings were found in these two materials by Skibo and co-workers (1977). The experiments by Matsumoto and Takemori (1982) and Mills and Walker (1980) which produced ϵ-DG bands had two common features: specimens were unnotched and the applied stress levels were greater than 25% of the material's yield strength. By contrast, the specimens tested by Skibo and co-workers (1977) were in the form of compact tension samples and contained long cracks; at the same time, the nominal stress was a very small fraction of the yield strength. It is believed that the key difference in test conditions that influences the formation or absence of shear bands along with DG bands rests with the magnitude of the nominal stress level. To verify this hypothesis, a crack was grown in single-edge notched sample and then cut back to a remaining length of 0.4 mm. Cyclic loading was then resumed at a cyclic stress range of about 22% of the yield strength to achieve a ΔK level of 0.4 MPa$\sqrt{m}$. Under this high nominal stress condition, ϵ-DG bands were formed as shown in Fig. 14. It may be concluded, therefore, that ϵ-DGB formation is strongly dependent on the applied stress level and restricted to the regime where σ_{max}/σ_{ys} is large. To confirm this finding, the short crack length sample was cycled further under constant ΔK conditions (0.4 MPa$\sqrt{m}$). To maintain a constant ΔK level with increasing crack length, the stress level was reduced accordingly. As a result, the ratio of σ_{max}/σ_{ys} decreased

continuously and ϵ-DG bands no longer developed. Instead, normal DG bands were found on the fracture surface.

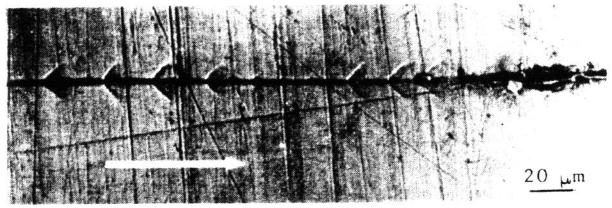

Fig. 14. Fatigue fracture surface profile of polycarbonate showing ϵ-DG bands when $\sigma/\sigma_{ys} \approx 0.22$ (Rimnac, Hertzberg and Manson, 1983a).

It is most intriguing to relate this finding to the overall fatigue crack growth response of short cracks in metal alloys. Numerous studies have shown that short cracks behave differently from long cracks at a given ΔK level with many investigators attributing at least part of this difference to large plastic zone size-crack length ratios in the short crack samples. The formation of ϵ-DG bands at high σ_{max}/σ_{ys} ratios in engineering plastics provides dramatic proof that deformation processes can change markedly when the nominal stress/yield strength ratio is large. Further confirmation of this point was obtained when attempts were made to infer the craze yield strength based on discontinuous growth band widths. That is, Eq. 4 can be used to infer the material's yield strength at a given stress intensity level where the plastic zone dimension is given by the band width. Computed σ_{ys} results based on Eq. 4 were found to be unreasonably low for the short crack sample described above. However, when the more general form of the Dugdale model was used (Dugdale, 1960)

$$\text{Plastic zone length} = [\sec\left(\frac{\pi}{2}\frac{\pi}{\sigma_{ys}}\right) - 1]\, a \qquad (9)$$

a far more realistic estimate of the yield strength was computed.

CONCLUSIONS

1. Fatigue fracture processes in engineering plastics include the formation of fatigue striations which have spacings in excellent agreement with the incremental advance of the crack resulting from a single loading cycle. The ΔK dependence of polymer striation width is not constant but varies greatly with materials. Hence, a simple correlation between striation width and crack opening displacement is not found.

2. A second fracture band--referred to as a discontinuous growth band--corresponds to continuous growth of a crack tip craze followed by its rapid breakdown. The width of these fracture bands corresponds to the plastic zone size as computed from the Dugdale plastic strip model. The growth rate of the precursor craze varies inversely with the number of cycles with a relationship of the form analogous to Andrade-type transient creep.

3. A second type of discontinuous growth markings containing a pair of

shear bands (so-called ϵ-DG bands) has been reported and the presence of these shear bands related to the ratio of applied stress to yield strength levels.

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