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Polymer 48 (2007) 3512–3519
polymer
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Peak stress intensity dictates fatigue crack propagation in UHMWPE

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Received 11 December 2006; received in revised form 31 March 2007; accepted 3 April 2007
Available online 8 April 2007

Abstract

The majority of total joint replacements employs ultra-high molecular weight polyethylene (UHMWPE) for one of the bearing components. These bearings may fail due to the stresses generated in the joint during use, and fatigue failure of the device may occur due to extended or repeated loading of the implant. One method of analysis for fatigue failure is the application of fracture mechanics to predict the growth of cracks in the component. Traditional analyses use the linear elastic stress intensity factor K to describe the stresses near a loaded crack. For many materials, such as metals, it is the range of stress intensity, ΔK, that determines the rate of crack propagation for fatigue analysis. This work shows that crack propagation in UHMWPE correlates to the maximum stress intensity, K_{max}, experienced during cyclic loading. This K_{max} dependence is expected due to the viscoelastic nature of the material and the absence of crazing or other cyclic load dependent crack tip phenomena. Such a dependence on a non-cyclic component of the stress allows cracks to propagate under load with little or no fluctuating stresses. Consequently, traditional fatigue analyses, which depend on the range of the stress to predict failure, are not always accurate for this material. For example, significant static stresses that develop near stress concentrations in the component locking mechanisms of orthopedic implants make such locations likely candidates for premature failure due the inherent underestimate of crack growth obtained from conventional fatigue analyses.

Keywords: UHMWPE; Fatigue crack propagation; Peak stress intensity

1. Introduction

1.1. Fatigue in orthopedic implants

Total joint replacements (TJRs) generally comprise metallic components fixed to the underlying bone and two or more articulating bearing surfaces. Ultra-high molecular weight polyethylene (UHMWPE) is the predominant compliant counter-bearing material used in TJRs, articulating against either a metal or ceramic part [1]. The bearing surfaces move over one another as the joint is articulated, and this articulation generally results in a maximum force well in excess of the body weight of the individual. This force is communicated between the articular surfaces through a relatively small area, resulting in substantial contact stresses. Orthopedic implants experience tens of millions of such stress cycles in their years of service, and these stresses can exceed the yield strength of UHMWPE [2]. The application of repeated contact stresses or variations in far-field loads leads to a process known as fatigue, whereby flaws can develop or propagate in the material leading to premature failure of the device. Further, in addition to superficial asperity wear mechanisms, UHMWPE has been observed to undergo delamination in service. Delamination is characterized by damage to the material below the surface allowing large plate-like debris to break-off, causing severe wear, and is most often observed in tibial bearing components of total knee replacements (TKRs) [3]. Production of UHMWPE wear debris may lead to osteolysis, which is a biological foreign body immune response that drives bone loss

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0032-3861/$ - see front matter © 2007 Elsevier Ltd. All rights reserved.
doi:10.1016/j.polymer.2007.04.006

near the implant. This leads to aseptic implant loosening, necessitating replacement of one or more components of the implant [1,3]. Both fatigue and delamination depend to a significant extent on fracture processes, so understanding the fracture properties of UHMWPE is paramount to proper design of orthopedic implants [1,3–8].

1.2. Fatigue crack propagation

The fatigue resistance of UHMWPE is usually investigated by measuring the rate of fatigue crack propagation (FCP) in a previously cracked specimen. Linear elastic fracture mechanics (LEFM) quantifies the stress near the tip of a crack through the stress intensity factor, K, defined by Eq. (1), where Y is a factor that accounts for geometric effects, σ^{∞} is the nominal far-field stress, and a is the crack length. The stress near the crack tip (in the opening mode), described by Eq. (2), is schematically shown in Fig. 1, where r is the distance from the crack tip in the plane of the crack, and σ_{yy} is the stress perpendicular to the crack plane. Opening mode cracks are the only type of flaw considered in this work, as they are generally the most critical species of crack with respect to part failure.

$$K = Y\sigma^{\infty}\sqrt{\pi a} \tag{1}$$

$$\sigma_{yy} = \frac{K_I}{\sqrt{2\pi r}} \tag{2}$$

The local stresses near the crack tip govern the growth of the crack and so the stress intensity factor or some function of the stress intensity factor, K, is used as a metric of the driving force for crack propagation. LEFM assumes that the material is linear elastic, but in response to the singular stresses immediately in front of the crack elasticity breaks down. When this inelastic region is small, that is if there is small scale yielding, K remains a good parameter for the description of crack propagation behavior. If the region is large or changes with time, other models are required.

FCP empirically can be separated into three regimes of behavior, as shown in Fig. 2. In Regime I, corresponding to sub-threshold propagation, the stress intensity is insufficient to propagate a crack to a significant degree. In the third regime, the crack accelerates rapidly and propagates through the entire specimen or part in an unstable, catastrophic manner. This occurs at a critical stress intensity, K_c. Beyond the threshold for crack propagation but prior to catastrophic fracture, in Regime II, cracks propagate in a stable manner, i.e. there is significant propagation that does not immediately cause catastrophic failure. Crack growth rates in this regime typically range from 10^{-6} to 10^{-4} (mm/cy). For many engineering materials, the rate of stable crack propagation fits a power law relationship with K, such as in the classic Paris equation,

$$\frac{da}{dN} = C\Delta K^m \tag{3}$$

where $\Delta K = (K_{max} - K_{min})$, a is the crack length, N is a load excursion cycle, and C and m are experimentally determined parameters. As cracks grow under fatigue conditions, the

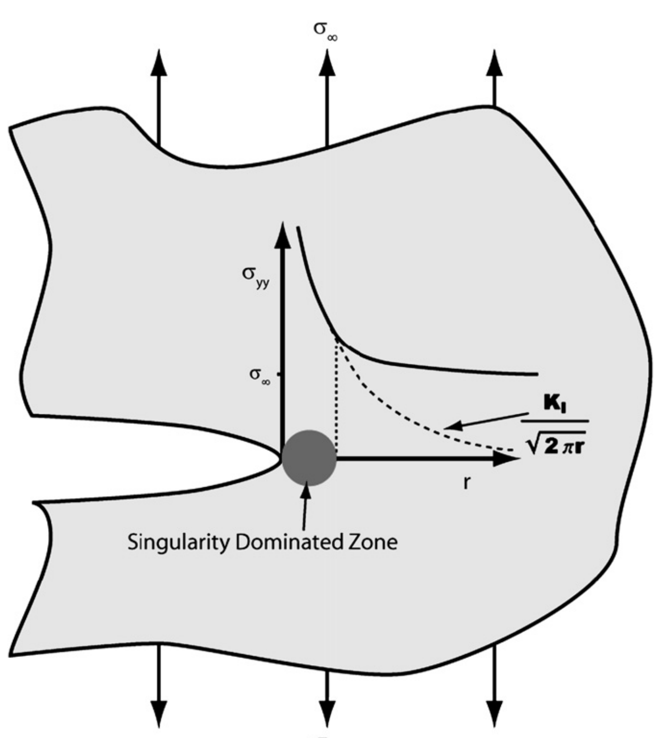

Fig. 1. The stress intensity factor, K_I, shown as the parameter characterizing the stress as it becomes singular near the crack tip. After Ref. [26].

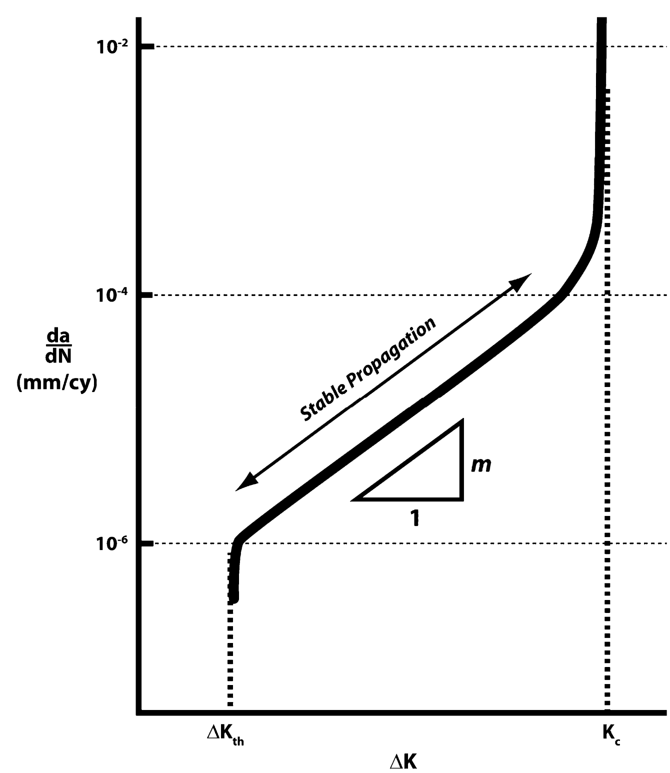

Fig. 2. Three regimes of fatigue crack propagation (FCP), shown on log–log axes. Stable, or Paris, regime of crack propagation is shown with its characteristic log–log "slope", m.

stress intensity for a given load steadily increases, until eventually the stress intensity exceeds the critical value and the specimen fails due to fatigue fracture.

The Paris equation describes the rate of stable crack propagation in materials known to grow cracks due to cyclic mode phenomena. Therefore, the excursion of the stress intensity due to a cyclically varying applied load, ΔK, is employed as the crack growth driving parameter. On the other hand, crack growth in a static or creeping mode may be described by:

$$\frac{da}{dt} = QK^n \tag{4}$$

which is analogous to the Paris equation, except that the process depends on the instantaneous K rather than some excursion or cyclic component of it [9]. This may be designated as static mode crack propagation since the process is not sensitive to cyclic phenomena *per se*.

Early attempts to account for both static and cyclic mode components of crack propagation in one model blended the relationships shown in Eqs. (3) and (4). Two basic approaches involve either superposition of the two processes or a coupled fatigue system where each process (static or cyclic mode) affects the other. The former of these two approaches holds crack propagation rates for the two mechanisms to be additive [10],

$$da = \left[\frac{\partial a}{\partial N}\right]_t dN + \left[\frac{\partial a}{\partial t}\right]_N dt \tag{5}$$

where each bracketed differential has the non-differentiated parameter fixed. These bracketed expressions correspond to the above Eqs. (3) and (4). Since this formulation only involves the addition of each mechanism, there is no ability for it to predict any synergistic effect of superimposed static mode and cyclic mode FCP. Another approach explicitly uses the product of the right hand side of Eqs. (3) and (4) in order to enforce the coupled effect of both mechanisms [11,12].

$$\frac{da}{dN} = C\Delta K^p K_{\max}^q \tag{6}$$

For materials that fatigue via a dominant cyclic mode, $p \gg q$, while static mode fatigue implies $q \gg p$. Materials that are essentially ductile tend to undergo cyclic FCP, while brittle materials undergo static mode FCP. Eq. (6) allows only coupled effects, i.e. there must be a contribution from both cyclic and static modes for any crack propagation, which is a potential limitation. While Eq. (3) is often used for fitting of FCP data regardless of the material; it is most appropriate when the failure process is essentially cyclic. Eq. (3) is the result from setting $q \ll p$ in Eq. (6), meaning that there are no static effects active in the process. Similarly, if $q \gg p$ in Eq. (6), and therefore the process is essentially static, crack growth can be expressed as dominated by $K_{\max}$, as shown in Eq. (7).

$$\frac{da}{dN} \approx CK_{\max}^q \tag{7}$$

The battery of experiments employed in this work differentiates between cyclic and static mode fatigue by imposing a load fluctuating with a low magnitude superimposed on a relatively large static load. Cyclic mode propagation is then inhibited, but static propagation is preferentially enabled. Such a loading scenario is clinically feasible in orthopedic implants, where cyclic stresses develop due to the cyclic nature of normal human activity, but static stresses due to body weight or interfacial forces between tight fitting components are also significant. If cracks in UHMWPE indeed propagate in the static mode, then Eqs. (6) and (7) imply that cracks could therefore propagate much more quickly than expected from Eq. (3), leading to substantially shorter lifetimes for total joint replacements.

This work investigates the effects of cyclic and static loading on fatigue crack propagation in UHMWPE. This work shows that the fatigue crack propagation in UHMWPE correlates to peak stress intensity. This result has broad applicability to orthopedic implants, particularly those devices that experience high static stresses such as tibial bearing components in total knee replacements or at locking mechanisms in acetabular components of total hip replacements.

2. Experimental

2.1. Material and specimen preparation

GUR 1050 UHMWPE resin (Ticona, Bishop, TX) was compacted and consolidated into billets, and then thermally annealed in air according to a standard commercial practice by Poly Hi Solidur (Fort Wayne, IN). GUR 1050 has a molecular weight of 4–6 million g/mol, and has a semi-crystalline microstructure comprising 40–60% by volume plate-like crystalline lamellae. Material properties of UHMWPE are widely available in the literature [13,14]. The consolidated blocks were subsequently machined into compact tension (CT) specimens. The details of the CT geometry are described elsewhere [7]. The tip of the notch was sharpened with a razor blade, and then fatigue precracked using a fully compressive sinusoidal waveform for 20,000 cycles, in order to ensure a maximally sharp crack at the outset of the experiment [7]. Samples were allowed to rest at room temperature for at least 24 h to allow the residual stresses from compressive loading to relax [15].

2.2. Fatigue crack propagation testing

Fatigue crack propagation experiments consisted of a series of sinusoidal load-controlled tests, run on an Instron 8800 servohydraulic load frame (Instron, Norwood, MA). Each test employed a waveform at constant minimum and maximum load for 10,000 cycles, after which the specimen was removed from the machine and the crack advance measured. Crack advance was obtained by measuring to the crack tip from a datum

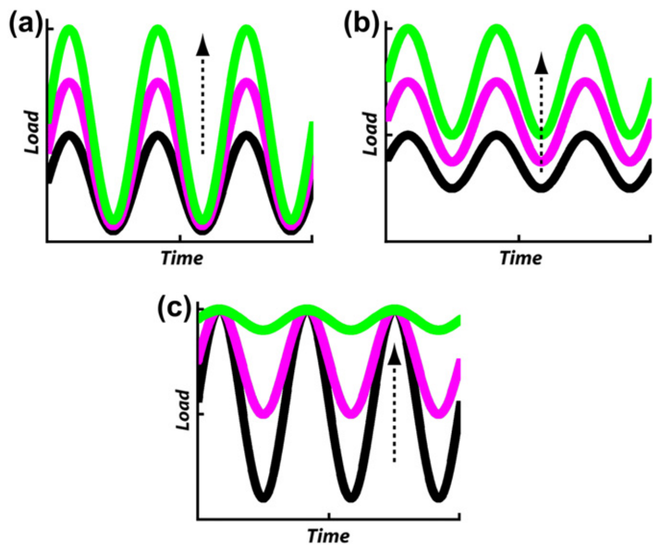

Fig. 3. Load-controlled waveforms for (a) $R = 0.1$, (b) $R = 0.5$, and (c) P_{max} constant (R variable) experiments. Arrows indicate progression of waveform amplitude. Since stress intensity for a given crack length is proportional to applied load, these curves describe the stress intensity as well.

line scored on the surface of the specimen, using an optical microscope (Olympus, Melville, NY) with a graduated ocular reticule with a minimum gradation spacing of 5 μm.

Three types of experiments were performed, each with a different load range sequence. The first test for FCP employs a fixed load ratio $R = 0.1$,

$$R = \frac{\sigma_{min}}{\sigma_{max}} = \frac{P_{min}}{P_{max}} \tag{8}$$

where P_{min} and P_{max} are, respectively, the minimum and maximum loads experienced in the load-controlled waveform. The initial range of the load waveform is below the threshold for propagation, and is then incremented after each test up to failure. A second experiment with fixed load ratio $R = 0.5$ was run in similar fashion to assess the effect of increased static stress on FCP in this material. A third experiment with P_{max} held constant probed the overall effect of varying load ratio, with the load ratio changed by increments of 0.1 from $R = 0.1$ to $R = 0.9$. Tests with R decreasing from 0.9 to 0.1 showed similar results (see Section 5). Schematics of the waveforms employed in these experiments are shown in Fig. 3. Eighteen samples were used for the $R = 0.1$ experiment, while 4 were used for $R = 0.5$, and 5 specimens were used for the K_{max} fixed experiment.

3. Results

3.1. Fixed load ratio

The results for the $R = 0.1$ experiment are shown in Fig. 4. These data show both a prominent crack inception stress intensity range and a stable crack propagation regime that fits to the classic Paris equation (Eq. (3)), with a power law exponent $m = 9.5$. It is interesting to note that ductile engineering materials typically exhibit $m < 4$ and brittle materials typically exhibit $m > 15$ [10]. Thus, the stable crack propagation data approach expectations for brittle materials in this regard, even though UHMWPE engineering strain to failure can exceed 300% [14].

Fig. 4 also shows the data from the $R = 0.5$ experiment. These data show a stable propagation region throughout the tested range, however, without apparent Regime I (sub-threshold) propagation. These data also show markedly higher rates of crack propagation for any given stress intensity range compared to those from the $R = 0.1$ experiments. For $\Delta K = 1.5$, $da/dN = 2.5 \times 10^{-6}$ mm/cy for $R = 0.1$, compared to $da/dN = 2.5 \times 10^{-4}$ mm/cy for $R = 0.5$. This demonstrates a deleterious effect of a superimposed additional static mean stress applied in concert with the cyclic stress.

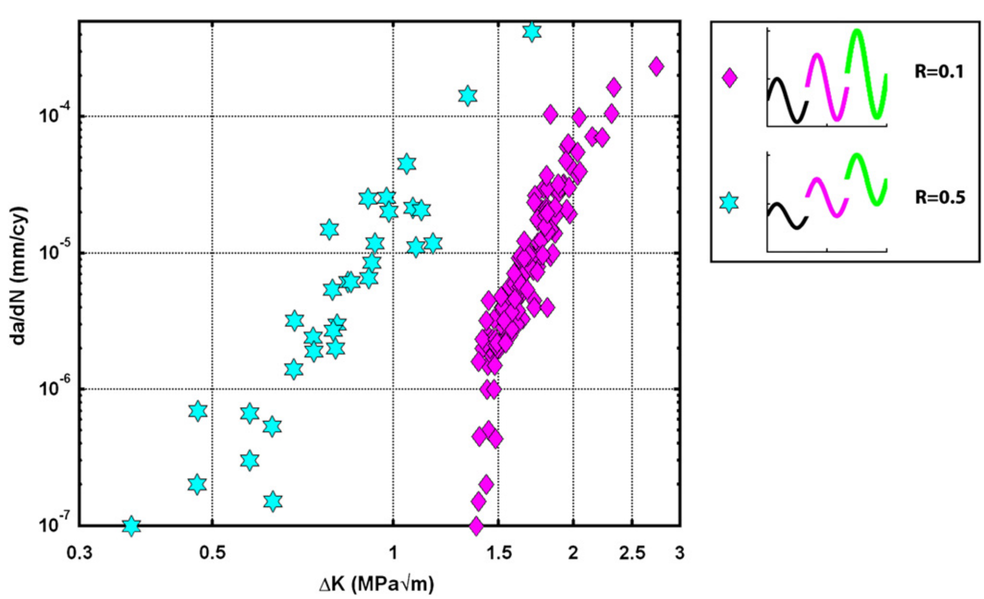

Fig. 4. FCP data for $R = 0.1$ and $R = 0.5$. Data show stable crack propagation and $R = 0.1$ exhibits threshold behavior.

3.2. Fixed maximum load

The data in Fig. 5 correspond to the fixed P_{max} (variable R) experiments. At high stress intensity ranges, the data in Fig. 6 overlay those in Fig. 4. This is expected as data in Fig. 6 are from experiments with approximately $R = 0.1$. However, at lower stress intensity ranges, the data appear to converge to a constant rate of crack propagation. The low stress intensity range (high R) waveforms employed, with P_{max} fixed, approach a steady state applied load of magnitude P_{max}, with the cyclic component negligible compared to the static component of the load, thus motivating the expectation of static behavior.

It is clear from the results for $R = 0.1$ and $R = 0.5$ that there is a considerable stable crack propagation regime for FCP in UHMWPE. Further, when the data from these two protocols are compared certain trends become apparent. The stable crack propagation regime appears linear on a log–log plot, since it follows a power law relationship to ΔK. Fig. 7 shows the two stable regime lines to be approximately parallel. This implies that applying an additional mean stress merely translates the crack propagation data to lower values of ΔK on

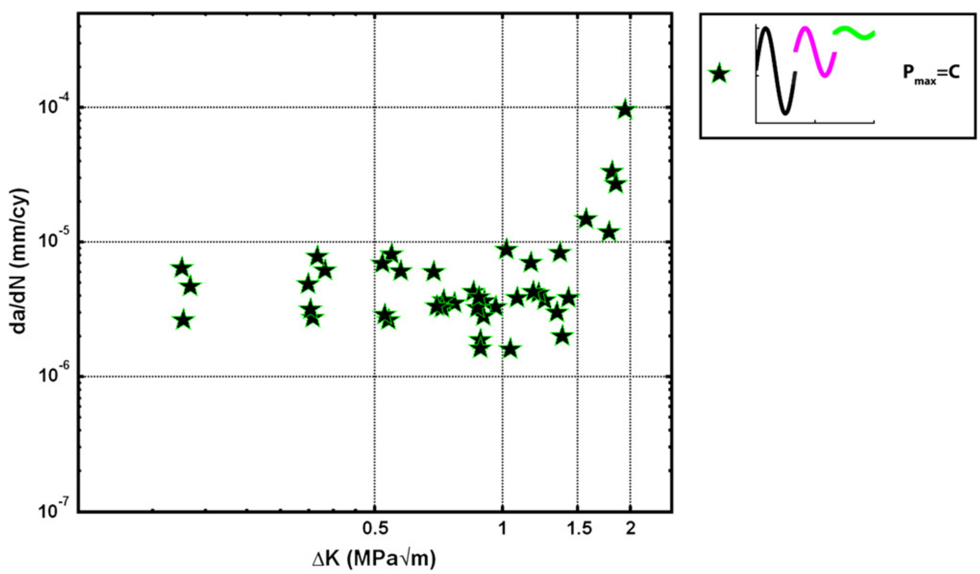

Fig. 5. Data for P_{max} experiment. Crack propagation rates at higher ranges of stress intensity behave in typical manner. Lower ΔK (high R) approaches a steady state asymptote.

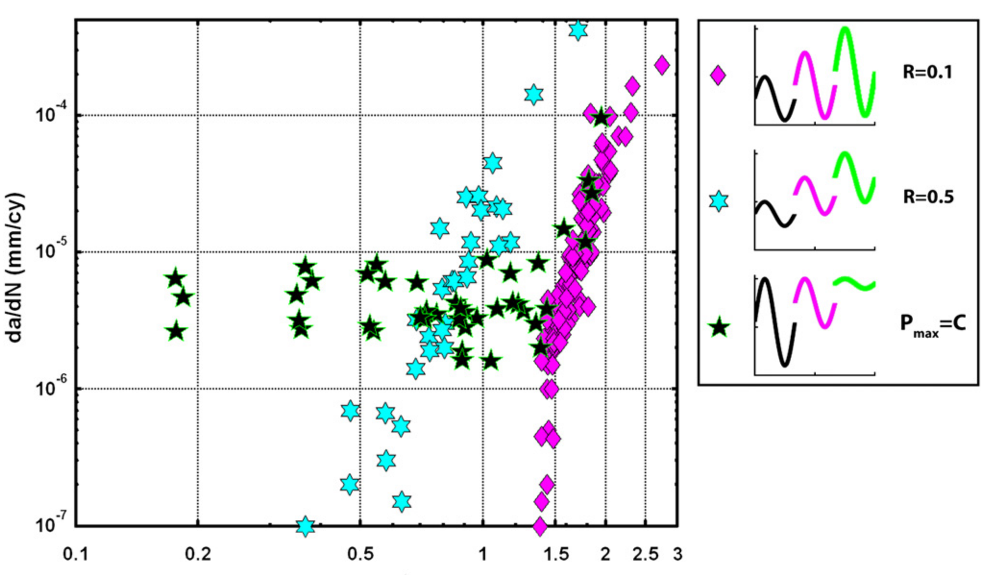

Fig. 6. FCP data from $R = 0.1$, $R = 0.5$, and P_{max} constant experiments. Qualitative disagreement between these data over full range on stress intensity range implies that ΔK is not a sufficient parameter to predict crack propagation behavior.

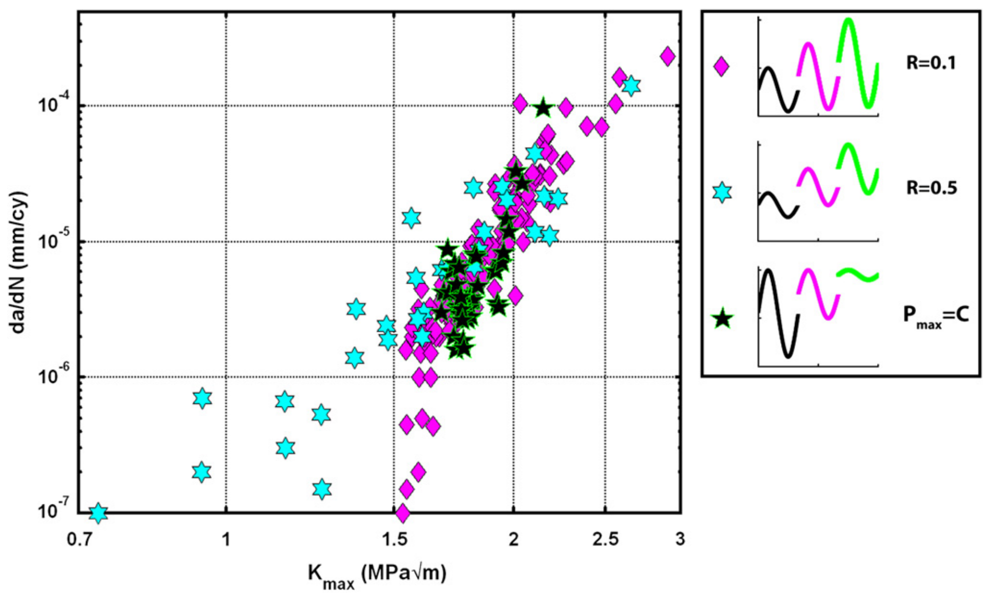

Fig. 7. Data from all experiments, also shown in Fig. 6, plotted with K_{max} as the presumed FCP parameter. All data in the stable crack propagation regime collapse to the same curve, confirming K_{max} as the FCP parameter for this regime of crack growth.

the plot without substantially altering the log–log slope of the stable propagation regime.

3.3. K_{max} dominated FCP

The data for the P_{max} experiment are plotted along with the fixed R experiments with ΔK as the presumed FCP parameter in Fig. 6. This figure demonstrates qualitatively different crack growth rates between the experiments at low ΔK, where the mean stress in the P_{max} experiment is greater than the total range of the waveform. When these same data are plotted using K_{max} as the FCP parameter in Fig. 7, the stable propagation regions of all three experiments collapse to one common trend. This provides clear indication that K_{max}, rather than ΔK, is the dominant FCP parameter for UHMWPE. This result demonstrates that stable crack growth in UHMWPE, under these experimental conditions, is described accurately by Eq. (7).

4. Discussion

Fatigue crack propagation in many materials is known to depend on both cyclic and static mechanisms. Specifically for polymers, which have the capability in many cases to propagate cracks with a static applied load, crack propagation due to a variable load has been modeled as the concerted action of cyclic and static mode phenomena [16]. These results clearly show that K_{max} is the dominant FCP parameter for UHMWPE under the tested conditions, implying that cracks can propagate in this material in the absence of significant cyclic loading. Also, m is of the magnitude expected for brittle materials such as some ceramics. This is significant, as large power law growth exponents imply rapidly accelerating stable crack growth in a fatigue scenario, causing the material to be less tolerant to damage and generally less safe under fatigue loading conditions since cracks are able to grow more rapidly to a critical length and cause unstable fracture. Further, the concept of a time dependent ductile–brittle transition in the failure of polymers is documented, where during short or monotonic tests the material fails in a ductile yielding manner, while in at long duration tests the failure mechanism resembles brittle fracture [17,18]. However, these outcomes do not necessarily shed any light on the physical mechanisms responsible for the underlying behavior.

The salient mechanistic differences between static mode and cyclic fatigue are detailed in the recent work of Ritchie and co-workers [11,12]. Their work deals with the subject without direct application to polymers, but the philosophical framework can be extended to include them as well. FCP can be in general fit to a dual power law relation, shown above as Eq. (6), where ductile materials generally show ΔK dominance through a higher ΔK exponent, p, and brittle materials have a dominant K_{max} component through the corresponding exponent, q. Within the framework of their work, ''brittle'' is a term used to describe materials that fatigue and fracture in an intrinsically K_{max} dominant, static mode manner, while ''ductile'' implies that the material is ΔK dominant and cyclic load dependent, without explicit reference to monotonic properties like strain to failure that traditionally indicate material ductility. The work of Ritchie and co-workers separates factors that affect crack propagation into intrinsic and extrinsic phenomena, where intrinsic phenomena occur immediately at the crack front and govern material failure, and extrinsic effects occur behind the crack tip or in the bulk and influence the driving force for crack extension. Intrinsic phenomena reflect mechanisms necessarily active in the failing material ahead of the crack, and thus are indicative of the material behavior of principal concern. Ductile materials exhibit intrinsic ΔK dominance, as both loading and unloading are required to propagate a crack and keep it sharp enough (e.g. crack tip sharpening by unloading). Brittle materials crack in an intrinsically static, K_{max} controlled manner. UHMWPE undergoes FCP that is dictated by K_{max} (in the stable propagation regime), and so is phenomenologically similar in fatigue to brittle materials in this respect.

Other polymers that undergo other crack tip phenomena, such as crazing or shear banding, may therefore exhibit alternative overall fatigue crack propagation performance. For instance, craze fibrils are known to break down in compression due to a buckling process, giving rise to a unique unloading damage mechanism that may become the rate limiting phenomenon for crack extension, therefore dominant with respect to crack propagation and thus forcing FCP to be cyclic load dependent [19,20]. Both crazing and shear banding may be characterized as so-called extrinsic crack propagation phenomena. UHMWPE lacks these extrinsic phenomena and so crack propagation in it is governed by intrinsic material behavior, at least in the absence of crack tip closing. Thus, due to the lack of extrinsic crack propagation mechanisms in UHMWPE, no attempt to draw connections between these results and those in the literature for static mode fatigue crack growth in other polymers is made here.

Work by Schapery, Saxena, Williams, and others details the analysis of cracks in creeping, strain rate dependent materials, and provides a predictive basis for the apparently brittle nature of FCP in UHMWPE [21–25]. In viscoelastic solids, the ability of the material ahead of the crack to resist stress, and therefore crack propagation, diminishes continuously with time and thus provides an intrinsic mechanism for static crack growth. For cracks in a creeping solid, there are two extremes of behavior denoted creep-ductile and creep-brittle. In these materials, as a crack is loaded, the region that has undergone significant creep relaxation begins to spread beyond the immediate vicinity of the crack tip. If the crack grows faster than the creeping zone of material, then the crack tip leaves behind the creep zone of the previous instant, and insufficient time elapses for the crack tip to undergo significant creep. Consequently, the creep zone stays small, and the stress field appears roughly the same as if the material properties were time independent and elastic. Thus, fracture and FCP in creep-brittle materials correlate well with K, and the material is called creep-brittle [21]. More work must be done to establish whether UHMWPE indeed behaves as a creep-brittle material, but the observed FCP behavior indicates this as a likely mechanism that causes apparently brittle crack behavior in an otherwise ductile material. Further, Schapery states that cracks in viscoelastic materials are not sensitive to stress history, but rather to the instantaneous value of the stress intensity, which is also the presumption for brittle materials and Eq. (7) [25]. This further motivates the lack of an intrinsic cyclic driving force for FCP in UHMWPE, and therefore indicates its treatment in design for fatigue as a brittle material.

Note that viscoelastic crack propagation in this sense is purely intrinsic and dependent only on the continuum behavior of the bulk material in the presence of a crack. As mentioned above, other extrinsic phenomena may be dominant with respect to crack propagation, which may explain why other viscoelastic polymers do not always propagate cracks due to static loading or show substantial K_{max} dependence. Additionally, for viscoelastic materials, the power law exponent m in the stable growth regime is related to the relaxation time constant, such that rapidly relaxing polymers propagate cracks in a more accelerated fashion (high m) [24,26]. Thus, mildly viscoelastic materials could be K_{max} dependent in fatigue and still exhibit a low m.

5. Limitations

The P_{max} fixed experiment in general used an increasing R (decreasing ΔK, increasing K_{mean}), which is atypical for fatigue testing. Other tests with P_{max} fixed and decreasing R showed similar behavior, except there was substantially more gross creep in the specimen. Given the observed lack of crack closure in the stable regime, and the observed nominal equivalence of the two P_{max} test methods, the increasing R data are reported for the P_{max} fixed experiment.

There are a number of empirical relations for FCP that take into account the effect of changing R, i.e. mean stress effects [26]. Since K_{max} is a sufficient parameter in this case to predict crack propagation in UHMWPE, these are not considered here. However, lack of crack closure due to mean stress may explain the lack of Regime I propagation behavior in the $R = 0.5$ experiment. Further work with fully reversed tests that are relevant to clinical service conditions should be conducted to examine how crack closure affects crack propagation.

6. Conclusions

This work demonstrates that the fatigue crack propagation behavior of UHMWPE is governed primarily by the magnitude of the peak stress intensity rather than the stress intensity range. Moreover, the high power law exponent in the stable fatigue regime is characteristic of a brittle material. These results indicate that UHMWPE behaves in an intrinsically brittle manner when stress concentrations or cracks are present. The findings from this work qualitatively match predicted behavior from existing fracture models for viscoelastic solids. The current data indicate that static mode micromechanisms dominate the growth of cracks in UHMWPE and that cracks can grow without significant cyclic loading. This finding may have grave implications for the performance of polymer bearings in orthopedic applications since these components experience substantial static mode loading. Therefore, designing joint replacement bearings without considering UHMWPE as an intrinsically brittle material may lead to unexpected early catastrophic failures. This is especially important for the state of the art highly cross-linked resins, and those with otherwise reduced fracture toughness, e.g. due to oxidation, particularly in designs that contain sharp notches or stress concentrations that may initiate cracks. While uncross-linked UHMWPE total hip replacement bearings have been successful for decades, a number of recent case studies that have found catastrophic rim fractures and brittle failures in both lightly and highly cross-linked acetabular bearing components of total hip replacements, indicating that the understanding of brittle fracture mechanisms in both conventional and cross-linked UHMWPE is of immediate interest [27–29].

Acknowledgements

This work is funded by the National Science Foundation grant CMS 0505272 to the University of California at Berkeley.

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