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Plane stress local failure criterion for polycarbonate containing laser drilled microvoids

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ABSTRACT

The deformation and fracture mechanisms of voided polymers and rubber-toughened polymers are an active research area where accurate fracture models still need to be proposed. The reason for the lack of accurate models stems from a lack of controlled experiments at the microscale, the scale of voids and cracks forming in polymers. In this paper, controlled experiments were carried out on thin polycarbonate sheets containing controlled arrays of laser microvoids, loaded in-situ under an optical microscope. Void dimensions were extracted experimentally and compared to a simple finite element model that was able to accurately predict macroscopic stress strain curves as well as the microscopic behavior in terms of void growth. It is shown that the local equivalent plastic strain at failure is independent of the void configuration studied and can therefore be used as a criterion for predicting fracture in voided polycarbonate under plane stress. We propose that it is not the intervoid spacing that controls fracture but the competition between intervoid strain localization and the spread of plasticity through the propagation of shear bands.

1. Introduction

Polymers have engineering applications in various industries including electronic, automotive and medicine. However their usage has been largely constrained by their tendency to fracture in a brittle manner especially at low temperatures and high strain rates. In order to increase the fracture toughness of polymers, a rubber-like phase can be introduced. Previous experimental studies [1,2] have indicated the importance of void initiation, or "cavitation", and growth at such rubber-like phases in the toughening of polymers. Cavitation leads to a reduction in the hydrostatic component of the stress tensor and bulk modulus around the void thus leading to an increase in the deviatoric component of the stress tensor and an increased plastic flow, a phenomenon referred to as 'shear yielding'. Also, a very high resistance to cavitation might lead to crazing and in turn brittle fracture. Hence, it is imperative to build knowledge about the process of void nucleation and growth in order to critically understand fracture processes in a polymer.

Deformation of polymer-rubber blends can be split into two distinct processes, (i) cavitation occurring in the rubber particles and (ii) yielding in the cavitated polymer. It was suggested that plastic void growth is the most dominant toughening mechanism in rubber-toughened polymers irrespective of the particle size (even in the "nano" regime), particularly for high bulk modulus and high rigidity particles [3]. It was also demonstrated that tests on samples containing rubber modifiers with different cavitation resistance or containing hollow plastic micro-spheres which act as pre-existing microvoids toughen epoxies in the same manner [4]. Recent studies [5–7] have therefore replaced the rubber particles by equivalent spherical voids in their numerical models. This is because post-cavitation, the soft rubber particles do not contribute significantly to the plastic deformation of the matrix and hence can be replaced by spherical voids of void fraction equal to the initial rubber volume fraction. The polymer-rubber blend thus imitates the deformation behavior of a porous polymer where plastic deformation is encouraged in the ligaments between the voids. To predict polymer fracture, it is therefore necessary to understand void growth and thus plasticity in polymers. Unlike plasticity in metals, polymers show significant softening right after yield followed by drastic hardening at large strains. The softening behavior is attributed to the evolution of large free volume associated with certain metastable states and the hardening behavior is explained by the orientation of polymeric chains and the stretching of the entangled polymer network. Therefore polymers are highly prone to strain localization, propagation of shear bands and necking. There have been models to imitate this stress-strain behavior of polymers including piece-wise linear functions [8], visco-plastic constitutive equations with a back stress based on three-

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dimensional orientation distribution of molecular chains in a non-Gaussian network [7] and a flow stress model [9].

In terms of defining fracture criteria for polymers, various macroscopic criteria based on parameters like the J-integral [10], the equivalent elongation concept for rubber-like polymers under plane stress conditions [11], strain energy [12,13] and an invariant strain criterion [14] have been defined. Since the process of cavitation, void growth and toughening along the ligaments linking the voids are all microscopic processes dependent on local stress and strain conditions rather than the far field ones, it is imperative to estimate fracture using criteria derived locally. Previous attempts to study the local deformation behavior of polymers have been done with methods such as x-ray tomography [15], x-ray scattering [16,17], digital image correlation [18,19], Moiré interferometry [20], finite element simulations [15,20] and molecular modeling [21]. However, except for a few like [17], there have not been many in-situ studies of fracture processes in polymers.

In this paper, we conducted in-situ experiments and finite element simulations on polycarbonate to study void growth and coalescence in order to extract a local fracture criterion. Polycarbonate foils containing micron-sized voids machined using a femtosecond pulsed laser were subject to in-situ tensile deformation under an optical microscope. Finite element simulations were then performed in order to extract the local stress and strain values corresponding to the failure of the material.

2. Experimental methods

Tensile coupons were machined out of a 175 μm thick, extruded, isotropic, polycarbonate film purchased from Goodfellow (*Bayer Makrofol Grade DE 1-1*). The tensile coupons had a dog-bone shape with a gage length of 5 mm and a gage width of 1 mm. Each tensile coupon had a series of cylindrical through voids machined across the width of the gage length. The voids were machined with a femtosecond pulsed laser in order to produce voids with a negligible heat affected zone. This laser machining approach has recently been developed to study fracture in metals [22–25]. The voids diameter was measured to be on average 20 μm. Six different void configurations were tested and are summarized in Table 1. The void configurations consisted of either one line with voids along the width of the sample or two lines where the voids made an angle of 45° with respect to the tensile direction. In both cases, three void spacings were considered, 50 μm, 75 μm, and 100 μm in order to obtain volume fractions of voids similar to that found in rubber-toughened polymers, i.e. 20%–40% [27]. Samples were loaded in tension with a micro-tensile tester (*MTI Instruments SEM tester18246-SEM*) at a speed of 4.8 μm/s resulting in a strain rate of approximately 1×10^{-3} s^{-1}. It should be noted that the local strain rate between the voids will be higher than the global strain rate. When the shear band forms between the voids we can assume that all the deformation is localized in the voids region. The local strain rate is then the test speed (4.8 μm/s) divided by the voids diameter (20 μm) which gives a strain rate of 0.24 s^{-1}. Based on experimental data on polycarbonate obtained by Fleck et al. [26], a change in strain rate from 1×10^{-3} s^{-1} to 0.24 s^{-1} would result in only small changes in yield strength and failure strains. Strain rate effects have therefore not been included in our study. The micro-tensile tester also allowed for both load and displacement to be recorded during the test in order to obtain a stress strain curve for each void configuration. Tensile tests were carried out in-situ under an optical microscope (*Nikon Optiphot-100*) and images were acquired during the tensile test at a rate of 3 images per second. Images acquired for each tensile test were used to extract the dimensions of each void throughout the course of the test using the ImageJ program, a Java-based open-source software (http://imagej.nih.gov/ij/). Dimensions of a bounding rectangle around each void were exported allowing for the results to be easily analyzed and compared. The main values of interest were the initial void diameter a_0 in the tensile direction and the initial void diameter in the transverse direction b_0 as well as the instantaneous void diameters a and b in the tensile and transverse directions respectively.

Table 1
Void configurations machined in the gage of polycarbonate samples.

Sample name	Condition	Smallest void spacing [μm]	Void spacing normal to tensile direction [μm]	Number of voids in sample	Void volume fraction [%]
50mic	1 line	50	50	19	40.0
50mic45	2 lines	50	71	27	28.2
75mic	1 line	75	75	13	26.7
75mic45	2 lines	75	106	17	18.9
100mic	1 line	100	100	10	20.0
100mic45	2 lines	100	141	13	14.2

3. Experimental results

Engineering stress-strain curves were obtained for each sample and are presented in Fig. 1. They show typical polymeric behavior with a linear elastic region, a yield point, and a strain softening region followed by rehardening as straining progresses. The larger the spacing between the voids, the larger the engineering failure strain is. The 75 μm and 100 μm cases have similar failure strains while the 50 μm cases have much lower failure strains. There are also no significant differences between the 1 line case and the 2 lines case with voids at 45°. A typical deformation sequence is shown in Fig. 2 where images were taken from the center of the gage length for a sample with 2 lines of voids at 45° and smallest void spacing of 75 microns. The voids are initially circular (Fig. 2(a)) and elongate as the deformation band starts forming between the voids (Fig. 2(b)). Propagation of the band through the voids results in extensive void growth (Fig. 2(c and d)). Very little growth is observed during band propagation through the gage length as most of the deformation takes place away from the voids (Fig. 2(e)). Once the band has finished propagating through the entire gage length, cracks start forming at the equator of the voids leading to the final failure of the sample (Fig. 2(f)).

Fig. 3 shows the last image recorded before final failure of the material. All configurations fail by the formation of a crack at the equator of the voids. It is interesting to note that even though the voids are spaced closer to each other for the cases with voids at 45°, fracture takes place normal to the tensile direction. This effect is particularly striking for the 50 μm case (Fig. 3(b)) where the ligament between the voids has thinned down significantly at 45° compared to the ligament normal to the tensile axis and yet the sample fails normal to the tensile axis.

Major and minor void dimensions were extracted for all void configurations and are presented in Fig. 4. The growth of the major diameter (a/a_0) initially accelerates over a small amount of plastic strain (<0.1) leading to a very steep void growth corresponding to the nucleation and propagation of the deformation band through the voids region. During this process the major diameter of the voids experiences a 3–3.5 times increase. For the 50 μm case with 1 line of voids, failure takes place before the deformation band has time to propagate fully through the line of voids. For the 50 μm case with voids at 45°, the deformation band propagates through the voids but the sample fails shortly afterwards. For all other void configurations (75 μm and 100 μm), the initial band propagation through the voids is followed by a relatively slow void growth over a large amount of strain, up to an engineering