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Low speed impact of laminated polymethylmethacrylate/adhesive/polycarbonate plates

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ABSTRACT

We study three-dimensional finite transient deformations of transparent poly-methyl-methacrylate (PMMA)/adhesive/polycarbonate (PC) laminates impacted at low speed by a hemispherical nosed rigid cylinder using the commercial finite element (FE) software LS-DYNA. The two glassy polymers PMMA and PC are modeled as thermo-elasto-visco-plastic materials by using the constitutive relation proposed by Mulliken and Boyce and modified by Varghese and Batra. For the nearly incompressible viscoelastic bonding layer, the elastic response is modeled by the Ogden relation and the viscous response by the Prony series. Delamination at interfaces between the adhesive and the polymeric sheets is simulated by using the cohesive zone model incorporated in LS-DYNA. The effective plastic strain, the maximum principal stress, and the maximum stretch based failure criteria are used for delineating failure in PC, PMMA and the adhesive, respectively. Failed elements are deleted from the analysis domain. The three layers are discretized by using 8-node brick elements and integrals over elements are numerically evaluated by using a reduced Gauss integration rule. The coupled nonlinear ordinary differential equations obtained by the Galerkin approximation are integrated by using the conditionally stable explicit algorithm. Results have been computed for at least two different FE meshes. The computed number and configurations of cracks in the PMMA are found to qualitatively agree with the test observations. It is also found that the energy dissipated due to plastic deformations in the PC is considerably more than that due to cracks formed in the PMMA.

1. Introduction

Polymers are composed of long chains of monomers while a metal is generally a polycrystalline material. This difference in the microstructure explains why their thermo-mechanical response is quite different. Polymers usually exhibit strong strain-rate dependence in their mechanical response and are widely used as transparent armors because of their high specific impact performance, e.g., see Radin and Goldsmith [1]. Sands et al. [2] have reported that polymethylmetha-acrylate (PMMA) and polycarbonate (PC) polymers have better impact resistance than most glasses. Rabinowitz et al. [3] used a high pressure torsion test to experimentally investigate the effect of pressure on the quasi-static shear stress-shear strain response and on the fracture strain of PMMA and poly(ethylene terephthalate). They found that an increase in the hydrostatic pressure increases the yield strain, the yield stress and the fracture stress of the materials but decreases the fracture strain.

The mechanical behavior of glassy polymers has been experimentally and computationally studied by several investigators. Duckett et al. [4] experimentally studied the strain rate and pressure dependence of the yield stress of PMMA and poly (ethylene terephthalate) deformed in uniaxial compression at strain rates from 10^{-6}/s to 10^{-1}/s. They found that the yield stresses in compression and torsion increase monotonically with increasing strain-rate and decreasing temperature. This was confirmed by Arruda et al. [5] in the same range of strain rates and by Mullliken and Boyce [6] for low to high strain rates (up to 10^3/s). Similar conclusions about the dependence of the yield strain and the yield stress upon the strain rate and the temperature hold for the PC material [6–11]. Moy et al. [8] performed uniaxial compression tests on Lexan 9034A PC from 10^{-4}/s to 4600/s strain rates. They used an Instron test machine for the quasi-static tests and a split-Hopkinson (Kolsky) bar for the high strain rate experiments. Rittel et al. [12] studied heating of PMMA samples subjected to uniaxial compressive cyclic loading at stress levels above the yield stress of the PMMA and at strain rates up to 0.1/s. They related the temperature rise to the chain mobility of the polymers. Rittel [13] and Rittel and Rabinowitz [14] performed uniaxial tensile and

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cyclic tests on PMMA and PC at low strain rates and measured the temperature rise.

Schmachtenberg et al. [15] proposed a method to find values of material parameters for modeling the mechanical behavior of thermoplastics. Arruda et al. [5] performed uniaxial compression tests on PMMA samples at various strain rates and temperatures, and developed a material model for the PMMA based on the multiplicative decomposition of the deformation gradient into elastic, thermal and inelastic parts. Moreover, they assumed that a part of the energy dissipated due to inelastic deformations is converted into heating while the remainder contributes to the back stress. Richeton et al. [9–11] used shift factors for the strain rate and the stresses to model the temperature and strain rate dependence of the yield stress of amorphous polymers. Richeton et al. [16] and Tervoort et al. [17] employed a spring in series with a dashpot and a Langevin spring to derive constitutive equations for polymers. Buckley and Jones [18] modeled the behavior of amorphous polymers near the glass transition temperature by decomposing the stress into two components, a bond stretching stress (partially irreversible) and a reversible conformational stress, and introduced a shift factor for the temperature and the strain rate dependence of the elastic moduli. Alavarado-Contreras et al. [19] used a simple chemical description of the amorphous and the crystalline phases to derive equations describing the mechanical behavior of amorphous semicrystalline materials. An eight-chain model is used for the amorphous phase and the stress tensor for this phase is the sum of a conformational stress tensor and a back-stress tensor. The deformation of the crystalline phase is assumed to be driven by effective shear stresses along its eight slip systems. Furthermore, they introduced two scalar variables that represent damage in each phase, proposed their evolution relations, and considered the loss of stiffness of the material due to progressive damage prior to failure. Boyce et al. [20] and Mulliken and Boyce [6] introduced a rate dependent model for the behavior of glassy polymers that assumes the coexistence of three different phases with the same values of the deformation gradient at a material point. Two of these phases have similar constitutive equations with different values of material parameters, and the third phase contributes to the general stress via a back-stress component. This material model has been generalized by Varghese and Batra [21,22] to account for the dependence of the elastic moduli upon the temperature and temperature dependent strain softening. Predictions from this work for uniaxial tensile and compressive deformations compare well with the test results at low and high strain rates.

The damage initiation and evolution, and the failure of polymers have been studied experimentally and numerically. Lajtai [23] studied the failure of pre-cracked specimens subjected to compressive loading and used a modified Coulomb model to study the final stages of their failure. Saghafi et al. [24] proposed a criterion to predict mode II and mixed-mode fracture toughness of brittle materials. Using data obtained from three-point bend tests on marble specimens they found that their criterion predicted better the failure of specimens than that given by the maximum tangential stress (MTS) criterion. Seweryn [25] proposed a criterion for brittle fracture of structures with sharp notches based on modes I, II and III stress intensity factors. Seweryn and Lukaszewicz [26] analyzed crack initiation in brittle specimens with V-shaped notches under mixed mode loading and found good agreement between the experimental measurements and the J-integral computed with the boundary element method. Vandenberghe et al. [27] proposed a model of crack formation in PMMA plates impacted at normal incidence. They considered energy dissipated due to crack formation and for flexural deformations of petals formed in the cracked plate. They could accurately predict the number of cracks formed in the plate as a function of the impact speed and the plate thickness.

Schultz [28] has related the failure of semicrystalline polymers to the spherulitic structure of the material and the inter- and intra-spherulitic fracture. Zaïri et al. [29] proposed constitutive equations to describe the progressive void growth in elasto-viscoplastic polymers, and showed that their model could predict well the stress–strain response of rubber-modified PMMA deformed in uniaxial tension. Ayatollahi et al. [30] performed fracture tests on semi-circular PMMA specimens containing an edge crack and modeled crack propagation using the finite element method (FEM) and the MTS failure criterion. They accurately predicted the crack trajectory in semi-circular bend (SCB) specimens. For small pre-existing cracks in an infinite PMMA medium, Beaumont et al. [31] investigated the relation between the crack propagation speed and the stress intensity factor at the crack tip for mode I failure of sharply-notched specimens. Wada et al. [32] experimentally studied the impact fracture toughness of edge-cracked PMMA specimens and computationally analyzed their 3D deformations using the FEM. Marshall et al. [33] introduced a factor to account for the notch size in the calculation of the energy release-rate for the PMMA material, which enabled them to derive a material specific energy/area that is independent of the specimen geometry. In their work the ratio of the initial crack length to the width of the sample ranged between 0.03 and 0.5. Moy et al. [34] found that the failure of PMMA is ductile at strain rates $\ll 1/s$ and brittle at strain rates $\gg 1/s$. Weerasooriya et al.'s [35] test data have revealed that the fracture toughness of PMMA for strain rates >100/s is almost twice that for quasi-static loading.

The failure modes of PC are quite different from those of PMMA. Chang and Chu [36] study the effect of temperature and notch-tip radius on the fracture mode of PC and of a modified PC. They found the existence of a semi-ductile fracture mode at low temperatures ($-40\,^{\circ}C$) for some notch-tip radii and proposed a diagram describing the 2D-fracture mode of PC as a function of the temperature and of the notch-tip radius. Mills [37] performed Charpy impact tests on notched PC bars and studied the effect of annealing on the ductile to brittle failure transitions. Fraser and Ward [38] investigated the effect of the notch-tip radius on the impact fracture behavior of PC samples. Allen et al. [39] measured the Charpy impact strength of notched polydiancarbonate and found that the polymer exists in two different varieties with different yield and failure properties, but did not relate this difference to the morphology of the polymers. Rittel et al. [40] tested cracked specimens and delineated two failure mechanisms, opening and shear banding, in PC as a function of the mode-mixity. Rittel and Levin [41] used two different experimental set-ups to study crack propagation in PC with either mode-I dominant or mode-II dominant deformations. They found that the same mechanisms govern the failure of PC regardless of the mode-mixity. Curran et al. [42] analyzed the fracture of PC disks subjected to dynamic flat-plate impact and predicted the level of damage induced in the PC sample with a damage model based on the nucleation, growth and coalescence of cracks. Plati and Williams [43,44] compared the energy release rate of different polymers including PMMA and PC obtained with Charpy and Izod tests at different temperatures, and found that they gave essentially the same value of the energy release rate. Adams et al. [45] used three-point bend specimens made of five different polymers (PC, polyacetal, two nylons and PMMA) to measure the plane strain fracture toughness of these materials under impact loading. Ogihara et al. [46] examined fracture mechanisms of polymeric materials by high velocity impact and quasistatic perforation tests and compared the static and the dynamic perforation energies of PMMA plates with various edge lengths and thicknesses. Fleck et al. [47] pointed out that craze nucleation was the principal failure mechanism for both PMMA and PC at high strain rates despite PMMA exhibiting brittle failure and PC ductile failure. Livingstone et al. [48] and Richards et al. [49] used a bulk strain