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Orientation-dependent compression/tension asymmetry of short glass fiber reinforced polypropylene: Deformation, damage and failure

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ABSTRACT

Short glass fiber reinforced polypropylene (sgf-PP) is increasingly employed in structural components which are subjected to a variety of loading conditions including tensile, compressive and bending loading modes. Since typical industrial components exhibit a wide range of fiber orientation distributions, their mechanical response to these loading conditions is also highly anisotropic. In this paper, the compression/tension asymmetry in the stress–strain behavior of sgf-PP is investigated from a macroscopic engineering and a micro-mechanisms of deformation and failure point of view for specimens with varying, precisely defined fiber orientations. Furthermore, we performed volume strain measurements and two-cyclic tests. We used the results to deduce the onset of damage due to cavitational mechanisms under tension and compared this to the onset of deviation of the tensile from the compressive stress–strain behavior. The results showed a good correlation for specimens with high fiber orientation, whereas for specimens with low fiber orientation results deviate due to the high deviatoric matrix volume strain contribution.

1. Introduction

Using short glass fiber reinforced polypropylene (sgf-PP) in structural components which are subjected to a variety of loading conditions involving tensile, compressive and bending loading modes is becoming increasingly popular. The corresponding micro-modes of deformation and the associated damage onset and evolution are essential factors governing failure and lifetimes in composites such as sgf-PP. Regarding the different loading modes, it is well known that the stress–strain response of plastics generally differs under tension and compression [1–4]. This is frequently referred to as *compression/tension asymmetry* and may be caused by a number of different mechanisms depending on the specific type of material.

In neat PP, the main mechanism of irreversible (plastic) deformation under tension is the formation of cavitations by crazing, followed by crystallographic slip associated with shear yielding. In contrast, the higher stresses (around 30%, [5]) and strains reached in the irreversible deformation regime under compression are attributed to the negative pressure component, which inhibits the formation of cavitations by crazing [3]. Compared to unreinforced plastics, in the case of sgf materials, the situation may be different and is likely more complex for the following two main reasons. First, the triaxiality of the compressive stress state of the matrix on a local scale caused by the constraint of the fibers may also affect the compression/tension asymmetry. Second, and perhaps more importantly, the FOD both on a global and local scale is expected to influence the compression/tension asymmetry. The mechanical response of various discontinuous and short glass fiber reinforced polymers in tension and compression has been investigated and compared, with some studies also looking into the micro-mechanisms of deformation and failure associated with the two loading modes (e.g. [6–10]). Analogously, higher stresses and strains were found in compressive testing of sgf-PP. Several studies detected later onsets of damage for compression than for tension [9,11]. This asymmetric behavior may again be attributed to differences in the contribution of voiding and shear effects, in this instance at the fiber–matrix interface or in the matrix. Contrary to neat PP, the formation of cavitations can possibly occur in compression in sgf-PP by fiber buckling mechanisms [12].

In general, the main micro-mechanisms of deformation and failure in sgf-PP can be attributed to the matrix (i.e., shear yielding, crazing and micro-cracking), the matrix–fiber interphase (debonding and fiber pull-out), or the fiber itself (fiber breakage). The char-

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acteristic contribution of the respective damage mechanism depends on fiber length, interface, ductility of the matrix, fiber orientation and, as discussed above, on the loading mode. In sgf-PP, fiber length is typically shorter than the critical fiber length, even under perfect interface conditions. Thus, the stress cannot be fully transferred to the fiber, and failure will typically occur in the matrix and in the fiber–matrix interphase rather than in the fiber itself. The interface, which is usually improved by adding a coupling agent, plays a crucial role in the ability of the matrix to transfer stress to the stiffer fibers [13]. The complex and wide range of local fiber orientations in industrial components is a result of local melt flow conditions in terms of shear and extensional flow gradients developing in the injection molding process [14]. For simplicity, however, two major orientation states causing distinctly different deformation and damage behavior can be defined, which are main fiber orientation (1) in the loading direction and (2) perpendicular to the loading direction. The relevant micro-mechanisms of deformation and failure under tensile loading, on the one hand, have been investigated by several studies (e.g., [9,11,13,15,10,16,17]). For specimens with fiber orientation (FO) parallel to the loading direction, a non-planar macroscopic crack profile is typically observed. Microscopically, matrix cracks begin at fiber-ends because (i) local shear stresses are maximal there, and (ii) the fiber-ends are created by fracturing during processing and are thus inherently not surface-treated. In well-bonded systems, major matrix deformation is suppressed by the tri-axiality of the stress state, which frequently leads to brittle matrix failure by a coalescence of cracks circumventing the fibers, and final failure by fiber pull-out [12,18,19]. In tensile testing of specimens with fiber orientation perpendicular to the loading direction, macroscopic failure occurs parallel to the fibers. On a microscopic scale, cracks and crazes begin at fiber ends and come together accompanied by ductile deformation of the matrix around the fiber surfaces [12]. The deformation and failure behaviors of sgf-PP under purely compressive loads, on the other hand, have not yet been studied systematically and in detail. In a study by Greenhalgh [12] buckling was observed for single fibers oriented in the loading direction, but not on a global scale. Thus, similarly to the neat matrix, the material was observed to fail in a shear deformation mode.

Highly sophisticated, equipment-intensive approaches are suggested in the literature to directly characterize and identify the different types of damage micro-mechanisms and the onset and kinetics of damage evolution. Non-destructive methods based on acoustic emissions have been used to determine the evolution of damage, and acoustic emission pattern recognition techniques have been applied to investigate the corresponding damage mechanisms [13,15,20,21]. Among others, μ-CT experiments [22] and small-angle X-ray refraction techniques [18,23] have attracted attention. However, less cost-intensive methods for examining deformation behavior and drawing conclusions on application-relevant microscopic damage are available. Several authors [24–28] have studied the dilatometric behavior in order to characterize debonding in particulate- and fiber-filled polymers. Other studies [10,28,29] focused on the changes in global mechanical properties (such as stiffness, Poisson's ratio and residual strain) as functions of stress and strain.

The lack of a comprehensive investigation and description of orientation effects on the compression/tension asymmetry of sgf-PP prompted us to undertake the present study. The focus of this paper was on the mechanical response of sgf-PP under tension and compression, with special focus on the compression/tension asymmetric behavior as a function of mean fiber orientation. More precisely, we discuss differences in the macroscopic properties (E-modulus, strength and stress–strain behavior) and details of the micro-mechanisms of deformation and failure as well as damage evolution, by analyzing the volume strain trends, the multi-cycle test results and the fracture surfaces of specimens tested under tension and compression.

2. Experimental

2.1. Materials and specimens

The material investigated in this study was Fibremod™ GD301FE, a commercial sgf-PP supplied by Borealis Polyolefine GmbH (Linz, A) with a sgf content of 32 wt% and an added chemical coupling agent for fiber matrix adhesion improvement. For comparative purposes, specimens of the same sgf-PP but without coupling agent and the unreinforced neat PP matrix were also included in the investigation.

A 4 mm thick plate (150×210 mm) was injection-molded via a fan end-gate at the shorter plate edge. For tensile testing, 5A specimens according to ISO 572-2 were milled out at different positions of the plate and at different angles (0°, 45° and 90°) relative to the main melt flow direction (MFD) (see Fig. 1). Compression specimens with the dimensions $10 \times 10 \times 4$ mm were milled from the middle of the corresponding tensile specimens. At these locations μ-CT experiments were performed to accurately characterize the fiber orientation distribution (FOD). The out-of-plan fiber orientation was found to be negligible (significantly below 0.05) and thus a 2D FOD is assumed. The 2D FOD can be represented mathematically by the 2×2 orientation tensor, and is best illustrated by orientation ellipses [30,31]. The average fiber orientation probability in a given direction is the intercept point of the respective axis with the orientation ellipses (designated as p^0 in the loading direction). For 0° and 90° nominal fiber orientations, this parameter corresponds to the first entry of the orientation tensor A_{11}. The fiber orientation probability in loading direction was found to be a key property governing mechanical behavior in tensile testing both in terms of modulus and strength, irrespective of the detailed layer composition. Similar results were also shown by Hine et al. (2014) who reported a linear correlation of the modulus as well as of the strength with the averaged fourth order orientation tensor [32,33]. Fig. 1 shows the average orientation in terms of the first entry of the orientation tensor (A_{11}) and the corresponding orientation ellipses for the different plate positions, where the orientation is highest (i.e., the degree of fiber alignment to the loading direction is highest) at the plate edges and lowest at the center. Not only the average orientation decreases but also the degree of layer structure increases toward the center as shown in Fig. 2, where the FOD across the thickness is plotted in terms of the A_{11} entries relative to the MFD for all plate positions. This is reflected by the shape of the orientation ellipses with lower ratios of minor to major axis indicating a more perfect orientation with little layer structure (for further details as to FOD characterization see [33]). In addition, Table 1 gives an overview of the A_{11} and p^0 values of the specimen extracted from the different positions of the plate and at 0°, 45° and 90° to the main melt flow direction.

2.2. Mechanical testing and data reduction

All mechanical tests were performed on a Zwick Z020 universal testing machine (Zwick GmbH & Co. KG; Ulm, Germany) at a strain rate of 0.001 s^{-1} and at an ambient temperature of 23 °C. For all tension and compression tests, the longitudinal force, the full-field strain on the front surface (x and y direction) and also on the side surface of the specimen (x and z direction) were measured by means of digital image correlation (DIC) with two cameras (ARAMIS 12M by GOM; Gesellschaft für optische Messtechnik mbH; Braunschweig, Germany) (see Fig. 3a). At least three speci-