IMPACT OF THE PROCESS-INDUCED MICROSTRUCTURE ON THE MECHANICAL PERFORMANCE OF INJECTION MOLDED LONG GLASS FIBER REINFORCED POLYPROPYLENE

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Abstract

The deformation of the material during injection molding of fiber filled composites causes a process-induced change in the fiber configuration. The local fiber orientation, fiber concentration, and fiber length within the molded part varies in thickness direction and along the flow path. This heterogeneous fiber microstructure inevitably results in anisotropic and locally varying mechanical properties.

This paper presents a detailed experimental analysis of the microstructure of long glass fiber reinforced polypropylene (PP) plates and its influence on the mechanical properties. Large and thin center-gated plates are injection molded with three different nominal fiber concentrations (20, 40, and 60 wt%) and an initial fiber length of 15 mm. The analysis comprises local fiber orientation, fiber concentration, and fiber length measurements conducted by means of advanced measurement techniques, including micro-computed tomography (µCT) and digital image processing. Tensile test results reveal the correlation between the process-induced fiber configuration and the mechanical properties.

The results of this experimental study verify a distinct seven-layered fiber orientation pattern for industry relevant nominal fiber concentrations. Besides a nominal fiber concentration and flow length dependent reduction of the average fiber length, the measurements suggest a non-uniform fiber orientation and fiber concentration distribution through the part thickness and along the flow path. Tensile test results show that tensile modulus increases with nominal fiber concentration, whereas tensile strength does not increase above 40 wt%. The process-induced fiber configuration causes a larger degree of anisotropy of the mechanical performance in high fiber-filled components (40 wt% and 60 wt%).

Introduction

Over the last few years, injection molding of long fiber-reinforced thermoplastics (LFTs) has gained importance in the automotive industry due to their potential for lightweight applications. LFT components show advanced mechanical properties, for instance improved stiffness and strength, and can be produced cost-efficiently in large quantities. Their potential to replace metal components in order to reduce the vehicle weight makes this material class a key aspect for the industry’s endeavor of reducing CO₂ emissions.

However, the deformation of the material during processing results in fiber alignment, fiber-matrix separation, and fiber breakage, which can result into a heterogeneous microstructure and non-uniform mechanical properties [1]. These phenomena have been addressed by several research groups and the common conclusions are that the fiber configuration changes as a function of material combination, process conditions and part design. The three main microstructural properties are the local fiber orientation, fiber concentration, and the average fiber length. Although these properties have been addressed mostly separately, the results indicate a strong correlation, which is not yet fully understood [2]–[4].

Crucial mechanical characteristics in the LFT part design phase, such as tensile strength and modulus, are greatly dependent on fiber orientation, fiber concentration distribution and the average fiber length in the final part [5]–[9]. Since the impact of fiber configuration inside the part on the mechanical behavior is complex and also not yet fully understood, it is crucial to comprehensively study the process-induced fiber configuration [10], [11].

This paper aims to correlate the fiber configuration inside injection molded LFT components with the mechanical properties determined through tensile tests. Local changes in fiber orientation and fiber concentration through the part thickness and along the flow path were examined by means of µCT-analyses. The local average fiber length is studied through image processing analyses of dispersed fibers obtained through pyrolysis of cut-out samples.

Fiber Orientation

During injection into a center-gated cavity, fibers within the polymer melt are exposed to fountain, shear, and extensional flow [12]–[14]. Shear and velocity gradients force the fibers to rotate and align along lines of similar velocities (isotachs) and in the planes of shear [15], [16]. Hence, the majority of fiber is aligned in the flow plane [17]–[19].

The fountain flow effect at the propagating melt front transports fibers from the core outwards to the cavity walls, where the melt skin unrolls onto and solidifies immediately before shear flow is able to orient the fibers
in flow direction [12], [14]. This creates a thin skin layer at the cavity walls with a random in-plane fiber orientation [20], [21]. Following melt is forced to flow between the solidified skin layers [14]. Fibers shortly behind the flow front are exposed to high velocity gradients and the resulting shear stresses at the interface between the solidified skin and the moving, still fluid core [22]. The result of the occurring shear flow is a zone of very marked fiber alignment in flow direction, referred to as shell layer [12], [20], [21]. Extensional flow due to the radial melt expansion in a center-gated cavity causes a cross-flow oriented core layer [13], [20], [21], [23]. Between shell layers and the core layer are thin transition layers where fibers are aligned randomly [14], [23]. These hydrodynamic conditions inside the mold result in a specific fiber orientation pattern with seven layers through the thickness of a part [14], [24], [25].

The fiber orientation has a large impact on the mechanical properties of injection molded parts [18], [24], [26]–[31]. It induces substantial anisotropy due to the fact that fibers realize their superior reinforcing characteristics along their orientation axis. Consequently, a high fiber orientation in the direction of loading improves the modulus and strength of components. [28], [31], [32].

The two Eulerian angles \( \phi \) and \( \theta \) describe the orientation of a single fiber in a 3D space (Figure 1). The unit vector \( p(\phi, \theta) \) represents the direction of the fiber axis and is defined as follows:

\[
p = \begin{pmatrix} p_1 \\ p_2 \\ p_3 \end{pmatrix} = \begin{pmatrix} \cos \phi \sin \theta \\ \sin \phi \sin \theta \\ \cos \theta \end{pmatrix}
\]

Figure 1. Fiber orientation in three-dimensional space described by Eulerian angles \( \phi \) and \( \theta \) and the unit vector \( p \).

Since fiber reinforced parts contain a fiber population rather than single diluted fibers, it is cumbersome to describe the orientation of each single fiber in the volume of interest using the angle pair \( (\phi, \theta) \). The average orientation of a variety of fibers in a discrete volume has to be depicted more concisely, as proposed by Advani and Tucker III in terms of even-order orientation tensors [33]. One advantage of this method is that the orientation state at any point in space can be described by tensors, which immensely reduces the computational requirements needed for numerical calculations [14], [33].

The symmetric second order orientation tensor \( a_{ij} \) represents an average orientation of all fibers in a discrete volume and serves as the physical interpretation for the orientation state of the fibers [4].

\[
a_{ij} = \int_{\phi=0}^{2\pi} \int_{\theta=0}^{\pi} p_1 p_j \psi(\theta, \phi) \sin \theta \, d\theta \, d\phi
\]

\[
= \begin{pmatrix} a_{11} & a_{12} & a_{13} \\ a_{21} & a_{22} & a_{23} \\ a_{31} & a_{32} & a_{33} \end{pmatrix}
\]

with \( i, j = 1, 2, 3 \)

Due to the symmetry of the tensors \( (a_{ij} = a_{ji}) \) and the normalization condition \( (a_{11} + a_{22} + a_{33} = 1) \), the second order tensor has five independent components [12], [23], [34]. The six orientation tensor components are defined as follows:

\[
a_{11} = (\sin^2 \theta \cos^2 \phi) \quad a_{12} = a_{21} = (\sin^2 \theta \cos \phi \sin \phi) \\
a_{22} = (\sin^2 \theta \sin^2 \phi) \quad a_{13} = a_{31} = (\sin \theta \cos \phi \sin \theta \cos \phi) \\
a_{33} = (\cos^2 \theta) \quad a_{23} = a_{32} = (\sin \theta \cos \theta \sin \phi)
\]

The angle brackets \( ( ) \) indicate the average orientation state of \( p \), weighted by the probability distribution function \( \psi \) [14].

To present fiber orientation distributions through the part thickness, the tensor components \( a_{11}, a_{22}, \) and \( a_{33} \) suffice for a physical interpretation. By convention, the coordinate system is chosen so that axis 1 represents the flow direction and axis 2 the cross-flow direction. The thickness direction is represented by axis 3. The higher the value for \( a_{11} \), the more distinct is the fiber orientation grade in flow direction. A high \( a_{22} \) value stands for a high extent of fibers aligned in cross-flow orientation. For simple part geometries, such as a center-gated plate, fibers do not align in thickness direction. The tensor component \( a_{33} \) is nearly constant over the thickness and shows a low value in this work. Hence, the tensor component \( a_{33} \) is not shown to simplify the readability of the graphs. Since \( a_{11} \) and \( a_{22} \) inherently show symmetrical but mirrored curve progressions in a fiber orientation graph, the sole indication of \( a_{11} \) is considered to be sufficient for simple part geometries [12], [18]. According to Lafranche et al. [17], [24], the following ranges of values for \( a_{11} \) indicate the different orientation states:

\[
a_{11} > 0.7 \quad \text{in-flow orientation (shell layers)} \\
0.35 < a_{11} < 0.7 \quad \text{cross-flow orientation (core layer)} \\
0.35 < a_{11} < 0.7 \quad \text{random orientation (skin and transition layers)}
\]

**Fiber Concentration**

In injection molded LFT components, a spatial variation in the fiber concentration emerges [4], [5], [35]–[38]. On the one hand, there is a fiber distribution gradient through the part thickness. On the other hand, a fiber
distribution gradient along the flow path can be observed, which is assumed to be caused by shear-induced particle migration (fiber-matrix separation) [3], [39], [40]. Variations of fiber concentration along the flow path are attributed to the flow of a molten core through the channel formed by the solidified layer at the mold wall [37]. At the interface, protruding fibers are broken or pulled out and flow within the melt flow, so that they accumulate at further flow path sections [1], [17]. In high shear rate regions, the fiber concentration is lower than in regions with low shear rates [40].

Zeng et al. reported a fiber concentration up to 50 % above the nominal fiber concentration in the core layer of 30 wt% long glass fiber reinforced thermoplastic plaques [9]. Recently presented results by Goris et al. show 6 % lower fiber concentrations at the beginning and 14 % higher concentrations at the end of the flow path, as well as up to 42.3 % higher fiber concentrations in the core of 40 wt% long glass fiber-reinforced PP plaques [4].

Fiber Length

During all stages of the injection molding process, the reinforcing fibers are exposed to buckling forces, bending and shearing stresses [14]. The high hydrodynamic forces inside the flow and the present fiber-fiber, fiber-wall and fiber-polymer interactions cause severe fiber breakage [41].

In literature many authors report fiber breakage in LFT components in the range of approximately 80 % up to 97 % of the initial fiber length [31], [41]–[43]. The fibers in the molded part have various lengths resulting in a fiber length distribution with a large peak at shorter lengths followed by a tail towards longer fibers [27]. Short glass fibers with a length below 1 mm primary improve the composite’s elastic modulus (stiffness), whereas long fibers in the range of 1 to 10 mm increase the strength and impact properties [44]–[46]. Overall, fiber length is an important factor that defines the performance of the molded part and current trends in the field of discontinuous fiber-reinforced composites aim to use longer fibers in order to increase the reinforcing characteristics of this material class.

Materials

The parameters given in Table 1 were used to injection mold three trials of center-gated plates with a 1600 ton hydraulic KraussMaffei injection molding machine equipped with a fiber-preserving long fiber injection screw (Ø = 120 mm) designed for LFT injection molding. Each part was manufactured with long glass fiber reinforced PP with an initial fiber length of 15 mm and three different nominal fiber concentrations. The material used in this work is commercially available LFT material, provided by SABIC (SABIC, Riyadh, Saudi Arabia), which came pelletized at a fiber weight concentration of 20, 40, and 60 wt% (SABIC® STAMAX 20YM240, 40YM240 and 60YM240).

Table 1. Injection molding parameters used for all trials.

<table>
<thead>
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<td>Mold temperature</td>
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</tr>
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Figure 2 shows the 610 mm x 610 mm x 2.3 mm center-gated plate geometry and the spatial arrangement of the test specimens. For the microstructural analysis 12 round samples (Ø = 30 mm) and 15 ASTM D638 type 1 [47] tensile testing specimens per plate were cut out in a specific cutting pattern by means of a water jet cutter. The tensile test specimens were located so that their center has a distance of 200 mm to the gate and were used for mechanical characterization by means of tensile testing. They are oriented in three different cutting orientations, namely 0°, 45° and 90° to the radial flow direction. There are five specimens for each orientation per plate to show the repeatability of the mechanical tests. The round samples were used for the characterization of the fiber configuration in thickness direction as well as along the flow path. Four samples each were cut out in three directions along the flow path. Samples 1 to 4 are positioned in 50.00, 113.75, 177.50 and 241.25 mm distance to the gate.

Figure 2. Center-gated injection molded plate (top view) and displayed water jet cutting scheme with 12 round samples (Ø = 30 mm) and 15 tensile test specimens (ASTM D638 type 1 [47]) in three different orientations to the flow direction.
The measurements included one plate per trial and the local fiber orientation, fiber concentration and fiber length distribution was analyzed. Two samples from additional plates were investigated to validate the results. The mechanical properties were determined from five tensile test specimens for each orientation (0°, 45° and 90°). Six further tensile test specimens per orientation from additional plates were tested to validate the tensile tests.

Methods

Fiber Orientation

Fiber orientation measurements were non-destructively conducted by µCT-scan-analysis. The cut out round samples were scanned at a resolution of 8 µm with a Zeiss Metrotom 800 µCT-scanner (Carl Zeiss AG, Oberkochen, Germany). Table 2 shows the most important scan parameters.

<table>
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<td>Current</td>
<td>[A]</td>
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<tr>
<td>Voxel size</td>
<td>[µm]</td>
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</table>

The microstructure was examined from the resulting 3D-data set by means of the commercially available software package VGStudio MAX 3.0 (Volume Graphics GmbH, Heidelberg, Germany) [4]. The scanned 3D-volume was registered to a global coordinate system with the flow direction along the 1-axis. The sample plane was registered parallel to the 1-2-plane, so that the 3-axis represents the thickness direction of the sample. Fiber and matrix material are distinguished by appropriate thresholding. This step converts the greyscale representation into a binary file. Fibers are defined by white pixels and the polymer matrix is represented by black pixels.

The analyzed volume of each sample has a dimension of 10.5 x 10.5 x 2.3 mm³. The sample thickness (2.3 mm) was subdivided into 50 layers to obtain a finely resolved sampling of the fiber concentration and orientation distribution across the thickness. For each cell of this analysis discretization (mesh) the fiber orientation tensors $a_{11}$, $a_{22}$, and $a_{33}$ were evaluated. The fiber orientation of 12 round samples was analyzed and for each of the four positions along the flow path three samples were averaged.

Fiber Concentration Distribution

The same µCT-data sets were used to study the fiber concentration distribution across the part thickness. The µCT-data of all 12 round samples per plate was also analyzed with the previously described µCT-3D-file analysis software package. For this purpose, the same threshold and mesh generated for the fiber orientation analysis were used. From the resulting binary cross sectional images a quantitative comparison of white to black pixels leads to a fiber volume fraction for each cross section. Thus, an average fiber concentration distribution across the thickness can be visualized. Due to the ambiguity in determining the threshold value, the fiber volume concentration has to be normalized, as described in [4].

Fiber Length

Once the samples were scanned in the µCT-scanner, the destructive test method of measuring the fiber length was performed on the same samples. Furthermore, the fiber length distributions of purged material samples for each trial were measured. Hence, the process-induced fiber breakage that occurred during plasticating can be separated from the additional fiber attrition during mold filling.

The applied fiber length measurement technique in this study was developed at the Polymer Engineering Center (PEC) at the University of Wisconsin-Madison, USA. The technique is a further development of existing methods, based on ISO 22314 and is also derived from the method Kunc et al. introduced [48, 49].

For fiber length measurements fibers and polymer matrix material were separated by pyrolysis at 520 °C for 3 h. After the pyrolysis in an electrical muffle furnace the incinerated sample mat contains more than 500,000 fibers. Since this amount of fibers would imply a laborious measuring process with lots of measuring runs, the sample size was reduced by a downsampling step according to Kunc et al. [48]. This downsampling method ensures the extraction of a representative amount of fibers from the round samples at a reasonable measuring effort.

The downsampling begins with the insertion of ultraviolet (UV) light curable epoxy (Bondic GmbH, Cologne, Germany) exactly into the center of the fiber stack by means of an injection syringe. Once the needle has completely penetrated the fiber mat, it is slowly pulled out while simultaneously pushing out the epoxy at a constant rate. Immediately after the injection, UV light is used to cure the epoxy. A uniform epoxy cylinder with a diameter of approximately 2 to 3 mm remains inside the fiber bed. The cylinder-shaped epoxy column contains a specific amount of fibers from the entire sample thickness. After the removal of loose fibers, the epoxy/fiber bond contains from 10,000 to 50,000 fibers. A second 1.5 h long pyrolysis burns off the epoxy at 500 °C.
The fibers are then dispersed on glass plates (290 mm x 215 mm x 2.3 mm) by means of a turbulent air flow generated by a short air blast in an enclosed chamber and funnel system. This dispersion method offers an essential time advantage compared to manual fiber dispersion per brush [38]. Subsequently, the glass plate with the dispersed fibers is scanned with a flatbed scanner at a resolution of 1200 dpi and 8-bit greyscale. The resulting image is then used to measure the length of each fiber on the glass plate. The plate size allows analyzing several thousand fibers in one step.

A fully automated image processing algorithm, developed at the PEC, was applied to detect single fibers and to determine their individual lengths [50]. Even bent and intersecting fibers can be detected automatically with the PEC algorithm, which is particularly important for long fiber samples.

The downsampling step results in a biased measurement since the cured epoxy cylinder embeds longer fibers more likely than shorter ones. For this reason, a correction formula, proposed by Kunc et al. [48], was applied to the measured fiber length values.

Tensile Testing

The tensile test specimens were stored and tested at room temperature of 21 °C and 46 % ± 2 % relative humidity. The specimens were tested according to the ASTM D638-14 standard [47] using an Instron 5967 tensile testing machine (Instron, Norwood, MA, USA). The machine was fitted with a 30 kN load cell, and operated at a crosshead speed of 5 mm/min. An Instron 2630-113 extensometer with a gauge length of 50 mm was used to measure the exact strain until failure for modulus determination. Tensile modulus was determined from the slope of the stress-strain curve between 0.0005 and 0.0025 mm strain. Tensile strength was calculated at the point of maximum tensile stress.

Results

Fiber Orientation Analysis

All fiber orientation results exhibit the expected seven-layered orientation pattern, referred to as core-shell-skin structure (Figure 3).

The values of orientation tensor component $a_{11}$ indicate the fiber orientation in flow direction and have a minimum in the core layer and a maximum in each of the two shell layers for all investigated trials (Figure 4). The degree of fiber orientation decreases in the skin layer between the shell maxima and the surfaces. The core layer is bounded by a relatively rapid change in fiber orientation that manifests in a steep ascent of the orientation tensor $a_{11}$. Between the core edge and the shell layer the fiber orientation shows tensor values of approximately 0.5, which indicates a random in-plane orientation. Overall, this suggests that the samples have a fiber orientation layer structure with a pronounced cross-flow oriented core layer and a shell layer with high fiber alignment in flow direction.

Figure 3. Seven-layered microstructure illustrated with selected µCT images showing the fiber orientation in each layer of a 60 wt% sample.

The graphs in Figure 4 illustrate the fiber orientation distribution through the thickness and along the flow path for each trial. To improve the readability, the graphs solely show the value of orientation tensor component $a_{11}$ and the corresponding minimum and maximum values are displayed.

Figure 4. Fiber orientation patterns through the part thickness along the flow path of trial 1, 2, and 3 (average from three samples per position).
The shape in the core layer of the fiber orientation curves of each trial changes with propagating flow path. For each trial, the core layer is significantly lower and wider at position 1 than at position 4. Additionally, the core becomes narrower along the flow path. Apart from this, the shell peak in all trials is consistently sharper at the beginning of the flow path and becomes blunter with distance to the gate. The fiber orientation in flow direction in the shell layer is significantly higher in the 40 and 60 wt% trials, whereas the core layer exhibits comparable orientation values in all three trials.

Overall, the results suggest that the maximal orientation values seem to be independent from the flow path and that the nominal fiber concentration has an impact on the degree of fiber alignment. In particular, the curve shape of the fiber orientation in the core reveals differences between all three trials. It can be observed that the core layer is more extended with higher nominal fiber concentration. Furthermore, the curve slopes between core and shell layer become steeper along the flow path.

**Fiber Concentration Analysis**

Figure 5 summarizes the fiber distribution variations through the sample thickness and along the flow path of all trials. Independent from the nominal fiber concentration, all trials show fiber agglomerations in the core layer. Additionally, a secondary fiber concentration peak in proximity to the part surface can be identified. It can be observed that the region around the primary center peak becomes narrower along the flow path, whereas the area around the secondary peak becomes wider and the peak itself less sharp. All primary and secondary peak values are indicated in the figures.

The fiber distribution peaks in the center of the 20 wt% trial increase from position 1 to position 4 showing a maximum of 51 % more fibers in the core at position 4. At position 4 the primary peak is particularly pronounced, but narrow. Outside the core layer the fiber concentration is below the nominal fiber concentration with a minimal fiber concentration near to the surface in the skin layer. The secondary peaks are less pronounced than the primary peak in the core.

The 40 wt% trial exhibits the highest secondary peaks with a maximum of 72 % more fibers than the nominal fiber concentration close to the gate, but the height of the secondary peaks decreases along the flow path. The primary fiber concentration peak in the core layer slightly increases. In comparison to the 40 wt% trial, the fiber agglomerations and depletions are less distinct at 60 wt%, but show qualitatively the same pattern.

It is clearly visible that the primary peak becomes wider and less high (blunter) with increasing nominal fiber concentration and narrower along the flow path. The 20 and 40 wt% trials show increasing fiber concentrations in the core layer along the flow path. The 40 and 60 wt% trials exhibit decreasing fiber concentrations in the shell layer along the flow path.

**Fiber Length Analysis**

Figure 6 summarizes the measured weight average fiber length ($L_w$) of the purged material and along the flow path for all three trials. As expected, the fiber length of all 3 trials reduces when the material is injected into the cavity. A pronounced reduction in residual fiber length is visible for 20 wt% trial from the purged material (P) to the location close to the gate (1) with a reduction from 4.88 mm to 3.44 mm. The higher the nominal fiber concentration, the more severe the fiber length reduction in the plasticating zone. Along the flow path from position 1 to position 4 the average fiber length changes in each trial. The 20 and 40 wt% trials show an initial fiber length reduction in the cavity and a subsequent slight increase at the end of the flow path. The fiber length of trial 3 constantly declines inside the cavity flow.

The measured fiber length of the purged material shows that during the plastication the initial fiber length of 15 mm is severely reduced. The weight average fiber length of the 20 wt% trial is reduced to 4.88 mm, the 40 wt% trial and the 60 wt% trial have remaining fiber lengths before the cavity of 2.66 and 1.98 mm, respectively. This means that the initial fiber length is already reduced by more than 67 % in the 20 wt% trial. The initial fiber length reduction accounts for more than 82 % and nearly 87 % in the 40 wt% and 60 wt% trial, respectively.

The minimal fiber length measured in trial 1 at position 3 (2.77 mm) stands for a fiber length reduction of further 43 % and an overall reduction of approximately 82 %. In the trials with higher nominal fiber concentrations these results are even more pronounced. In
trial 2, the fiber length reduces from 2.66 mm to 1.31 mm at position 2. This is a reduction of approximately 51% inside the cavity and more than 91% with respect to the initial fiber length. Trial 3 shows overall the most severe fiber breakage. The fiber length decreases from 1.98 mm in the purged material to 0.92 mm at position 4, which represents a lengths reduction of about 54% inside the cavity and nearly 94% in total.

**Figure 6.** Fiber length variations along the flow path for trial 1, 2, and 3 (average from three measurements per position; P indicates the average fiber length of the respective purged material).

**Mechanical Properties**

The tensile test results are shown in Figure 7 and Figure 8. The tensile strength of the 20 wt% trial does not show a substantial anisotropy. In contrast, the highest tensile strength is reached with the 0° orientation specimens in the trials at higher fiber concentrations. The 0° specimens of the 40 wt% trial show the highest tensile strength values, whereas the 90° specimens of trial 3 reaches the lowest tensile strength. The 40 wt% trial shows a higher tensile strength in the 90° specimens than in the 45° specimens.

**Figure 7.** Tensile stress for three different tensile test specimen orientations (0°, 45°, and 90° to the flow direction) of trial 1, 2, and 3 (average of 11 specimens per orientation).

**Figure 8** shows that the tensile modulus increases at elevated nominal fiber concentrations. The measurements of the 20 wt% trial show isotropic results while the moduli for 40 wt% and 60 wt% are anisotropic with the highest values for specimens extracted in flow direction.

**Figure 8.** Elastic modulus for three different tensile test specimen orientations (0°, 45°, and 90° to the flow direction) of trial 1, 2, and 3 (average of 11 specimens per orientation).

**Discussion**

From the tensile test results, it can be seen that the mechanical performance of fiber-reinforced components depends on the nominal fiber concentration and specimen orientation to the load direction. Confirming Thomason’s results, it can be observed that the tensile modulus continuously increases with fiber concentration while the tensile strength does not increase above 40 wt% [6], [29]. In addition, it can be noted that the load direction (0°, 45° or 90°) affects the tensile strength and modulus of trials with nominal fiber concentrations of 40 wt% and 60 wt%, whereas the 20 wt% trial shows little to no anisotropic behavior.

The tensile modulus for trial 1 is on average approx. 39% lower when compared to trial 2 and 3. However, the tensile strength is only 18% lower. This might be explained by the fact that trial 1 exhibits substantially longer fibers. Tensile strength is more strongly affected by the length of the fibers than tensile modulus.

Additionally, the degree of fiber orientation and fiber concentration in the shell layers is less pronounced than in the high fiber-filled trials. Thus, 20 wt% plates have larger areas with random fiber orientation (α11 ≈ 0.5), which explains the more isotropic results for trial 1.

The 60 wt% trial is exposed to more severe fiber breakage due to elevated fiber-fiber, fiber-wall and fiber-polymer interactions. The average fiber length is below 1 mm and average aspect ratio of approx. 50, which is smaller than the critical fiber length [14]. This and a decreased interfacial adhesion between the fibers and the matrix at 60 wt% might explain the lower tensile strength compared to the 40 wt% trial [6], [29]. However, the high
amount of fibers consequently led to the highest elastic modulus.

In higher fiber-filled trials there is fiber depletion between the fiber agglomerations in core and shell layers. Especially in trial 2, the majority of fibers is located in the shell and has the highest measured orientation degree in flow direction. The other large part of the fibers is located in the core and aligned in cross-flow direction. Since fibers oriented perpendicular to the load direction do not contribute to the improvement of the tensile performance, the fiber orientation inside the tensile test specimens plays a major role.

Once a sample is water jet cut in 45° to the flow direction, the majority of the fibers is oriented in +45° or -45° to the load direction. This explains the varying tensile strength values measured for the different cut out orientations of the tensile test specimen in trial 2 and 3. In the 90° specimens, the fibers in the core layer are directly oriented in stress direction. For this reason, the 90° oriented specimens in the 40 wt% trial reach higher tensile strength than the 45° oriented ones of the same trial. Although the core becomes thicker, this explanation is not valid for trial 3, in which the 90° specimens show the lowest tensile strength. This might be also explainable by a more inferior fiber-matrix adhesion in the core layer at 60 wt%. Additionally, the fibers at the surface of the 90° specimens are oriented perpendicular to the load direction. Both effects could facilitate potential cracks through notch effect.

Conclusion and Outlook

The results of this study addressed the process-induced microstructure of LFT injection molded components at varying nominal fiber concentrations and its impact on the mechanical properties. The measurements confirm the seven-layered fiber orientation structure described in the literature. Besides flow length-dependent changes in fiber length, the results verify variations of fiber orientation and fiber concentration distributions across the part thickness and along the flow path. It has been shown that these microstructural properties are highly dependent on the nominal fiber concentration and indicate a correlation between one another.

Increased nominal fiber concentration generated wider core layers and caused elevated fiber breakage, which was most pronounced in the 60 wt% trial. Furthermore, the core layer thickness, the average fiber length and the fiber concentration in the shell layer diminished in all trials with propagating distance to the gate. Higher nominal fiber concentrations (40 or 60 wt%) also led to higher fiber orientation grades in the shell layer compared to 20 wt% fiber-filled plates. In contrast to the 60 wt% trial, the 20 and 40 wt% trials showed increasing fiber concentrations in the core along the flow path.

The tensile test results indicate the significant influence of the composite’s local inner microstructure on the mechanical performance. They show that the mechanical properties vary depending on nominal fiber concentration. Beyond that, the tensile performance shows a larger degree of anisotropy in high fiber-filled components (40 and 60 wt%).

Overall, the results of this work show the importance of comprehensive material characterization that also addresses the correlating microstructural properties. Full microstructure analyses support the understanding of the phenomena present in LFT injection molding, which would ultimately improve the crash worthiness prediction of automotive components.

A more comprehensive understanding of the interaction of the microstructure and the injection molding processing conditions could be acquired by investigation of further experimental trials at constant nominal fiber concentrations, but varying processing parameters. Eventually, this can be used to test available predictive tools in software packages that estimate the fiber microstructure for injection molded parts.

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