Mixed numerical-experimental identification of non-local characteristics of random-fibre-reinforced composites

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Abstract

In the present study, the relationship between the initial microstructure, the specific damage mechanisms and the length parameter which arises in higher-order continuum models, is investigated for three materials with a gradient-enhanced damage model. For this purpose, a series of experiments has been carried out on polypropylene-based composite materials with different glass-fibre lengths or microstructures. The process zone close to the crack tip has been investigated with a digital image correlation technique, in which images of the deformed process zone are digitized and stored. A correlation analysis between the successive images permits quantification of the displacements in the process zone. An inverse mixed numerical-experimental approach then permits the determination of the length parameter and the governing parameters for damage development. The influence of the microstructural characteristics of each material and the failure mechanisms on the parameters has been scrutinized. Conclusions are drawn with respect to the physical nature of some non-local aspects, which relate the transforming microstructure to the macroscopically observed behaviour. © 1999 Elsevier Science Ltd. All rights reserved.

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1. Introduction

Knowledge of the relationship between the microstructure and the macroscopic mechanical properties of a material is a prerequisite for modern material design. Many complex microstructural phenomena underlie the non-linear mechanical behaviour. The introduction of non-local models in non-linear mechanics was initially introduced as a remedy for the ill-posed boundary value problem which is a consequence of softening behaviour. However, experimental investigations have shown that non-locality has a physical origin which is closely related to the transforming microstructure of the material. A large class of non-local models is nowadays available: non-local damage [1,2], non-local plasticity [3], gradient-enhanced damage [4–7], gradient-dependent plasticity [8–10], Cosserat media and others. For an overview the reader is referred to de Borst et al. [5]. From a macroscopic point of view, these models have a common characteristic: the mechanical behaviour of the material is related to the microstructure through the incorporation of an intrinsic length scale. The present contribution focuses on the relationship between this length scale and certain microstructural aspects in composites.

The experimental validation of a constitutive model is an important issue in computational mechanics. The combined use of sophisticated measurement techniques, computational simulations and a numerical identification tool for the model parameters is necessary for proper material characterization.

The present application is the characterization of the ultimate behaviour of some random glass-fibre-reinforced polypropylene composites which are used commercially in the automotive industry. Mechanical characteristics and details regarding the failure behaviour of these materials can be found in the literature. [11–14] The microstructures of these materials differ by virtue of the different fibre lengths and distributions, as well as through the effects of the manufacturing process. For the analysis of these materials, a perfect random-in-plane distribution of the fibres will be assumed. In this
contribution three different materials have been investigated. A gradient-enhanced damage model has been used as proposed by Geers et al. [7] The parameter estimation has been carried out by using a mixed numerical–experimental method. For this purpose, experimental displacement fields have been determined in the process zone close to the notch of a compact-tension specimen by means of a digital image correlation technique. Mixed numerical–experimental techniques are frequently used for the characterization of metals, [15] but are rarely applied in combination with non-local constitutive models.

2. Computational model

The damage and fracture behaviour of the composites has been modelled with a transient gradient-damage model as presented by Geers et al. [7] In this transient gradient-enhanced model, the effect of the length parameter on the governing equations depends on the local strain state of the material. This technique has a stabilizing influence during the unloading stage of the material in the vicinity of the crack. The boundary value problem is defined by the following set of equations: [7]

\[ \nabla \cdot \sigma + \hat{f} = 0 \quad (1) \]

\[ \tilde{\varepsilon}_{\text{eq}} - \partial \nabla^2 \tilde{\varepsilon}_{\text{eq}} = \varepsilon_{\text{eq}} \quad (2) \]

complemented by the appropriate boundary conditions. The Cauchy stress tensor \( \sigma \) is related to the infinitesimal strain tensor \( \varepsilon \) via a classical isotropic continuum damage approach:

\[ \sigma = (1 - D) \mathbf{C} : \varepsilon \quad (3) \]

\( 4 \mathbf{C} \) being the fourth-order elastic stiffness tensor, \( D \) the internal scalar damage variable and \( \varepsilon_{\text{eq}} \) the local equivalent strain. If the non-local equivalent strain is denoted by \( \tilde{\varepsilon}_{\text{eq}} \) in Eq. (2), while \( \partial \) represents the gradient activity which is a variable with the dimensions of length squared, which is closely related to the internal length scale.

A third equation, \( \zeta = \partial \), has to be added to the system in order to carry out the discretization and linearization of the weak forms of Eqs. (1) and (2), since the gradient of \( \zeta \) needs to be computed. The solution is found by discretizing the displacements \( \tilde{u} \), the non-local equivalent strain \( \tilde{\varepsilon}_{\text{eq}} \) and the gradient activity \( \zeta \).

The local damage equivalent strain \( \varepsilon_{\text{eq}} \), is derived from the local strain tensor. The present definition of \( \varepsilon_{\text{eq}} \) is based on the positive principal strain components \( \varepsilon^+ \) [16]:

\[ \varepsilon_{\text{eq}} = \sqrt{\sum_{j=1,2,3} (\varepsilon^+)^2} \quad (4) \]

A history parameter \( \kappa \) is computed during the entire loading history, which equals the largest value of the non-local equivalent strain \( \tilde{\varepsilon}_{\text{eq}} \) that the material has experienced during its loading history. Next, the damage variable \( D \) can be computed through the use of a damage evolution law \( D(\kappa) \). The damage evolution is described by a power law:

\[ D = 1 - \left( \frac{\kappa / \kappa_c}{\kappa / \kappa_c} \right)^\beta \quad (5) \]

The threshold value for damage initiation \( \kappa_i \) is the ultimate non-local equivalent strain that characterizes the elastic material behaviour prior to the occurrence of damage. The critical history parameter \( \kappa_c \), characterizes the value of \( \kappa \), for which the damage \( D \) reaches its ultimate value 1 (0 \( \leq D \leq 1 \)). The exponent \( \beta \) influences the slope and the shape of the stress–strain softening curve [17]. The gradient activity \( \partial \) in Eq. (2), which controls the non-local effect, satisfies:

\[ \partial = \begin{cases} c \left( \frac{\varepsilon_{\text{eq}}}{\varepsilon_{\text{eq}}^*} \right) & \text{if } \varepsilon_{\text{eq}} \leq \varepsilon_{\text{eq}}^* \\ c & \text{if } \varepsilon_{\text{eq}} > \varepsilon_{\text{eq}}^* \end{cases} \quad (6) \]

The gradient activity \( \partial \) has an upper bound that equals the gradient parameter \( c \). An internal length scale, \( l_c \) (related to the microstructure), can now be defined as the square root of \( c \). This limit of \( \partial \) is reached at the local equivalent strain \( \varepsilon_{\text{eq}} \). The gradient parameter \( c \) is related to the square of the length parameter. Previous analyses have shown that this evolution of \( \partial \) avoids spurious damage widening, mainly during the unloading stage of previously damaged material [7,18]. It has been found that the model response is not very sensitive to small variations of this evolution law. A fixed value of \( \varepsilon_{\text{eq}} = 0.15 \) has therefore been adopted in the present analysis. More computational details can be found in Geers et al. [7].

3. Experimental analysis

3.1. Digital image correlation

The digital image correlation (DIC) method is a recently developed experimental technique which permits a fast and straightforward identification of displacement fields. The method is based on the correlation of grey values of successive digital images of the undeformed and the deformed specimen. During the experiment,
images are recorded using a CCD (charged coupled device) camera at different deformation stages. A schematic overview of the experimental set-up is shown in Fig. 1.

The images are stored and the actual processing of recorded data is done after the experiment. The specimen must have a random speckle pattern on its surface in order to obtain grey value distributions in the image. If such a speckle pattern is not present it can be created by lightly spraying some paint onto the surface. The displacement of an arbitrary material point of the image can be determined by correlating a subset of pixels between two images [19–21]. One of the important advantages of the method is that the discretization of the actual measurement points is carried out after the experiment. This provides a large flexibility with respect to the choice of the proper location of each measurement point. An example of the recorded DIC-images and the used measurement grid is shown in Fig. 2.

During data processing, the quality of the correlation is analysed and decorrelated measurement points are removed from the subsequent analysis. The accuracy depends on the resolution of the camera, the quality of the speckle pattern, surface conditions during deformation and other minor influences. The experiments described here were carried out with a Pulnix TM-765 CCD camera and a Matrox Magic framegrabber, which permits recording of images with a size of 768×512 pixels with 256 grey levels each. Additionally, rigid displacements of the specimen are subtracted from the measured displacement fields, as these phenomena (due to unavoidable effects in the loading clamps) are not present in the numerical simulations. The best accuracy which can be expected from this method is 0.05 pixels, but the accumulation of errors from image to image, as well as the grey pattern changes during deformation may reduce this accuracy considerably. In this case, an error of 1 μm to a maximum of 40 μm is expected. An exact error analysis of the measurement technique is difficult and estimates for the standard deviation of the error may be misleading, since the real errors are partially problem and material-dependent.

3.2. Experimental results

The experimental analysis has been carried out on glass-fibre-reinforced polypropylene (GFPP) plates with three different microstructures. Compact-tension specimens with a characteristic size of 50 mm and a total notch depth of 22.5 mm were used to carry out the fracture tests (see Fig. 3 for geometrical characteristics and coordinate axes). The deformations in this test are typically non-uniform, where high deformation gradients occur in front of the notch of the specimen. The test, described in the ASTM Standards E399, [22] and E647, [23] has been adapted for composite materials by Williams and Cawood [24]. The specimen is fixed at the top and the bottom by a clevis and pin assembly. The specimen and loading arrangement is used for tensile loads only. In the test, the lower pin is kept fixed while the displacement of the upper pin is prescribed in time.

![Fig. 1. Experimental set-up for the Digital Image Correlation method.](image1)

![Fig. 2. CDD-image of a composite compact-tension specimen used for digital image correlation. The measurement grid is indicated on the image.](image2)

![Fig. 3. Geometry of the compact-tension test with a graphical representation of a measured strain field at the notch.](image3)
In the present analysis, the displacement fields are determined on a square grid of 100 measurement points as shown in Figs. 2 and 3. Rigid rotations and translations of the specimen in the loading clamps are subtracted from the measured displacement fields, as these phenomena (which completely depend on the boundary conditions) are not present in the numerical simulations.

Table 1 summarizes the principal characteristics and denotations for the composites that have been investigated: two types of glass-mat thermoplastics (GMT) and one short-glass-fibre-reinforced injection moulding compound. All materials are characterized by a quasi random-in-plane fibre distribution with a glass fibre content of 30% weight. The material GMT–SM (Azdel® PM11300) is based on a swirled glass mat which is melt impregnated on a belt-press, while the material GMT–PM (Azdel® PD3243) is based on the paper-making process.[25] The third material, a short-fibre-reinforced thermoplastic SFRTP (DSM® Stanylan P), is a glass-fibre-reinforced-injection moulded compound.

The assumption of random fibre distribution is best fulfilled for the GMT–SM material. The GMT–PM plates may show a preferential fibre orientation in the rolling direction during the drying, heating, consolidation and cooling of the paper-making route for the production of the material. Simple tensile tests [26] showed that the GMT–PM material presents a pronounced orthotropic behaviour. All specimens were therefore tested perpendicular to the rolling direction, which is the weakest direction. The SFRTP material likely shows a different fibre distribution through the thickness of the specimen, but it is assumed that the in-plane isotropic behaviour holds. The thickness of the specimens equalled 4 mm for the GMT–SM material and 3.8 mm for the GMT–PM and SFRTP materials. As the thickness of the available SGFPP-plates is limited, it is assumed that all the specimens are loaded in a plane-stress state. All specimens were loaded at a rate of 2.5 mm/min applied to the upper loading pin.

The global load/displacement curves are shown in Figs. 4–6, where the relative displacement between the loading pins has been plotted on the horizontal axis. The onset of the crack is generally straight, but the further propagation depends highly on the inherent local anisotropy or heterogeneity of the composite. The experimental analysis is therefore focused on the onset of damage and the initial crack propagation. Several experiments were carried out for each material and the most representative experiment is used for the mixed numerical–experimental analysis. The material GMT–SM presented the most irregular crack and the most distributed cracks in the neighbourhood of the principal crack. The corresponding load/displacement curve (total force versus loading pin displacements) reflects the irregularity of this response. Although crack patterns

Table 1
GFPP-composites that have been investigated

<table>
<thead>
<tr>
<th>Tested materials</th>
<th>Processing</th>
<th>Average fibre length</th>
</tr>
</thead>
<tbody>
<tr>
<td>GMT–SM</td>
<td>Melt impregnation, swirled glass mat</td>
<td>Continuous</td>
</tr>
<tr>
<td>GMT–PM</td>
<td>Paper-making route</td>
<td>12 mm</td>
</tr>
<tr>
<td>SFRTP</td>
<td>Injection moulded</td>
<td>0.4 mm</td>
</tr>
</tbody>
</table>

![Fig. 4. GMT–SM: F-Δu_{pin}](image1)
![Fig. 5. GMT–PM: F-Δu_{pin}](image2)
![Fig. 6. SFRTP: F-Δu_{pin}](image3)
may differ from one experiment to another, the failure mechanisms are the same ([11,13,18]). Examples of the crack paths for GMT–PM and SFRTP are given in Figs. 7 and 8, respectively.

The time interval between two samples was 4 s, which corresponds to a pin displacement of 1/6 mm. The number of samples used in the mixed numerical–experimental analysis was limited in order to avoid erroneous simulations caused by the obliqueness of the crack. These would occur, since asymmetry is not present, in the numerical simulations.

4. Mixed numerical-experimental parameter identification

Fig. 9 shows a block diagram of the mixed numerical–experimental method. In this Figure, \( y \) denotes the input to the experiment and the numerical analysis, \( \mathbf{m} \) the measured response, \( \mathbf{h} \) the calculated response, \( \mathbf{\theta} \) the parameter set and \( \mathbf{\theta}^{(0)} \) the initial estimates of the parameters.

Generally, a set of measurements (e.g. displacement fields) is collected from the experiments. These measurements (forces, displacements etc.) are stored in a column \( \mathbf{m} = [m_1, \ldots, m_N]^T \), where \( N \) is the total number of measurements.

The unknown material parameters in the constitutive model are stored in a column \( \mathbf{\theta} = [\theta_1, \ldots, \theta_P]^T \), where \( P \) is the number of parameters. A numerical model is used to calculate the response \( \mathbf{h} \) for a given parameter set \( \mathbf{\theta} \) and input \( \mathbf{y} \). The relation between the parameters \( \mathbf{\theta} \) and the measurements \( \mathbf{m} \) is written as

\[
\mathbf{m} = \mathbf{h}(\mathbf{y}, \mathbf{\theta}) + \mathbf{\xi}
\]

where \( \mathbf{\xi} \) is an error column. The input \( \mathbf{y} \) consists of prescribed forces and/or displacements. In the estimation algorithm, it is assumed that the input to the model is known exactly. In practice, the input is determined from measurements and is consequently contaminated with errors.

The aim of an estimation algorithm is the determination of a set of parameters \( \mathbf{\theta} \) for which the model response \( \mathbf{h}(\mathbf{\theta}) \) is in closest agreement with the measurements \( \mathbf{m} \). An objective function must be defined which quantifies the quality of the agreement. In the present application, a quadratic form is used:

\[
J(\mathbf{\theta}) = (\mathbf{m} - \mathbf{h}(\mathbf{y}, \mathbf{\theta}))^T \mathbf{V} (\mathbf{m} - \mathbf{h}(\mathbf{y}, \mathbf{\theta}))
\]

Here, \( \mathbf{V} \) is a positive definite symmetric weighting matrix. The entries of the matrix \( \mathbf{V} \) are used to express the confidence in the measurements and the a priori estimates, respectively, where larger values correspond to a higher confidence.

In the context of the present mixed numerical–experimental application a gradient algorithm is used for the minimization of the objective function \( J(\mathbf{\theta}) \). A Gauss–Newton algorithm is applied to minimize \( J(\mathbf{\theta}) \).

[27]

A necessary condition for a (local) minimum of \( J(\mathbf{\theta}) \) is given by

![Fig. 7. Crack at the notch of a GMT–PM CT specimen.](image)

![Fig. 8. Crack at the notch of a SFRTP CT specimen.](image)

![Fig. 9. Block diagram of the mixed numerical–experimental method.](image)
In order to obtain a true (local) minimum in the parameter space, the second derivative of \( J(\hat{\theta}) \) must be positive definite. Condition (9) results in a set of \( P \) nonlinear equations for the \( P \) unknown parameters:

\[
\frac{\partial J(\hat{\theta})}{\partial \theta}^T \left[ H(\hat{\theta}) \right] = 0
\]  

(9)

where the explicit dependence of the response \( h(\hat{\theta}) \) on the input \( y \) is omitted for notational convenience and the sensitivity matrix \( H(\hat{\theta}) \) is given by

\[
H(\hat{\theta}) = \frac{\partial h(\hat{\theta})}{\partial \theta}
\]  

(10)

A finite-difference approach [27–29] is used to approximate the sensitivity matrix \( H^{(i)} \), which must be determined for each new estimate \( \hat{\theta}^{(i)} \) of the parameter set. The parameters are perturbed one after the other and for each perturbed parameter, the response of the model is calculated. Next, the sensitivity matrix is computed from the differences between the response of the initial parameter set and the response of each perturbed parameter set.

The response of the model \( h(\hat{\theta}) \) is a nonlinear function of the parameters, and thus (10) has to be solved iteratively. It is, therefore, assumed that an estimate of the optimal parameter set is available. This set is denoted by \( \hat{\theta}^{(0)} \) where the index \( i \) is the iteration counter. Using

\[
\hat{\theta}^{(i+1)} = \hat{\theta}^{(i)} + \delta \hat{\theta}^{(i)}
\]  

(12)

and assuming that \( \delta \hat{\theta}^{(i)} \) is small, the following approximations are made

\[
b(h^{(i)} + \delta h^{(i)}) \approx h^{(i)} + H^{(i)} \delta h^{(i)}
\]

\[
H(h^{(i)} + \delta h^{(i)}) \approx H^{(i)}
\]  

(13)

In here, \( h^{(i)} \) and \( H^{(i)} \) are used as abbreviated notations for \( h(\hat{\theta}^{(i)}) \) and \( H(\hat{\theta}^{(i)}) \). The following iterative scheme for updating the parameters \( \hat{\theta}^{(i)} \) is thus obtained:

\[
\delta \hat{\theta}^{(i)} = K^{(i-1)} \left[ H^{(i)} \left[ m - y(\hat{\theta}^{(i)}) \right] \right]
\]

\[
K^{(i)} = H^{(i)} \left[ V H^{(i)} \right]^{-1}
\]  

(14)

where \( \delta \hat{\theta}^{(i)} \) is an estimate of the least-squares solution. The identifiability condition [27] requires that the matrix \( K^{(i)} \) is regular. In the neighbourhood of the optimal solution \( \hat{\theta}_{LS} \), this scheme has quadratic convergence.

The iterative procedure is normally terminated if the change in the parameter updates is smaller than a critical value [15]. In the present analysis, it was noticed that a minimum was systematically found, but that convergence in the immediate neighbourhood of this minimum is slow. The anisotropy and heterogeneity of the used materials is definitely a complicating factor in the application of the mixed numerical–experimental method.

5. Results

In order to carry out a mixed numerical–experimental approach, the input to the experiment and the simulation \( y \) must be chosen properly. A bad choice may lead to an undesired influence of the incomplete modelling of the boundary conditions in the loading clamps. This is particularly the case for CT specimens, where ovalization and contact phenomena play an important role around the loading pin. It is therefore necessary to measure the input \( y \) directly from the specimen during the experiment. In this case the relative \( Y \)-displacement between the upper left and the lower left measurement point has been taken. Next, the numerical analysis is carried out with a Finite-element-model for the CT test [18] and the value of the input \( y \) is strictly respected through the use of an arc-length control scheme which keeps the input equal to the measured value during the incremental-iterative process. Intermediate steps are only taken if this is required for the convergence of the non-linear finite-element analysis.

One of the major difficulties in the optimization process resides in the highly nonlinear character of the constitutive model. Bad parameter estimates may lead to divergence of the minimization process of the objective function \( J(\hat{\theta}) \) and the calculation of the sensitivity matrix \( H(\hat{\theta}) \) becomes difficult. If divergence is encountered during the optimization process before the desired number of steps, the number of samples is automatically reduced to the actual number of converged steps, and the objective function is rescaled accordingly in order to permit a continuation of the optimization process. In the next iteration of the optimization process, the initial number of steps is automatically re-established.

Four parameters were identified with the presented mixed approach, namely the gradient parameter \( c \), the damage initiation history value \( k_p \), the ultimate value \( k_c \) (corresponding to \( D = 1 \)), and the parameter \( \beta \) in the damage evolution law Eq. (5). This implies that one iteration in the optimization process requires the solution of \( 5 \) finite-element-problems. The initial guess for the parameters was taken as follows: \( c^0 = 1 \) \( \text{mm}^{-2} \), \( k_p^0 = 0.01 \), \( k_c^0 = 0.5 \) and \( \beta^0 = 1 \). From the gradient parameter \( c \), a characteristic length scale \( l_c \) can be defined as the square root of \( c \). The computed values for these parameters are shown in Table 2. The remaining residues can be easily computed and the standard deviations \( s \) (for the force and \( s_y \) for the displacements) can be tabulated as shown.
in Table 3. \( F_{\text{max}} \) and \( u_{\text{max}} \) represent the maximum force and displacement, respectively.

Taking into account the heterogeneity of the materials, it may be concluded from this analysis that the method is successful in finding an optimal solution with acceptable relative errors. A more accurate prediction requires a statistical analysis of the results on a large series of experiments. In this contribution, attention is focused on the differences between the different materials. As expected, PP composites with long fibres have a much larger intrinsic length scale than short fibred composites. On the other hand, the failure mechanisms also have an important influence on the damage evolution law and \( l_c \). Fibre fracture is important for the rupture of long-fibre-reinforced PP, \([11,13]\) which has been observed in the final stage (large crack openings) with GMT–SM after the debonding and matrix failure stage. In the case of GMT–PM, matrix rupture and fibre pullout are the governing failure mechanisms \([11,13,18]\).

Figs. 10, 12 and 14 compare the computed force and the measured force versus the input \( \tilde{u} \). It is emphasized that the displacement on the horizontal axis, i.e. the input \( \tilde{u} \), is not the relative displacement between the two loading pins but the relative displacement between the upper left and the lower left measurement points indicated in Fig. 2. The forces in Figs. 10, 12 and 14 are only a limited number of the sample points that were shown in Figs. 4–6, where the pin displacement was used in the horizontal axis. The limitation of the number of samples (given in Table 2) avoids the erroneous estimation of the parameters if the crack becomes oblique (Fig. 7), since this obliqueness is not captured in the computational model. The difference in the number of samples and the different horizontal axes, explains why Figs. 10, 12 and 14 differ from Figs. 4–6 and 6. Figs. 11, 13 and 15 represent the non-decorrelated \( Y \) displacements of the points in a specific vertical cross-section, which in this case corresponds with the leftmost column of measurement points (see Figs. 2 and 3, \( X = 25 \text{ min} \)).
The macroscopic response is dominated by the intrinsic length scale and the damage evolution. A large value for \( l_c \), delays global rupture and softening, and gives a distributed damage pattern. The effect on the macroscopic load becomes clear in Fig. 10. The failure of the injection moulded specimen is matrix dominant combined with pull-out of the very short fibres. A small length scale is found, while the damage evolution law seems not so brittle. However, the macroscopically observed toughness or brittleness of the material depends both on the length scale and the damage evolution. This can be noticed in Fig. 12 where a more brittle response is found for GMT–PM than for GMT–SM in Fig. 10. The initial damage threshold, \( \kappa_i \), appears to be larger for GMT–PM than for the two other materials. This observation becomes apparent by comparing Fig. 13 with Figs. 11 and 15. From this analysis, it may be concluded that the intrinsic length scale depends on the length of the fibres in the material. Yet, the observed correlation between \( l_c \) and the fibre length cannot be explained from geometrical considerations only. The dominant failure mechanism in the composite plays an important role in the modification of the
microstructure and its mechanical resistance, and affects all parameters in the model.

It can be noticed from the evolution of the displacements in Figs. 11, 13 and 15 that symmetry is lost in the experiment, while it remains preserved in the simulation. Evidently, loss of symmetry is due to the heterogeneous character of a composite which is more pronounced in the long fibre reinforced GMT–SM. This difference is one of the major causes of the remaining deviations in Table 3.

The quality of the agreement can also be checked visually by representing some full field results. Complete displacement fields can be easily compared as shown in Figs. 16 and 17. Strain fields can be computed from these fields, [30] which gives the results shown in Figs. 18 and 19.

6. Conclusions

Using a gradient-enhanced damage model, a mixed numerical–experimental approach has been set up in order to identify the unknown model parameters which govern the damage evolution and the fracture process in a material. The procedure has been applied to three glass-fibre-reinforced polypropylene composites with different fibre lengths, microstructures and processing history. The characterization of the macroscopic behaviour of these composites with a gradient-enhanced damage model gives reliable results and provides a better insight into the relationship between macroscopic non-local constitutive models and microstructural characteristics of a material. The converged values of the model parameters have been compared with the geometrical characteristics of the fibres and the different failure mechanisms.

It has been shown that the intrinsic length scale, which is inherent in all non-local approaches, is correlated to the fibre length. The dominant failure mechanism influences both the internal length scale and the damage evolution law. The internal length scale depends on the initial undamaged microstructure and on all deformation mechanisms during the damage process which alter the microstructure in a progressive way.

References


