VISCOLASTIC BEHAVIOUR OF SHORT-FIBRE REINFORCED POLYAMIDE WITH DAMAGE

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Abstract

This article concerns the characterization of the viscoelastic behaviour of short-fibre reinforced polyamide taking into account microstructural damage. The purpose of the work is to identify the characteristic material damage function and the damage evolution during constant strain-rate tensile loading. Samples with different fibre orientations are tested in order to analyse the effect of the fibre alignment. This article focuses on the comparison of the analysis at two temperatures: 23 °C and 80 °C. A continuum damage model based on a work done by Schapery is applied. The results show that the damage is nearly zero under a certain stress and strain respectively and it intensively grows beyond this threshold. Further damage depends on both strain-rate and loading duration. The strain-rate dependency at 80 °C is similar to the one at 23 °C and lower strain-rates. The measured stress can be accurately predicted by the model.

1 Introduction

The technical use of polymeric materials demands the appropriate knowledge of their mechanical properties and deterioration. Short-fibre reinforced polyamide is widely used in technical parts as for example in the automotive industry. It is of great importance therefore to analyse the critical loading thresholds, from which on damage clearly evolves, as well as the influence of damage on the material response.

The purpose of the present work is to characterise the viscoelastic behaviour of short-glass-fibre reinforced polyamide taking into account microstructural damage due to loading. The kind of loading considered is tensile loading. The study is performed at two different temperatures: 23 °C and 80 °C.

In [1] it was shown in a qualitative damage study that the Young's modulus decreases with increasing imposed load, which is an indicator of the presence of damage.

The present article focuses on the temperature influence on the material behaviour and the damage evolution. Further it discusses the characteristic damage function of the reinforced material comparing it to the unreinforced material.

2 Constitutive Model

'Continuum Damage Mechanics' (CDM) is particularly appropriate for modelling the damage evolution in a material preceding the macroscopic fracture. Using the CDM approach the heterogeneous material is considered as homogeneous solid and the state of damage is described by internal state variables [2, 3].

Schapery [4, 5] established a basic formulation for viscoelastic response with growing damage. The microstructural changes are represented by a set of internal state variables whose evolutionary laws are motivated by considerations of viscoelastic fracture mechanics.

In Schapery's formulation it is assumed that the viscoelastic properties (creep compliance or relaxation modulus) are not influenced by damage. Damage is introduced through the so called effective stress, which bases on the principle of strain equivalence and is often used in the framework of continuum damage mechanics. The modified elastic-viscoelastic correspondence principle based on the introduction of pseudo variables is used to eliminate the time-dependence of the material. The one-dimensional formulation of this model is applied in this work to describe the mechanical behaviour of short-glass-fibre reinforced polyamide under tensile loading.

The intrinsic material behaviour of the material under study is modelled as linear viscoelastic. This approximation is considered as appropriate for this kind of loading. The strain range considered extends at room temperature up to 5 % and at 80 °C up to 3 %. The strain considered at 80 °C is smaller because the assumption of the inherent material behaviour as linear viscoelastic is not further appropriate over this limit.

At first the correspondence principle is here outlined; afterwards the problem of coupling viscoelasticity and damage is presented.

The constitutive equation for a linear viscoelastic body is characterised by the following hereditary integral:

$$\sigma = \int_0^t E(t-\tau)\frac{d\varepsilon}{d\tau}d\tau \tag{1}$$

where σ is the stress, ε is the strain and $E(t)$ is the relaxation modulus. With the introduction of a new variable, the so called pseudo-strain

$$\varepsilon^R \equiv \frac{1}{E_R}\int_0^t E(t-\tau)\frac{d\varepsilon}{d\tau}d\tau \tag{2}$$

Eq. 2 can be written as

$$\sigma = E_R \varepsilon^R \tag{3}$$

E_R is an arbitrary constant with the same dimension as the relaxation modulus and is called the reference modulus. Eq. 3 corresponds formally to the constitutive equation of an elastic material:

$$\sigma \equiv \frac{\partial W}{\partial \varepsilon} = E\varepsilon \tag{4}$$

where W is the strain energy density and E is the elastic modulus.

In order to describe the effect of damage, one internal state variable, S, is introduced and the following form for the pseudo-strain energy density is taken:

$$W^R = \frac{1}{2}C(S)\left(\varepsilon^R\right)^2 \tag{5}$$

$C(S)$ is the characteristic damage function. It describes the material degradation with increasing damage. The constitutive equation for the viscoelastic material with damage is then:

$$\sigma \equiv \frac{\partial W^R}{\partial \varepsilon^R} = C(S)\varepsilon^R \tag{6}$$

The damage evolution's equation, which was derived by Schapery from studies on fracture mechanics in viscoelastic materials, is [6]:

$$\dot{S} = \left(-\frac{\partial W^R}{\partial S}\right)^{\alpha} \tag{7}$$

where α is a material parameter, which is related to the viscoelastic properties. The damage-related constitutive parameters to be determined are the function $C(S)$ and the constant α. Furthermore the viscoelastic material function, the linear relaxation modulus $E(t)$, is to be determined in order to calculate the pseudo-strain.

3 Material and Specimen Fabrication

The polyamide type under study is PA 6.6 reinforced with 35 % of short-glass-fibres (PA 6.6-GF35).

The specimens are milled out from injection moulded plates. Due to the flow profile in the plate during the fill process the specimens show a flow induced fibre-alignment in the flow direction.

The samples are taken in three different angles with respect to the injection direction in the plate: 0 °, 45 ° and 90 °. This enables the analysis of the influence of the fibre orientation on the material behaviour and damage evolution.

The unreinforced polyamide is also tested for better understanding of the basic influence of the fibres on the behaviour of the composite.

The polyamide matrix absorbs water due to the carbon amide group in the molecule chain. The initial state of the material in the present study is dry. Therefore and to guarantee a homogeneous state of the material, the specimens were annealed and, at the same time, dried. Afterwards they were kept in an atmosphere with drying agent until the test.

4 Determination of Model Parameters

4.1 Determination of the Pseudo-Strain

In order to calculate the pseudo-strain (Eq. 2) the linear relaxation modulus is needed. This can be determined from relaxation tests or from creep tests by means of the interconversion between the linear creep compliance and the linear relaxation modulus.

For the relaxation modulus a prony series representation is taken:

$$E(t) = E_\infty + \sum_{i=1}^{N} E_i e^{-t/\tau_i} \tag{8}$$

where E_∞ is the equilibrium modulus ($\lim_{t\to\infty} E(t) = E_\infty$), E_i are the relaxation strengths and

τ_i the relaxation times. The pseudo-strain (Eq. 2) is then given by

$$\varepsilon^R = \frac{\dot{\varepsilon}}{E_R}\left\{E_\infty t + \sum_{i=1}^{N} E_i \tau_i \left(1 - e^{-t/\tau_i}\right)\right\} \qquad (9)$$

4.2 Determination of the Damage Variable and the Characteristic Damage Function

The characteristic damage function *C(S)* was determined from tensile tests at different strain-rates varying from 0,001 %/min up to 1 %/min at room temperature and from 0,01 %/min up to 1 %/min at 80 °C. In this work the following ansatz with two material parameters (*b* and *c*) is proposed:

$$C(S) = \frac{E_R}{1 + bS^c} \qquad (10)$$

Fig. 1 shows the calculated damage function *C* against the damage variable for the reinforced and the unreinforced polyamide at room temperature.

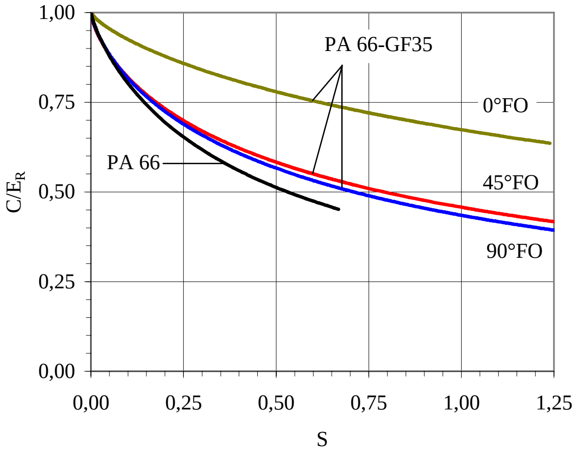

Fig. 1. Characteristic damage function versus damage variable. PA 6.6 and PA 6.6-GF35 with 0 °, 45 ° and 90 ° fibre orientation (FO), 23 °C

The damage function for the reinforced polyamide lies over the one for the unreinforced polyamide. This means that for a certain damage level due to the fibres the stiffness does not fall as much as in the case of the unreinforced polyamide.

In this regard however the fibre orientation has a major influence. The damage function for the samples with 90 ° fibre orientation is close to the one of the unreinforced polyamide. A certain damage level causes almost the same stiffness reduction in the case of loading the reinforced material transversely to the main fibre axis as in the case of the unreinforced polyamide. In contrast the damage function for the samples with 0 ° fibre orientation lies considerably over the damage functions for the samples with 45 and 90 ° fibre orientation and for the unreinforced polyamide. As long as the fibres themselves do not break down, they prevent the strong stiffness reduction as it can be seen for the other samples.

Fig. 2 and 3 show the evolution of the damage variable versus the strain for the samples with 0 ° and 90 ° fibre orientation respectively at room temperature.

It can be seen that the damage is nearly zero under a strain of approx. 1 %. From this strain value on damage grows intensively.

Moreover there is no direct dependence of the damage variable upon the strain-rate. At the low strain-rates between 0,001 %/min and 0,01 %/min the samples with 0 ° fibre orientation show no strain-rate dependency (Fig. 2) and in the case of the samples with 90 ° fibre orientation the damage grows with increasing strain-rate (Fig. 3). However at strain-rates higher than 0,01 %/min the damage variable at a certain strain decreases with increasing strain-rate.

This was already pointed out by Park et al. in [6]. The damage variable can grow with increasing strain-rate since it depends on the strain-rate through the pseudo-strain (Eq. 9). However damage depends also on the loading duration (Eq. 7). Since the loading duration decreases with increasing strain-rate, it depends on the material and the conditions which effect predominates.

Thus at the higher strain-rates over 0,01 %/min the loading duration up to a certain strain is not enough for the damage to evolve.

The strain-rate spectrum tested at room temperature in this work ranges over 3 decades. Both effects, damage increasing with strain-rate and loading duration can be observed.

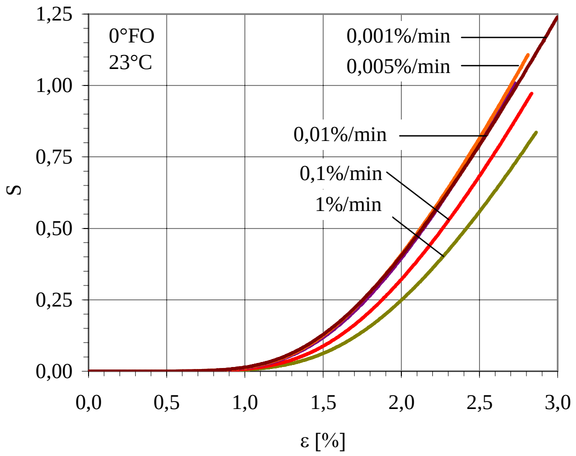

Fig. 2. Damage variable versus strain. PA 6.6-GF35, 0 ° fibre orientation, 23 °C

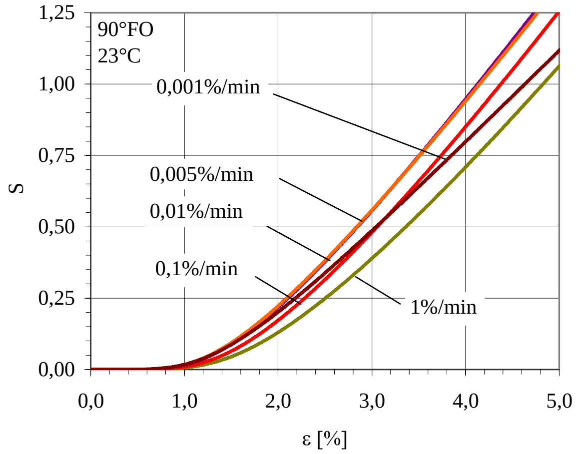

Fig. 3. Damage variable versus strain. PA 6.6-GF35, 90 ° fibre orientation, 23 °C

Fig. 4 and 5 show the damage evolution versus the strain for the samples with 0 ° and 90 ° fibre orientation respectively at 80 °C. At the higher temperature of 80 °C damage increases with increasing strain-rate in the tested range between 0,01 %/min and 1 %/min. The effect of the strain-rate prevails over the effect of the loading duration. This seems to be consistent with the results at room temperature. The behaviour at higher temperatures at a certain strain-rate or time is expected to be similar to the behaviour at lower temperatures and lower strain-rates or higher times respectively, since time and temperature are coupled variables. At 80 °C and strain-rates from 0,01 %/min up to 1 %/min damage shows the same strain-rate dependency as at 23 °C and strain-rates from 0,001 %/min up to 0,01 %/min.

Furthermore the damage variable at constant strain is considerably smaller at 80 °C than at 23 °C. Fig. 6 shows the comparison of the damage variable calculated for the strain-rate of 1 %/min at both temperatures. The temperature of 80 °C lies over the glass transition temperature, where a higher molecular motion is possible. This enables the material to elongate up to a higher strain than at room temperature suffering less damage.

At the same stress however the damage variable is higher at 80 °C than at 23 °C (see Fig. 7). The material withstands a higher stress loading at room temperature suffering less damage.

Furthermore the same damage level causes a higher stiffness reduction at 80 °C than at 23 °C. This can be seen in Fig. 8, where the characteristic damage function for the reinforced polyamide with 0 ° fibre orientation is plotted versus the damage for the two temperatures.

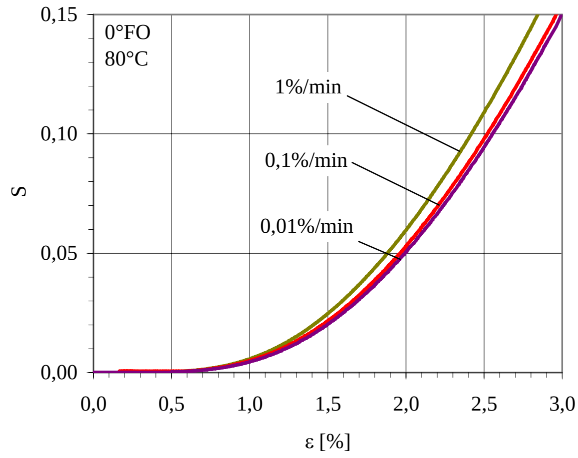

Fig. 4. Damage variable versus strain. PA 6.6-GF35, 0 ° fibre orientation, 80 °C

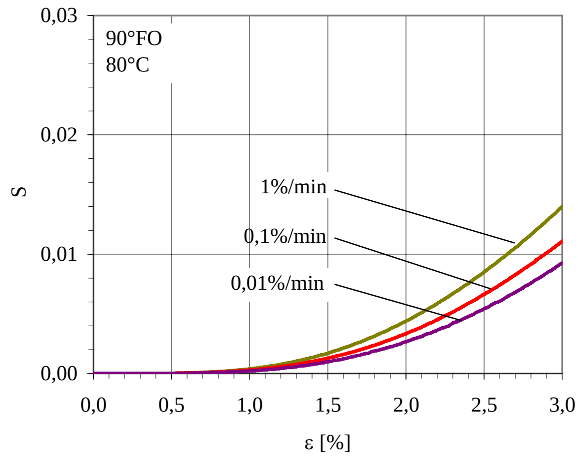

Fig. 5. Damage variable versus strain. PA 6.6-GF35, 90 ° fibre orientation, 80 °C

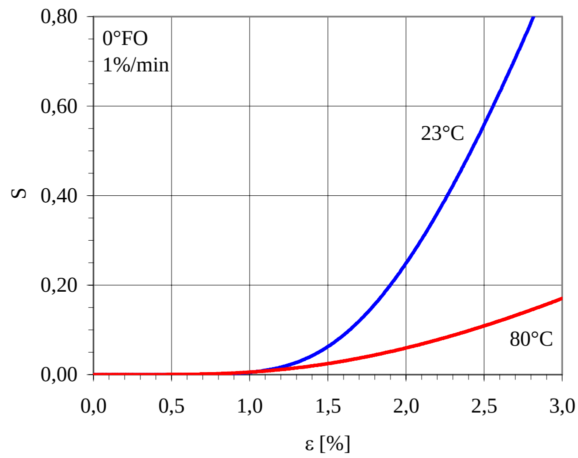

Fig. 6. Damage variable versus strain at a strain-rate of 1 %/min and two different temperatures: 23 °C and 80 °C, PA 6.6-GF35, 0 ° fibre orientation

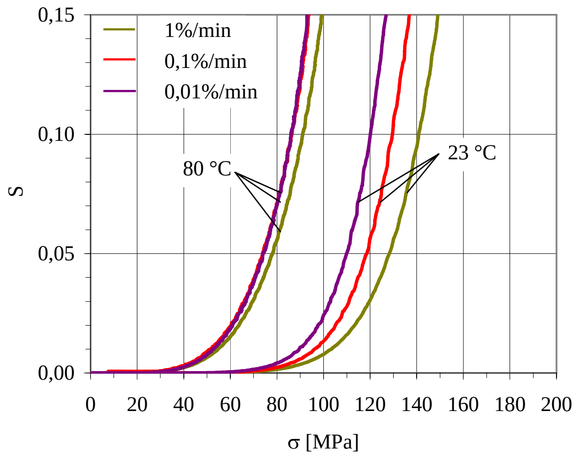

Fig. 7. Damage variable versus stress at 23 °C and 80 °C, PA 6.6-GF35, 0 ° fibre orientation

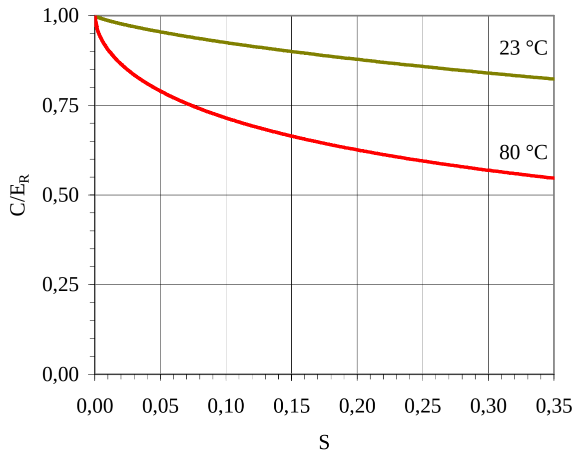

Fig. 8. Characteristic damage function versus damage variable at 23 °C and 80 °C. PA 6.6-GF35, 0 ° fibre orientation

5 Comparison of the Model Predictions and the Test Data

Fig. 9 and 10 show the comparison of the calculated and the measured stress for the samples with 0 ° and 90 ° fibre orientation respectively at room temperature. The stress can be accurately predicted by the model for the samples with 0 ° fibre orientation in the whole strain and strain-rate range. For the samples with 90 ° fibre orientation however the calculated stress deviates from the measured for the higher strain-rates at higher strains. From 3,5 % strain on the material deforms with very little increase in stress, and the model overestimates this stress. As in [1] pointed, we assume that this is due to the presence of plastic deformations at these strain-rates. Fig. 11 shows the comparison between the experiment and the model at 80 °C. The stress-strain behaviour can still be accurately described by the model at this higher temperature in the considered strain range.

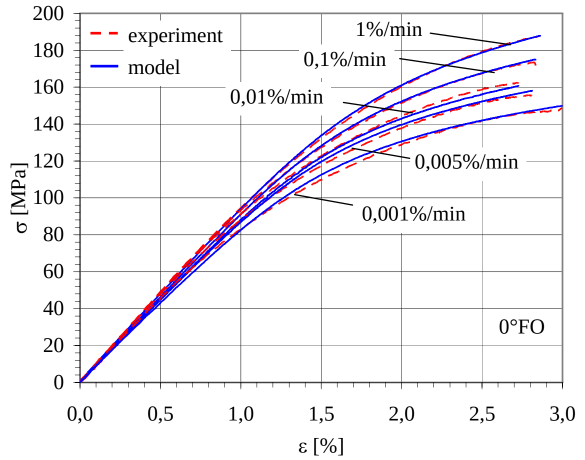

Fig. 9. Stress versus strain. Comparison of model with the experiment, PA 6.6-GF35, 0 ° fibre orientation, 23 °C

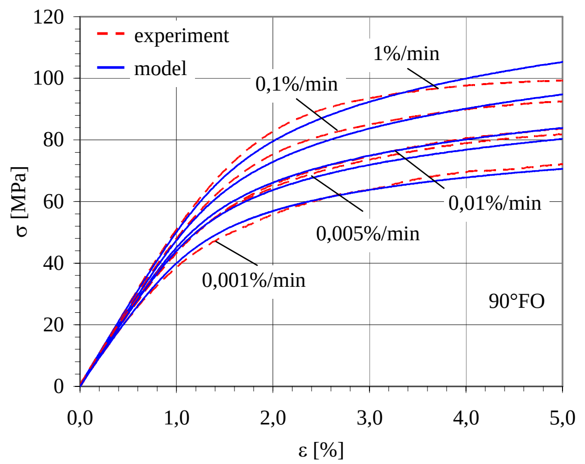

Fig. 10. Stress versus strain. Comparison of model with the experiment, PA 6.6-GF35, 90 ° fibre orientation, 23 °C

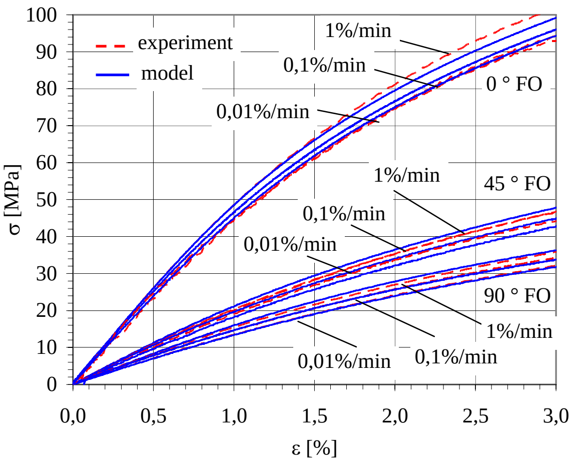

Fig. 11. Stress versus strain. Comparison of model with the experiment, PA 6.6-GF35, 0 °, 45 ° and 90 ° fibre orientation (FO), 80 °C

6 Concluding Remarks

The viscoelastic behaviour of short-glass-fibre reinforced polyamide with damage was investigated in this work. A continuum damage model was applied to describe the mechanical behaviour under extensional monotonic loading at different strain-rates. The study focuses on the comparison of the results at 23 °C and 80 °C.

It was found out that the characteristic damage function of the reinforced polyamide with 0 ° fibre orientation lies clearly over the one of the samples with 45 and 90 ° fibre orientation and the one of the unreinforced polyamide. The stiffness reduction for a given extent of damage is in the case of loading the material in the direction of the fibre axis considerably fewer.

It was shown that the damage depends on both strain-rate and loading duration. At room temperature damage grows with increasing strain-rate for rates from 0,001 %/min up to 0,01 %/min. From 0,01 %/min to 1 %/min damage decreases with increasing strain-rate, since the loading duration at these rates is not enough for the damage to evolve. In contrast, at 80 °C and strain-rates from 0,01 %/min to 1 %/min damage grows with increasing strain-rate. The effect of the higher temperature on the damage evolution corresponds to the effect of the lower strain-rates at room temperature.

Furthermore it was observed that for constant strain the damage variable is smaller at 80 °C than at room temperature. This is explained by the higher molecular motion at this temperature. The material is able to deform up to a higher strain suffering less damage. At constant stress however the damage is lower at 23 °C. The material withstands a higher stress at room temperature with less damage.

The stress-strain behaviour is accurately described by the model at room temperature. In the case of the reinforced polyamide with 90 ° fibre orientation from nearly 3,5 % strain on the model fails at the higher strain-rates since plastic deformations become relevant.

At 80 °C the model is still adequate to describe the strain-stress behaviour of the reinforced polyamide up to 3 % strain.

Acknowledgements

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