



Can plant fiber composites outperform glass and carbon fiber composites in terms of stiffness-damping trade-off?

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ABSTRACT

This study investigates the influence of fiber volume fraction on the viscoelastic behavior of flax–GreenPoxy composites, with a focus on damping characteristics relevant to vibration sensitive structural applications. Composites with 25%, 35%, and 47% fiber volume fractions are characterized using dynamic mechanical analysis, modal analysis, and quasi-static measurements, and the experimental results are interpreted in parallel with multiscale homogenization simulations. The results show that increasing the plant fiber volume fraction leads to enhancement of the longitudinal storage modulus as expected, while the damping factor remains nearly unchanged at about 1.5% over the tested range. This behavior contrasts with that of conventional synthetic fiber composites, in which increasing fiber content typically results in a significant reduction in damping. The combined experimental and numerical approach provides a consistent understanding of how fiber content governs the viscoelastic response of plant fiber composites. Furthermore, analysis of the loss modulus demonstrates that plant fiber composites exhibit a more favorable stiffness–damping trade-off, outperforming glass fiber composites while approaching the dynamic performance of carbon fiber composites, highlighting their potential for vibration-sensitive structural applications.

1. Introduction

The growing demand for sustainable materials has led to increased interest in plant fibers as reinforcements in polymer composites. Compared to synthetic fibers such as glass or carbon, plant fibers offer advantages including low density, renewability, and reduced environmental impact. However, these benefits come at the expense of lower modulus, moisture sensitivity, and greater variability in properties [1–4]. Despite their limitations, plant fiber composites have shown promise in structural applications. While current industrial use is mostly limited to non-critical components like automotive interiors and sporting goods [5], recent studies suggest that bast fiber composites, particularly those made from flax, hemp, jute, and kenaf, are suitable for load bearing applications such as wind turbine blades due to their relatively high specific strength and stiffness [6–8].

Beyond static strength and stiffness, many advanced applications require materials with well controlled dynamic mechanical behavior. Properties such as damping capacity influence vibration resistance, noise suppression, and fatigue performance, making plant fiber composites attractive for structural applications. They offer an advantage in this regard, as they exhibit higher intrinsic damping than glass or

carbon based composites [9–12]. However, this benefit strongly depends on microstructural parameters such as fiber orientation, volume fraction, fiber–matrix adhesion, and processing quality [13–18].

Among these parameters, fiber volume fraction has been the subject of several studies. In most cases, increasing fiber content enhances the stiffness of fiber reinforced composites [19,20], but its effect on damping is more complex. In synthetic fiber composites, Adams et al. [21] reported that the longitudinal damping of unidirectional carbon and glass fiber composites decreases rapidly with increasing fiber volume fraction. Etaati et al. [22] reported that in short hemp/PP composites, damping initially increases with fiber content up to around 30 wt% due to enhanced dissipation at poorly bonded fiber–matrix interfaces, but decreases at higher fractions as the behavior becomes increasingly governed by the fibers. Rahman et al. [15] studied flax/PP composites where no significant differences are observed between 31%, 40%, and 50% volume fractions, likely due to the presence of more fibers creating additional fiber–matrix interfaces. Da Silva et al. [23] reported that the loss factor of kenaf/epoxy composites decreased as the fiber volume fraction increased from 10 to 30%, owing to improved fiber–matrix interfacial adhesion and the restricted mobility of the epoxy matrix. Nevertheless, the influence of fiber volume fraction in continuous fiber reinforced thermoset composites remains far less explored than

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in thermoplastic counterparts, raising the question of whether similar trends apply.

To fully exploit the potential of plant-based composites, quantitative links between microstructure and macroscopic behavior must be established. Experimental characterization provides valuable insight into the viscoelastic response, with dynamic mechanical tests covering broad frequency ranges and revealing sensitivities to defects such as voids or fiber misalignment, which are often overlooked in simulations [11]. Nevertheless, these results are typically limited to specific materials and conditions, making it difficult to generalize the observed behavior or to explore a wider range of design parameters. In addition to experimental approaches, analytical models have been developed to predict the viscoelastic behavior of composite materials. Notably, Berthelot and Sefrani [24] proposed formulations to estimate transversal damping based on the respective contributions of the fiber and matrix phases; within the same framework, the contribution of the fibers to longitudinal damping is commonly neglected in unidirectional polymer–matrix composites reinforced with glass or carbon fibers due to their very low intrinsic damping, although this assumption leads to a significant underestimation of the damping observed experimentally at the composite scale. These models provide insight into the role of constituent properties and fiber volume fraction in determining the composite response, yet their validity for plant fiber composites remains uncertain, as they were primarily developed for conventional synthetic composites and rely on simplifying assumptions that may not hold for plant-based materials, whose complex microstructure, intrinsic variability, and interfacial effects can significantly influence damping behavior.

As a result, a gap remains between experimentally observed behavior and model predictions, particularly for plant fiber composites. Existing approaches struggle to capture the wide variability of damping properties and to predict their evolution over a broader range of material parameters. This limitation motivates the use of multiscale modeling, which provides a framework to predict viscoelastic properties from microstructural data and to assess the influence of material and processing parameters [25], in combination with experimental data to evaluate the stiffness–damping response across different composites.

The aim of this study is to investigate the influence of fiber volume fraction on the dynamic mechanical behavior of plant fiber reinforced composites. Experimental measurements are used to quantify the macroscopic viscoelastic response of flax fiber reinforced GreenPox composite and to validate the modeling approach. The multiscale framework is subsequently employed to relate microstructural features to the viscoelastic behavior of plant fiber composites through homogenization based numerical modeling, enabling a direct comparison between plant and conventional fiber-reinforced composites. The novelty of this work lies in the proposed multiscale modeling framework, which establishes the relationship between the composite microstructure and its macroscopic viscoelastic behavior through homogenization-based numerical modeling. The paper is organized as follows: Section 2 describes the materials, fabrication process, and experimental methods; Section 3 details the multiscale modeling approach; Section 4 presents and discusses the results; finally, Section 5 concludes with the main findings and perspectives.

2. Materials and methods

2.1. Materials and samples preparation

The materials investigated in this study include a neat epoxy resin and an epoxy reinforced with unidirectional flax fibers. The epoxy consists of SR GreenPox 56 resin and SD 7561 hardener, both supplied by Sicomin®. The resin and hardener are mixed at a mass ratio of 100:37, poured into a mold, and cured in a thermocompression device at 60 °C for 1 h. After curing, the samples are cooled at room temperature for 18 h and then post-cured in an oven at 130 °C for 1 h.

The unidirectional flax reinforced GreenPox composite (FGUD) is fabricated using eight layers of unidirectional FlaxTape 110 supplied by Ecotechnilin®, combined with the same epoxy. Prior to processing, the flax fibers are dried at 50 °C for 1 h to reduce moisture content. The laminate is manufactured by hand impregnation and cured under thermal compression at 60 °C and 3 bar on the mould for 1 h, followed by post-curing at 130 °C for 1 h.

Three fiber volume fractions, 25%, 35% and 47%, are targeted by using wedges with thicknesses of 1.8 mm, 1.45 mm, and 1.2 mm, respectively. After laser cutting, the FGUD samples are stored in a climatic chamber at 23 °C and 50% relative humidity for at least two weeks to ensure moisture equilibrium prior to testing.

2.2. Microstructural analysis

X-ray microtomography is performed to characterize the internal structure of flax–epoxy laminates. Measurements are carried out on an EasyTom RX Solutions system (Chavanod, France) equipped with a 160 kV Hamamatsu tube (L10711). The tube voltage and current are set to 60 kV and 130 μ A, respectively. A total of 1440 projections are acquired in step-by-step mode over 360° using a 2530DX detector (2176×1792 pixels²). The exposure time is 330 ms, with four images averaged per projection, resulting in a scan time of approximately 1 h per sample. Volume reconstruction is performed using Xact Software (RX Solutions) with a filtered back-projection algorithm, yielding a voxel size of 0.8 μ m.

Each reconstructed volume is analyzed in VG StudioMax using a cubic Region of Interest (ROI) of approximately $400 \times 400 \times 500 \mu\text{m}^3$ to avoid edge effects. A Gaussian filter is applied to enhance gray-level contrast between phases. Fiber segmentation is performed via inclusion analysis to create a dedicated ROI fiber mask, which is refined to fully capture the fiber geometry. For porosity analysis, segmentation is carried out on the ROI using a gray-level threshold algorithm implemented in the software.

Further tomographic sections are analyzed in plane projection to quantify fiber alignment. Voxel-based orientation histograms are generated to identify fiber angles, confirming the extent of unidirectional alignment and local deviations due to manual lay-up and bundle waviness. Fiber spacing is also analyzed to assess the uniformity of fiber distribution within the matrix, as variations in spacing can influence stress transfer and resin-rich regions. This analysis is performed by subtracting the ROI fiber mask from the ROI to isolate the matrix phase, followed by a wall thickness analysis in sphere mode (VG StudioMax).

2.3. Quasi-static tensile tests

For each fiber volume fraction, six rectangular specimens are tested in quasi-static tension: three aligned with the fiber direction and three in the transverse direction. The specimen dimensions are 200 mm $\times$ 15 mm, with thickness depending on the fiber volume fraction (see Section 2.1). Tests are conducted in accordance with ASTM D3039/D3039M – 00e1 [26], using a universal testing machine (MTS, 100 kN load cell). Specimens are clamped using wedge action grips and loaded under displacement control at a crosshead speed of 5 mm/min. Longitudinal and transverse strains are measured using two extensometers. The quasi-static tensile modulus is determined from the initial linear portion of the stress–strain curve.

2.3.1. Dynamic mechanical analysis

Dynamic mechanical analysis (DMA) is performed to evaluate the viscoelastic properties of both neat GreenPox and flax–GreenPox composites. Neat GreenPox specimens with dimensions of $49 \times 10 \times 5 \text{ mm}^3$ are tested in tension mode. FGUD specimens are tested under both tension and three-point bending configurations using a METRAVIB® DMA+300 apparatus. The span length for bending tests is 43 mm, and the effective length for tension tests is 40 mm. In both configurations,

the specimen width is 10 mm, and the thickness varies according to the fiber volume fraction (see Section 2.1).

All measurements are carried out at a temperature of 25 °C. Neat GreenPoxy specimens are tested in tension mode at 1 Hz with a static force of 1 N. FGUD specimens are tested over a frequency range of 6.5–42 Hz with a static force of 10 N in tension and –2 N in bending. For all tests, the dynamic displacement amplitude is set to 5 μm , corresponding to a mean strain below 0.05%. The storage modulus (E') and the loss factor (η) are extracted from the measured force displacement response, as processed by the METRAVIB® software.

2.4. Modal analysis

Modal analysis is used to determine the storage modulus (E') and loss factor (η) over a broad frequency range. A neat GreenPoxy plate with dimensions $195 \times 125 \times 4.95 \text{ mm}^3$ is tested as a reference. Square plates are cut from the same flax–GreenPoxy laminates used for other tests, with dimensions of $100 \times 100 \times 1.2 \text{ mm}^3$ for 47% V_f , $110 \times 110 \times 1.45 \text{ mm}^3$ for 35% V_f , and $125 \times 125 \times 1.8 \text{ mm}^3$ for 25% V_f , to ensure comparable natural frequencies across the different V_f .

Each plate is suspended using soft springs to approximate free–free boundary conditions. Excitation is applied via a speaker positioned at one corner, and the vibrational response is measured using a scanning laser vibrometer. Modal frequencies, damping coefficients, and mode shapes are extracted from the measured frequency response functions using MODAN®, an in-house modal analysis software. To determine the elastic properties of the materials, a model updating procedure following the method proposed by McIntyre and Woodhouse [27] is applied, in which an orthotropic finite element model is used instead of the original analytical orthotropic plate formulation. In this approach, initial estimates of the elastic constants are obtained from the measured modal frequencies together with the specimen dimensions and density. These values are then refined through an iterative correction step, in which the numerical and experimental frequencies are compared until convergence is achieved. The final updated model yields the effective orthotropic moduli and associated damping coefficients for each plate.

2.5. ANOVA analysis

A one-way analysis of variance (ANOVA) is performed to assess whether statistically significant differences in modulus and damping exist among flax–GreenPoxy composites with different fiber volume fractions for each experimental method. The data within each group are assumed to follow a normal distribution, which allows the application of ANOVA. For the vibration tests, only three samples are available for each fiber volume fraction, which is insufficient to reliably evaluate normality. Therefore, the normality assumption is accepted for the purpose of this analysis. The analysis is conducted at a significance level of $\alpha = 0.05$. The statistical results are presented using letter annotations in Fig. 8, where different letters (a, b, c) indicate statistically significant differences between composites with different fiber volume fractions within the same experimental method, and identical letters indicate no significant difference.

3. Analytical and numerical homogenization models

While the experimental procedures described above provide direct access to the macroscopic stiffness and damping behavior of the composites, they do not by themselves explain how these properties arise from the underlying microstructure. Classical rule of mixture approaches provide first order estimation of the elastic properties based on the averaged constituent behavior and can be extended, through strain energy formulations, to approximate damping properties [28–30]. However, such analytical models rely on idealized assumptions regarding stress and strain distributions and cannot capture local heterogeneities that govern energy dissipation. To address this limitation, a multiscale modeling framework is introduced in the following section, enabling the prediction of effective viscoelastic properties from microstructural characteristics and their transfer to the structural scale.

3.1. Analytical model

The elastic properties of the composite are evaluated using classical micromechanical models, including the Voigt [28], Reuss [29], and Halpin–Tsai [31] formulations. The damping response is estimated using a strain energy approach [30] and a modified transverse formulation proposed by Berthelot and Sefrani [24].

3.1.1. Longitudinal properties

The effective longitudinal modulus of the composite is calculated assuming an iso-strain condition in the fiber direction. Under this assumption, the rule of mixtures gives:

$$E_L = V_f E_{fL} + V_m E_m \quad (1)$$

where E_L is the longitudinal modulus of the composite, V_f and V_m are the fiber and matrix volume fractions, respectively, and E_{fL} and E_m are the longitudinal elastic moduli of the fiber and matrix. The longitudinal loss factor is computed as:

$$\eta_L = \frac{\eta_{fL} V_f E_{fL} + \eta_m V_m E_m}{V_f E_{fL} + V_m E_m} \quad (2)$$

where η_L is the longitudinal loss factor of the composite, and η_{fL} and η_m are the longitudinal loss factors of the fiber and matrix, respectively.

3.1.2. Transverse properties

The transverse modulus of the composite can be estimated using the Reuss model, which assumes a uniform stress distribution within the composite constituents. Under this assumption, the transverse modulus is given by:

$$E_T = \frac{E_{fT} E_m}{V_f E_m + V_m E_{fT}} \quad (3)$$

where E_T and E_{fT} are the transverse elastic moduli of the composite and fiber, respectively.

Alternatively, the Halpin–Tsai model, a semi-empirical formulation, can be employed to account for the influence of fiber geometry. This model provides an alternative estimation of the transverse modulus, expressed as:

$$E_T = E_m \frac{1 + \zeta_{22} \gamma_{22} V_f}{1 - \gamma_{22} V_f} \quad (4)$$

where γ_{22} is:

$$\gamma_{22} = \frac{\left(\frac{E_{fT}}{E_m}\right) - 1}{\left(\frac{E_{fT}}{E_m}\right) + \zeta_{22}} \quad (5)$$

and ζ_{22} is a shape parameter. For cylindrical fibers, it is commonly taken as $\zeta_{22} = 2$.

The transverse loss factor is estimated using a strain energy approach:

$$\eta_T = \frac{\eta_{fT} V_f E_m + \eta_m V_m E_{fT}}{V_m E_{fT} + V_f E_m} \quad (6)$$

where η_T and η_{fT} are the transverse loss factors of the composite and fiber, respectively.

In addition, the formulation proposed by Berthelot and Sefrani [24] is used to estimate the transverse damping. In this approach, the composite is modeled as a layered medium composed of alternating fiber and matrix phases, subjected to transverse tension perpendicular to the fiber direction. The transverse loss factor is then expressed as:

$$\eta_T = \eta_{T_m} + \eta_{T_f} \quad (7)$$

$$\eta_{T_m} = \eta_m \left[\frac{E_T}{E_m} (1 - V_f)^3 + \frac{9}{4} \frac{E_T}{E_{fT}} (1 - V_f)^2 V_f + \frac{3}{2} \frac{E_T E_m}{E_{fT}^2} V_f^2 (1 - V_f) \right] \quad (8)$$

$$\eta_{T_f} = \eta_{fT} \left[\frac{E_T}{E_m} V_f^3 + \frac{3}{2} \frac{E_T}{E_m} (1 - V_f) V_f^2 + \frac{3}{4} \frac{E_T E_{fT}}{E_m^2} (1 - V_f)^2 V_f \right]. \quad (9)$$

These analytical expressions provide estimates of the effective longitudinal and transverse stiffness and damping properties of the composite based on the constituent properties and volume fractions. The different models are used in parallel to capture both bounding and refined predictions of the viscoelastic behavior. The calculated values are used as theoretical references for comparison with experimental results and finite element predictions in the following sections.

3.2. Multiscale modeling

Modeling plant fiber reinforced composites remains a significant challenge due to their anisotropic and heterogeneous behavior [32], which makes it difficult to accurately determine their effective or homogenized properties using traditional analysis methods that rely on macroscopic constitutive models originally developed for more uniform materials such as metals. Consequently, more advanced and efficient techniques are required to account for the intricate microstructure of composite materials. Among the available approaches, multiscale modeling offers a framework for analyzing the dynamic behavior of complex structures by implementing the concept of homogenization [33]. This technique simplifies a heterogeneous material into an equivalent one with macroscopic properties that reflect its overall behavior. To implement this approach, multiscale simulations are carried out using the finite element software COMSOL®, following the step-by-step modeling process shown in Fig. 1. The microscale study focuses on the detailed interactions within the composite's microstructure, while the macroscale study evaluates its global performance, which are explained in the following sections.

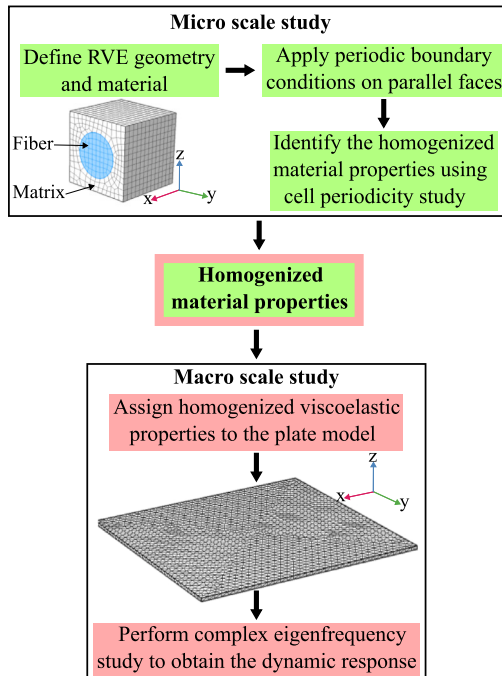


Fig. 1. Multiscale modeling steps in COMSOL®.

3.2.1. Microscale modeling

A Representative Volume Element (RVE) is introduced to estimate effective continuum properties from the local microstructure and constituent phases [34] under the assumptions of ergodicity and scale separation. The ergodicity assumption implies that spatial averages

of the mechanical response over a sufficiently large microstructural domain are equivalent to ensemble averages over multiple realizations, ensuring statistical representativity of the RVE [35,36]. In addition, the scale separation hypothesis assumes a clear distinction between the characteristic length scale of the microstructure and that of the macroscopic structure, allowing a consistent transition from microscale fields to macroscopic behavior [37]. The objective of microscale modeling is to predict the visco-elastic constants of the composite by capturing the interactions between fiber and matrix, taking into account their individual properties and spatial arrangement [38]. Within this framework, the RVE is treated as a heterogeneous medium with known constitutive properties, subject to boundary conditions that correspond to uniform local continuum fields. Homogenization then provides effective properties that are transferred to the macroscale model, where the composite is represented as an equivalent homogeneous medium.

The RVE used in this study consists of a unidirectional fiber embedded in a square array, chosen so that its fiber volume fraction (V_f) and geometry match those of the bulk laminate. To account for damping behavior, the fiber and matrix are assigned complex valued properties whose imaginary parts represent internal energy dissipation, allowing simultaneous prediction of storage and loss components of the response. Periodic boundary conditions (PBCs) are applied to enforce displacement continuity across opposing faces of the RVE, using the cell periodicity feature in COMSOL®, which enforces PBCs between corresponding points on a source face (*src*) and a destination face (*dst*) as:

$$u_{dst} - u_{src} = \varepsilon(X_{dst} - X_{src}) \quad (10)$$

where ε is the average strain, and u and X represent the displacement and position vectors on opposing RVE faces, respectively. Once the boundary conditions are applied and the simulation completed, the homogenized complex elasticity matrix D^* is extracted, containing stiffness components corresponding to each principal direction. For a linear orthotropic material, D^* takes the form:

$$D^* = \begin{bmatrix} D_{11}^* & D_{12}^* & D_{13}^* & D_{14}^* & D_{15}^* & D_{16}^* \\ D_{21}^* & D_{22}^* & D_{23}^* & D_{24}^* & D_{25}^* & D_{26}^* \\ D_{31}^* & D_{32}^* & D_{33}^* & D_{34}^* & D_{35}^* & D_{36}^* \\ D_{41}^* & D_{42}^* & D_{43}^* & D_{44}^* & D_{45}^* & D_{46}^* \\ D_{51}^* & D_{52}^* & D_{53}^* & D_{54}^* & D_{55}^* & D_{56}^* \\ D_{61}^* & D_{62}^* & D_{63}^* & D_{64}^* & D_{65}^* & D_{66}^* \end{bmatrix}.$$

Thanks to the symmetry of the elasticity and compliance matrices for orthotropic materials, the complex constants E_i^* , G_{ij}^* , and ν_{ij}^* can be retrieved from the inverse of the elasticity matrix, i.e., the compliance matrix:

$$D^{*-1} = \begin{bmatrix} \frac{1}{E_x^*} & -\frac{\nu_{xy}^*}{E_x^*} & -\frac{\nu_{xz}^*}{E_x^*} & 0 & 0 & 0 \\ -\frac{\nu_{yx}^*}{E_y^*} & \frac{1}{E_y^*} & -\frac{\nu_{yz}^*}{E_y^*} & 0 & 0 & 0 \\ -\frac{\nu_{zx}^*}{E_z^*} & -\frac{\nu_{yz}^*}{E_z^*} & \frac{1}{E_z^*} & 0 & 0 & 0 \\ 0 & 0 & 0 & \frac{1}{G_{xy}^*} & 0 & 0 \\ 0 & 0 & 0 & 0 & \frac{1}{G_{yz}^*} & 0 \\ 0 & 0 & 0 & 0 & 0 & \frac{1}{G_{xz}^*} \end{bmatrix}.$$

To characterize the damping, the directional loss factor η_i is calculated as the ratio between the imaginary and the real parts of the corresponding complex moduli:

$$\eta_i = \frac{\text{Im}(E_i^*)}{\text{Re}(E_i^*)} \quad (11)$$

where E_i^* and η_i represent the complex modulus and associated loss factor in the $i = x, y, z$ directions.

3.2.2. Macroscale modeling

The homogenized properties E'_i and η_i , obtained from the RVE serve as input to the finite element model, where the composite plate is treated as an equivalent homogeneous medium. While the microscale modeling provides effective elastic and damping properties, structural applications require plate scale analysis where specimen dimensions, stress distributions, and structural complexity influence the global response. This macroscale step bridges material behavior and structural performance, enabling direct comparison with vibration experiments and validation of the multiscale approach. Furthermore, the same methodology can be extended to more complex geometries allowing the prediction of stiffness and damping behavior within the same computational framework.

The multiscale framework described above provides effective viscoelastic properties that enable the role of microstructural parameters on the macroscopic response of both plant and synthetic fiber reinforced composites to be examined. The corresponding results are presented in the following section.

4. Results and discussion

4.1. Microstructural characterization

The 3D reconstructed volumes (Fig. 2) show fibers in gray and pores in blue, a half-section view is included to visualize the internal porosity distribution. X-ray microtomography confirmed that all laminates exhibit low porosity levels, with the values of 2.8%, 0.86%, and 1.2% for the plates with fiber volume fractions of 25%, 35%, and 47%, respectively. These porosity values are well below the typical range reported for thermocompression-moulded natural fiber composites, 3%–8% on average, and sometimes exceeding 10% when processing is not fully optimized [39]. Porosity levels below 3% are commonly considered acceptable in plant fiber composite manufacturing and are known to have only a minor effect on the global property. Therefore, the relatively low porosity observed here suggests that voids are unlikely to influence the subsequent analysis of the effect of fiber volume fraction on the mechanical and damping behavior of the laminates.

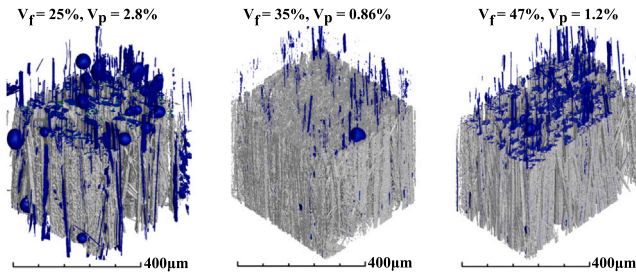


Fig. 2. 3D reconstructed images from microCT of specimens cut in plates (fibers in gray and pores in blue) with different fiber volume fractions.

Fig. 3 illustrates the fiber spreading and orientation analysis performed on the laminate with a 25% fiber volume fraction, using the front-plane micro-CT slice as an example. The raw image (Fig. 3a) shows the distribution of fibers (light gray), porosity (black), and matrix (dark gray). The fiber spreading across the slice, defined as the distance between adjacent fibers, ranges from 65 to 124 μm . Using the same projection, the local fiber orientations are also extracted, and the corresponding histogram (Fig. 3b) shows a narrow angular distribution centered around the nominal reinforcement direction, indicating limited fiber waviness. Applying this procedure to both the front and right planes of all laminates yielded mean angles between 88.5° and 93.2° , with standard deviations of 4.7° – 10.5° , confirming comparable alignment quality across volume fractions. These deviation levels are consistent with previous micro-CT studies of unidirectional natural fiber composites [40]. Orientation variability can therefore be considered negligible in this study, and differences in mechanical and damping behavior may be attributed primarily to changes in fiber volume fraction.

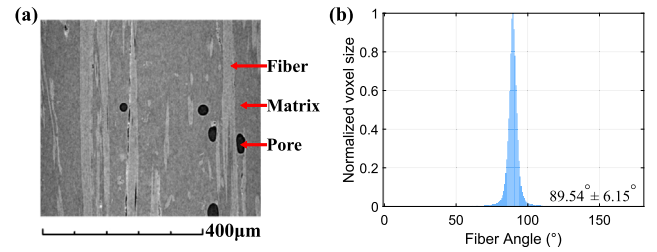


Fig. 3. Fiber spreading and orientation analysis on a laminate with 25% V_f (a) Raw micro-CT slice of the front plane; (b) Fiber orientation histogram.

4.2. Quasi-mechanical properties

Results from quasi-static tensile tests performed on flax–GreenPoxylaminates in both the longitudinal and transverse directions are shown in Fig. 4. As expected, the longitudinal modulus increases from an average value of approximately 16 GPa at a fiber volume fraction of 25% to about 24 GPa at 47%. The error bars reflect the variability observed among the nine tested specimens. In this figure, values obtained using a rule of mixtures model based on constituent properties available in the literature are also shown. For flax fibers, elastic moduli of 59 GPa in the longitudinal direction and 8 GPa in the transverse direction [41] are considered. For the GreenPoxyl resin, an elastic modulus of 3 GPa is assumed [42]. The order of magnitude of the experimentally measured values is consistent with the model predictions. The experimental longitudinal modulus is slightly lower than the predicted value, which may be attributed to the variability of flax fiber properties [43], as the values used in the model are taken from the literature and were not measured on the fibers used in the present reinforcement. Nevertheless, it is worth noting the very good agreement in terms of overall trend. The values obtained are consistent with results reported in the literature, with a longitudinal modulus of approximately 20 GPa reported at a fiber volume fraction of 40% [44].

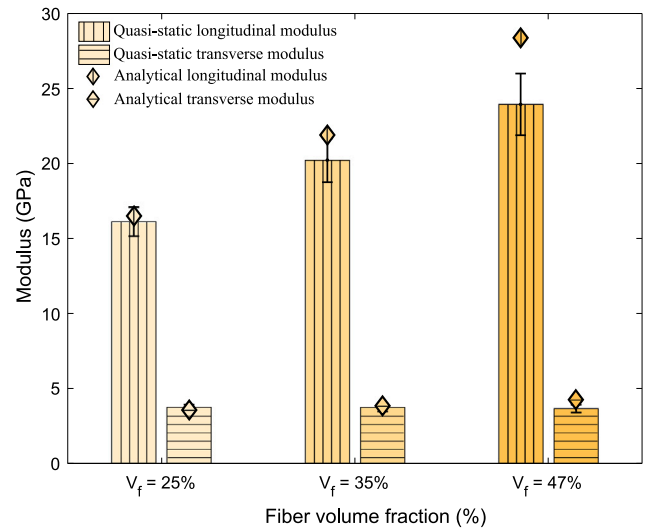


Fig. 4. Quasi-static tensile modulus of flax–GreenPoxylaminates in the longitudinal (E_L) and transverse (E_T) directions as a function of fiber volume fraction.

In contrast, the transverse modulus remains nearly constant at approximately 4 GPa across all fiber volume fractions. This behavior is also captured by the analytical model, which predicts a much weaker sensitivity of the transverse elastic response to fiber volume fraction compared to the longitudinal direction. While quasi-static measurement provides information on stiffness, it offers no insight into energy dissipation. Therefore, damping behavior is examined using DMA and vibration based measurements.

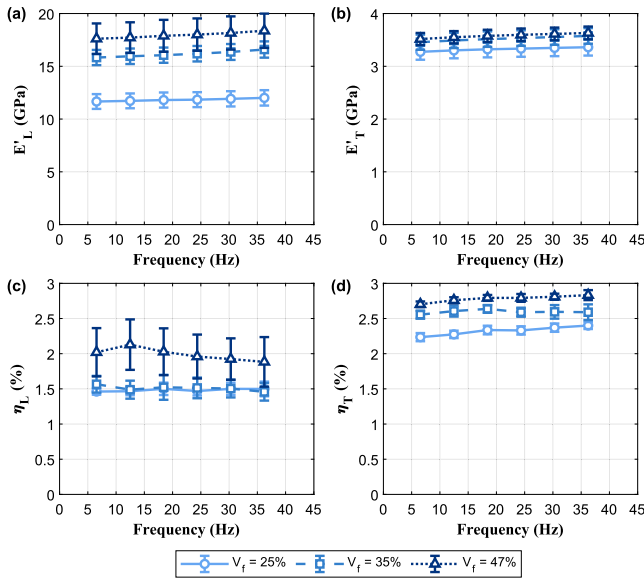


Fig. 5. DMA of flax-GreenPox composites in tension mode at 25 °C. (a) Longitudinal storage modulus E'_L , (b) transverse storage modulus E'_T , (c) longitudinal damping η_L , and (d) transverse damping η_T , as a function of frequency for different fiber volume fractions.

4.3. Dynamic mechanical analysis

DMA measurements are performed in both tension and bending modes to evaluate the viscoelastic response of the flax-GreenPox composites at 25 °C over the frequency range 6.5–42 Hz. For each fiber volume fraction, three specimens are tested in the longitudinal direction and three in the transverse direction. The curves in Figs. 5 and 6 correspond to mean values, and the error bars represent the standard deviation across the nine specimens.

As expected, in the longitudinal direction the storage modulus increases with fiber volume fraction in both loading modes (Figs. 5a and 6a). In tension, E'_L increases from approximately 12 GPa to 18 GPa, while in bending it increases from about 8 GPa to 16 GPa as the fiber volume fraction increases, in agreement with the trends observed in the quasi-static tensile results. E'_L from bending tests is lower than that measured in tension, as shear contributions in bending are not considered during post-processing, leading to an underestimation of the modulus. The longitudinal modulus shows no frequency dependence in bending and only a negligible change (7%) in tension. Consistent with the quasi-static measurements, the transverse storage modulus remains nearly constant with fiber volume fraction in both loading modes (Figs. 5b and 6b), with average values of approximately 3.4 ± 0.15 GPa in tension and 2.4 ± 0.1 GPa in bending.

Regarding damping behavior, no clear systematic dependence on fiber volume fraction is observed in either loading mode. In tension (Fig. 5c), the longitudinal damping (η_L) remained in the range of 1.5% to 2.2%, and the composite with 47% fiber volume fraction shows slightly higher damping, but this deviation is likely due to measurement uncertainty rather than a real material effect. The transverse damping (η_T) remains within a range of 2.2 to 2.9% over the frequency range, with a slight increase observed as the fiber volume fraction increases. Additionally, a general decrease in η_L and a slight increase in η_T with increasing frequency are observed.

In bending (Fig. 6c), the longitudinal damping (η_L) is higher than in tension, ranging from approximately 1.9% to 2.4%, due to frictional losses at the specimen–fixture contact points, leading to enhanced energy dissipation compared to pure tensile loading. A similar magnitude and trend have been reported in the literature [45], where the loss

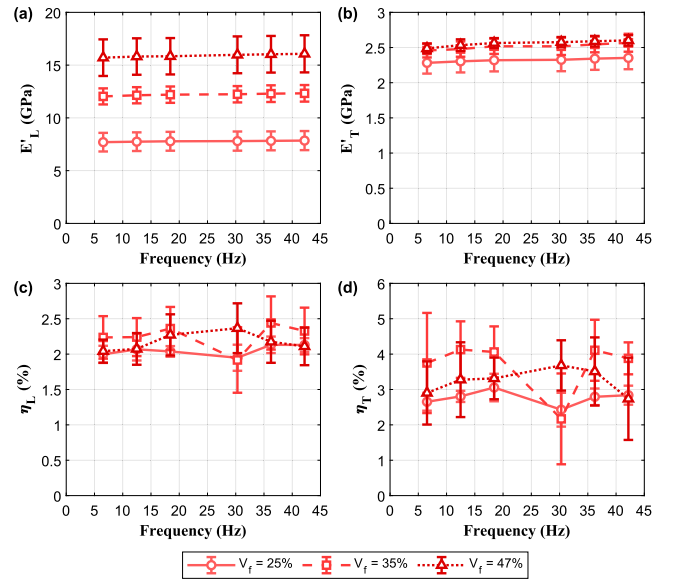


Fig. 6. DMA of flax-GreenPox composites in bending mode at 25 °C. (a) Longitudinal storage modulus E'_L , (b) transverse storage modulus E'_T , (c) longitudinal damping η_L , and (d) transverse damping η_T , as a function of frequency for different fiber volume fractions.

factor measured in bending mode (2.9%) was found to be higher than that obtained in tension mode (1.59%). The transverse damping (η_T) values are higher overall, lying between approximately 2.5% and 4.2% (Fig. 6d). The damping values fall within the range reported in the literature for flax fiber reinforced epoxy composites [44,46].

4.4. Modal analysis

The frequency response functions (FRFs) obtained from the vibration tests exhibit distinct resonance peaks, each corresponding to a structural mode shape of the specimens as shown in Fig. 7. From these peaks, both the natural frequencies and the associated damping ratios are identified. The experimentally identified modal frequencies are then used to determine the complex properties through model-updating procedure described in Section 2.4, which combines the measured vibrational response to the orthotropic stiffness and damping characteristics of the composites.

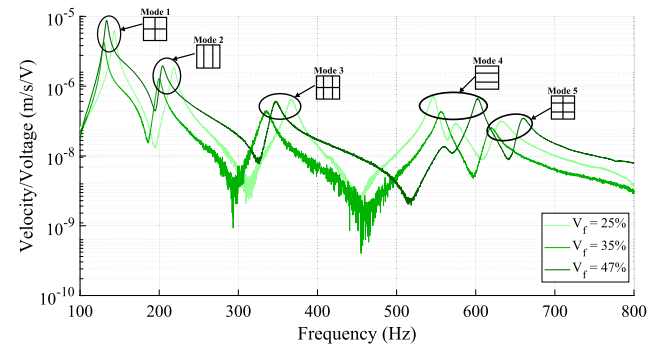


Fig. 7. FRFs of flax-GreenPox composite plates for different fiber volume fractions, where each peak corresponds to a distinct vibration mode. Mode 1 corresponds predominantly to a torsional response; Mode 2 to bending in the transverse direction; Mode 3 to a mixed bending–torsional response; Mode 4 to bending in the longitudinal direction; and Mode 5 to a higher-order mixed mode.

As expected, the longitudinal modulus increases significantly with fiber volume fraction, rising from 22 GPa to 29 GPa, showing consistent trends with the Quasi-static and DMA results. Regarding damping, η_L

remains nearly constant at around 1.5% across all fiber volume fractions. The transverse modulus exhibits negligible variations, remaining within the range of approximately 3.6–3.8 GPa. η_T range between 2.8% and 3.7%, with no clear systematic dependence on fiber volume fraction. Overall, both the longitudinal and transverse damping levels identified in the present study are comparable to values reported in the literature for flax fiber–reinforced composites [45], with damping ratios of approximately 1.3% in the longitudinal direction and 2.64% in transverse direction.

4.5. Global discussion

Fig. 8 provides a summary of the experimental results obtained for neat GreenPoxy and flax–GreenPoxy composites with three fiber volume fractions. For the neat GreenPoxy resin, the storage modulus and damping factor measured by DMA in tension mode at 1 Hz are 2.5 GPa and 2.7%, respectively. The vibration measurements shows higher values (3.4 GPa and 3.2%) coming from higher excitation frequency and lower strain amplitude. Nevertheless, the results are in agreement with the values reported in the literature [42,44]. Compared with the neat GreenPoxy resin, the flax–GreenPoxy composites exhibit a clear increase in stiffness. In the longitudinal direction, this increase becomes more pronounced with increasing fiber volume fraction across all experimental techniques, highlighting the dominant role of fibers in load transfer along their alignment. Differences are observed between the values of E_L obtained using the different measurement techniques. Compared with quasi-static tensile tests performed on the MTS machine, lower values of E_L are measured using DMA in tension mode, due to the low compliance of the DMA setup and to the fact that, unlike the quasi-static tests where strain is measured using an extensometer, the strain in DMA is estimated from the machine displacement. This displacement includes set-up deformation, leading to an overestimation of strain and consequently to an underestimation of the elastic modulus. Within DMA measurements, the values of E_L obtained in bending are lower than in tensile mode, as the deformation involves both bending and shear contributions, whereas only the bending component is considered during post-processing, leading to an underestimation of the elastic modulus. Higher values of E_L

are obtained from vibration measurements due to higher excitation frequencies and lower strain amplitudes associated with vibration tests, which lead to a stiffer apparent response of the viscoelastic composite. In contrast, the transverse modulus shows no significant dependence on fiber volume fraction, indicating a matrix-dominated response in this direction. For the same reasons discussed above, differences between the measurement techniques are also observed in the transverse case.

In the case of damping, no statistically significant variation is observed in the longitudinal direction with increasing fiber volume fraction. Although a slight increase in damping is measured at 47% fiber volume fraction in the DMA (tension) results, is attributed to spatial heterogeneity of the material, as the use of small specimens can lead to local variations in the measured response. The higher damping values observed in DMA bending are consistent with the contact effects discussed in the corresponding DMA results section (Section 4.3). In the transverse direction, damping values exhibit variations for all testing methods; however, these variations do not follow a consistent increasing or decreasing trend with fiber volume fraction. This behavior suggests that transverse energy dissipation is governed by mechanisms that are largely insensitive to fiber content and are instead dominated by the matrix and interfacial effects.

Overall, the longitudinal stiffness increases with fiber volume fraction (8–30 GPa across the different measurement techniques), whereas the transverse stiffness remains nearly constant (2.3–3.9 GPa). Damping shows limited sensitivity to fiber content, with η_L in the range 1.4%–2.3% and η_T 2.3%–4.1%, depending on the loading mode. A deeper understanding of these trends requires insight at the microstructural level. The multiscale modeling framework described in Section 3.2 offers a more general approach, applicable to fiber reinforced composites, and is further used to investigate these effects.

4.6. Numerical prediction

The numerical analysis is conducted to evaluate the effective viscoelastic properties of fiber reinforced composites and to compare the predictions with the analytical models presented in Section 3.1. As a reference case, a classical unidirectional glass fiber/epoxy composite is considered, with constituent properties taken from Berthelot and

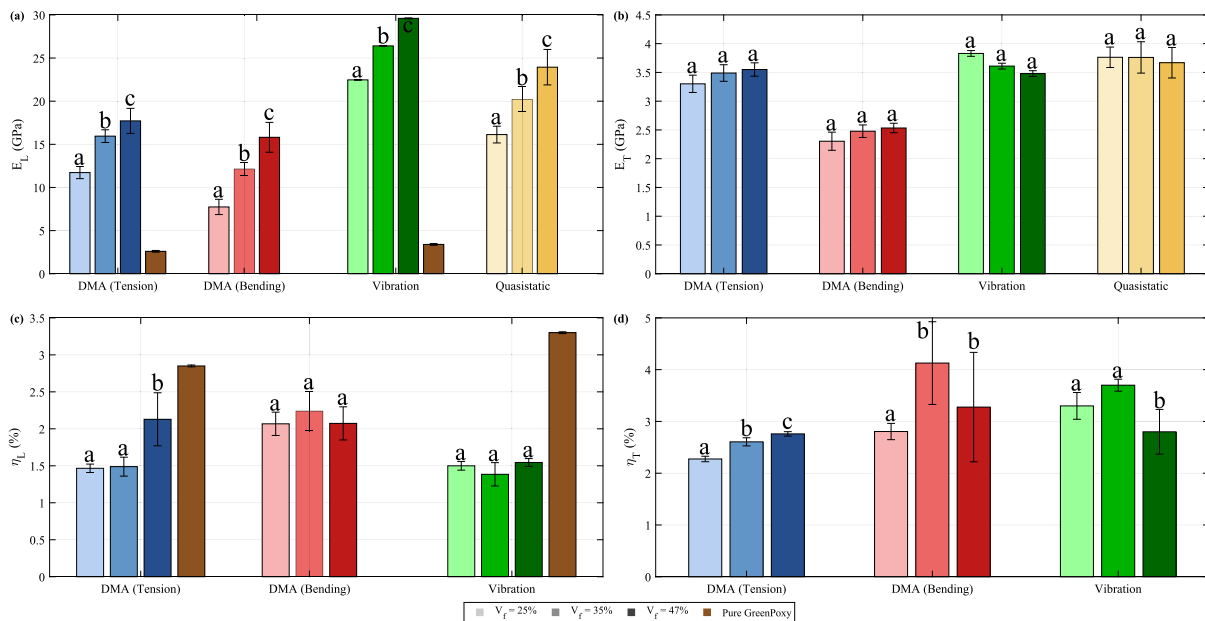


Fig. 8. Summary of elastic and damping properties of neat GreenPoxy and flax–GreenPoxy composites obtained from quasi-static, DMA, and vibration measurements for different fiber volume fractions. (a) Longitudinal modulus E_L , (b) transverse modulus E_T , (c) longitudinal damping η_L , and (d) transverse damping η_T . Different letters (a, b, c) indicate statistically significant differences between composites with different fiber volume fractions within the same experimental method, whereas identical letters indicate no statistically significant difference, based on one-way ANOVA.

Table 1
Material properties of glass fibers and epoxy matrix [47].

Material	E_L (GPa)	E_T (GPa)	G_{LT} (GPa)	ν	η_L (%)	η_T (%)
Epoxy (DX 210)	3.21	3.21	1.20	0.338	1.35	1.35
Glass	61	61	30	0.22	0.19	0.083

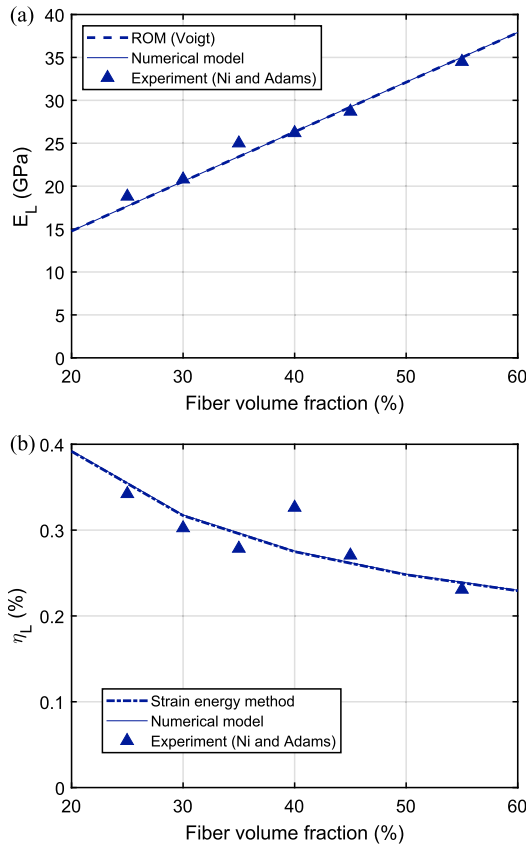


Fig. 9. Numerical predictions and analytical comparison as a function of fiber volume fraction for glass epoxy composite: (a) longitudinal modulus E_L and (b) longitudinal loss factor η_L .

Sefrani [24] and summarized in Table 1. These properties serve as input parameters for both the analytical formulations and the finite element model.

Fig. 9 presents the evolution of the longitudinal modulus E_L and loss factor η_L as a function of fiber volume fraction for the glass/epoxy composite. The numerical model accurately captures the experimental values reported by Ni and Adams [47], with the longitudinal modulus increasing from 18.8 to 37.9 GPa and the loss factor decreasing from 0.34% to 0.23% as the fiber volume fraction increases from 25% to 59%. The decrease in η_L is attributed to the lower intrinsic damping of glass fibers compared to the epoxy matrix, leading to reduced energy dissipation as the fiber contribution becomes dominant.

For the transverse modulus, the Reuss model, the Halpin–Tsai formulation, and the numerical model are compared with experimental results from Berthelot and Sefrani [24], as shown in Fig. 10(a). The Reuss model underestimates E_T due to the iso-stress assumption. In contrast, the Halpin–Tsai model and the numerical model show similar predictions and overestimate the experimental values, particularly at higher fiber volume fractions. This discrepancy may be attributed to idealized assumptions regarding stress transfer mechanisms.

The transverse damping η_T is presented in Fig. 10(b). The experimental results reported by Berthelot and Sefrani [24] show that the matrix phase contributes significantly to the transverse damping of the

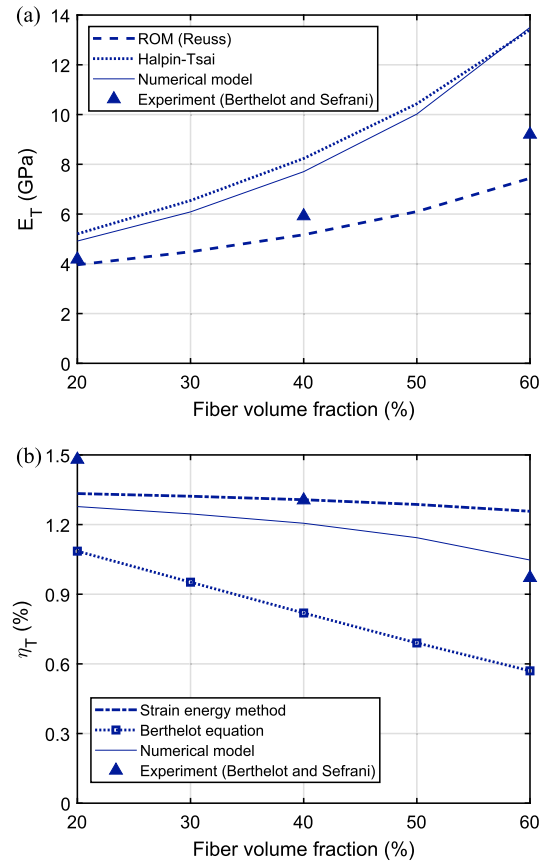


Fig. 10. Numerical predictions and analytical comparison as a function of fiber volume fraction for glass epoxy composites: (a) transverse modulus E_T and (b) transverse loss factor η_T .

Table 2
Material properties of flax fibers and epoxy matrix.

Material	E_L (GPa)	E_T (GPa)	ν	η_L (%)	η_T (%)
Epoxy	3 [42]	3 [42]	0.35 a	1.5 [44]	1.5 [44]
Flax fiber	57 [48]	1.7–8 [41,49]	0.3 a	0.66–1.23 [50]	0.62–4.5 a

a Arbitrary value

composite. As the fiber content increases, the relative contribution of the matrix is reduced, leading to a decrease in the overall damping. The numerical model is able to capture this trend well, while the strain energy method predicts an almost constant damping. The model proposed by Berthelot and Sefrani [24] also predicts a decreasing trend but significantly underestimates the values. This discrepancy may be attributed to the sensitivity of the model to the transverse damping properties of the fibers, which are difficult to accurately characterize; adjusting these parameters could improve the agreement with experimental data.

The applicability of the analytical and numerical models to plant fiber composites is then assessed using a flax/epoxy composite, with properties summarized in Table 2.

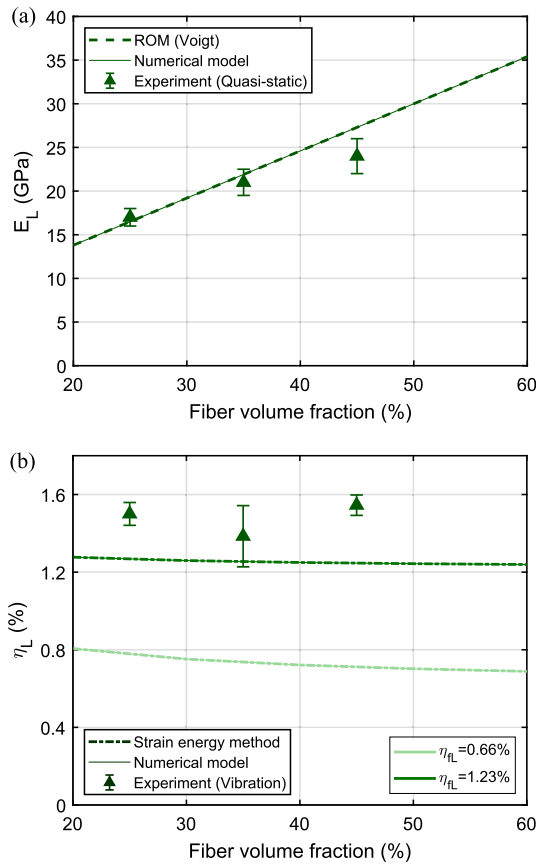


Fig. 11. Numerical predictions and analytical comparison as a function of fiber volume fraction for flax epoxy composites: (a) longitudinal modulus E_L and (b) longitudinal loss factor η_L for $\eta_{fL} = 0.66\%$ and 1.23% .

As shown in Fig. 11(a), for longitudinal modulus, the numerical model remains in good agreement with the Voigt prediction. For the longitudinal damping, different values of fiber damping are considered. The lower value ($\eta_{fL} = 0.66\%$) corresponds to measurements obtained from vibration test performed on single fibers in the dry state [50]. The higher value ($\eta_{fL} = 1.2\%$) is derived from back-calculation from composite level vibration tests in the present study and may therefore include additional contributions, such as interphase effects and other microstructural dissipation mechanisms. As shown in Fig. 11(b), the predicted evolution of η_L strongly depends on the assumed fiber damping. For low damping values, η_L decreases with increasing fiber content, while for higher damping values, the evolution becomes nearly constant, in better agreement with experimental observations. This highlights the strong sensitivity of the longitudinal damping to fiber properties.

For the transverse modulus (Fig. 12(a)), several values of the transverse fiber modulus are considered to account for the large variability reported in the literature. The lower value ($E_{fT} = 1.7$ GPa) corresponds to result coming from single fiber transverse compression tests [49], while the intermediate value ($E_{fT} = 4.75$ GPa) is obtained from back-calculation from the quasistatic tensile test performed in the present study. The higher value ($E_{fT} = 8$ GPa) is taken from literature data and also corresponds to values identified through inverse approaches [41]. This range reflects the uncertainty associated with the transverse properties of plant fibers. The experimental modulus shows an almost constant trend with increasing fiber volume fraction, indicating that the transverse response is mainly governed by the matrix. When $E_{fT} < E_m$, the predicted modulus decreases with increasing fiber content, whereas for $E_{fT} > E_m$, an increasing trend is observed. The numerical model

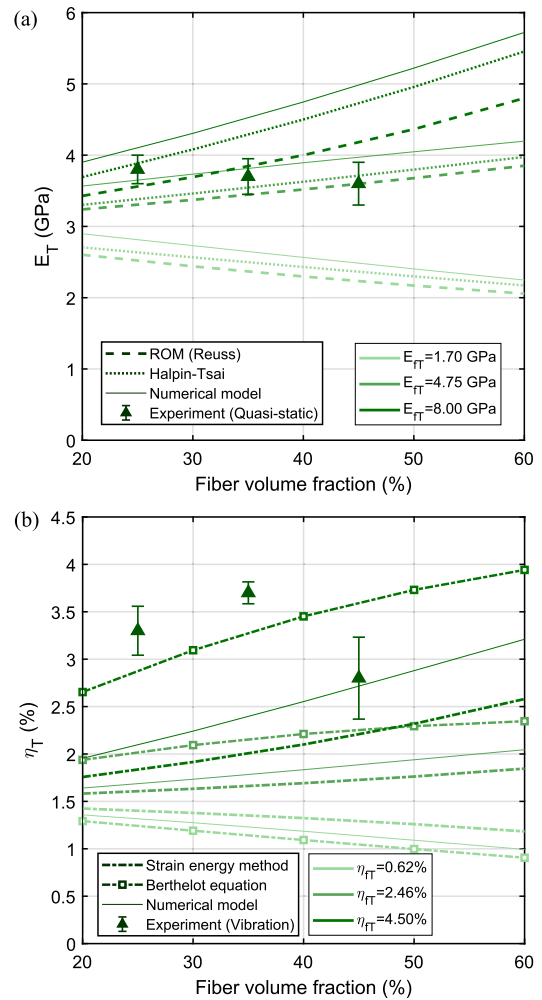


Fig. 12. Numerical predictions and analytical comparison as a function of fiber volume fraction for flax epoxy composites: (a) transverse modulus E_T for $E_{fT} = 1.7$ GPa, 4.75 GPa and 8 GPa and (b) transverse loss factor η_T for $\eta_{fT} = 0.62\%$, 2.46% and 4.5% .

tends to overestimate E_T at high fiber volume fractions, likely due to idealized assumptions such as perfect fiber alignment and interface bonding. In addition, the transverse response is strongly influenced by shear deformation in the matrix, which may be over predicted in the numerical model compared to the actual heterogeneous microstructure of plant fiber composites.

Considering the transverse damping, the fiber properties remain largely unknown, and several assumptions are therefore investigated. Following the approach proposed by Berthelot and Sefrani [24], the transverse fiber damping is assumed to be lower than the longitudinal one, reflecting the anisotropic behavior of the fibers. However, given the complex and amorphous structure of plant fibers, higher transversal damping values compared to the longitudinal damping may also be expected. In addition, values obtained from back-calculation based on composite experimental responses suggest higher transverse damping. Therefore, a range of values is considered in the model to account for this uncertainty.

As shown in Fig. 12(b), the experimental results from vibration test exhibit relatively high damping values and do not follow a clear monotonic trend with increasing fiber volume fraction, in contrast with the behavior typically observed in glass fiber composites. For both lower and higher assumed values of fiber damping, the models fail to capture the experimental response. These discrepancies indicate

that additional dissipation mechanisms are involved in plant fiber composites. One possible explanation is that the present model assumes perfect fiber–matrix bonding and therefore neglects energy dissipation associated with the fiber–matrix interface. Although recent nanomechanical characterization of flax/GreenPox composite has shown that the interphase is extremely thin and does not form a distinct third phase [51], localized interfacial dissipation may still contribute to the transverse damping response. In addition, the complex microstructure of natural fibers, including lumen structure, fiber heterogeneity, and moisture-related phenomena, contributes significantly to the transverse damping and is not fully captured by the current modeling approaches.

4.6.1. Comparative numerical analysis across different fiber types

To further investigate the stiffness–damping relationships, a broader numerical study is performed on synthetic and plant fiber composites, considering a wide range of material properties and fiber volume fractions for the material properties given in Table 3. As shown in Fig. 13(a), carbon fiber composites exhibit the highest longitudinal stiffness, while glass and flax composites display comparable E_L values over the investigated range.

The evolution of the longitudinal damping η_L differs significantly depending on the fiber type as shown in Fig. 13(c). For glass composites, η_L decreases with increasing fiber volume fraction, reflecting the low intrinsic damping capacity of glass fibers. This trend is consistent with literature data [47,53]. For carbon composites, the numerical

model predicts nearly constant η_L values with increasing fiber volume fraction. However, experimental results exhibit significant variability depending on the carbon type. While some datasets show an approximately constant damping [47], others report a much stronger reduction with increasing fiber volume fraction [53]. This variability highlights the sensitivity of damping to fiber grade and microstructural characteristics. In contrast, plant fiber composites exhibit significantly higher damping levels, with η_L remaining approximately constant or only weakly dependent on fiber volume fraction over the range of fiber damping considered. Experimental measurements confirm this behavior, showing nearly constant damping values around 1.5%. Overall, plant fiber composites exhibit damping levels approximately three times higher than those of glass and carbon composites, while maintaining stiffness comparable to glass fibers. This highlights a favorable stiffness-to-damping trade-off for plant-based composites.

For the transverse response (Figs. 13(c) and (d)), glass and carbon composites exhibit an increase in E'_T with increasing fiber volume fraction, reflecting the contribution of the fiber phase to the overall stiffness. In contrast, plant fiber composites show an approximately constant or slightly varying transverse modulus, depending on the relative magnitude of the fiber transverse stiffness compared to the matrix.

The transverse damping η_T exhibits different trends depending on the fiber type. For glass and carbon composites, η_T generally decreases

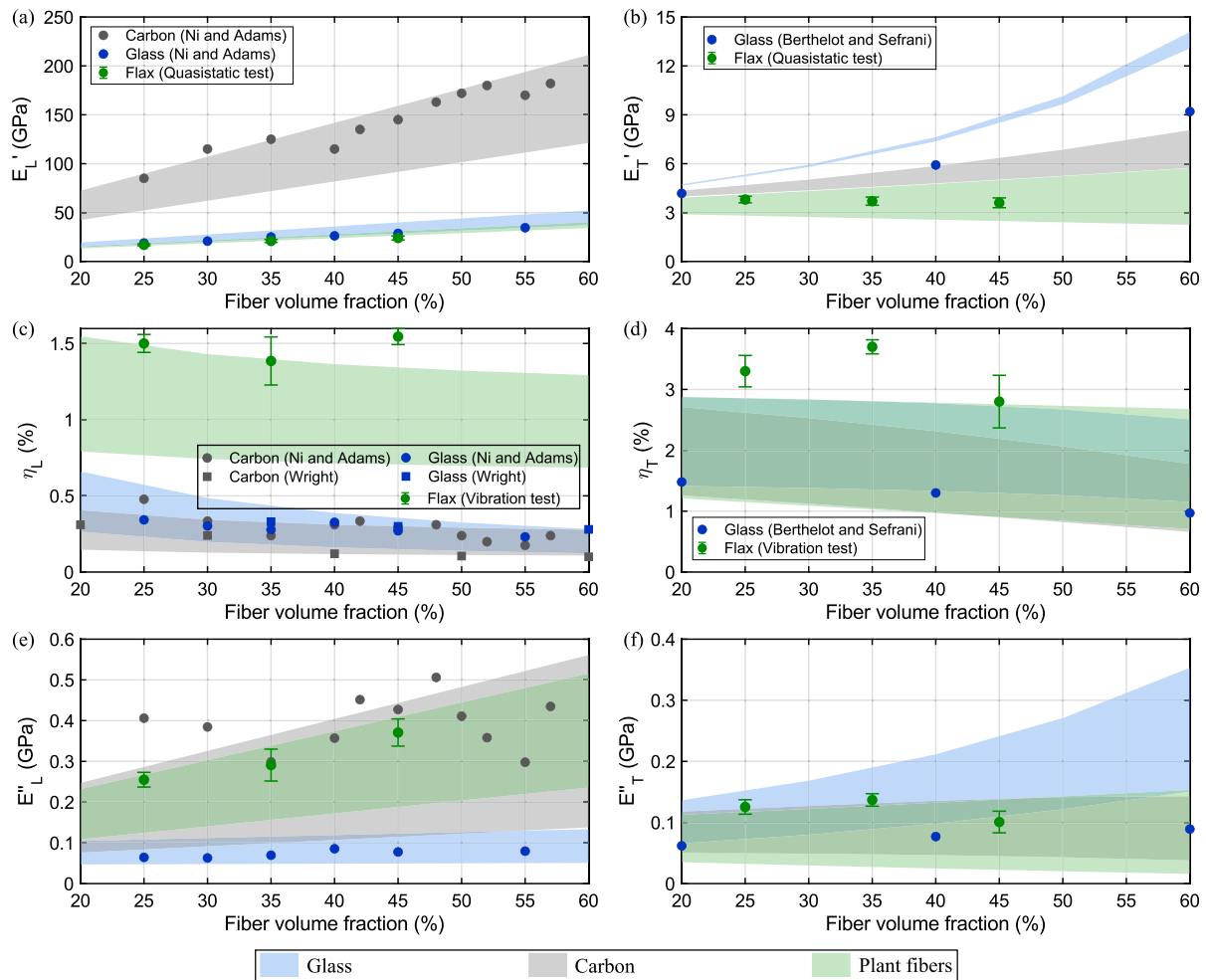


Fig. 13. Numerical and experimental comparison of stiffness and damping over a range of properties for glass, carbon, and plant fiber composites: (a) longitudinal storage modulus E'_L , (b) transverse storage modulus E'_T , (c) longitudinal loss factor η_L , (d) transverse loss factor η_T , (e) longitudinal loss modulus E''_L and (f) transverse loss modulus E''_T . Markers represent experimental values, and shaded regions indicate the range of numerical predictions for different fiber property values.

Table 3
Ranges of fiber properties used in the numerical model.

Material	E_{fL} (GPa)	E_{fT} (GPa)	η_{fL} (%)	η_{fT} (%)
Epoxy	3 [42]	3 [42]	1.5–3 [44]	1.5–3 [44]
Glass	60–85 [21,24,47]	60–85 [21,24,47]	0.09–0.19 [21,24,47]	0.045–0.38 [21,24,47]
Carbon (P-25, T300, T400)	200–350 [24,47,52]	8–15 [52]	0.10–0.25 [24,47]	0.05–0.50 [24,47]
Plant fibers (flax, hemp, nettle)	55–65 [48]	1.7–8 [41,49]	0.66–1.2 [50]	0.33–2.5 ^a

^a Arbitrary value.

with increasing fiber volume fraction, consistent with the reduced contribution of the matrix to energy dissipation. For plant fiber composites, the numerical model predicts either increasing or decreasing trends depending on the assumed fiber transverse damping η_{fT} . However, experimental observations for flax composites indicate higher damping levels and a weak dependence on fiber content. This suggests that additional dissipation mechanisms, such as interfacial effects and microstructural heterogeneity, are not fully captured by the model. This highlights the need for improved modeling approaches capable of accounting for additional dissipation mechanisms specific to plant fiber composites.

To evaluate the overall stiffness–damping performance of composites, the loss modulus E'' (Fig. 13(e) and (f)) is one of the most relevant performance indicators, as it combines the effects of stiffness and damping through the product of the storage modulus and the loss factor. In the longitudinal direction, E''_L (Fig. 13(e)) shows that plant fiber composites exhibit a more favorable stiffness–damping trade-off than glass fiber composites, despite their comparable stiffness. Furthermore, compared with carbon fiber composites, E''_L shows that plant fiber composites achieve a comparable stiffness–damping balance despite their substantially lower stiffness, owing to their higher intrinsic damping. A similar trend is also observed in the transverse direction (Fig. 13(f)).

Generally, the combined experimental and numerical analyses highlight the key role of fiber type and volume fraction on the stiffness and damping behavior of composites. The main findings are summarized in the following section.

5. Conclusions

This study investigated the influence of fiber volume fraction on the stiffness and damping behavior of unidirectional flax–GreenPoxy composites through a combined experimental and multiscale numerical approach. As expected, a clear and consistent increase in longitudinal stiffness with increasing fiber volume fraction is observed across all testing methods, in agreement with the rule of mixtures, confirming the dominant role of fibers in load transfer along their alignment, which is well captured by the numerical model. In contrast, the transverse modulus showed negligible sensitivity to fiber content. With respect to damping, no statistically significant dependence on fiber volume fraction is identified in the longitudinal direction, with η_L remaining approximately constant (around 1.5%). In the transverse direction, η_T lay in the range of about 2.2%–4.0% and exhibited noticeable variability without a clear monotonic trend. These results highlight that, unlike stiffness, damping in flax fiber composites is not governed solely by fiber content.

The multiscale numerical predictions provide a broader perspective by extending the analysis to other fiber types. In the longitudinal direction, the evolution of damping is strongly dependent on the intrinsic viscoelastic properties of the fibers. Glass fiber composites exhibit a consistent decrease in η_L with increasing fiber volume fraction, reflecting the low damping capacity of the fibers. Carbon fiber composites show a more complex behavior, with numerical predictions indicating weak variations, while experimental data reveal significant variability

depending on the fiber type and composite. In contrast, plant fiber composites maintain relatively high and nearly constant longitudinal damping levels with increasing fiber content.

In the transverse direction, the behavior differs significantly. For glass and carbon composites, η_T generally decreases with increasing fiber volume fraction, consistent with the reduced contribution of the matrix. For plant fiber composites, the numerical predictions show a strong sensitivity to the assumed transverse fiber damping, while experimental observations indicate higher damping levels with limited dependence on fiber content. This discrepancy highlights the influence of additional dissipation mechanisms that are not fully captured by the current modeling approaches, and emphasizes the need for improved characterization of the transverse damping properties of natural fibers. In future work, the model will be extended to include the viscoelastic response of the fiber–matrix interphase, enabling the contribution of interfacial dissipation mechanisms to be evaluated.

The stiffness–damping trade-off has long been recognized as a major challenge in the design of advanced engineering materials. Achieving a combination of high stiffness and high damping remains particularly demanding, yet is essential for vibration-sensitive structural applications. In this context, this study shows that plant fiber-reinforced composites offer a promising pathway to overcome this limitation, as their stiffness can be significantly enhanced through the development of highly fiber-loaded architectures while preserving their inherent damping capabilities. Increasing the fiber volume fraction therefore represents a key strategy to shift the stiffness–damping balance toward higher structural performance. However, achieving fiber volume fractions in the range of 60% and beyond remains a major processing challenge for the composite community, requiring the development of manufacturing routes capable of producing highly compact and defect-free plant fiber composite structures. The proposed experimental and numerical framework provides a robust tool for guiding the development of sustainable composite materials with tailored dynamic mechanical performance. Future work will focus on incorporating interfacial effects and broader frequency–temperature conditions to refine damping predictions and further optimize the design of plant-based structural composites.

CRediT authorship contribution statement

Tsilat Shiferaw: Writing – original draft, Methodology, Investigation, Formal analysis. **Pauline Butaud:** Writing – review & editing, Writing – original draft, Supervision, Resources, Methodology, Investigation, Funding acquisition, Conceptualization. **Morvan Ouisse:** Writing – review & editing, Supervision, Methodology, Investigation, Conceptualization. **Xavier Gabrion:** Writing – review & editing. **Vincent Placet:** Writing – review & editing, Supervision, Methodology, Investigation, Conceptualization.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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Data availability

Data will be made available on request.

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