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Flexural response of polypropylene/E-glass fibre reinforced unidirectional composites.

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Abstract

This paper presents a study of the flexural response of continuous E-glass fibre reinforced polypropylene composites. Experiments were designed to investigate monotonic and cyclic flexural response using three point bending test for laminates with different angle-ply and cross-ply arrangements. Results show that the monotonic and cyclic flexural response of the composites are influenced by the plastic deformation of the matrix. The study observed that increasing numbers of cyclic loads led to significant energy dissipation, stiffness reduction and micro-damage accumulation within the composite and especially at the matrix-fibre interface. Significant energy dissipation and damage were observed to dominate the first load-unload cycle. With subsequent cycles, the magnitude of energy dissipation and global damage reduces to a threshold value which is cycle independent. This study has also developed a phenomenological model to predict the dependence of energy dissipation with number of cycles. The experimental data generated here will be useful in the development of holistic macroscale constitutive models and finite element studies of the chosen test composite.

Keywords: A. Polymer Matrix Composites, A. PlytronTM, B. Three Point Bending, B. Cyclic Flexural Response

1. Introduction

In automotive and aerospace industries, the use of thermoplastics as matrix systems of fibre reinforced composites has continued to grow steadily. This is largely due to the materials' recyclability and ability to be processed rapidly. Higher strength-to-weight ratios, better chemical and impact resistance, improved fracture toughness over thermosets and enhanced fatigue strength are some other reasons why thermoplastic composites are becoming the material of choice for replacing traditional materials as steel, aluminum, wood, etc

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[1–4]. One particularly promising type of thermoplastic composites is cost-effective, continuous E-glass fibre reinforced polypropylene-matrix composites. These have potential as thermoformable automotive body parts [5, 6], and have attracted attention for example as possible components of composite integral body armours [7, 8], and of fibre-metal laminate sandwich structures designed for ballistic protection [9–11].

Although the use of thermoplastic composites is growing steadily, its application in structural components is inhibited by limited set of reliable experimental data about their mechanical response, especially the cyclic flexural response [12, 13]. Bending collapse of vehicles is a dominant failure mode experienced during oblique or side collisions of vehicles hence the interest in them in this paper [14]. In order to encourage the widespread adoption of thermoplastic matrix composites in different engineering applications, it is essential to widen the current understanding of their mechanical response. This demands research on both experimental investigations and numerical modelling of these composites. This work aims to generate experimental data on monotonic and cyclic flexural response of a continuous E-glass fibre reinforced polypropylene matrix composites marketed by the trade name PlytronTM [15, 16]. The interest in PlytronTM stems from the fact that it is widely used in automotive parts and has an unusually high volume fraction of matrix (i.e. 65%) compared with other thermoplastic composites.

A few authors have carried out experiments on polypropylene-based composites and such studies were focussed on the effect of processing histories on the mechanical performance of the test materials. Al-Zubaidy and co-workers [17] investigated the tensile and shear properties of orthotropic glass-polypropylene composites made under different processing conditions and found that crossply laminates had excellent tensile and shear properties with interlaminar shear strength (ILSS) being 2.5 times the ILSS of other thermoplastics composites made using the same processing history. Al-Zubaidys work also showed that PlytronTM had comparatively higher fracture toughness compared to similar composites and this was attributed to the high coupling between the fibre and the matrix components.

Data obtained from impact studies of commingled E-glass fibre polypropylene composites by Santulli and co-workers [18] were compared with those of PlytronTM and the authors reported better mechanical properties of PlytronTM than similar polypropylene-based composites. In particular, the flexural response of PlytronTM was comparable to the properties of the more aligned 4:1 weave TwintexTM composite. This observation was attributed to both the non-crimped form of the fibre reinforcement as well as improved impregnation achieved in PlytronTM composites. These excellent properties of make PlytronTM an interesting material to study.

As well as Al-Zubaidy's [17] and Santulli's [18] works, another study by Rijdsdijk and co-workers [19] assessed the role of interphase modification on the mechanical response of polypropylene-based composites. The work concluded that any modifications affected significantly those composite properties that depend on the interphase, like transverse, shear and compressive strength. Thomason and colleagues [20, 21] confirmed this by carrying out a single-fibre pullout test on composites modified using different interface coatings. The flexural strength of the glass fibre/polypropylene composite was found to vary by a factor of two depending on the type of glass fibre coating used.

Most recently, Hagstrand and co-workers [22] investigated the effect of percentage void volume fraction to the mechanical behaviour of commingled E-glass fibre polypropylene-

53 matrix composites. They hoped that by optimizing manufacturing techniques to minimize
 54 formation of voids, composites of improved mechanical responses could be obtained. The
 55 authors found the presence of voids had negative effects on the flexural modulus and strength
 56 except surprisingly the flexural rigidity, EI which was found to increase by 2% with each 1%
 57 increase in percentage void volume fraction. It was thought that the presence of voids led
 58 to increase in dimensions of tested specimen hence increasing the moment of inertia.

59 Simeoli and colleagues [23] reported that changes to the interface strength of PP/E-glass
 60 fibre laminates affected their low velocity impact behaviour. By incorporating a compatibi-
 61 lizer, the authors found that the flexural modulus and the strength of the composite were
 62 significantly increased. The authors observed that interface failure occurred at low strains
 63 for non-compatibilizer-strengthened composites. The authors explained that the failure re-
 64 sulted from large energy dissipation occurring at the polymer/fibre matrix. This conclusion
 65 was confirmed using what they described as locked-in thermographic analysis of the tested
 66 specimen [24].

67 This short review into mechanical (especially flexural) response of PP/E-glass fibre com-
 68 posites highlights the sustained research interest in such composites but most significantly
 69 shows the importance of accumulating more experimental data on thermoplastic composites
 70 with dominant matrix composition as PlytronTM. The development of micromechanical and
 71 macroscale constitutive models of thermoplastic composites will benefit immensely from re-
 72 liable experimental data generated through uniaxial, flexural, shear and fatigue testing of
 73 such composites. Motivated by this need, this paper presents experimental data on a series
 74 of experiments carried out with specific focus of understanding the monotonic and short
 75 cycle flexural behaviour of continuous E-glass fibre polypropylene matrix composite.

76 2. Test material

77 2.1. PlytronTM : a continuous polypropylene/E-glass fibre composite

78 The test material under investigation is continuous polypropylene/E-glass fibre reinforced
 79 composite. PlytronTM is the registered trademark for this continuous unidirectional glass
 80 fibre reinforced polypropylene composite made by a Swiss company called Gurit Suprem
 81 but now trades as Gurit. It is a 100% consolidated, thermoplastic composite which is
 82 commercially available as prepreg tapes of 300 mm wide, 0.25 - 0.28 mm thick and roll
 83 length of 400 m. The reinforcement is obtained with continuous unidirectional glass fibres
 84 with a weight ratio of 60-wt% (or 35-vol% fibre)[15, 16]. The matrix phase is a blend of
 85 standard polypropylene and 5% master-batch compound. The master-batch contains carbon
 86 black among other proprietary ingredients, hence giving the composite a black appearance.
 87 Plytron is commonly used in the automobile industries. Table 1 gives the manufacturer's
 88 data for typical mechanical properties of PlytronTM based on experiments performed on
 89 unidirectional and symmetric cross-ply laminates.¹

90 2.2. Manufacture of test material

91 Composite laminates were prepared from PlytronTM prepreg tapes through compression
 92 moulding in a heated press. In preparing test specimens, plies of dimensions of 140 × 140

¹The n in $[0_n]$ and $[(0/90)_n]_s$ within Table 1 represents the number of plies and takes values ranging from 5 - 12, whilst s represents *symmetric*.

Table 1: Physical and Mechanical Properties of PlytronTM [15, 16].

Properties	Units	$[0_n]$	$[(0/90)_n]_s$
Density, ρ	g/cm^3	1.48	1.48
Fibre Content	gew%	60	
	vol%	35	
Tensile Strength, X_t	MPa	680	360
Tensile Modulus, E_{xx}	GPa	22.5	16
Elongation at break, ϵ_f	%	2.1	2.5
Flexural Strength, $\sigma_{b,max}$	MPa	570	350
Flexural Modulus, E_b	GPa	22	16.5
Thermal Expansion Coefficient, α_T	$\mu\text{m/mK}$	7	20

93 mm were cut from the reel of prepreg tapes. The plies were then arranged according to
94 a desired stacking sequence and placed inside a picture-frame mould specially designed for
95 the moulding of test specimens. A PTFE spray was applied onto the inside lid and inside
96 base plates of the mould to help in easy removal of the laminates after moulding. Each
97 moulding was made under the optimal processing conditions of: (a) temperature of heated
98 press platens: $22^\circ - 250^\circ \text{C}$, (b) pressure applied on top and bottom of the mould: 1.5 - 2.0
99 MPa, (c) number of plies per laminate: 12 plies (for a laminate of thickness 3.0mm), and
100 (d) processing cycle: 25 mins (comprising 10 mins heating up, 5 minutes dwell time and
101 10 mins cooling period). Typical heating and cooling rates were 15°C/min and 20°C/min
102 respectively. Successive loading and unloading cycles, over 30-second intervals from start of
103 heating, were applied to ensure that any trapped air pockets that would cause voiding were
104 forced out. The mould was cooled from the set platen temperature to room temperature by
105 water cooling of the heated platens.

106 2.3. Microscopy of test composite

107 The microstructure of the manufactured test material was assessed using optical and
108 scanning electron microscopy studies. These studies were carried out to explore the nature
109 of the interaction/bonding between the matrix and the glass fibre reinforcement and also
110 check for the absence of voids in order to ensure the suitability of the test specimens for
111 flexural tests. The laminates under investigation have a unidirectional layup. They were
112 cut into specimens of dimensions of $10 \times 10 \text{ mm}^2$ using a band-saw with fine blades. The
113 specimens were mounted on Bakelite and polished for between 40-60 mins using Kemet
114 self-adhesive cloths diamond compound of grade 6-KD-C2 and a lubricating fluid. Once
115 satisfactory polishing was achieved, the test specimens were imaged in an Alicona Infinite
116 Focus profilometer set in 2D imaging mode. Micrographs were obtained in the through-
117 thickness section of the test materials. This assessment reveals that the microstructure of
118 PlytronTM consists of clearly defined regions of matrix-rich and fibre-rich zones (see *Figures*
119 *1(a) to 1(c)*) - typical of laminated composites. Scanning electron microscopy was used
120 to obtain micrographs demonstrating excellent bonding between the matrix and fibre as
121 shown in *Figure 1(d)*. This confirms that the laminate making process achieved high level

122 of consolidation. In all the assessed micrographs, voiding was very minimal.

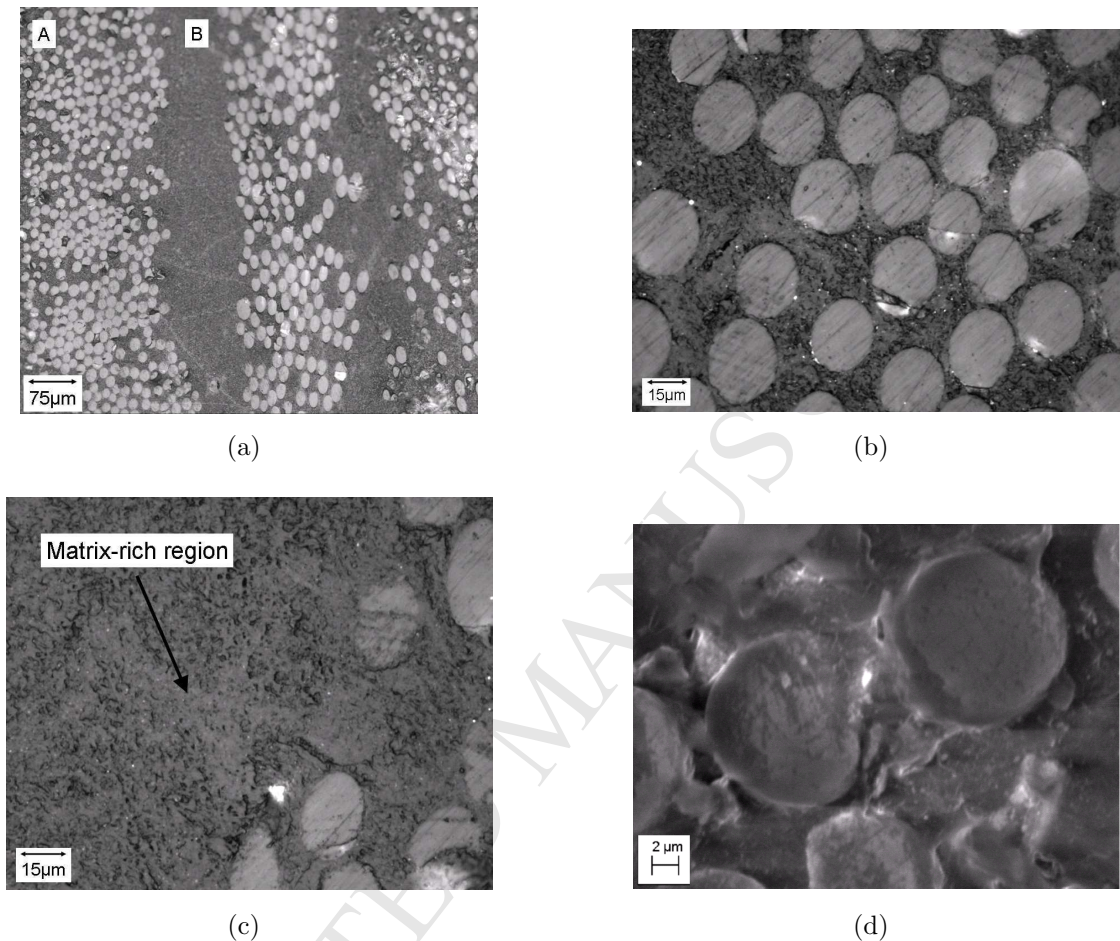


Figure 1: Optical micrographs of a typical test specimen showing: (a) banded arrangement of fibre (grey circles) within the matrix (black region); (b) random arrangement of fibres within fibre-rich zone (A); and (c) limited distribution of fibres within matrix-rich zone (B); as well as (d) zoomed-in view of two individual fibres: showing good fibre-matrix consolidation, and absence of voids.

123 3. Experimental test setup and specimen design

124 Compression moulded laminates with different fibre orientations were cut into beam spec-
 125 imens and tested by three-point bending test. Symmetric laminates of stacking sequences
 126 were studied where $\theta = 0^\circ, 15^\circ, 30^\circ, 45^\circ, 60^\circ, 75^\circ$, and 90° and $n = 5$. Also, cross-ply speci-
 127 mens of stacking sequence $[(0/90)_5]_s$ were also tested. *Figure 2* shows the geometry of three
 128 point bending test specimens.

129 According to the [ISO 14125:1998 test standard for flexural test of fibre-reinforced plastic](#)
 130 [composites](#), the dimension of the test specimen was chosen as $70 \times 20 \times 2.8 \text{ mm}^3$. All tests
 131 were carried out using an Instron Series IX Automated Materials Testing System 4204. A
 132 three-point bending test rig, shown in *Figure 3*, was fitted to the Instron machine such
 133 that when tensile load is applied on the end supports of the test specimen, the specimen
 134 responds in bending around a mid-span fulcrum support of radius 2.5 mm. Flexural forces

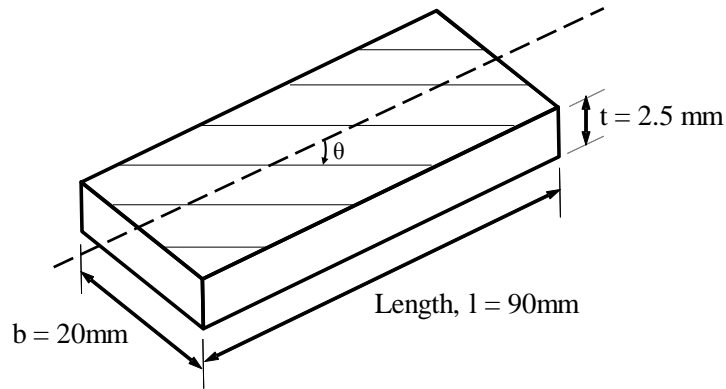


Figure 2: Design of a three-point bending test specimen.

135 were measured using a 5 kN load cell while the deflection of the beam was measured by
 136 crosshead displacement. Tests were carried at crosshead speed of 5 mm/min.

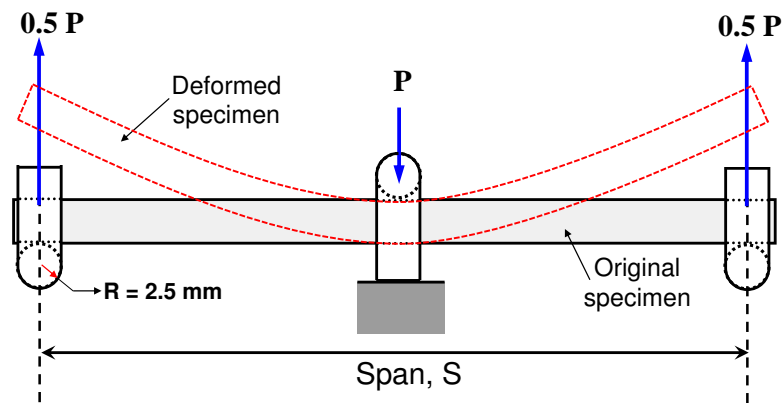


Figure 3: Schematic representation of a three-point bending test rig highlighting the test specimen and loading arrangement that enforces the bending response of the test material.

137 4. Test Results

138 4.1. Monotonic flexural tests

139 The flexural response of PlytronTM is presented in plots of Force, F [N] against mid-span
 140 deflection of the beam, δ [mm]. The maximum flexural force, F_{max} is identified as the peak
 141 force on the flexural force-deflection plot. Figures 4(a) - 4(h) show plots of flexural responses
 142 of PlytronTM laminates for fibre orientations from 0° to 90° . The 0° test specimens refer to
 143 unidirectional composites where the fibre-axis aligns with the longitudinal neutral axis of
 144 the test specimen. The plot for the 0° laminates was dominated by the linear elasticity of
 145 the fibre, as shown in Figure 4(a). Similarly, the flexural response of 90° test specimens was
 146 determined transverse to the fibre's longitudinal direction. In this later case, the flexural
 147 response was dominated by the nonlinear viscoelasticity of the matrix, as shown in Figure
 148 4(f). Also, three-point bending tests were carried out on cross-ply laminates and the result
 149 is shown in Figure 4(g). The comparison showing the full range of the monotonic flexural

150 response of this class of composite is shown in *Figure 4(h)* which shows the dependence of
 151 the composite flexural response with fibre orientations. These results show, as expected that
 152 the flexural response of the thermoplastic matrix composite was linear elastic till yield for
 153 fibre-dominated directions (i.e. for fibre orientation, $\theta = 0^\circ, 15^\circ$ and $[(0/90)_5]_s$) while for the
 154 other fibre orientations, the plasticity of the matrix dominated the flexural response.

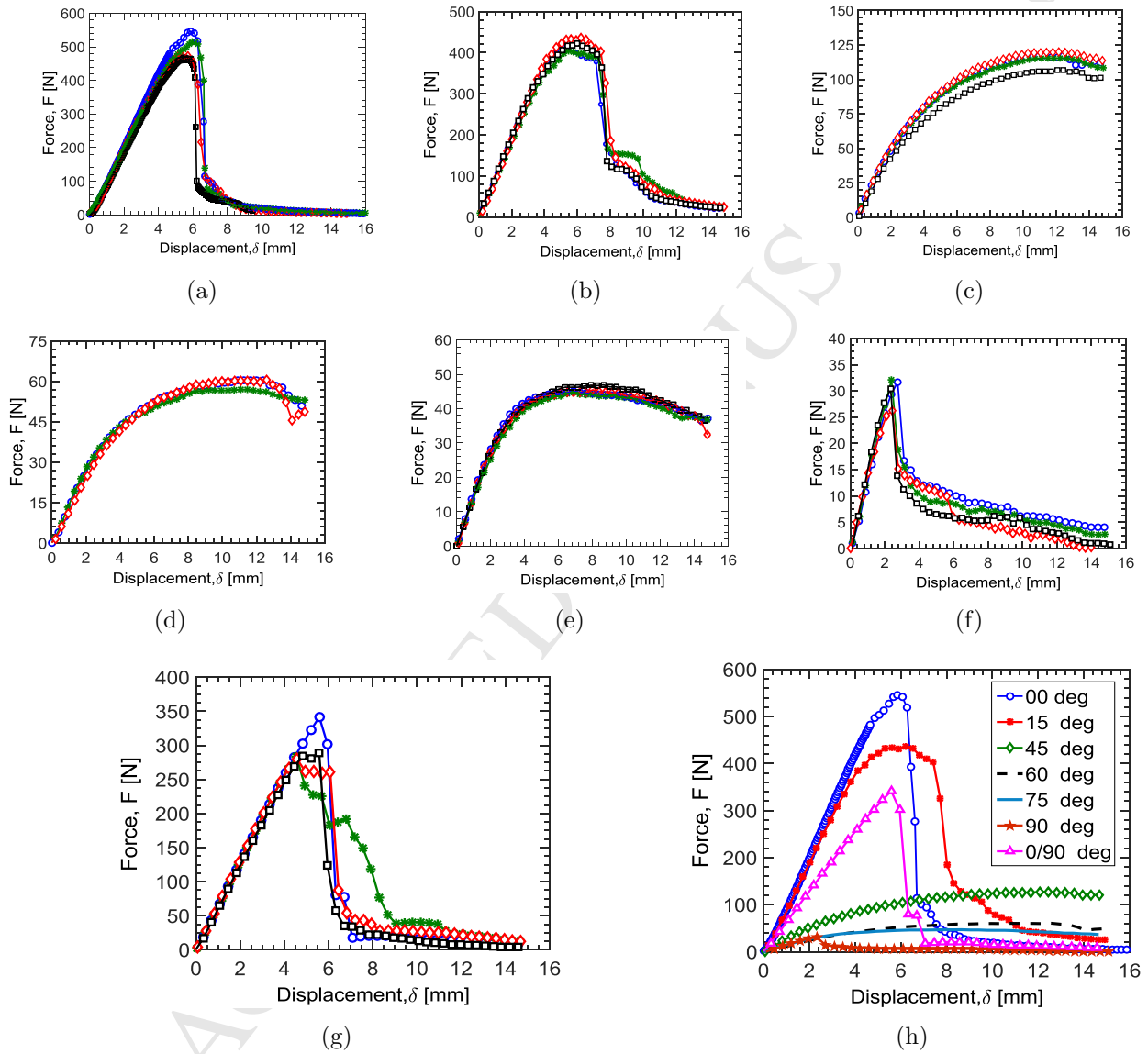


Figure 4: Monotonic flexural response of PlytronTM laminates for the following stacking sequences: (a) $[0]_{10}$, (b) $[(\pm 15)_5]_s$, (c) $[(\pm 45)_5]_s$, (d) $[(\pm 60)_5]_s$, (e) $[(\pm 75)_5]_s$, (f) $[90]_{10}$, (g) $[(0/90)_5]_s$, (h) comparison of all fibre orientations.

155 4.2. Short cycle flexural tests

156 The aim here was to investigate the flexural response of PlytronTM under short cycle
 157 loading. Similar specimen design and test rig used in the previous section were used in these
 158 tests. The test specimens were tested in the Instron Machine in displacement-control mode.

159 A limiting deflection, δ_{limit} (herein referred to as *cyclic deflection limit* (CDL)), was chosen
 160 such that deflection does not exceed 80% of the flexural peak load (i.e. $\delta_{limit} \leq 80\%$) for
 161 each laminate under consideration. This δ_{limit} also ensures that the loading regime exceeds
 162 the elastic limit of the laminates but not approach the unstable region of onset of failure.
 163 As a result, it was possible to investigate as reported in *Section 5.4*, the accumulation of
 164 microscopic damage until eventual failure.

165 All test specimens were subject to cyclic loads of up to 5 cycles and typical force-deflection
 166 plots for a cross-ply and four angle-ply laminates are shown in *Figure 5(a) - 5(e)*. Only 5
 167 cycles were chosen as this study was aimed at short cycle fatigue. **Traditionally, during**
 168 **fatigue tests, tests specimens are subjected to hundreds of thousands of cycles, in order**
 169 **to assess the cyclic response of the test material.** For the purpose of this work, it has
 170 been observed, and reported later in *Section 5.3* that after the first three - five cycles, the
 171 composite experiences a cycle independent energy dissipation which is the same irrespective
 172 of increasing number of cycles. **As a result, this work has focussed on the short cycle fatigue**
 173 **of the test composite - in the region within which significant changes in the mechanical**
 174 **response of the composite - is observed.**

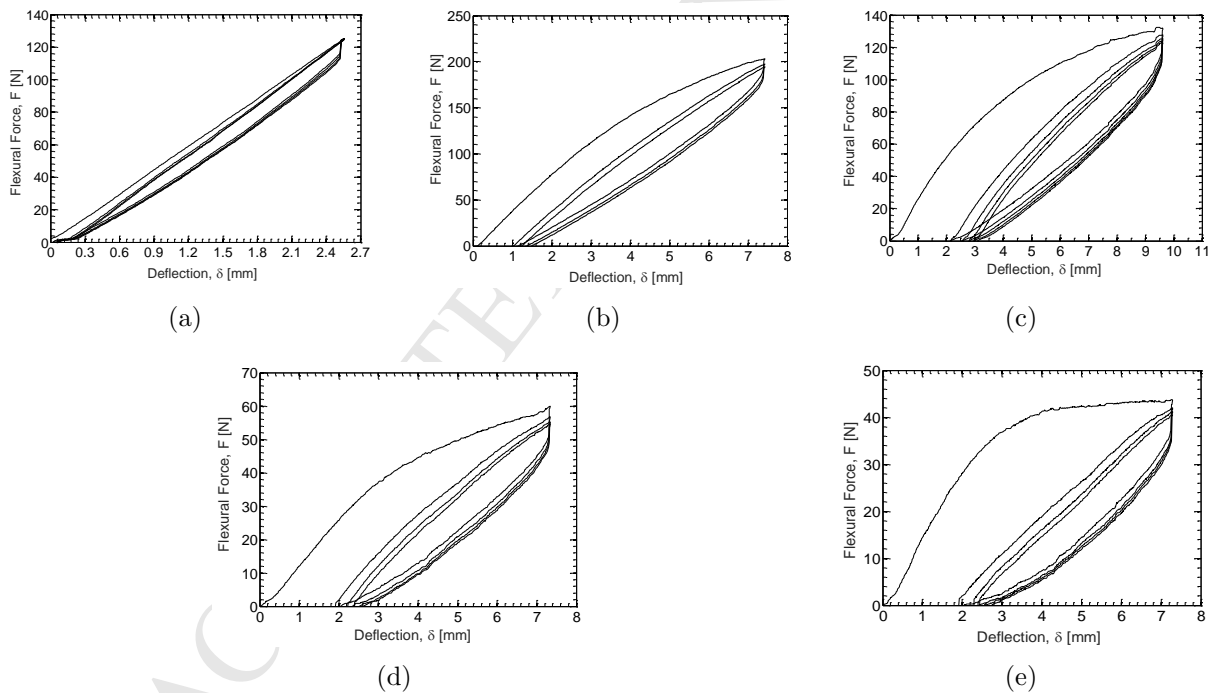


Figure 5: Short cycle flexural response of PlytronTM laminates for the following stacking sequences: (a) cross-ply $[(0/90)_5]_s$; angle-ply: (b) $[(\pm 30)_5]_s$, (c) $[(\pm 45)_5]_s$, (d) $[(\pm 60)_5]_s$, and (e) $[(\pm 75)_5]_s$ laminates.

175 5. Discussions

176 5.1. The mechanics of the composite's flexural response

177 At very small deflections, $\delta \ll t$ where $t =$ thickness, the principles of classical lamina-
 178 tion theory [25, 26] can be applied to the analysis of the linear elastic flexural response of

179 the symmetric laminates tested here. Consider a simply supported rectangular plate of di-
 180 mensions: length, L , width, b and thickness, t subjected to bending. The moment-curvature
 181 relationship [25] of a laminated composite plate is:

$$\begin{bmatrix} M_x(t) \\ M_y(t) \\ M_{xy}(t) \end{bmatrix} = \begin{bmatrix} D_{11}(t) & D_{12}(t) & D_{16}(t) \\ D_{12}(t) & D_{22}(t) & D_{26}(t) \\ D_{16}(t) & D_{26}(t) & D_{66}(t) \end{bmatrix} \begin{bmatrix} \kappa_x(t) \\ \kappa_y(t) \\ \kappa_{xy}(t) \end{bmatrix}. \quad (1)$$

182 Here, $M_x(t)$ is a rate-dependent bending moment per unit width of the beam about
 183 the plane containing the bar axis; $D_{ij}(t)$ is rate-dependent laminate bending stiffness, for
 184 $i, j = 1, 2, 6$ and $\kappa_x(t)$ is the resulting rate-dependent curvature about the x-axis. For
 185 ease of writing, the rate-dependence is assumed and not explicitly written, hence $M_x(t)$ is
 186 written as simply M_x , as well as flexural stress, $\sigma(t)$ becomes σ . In a pure bending test,
 187 $\kappa_y = \kappa_{xy} = 0$. Therefore, the moment-curvature expression - based on Equation 1 - for
 188 analysing the three-point bending tests becomes:

$$M_x = \kappa_x D_{11} \quad \Rightarrow \quad M = b \kappa_x D_{11} = \frac{b D_{11}}{\rho_x}, \quad (2)$$

189 where b = width of beam, M = total bending moment, and radius of curvature, $\rho_x = \kappa_x^{-1}$.
 190 For a simply supported composite beam of span, L subjected to a three-point bending by a
 191 concentrated force, F - imposed on the middle of the beam, the expression of the force per
 192 unit deflection is:

$$\frac{F}{\delta} = \frac{48EI}{L^3} \quad \Rightarrow \quad EI = \frac{L^3}{48} \left(\frac{F}{\delta} \right) = D_{11}. \quad (3)$$

193 According to simple beam theory, $M_x = \kappa_x EI_x$ which implies that the expression for
 194 the D_{11} element of the bending stiffness matrix, \mathbf{D} of the composite becomes equation 3.
 195 The expressions for calculating the flexural strength, σ_{max} and flexural modulus, E_f of the
 196 PlytronTM laminates are given in Equation 4. For calculating the strength and flexural
 197 modulus of the test composite, the x-axis moment of inertia for a rectangular cross-section
 198 beam is $I_x = \frac{bt^3}{12}$ and the following equation was used:

$$\text{Strength : } \sigma_{max} = \frac{3PL}{2bt^2}, \quad \text{and} \quad \text{Flexural Modulus : } E_f = \frac{1}{4b} \left(\frac{L}{t} \right)^3 \frac{F}{\delta}. \quad (4)$$

199 A MATLAB script called *LamPower* was created based on the principles of the com-
 200 monly used classical lamination theories[25] namely: *rule of mixtures*, *Halpin-Tsai equations*,
 201 Spencer's *square array model* as well as *simplified micromechanics equations* of Chamis[27].
 202 Also, the much recently published closed-form micromechanics equations herein called Morais'
 203 *self-consistent model*[28] was also implemented within *LamPower*. The script was used to
 204 generate the \mathbf{D} -matrix for a given laminated stacking sequence, having specified the elastic
 205 properties and volume fractions of the constituents as well as the ply size and the stacking
 206 sequence of the laminates. The output from *LamPower* was used to derive predicted flexural
 207 modulus used in the next section.

208 *5.2. Comparison of experiments with laminate theory predictions*

209 The objective here is to compare experimentally determined laminate bending stiffness
 210 of PlytronTM with predictions based on the classical laminate theory established in *Section*
 211 *5.1* above. In the expression for D_{11} in *Equation 3*, the term $\frac{F}{\delta}$ (unit = [N/mm]) is the
 212 slope (for $\delta \ll t$) of the force-deflection plots of *Figure 4*. Using the deflection of $\delta = 0.5$
 213 mm and the Modulus expression of *Equation 4*, *Figure 4* data were used to determine the
 214 laminate bending stiffness for all six fibre orientations. The properties of the fibre and
 215 matrix constituents, given in *Table 2*, were used to determine the D_{11} term of the \mathbf{D} -
 216 matrix. Predicted values of the bending stiffness were determined using the *LamPower*
 217 implementation of the classical laminate theory.

Table 2: Typical room temperature properties of E-glass fibre reinforcement and isotropic semicrystalline polypropylene matrix used in PlytronTM.

Property	Fibre	Matrix	Units
Density, ρ	2600	900	kg/m ³
Young's Modulus, E	73	1.308	GPa
Tensile Strength, X	2250	40	MPa
Shear Modulus, G	31	0.46	GPa

218 In determining the predicted values, the size of each laminate ply was found to be
 219 0.25 ± 0.002 mm. This was calculated from the plot of number of plies, N used per laminate
 220 against thickness, t of the laminate. Six mouldings of different thicknesses ranging from
 221 $t = 2.5$ mm to $t = 10$ mm were used. The ply size: 0.25 ± 0.002 mm, was determined
 222 by substituting $N = 1$ (for one ply) into the linear function of plot of number of plies per
 223 laminate versus laminate thickness. *Figure 6* shows the comparison of the experimental and
 224 predicted variation of the laminate flexural modulus, E_f with ply orientation, θ . The results
 225 in *Figure 6* show a good agreement, within experimental error, between the experimental
 226 and predicted E_f values for all fibre orientations based on the Morais *self-consistent model*
 227 [28]. According to this result, the Spencer's *square array model*[25] and Chamis' *simplified*
 228 *micromechanics model*[27] established upper and lower bounds to the E_f values.

229 *5.3. Quantifying the energy dissipation for cyclic flexural response*

230 The stress-strain cyclic profile following a uniaxial fatigue test gives a measure of the
 231 stiffness degradation or energy dissipation experienced by a test material. Similarly, the
 232 force-deflection plot of a three-point bending tests as reported here can give information
 233 of degradation of stiffness or energy dissipation as the uniaxial tests [29–31]. In order to
 234 quantify the cyclic flexural response, let us define an *energy dissipation factor*, ζ_d as in
 235 *Equation 5*:

$$\text{Energy dissipation factor, } \zeta_d = 1 - \frac{E_{unloading}}{E_{loading}}, \quad (5)$$

236 where $E_{loading}$ is the strain energy of the loading curve segment whilst $E_{unloading}$ is the
 237 strain energy of the unloading curve segment. These strain energies were determined as the

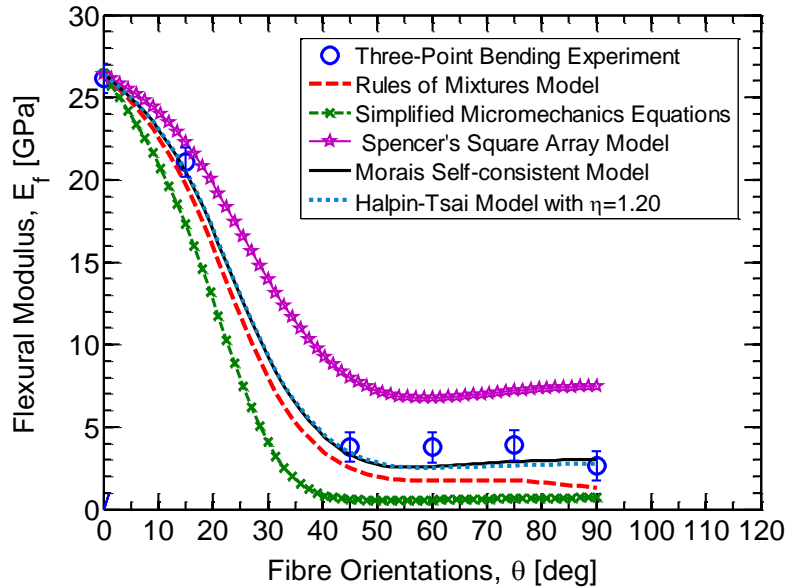


Figure 6: Comparison of experimental and predicted flexural modulus, E_f . The prediction was based on selected composite lamination theories. *Note: The error bars was determined by calculating the standard deviation of the slope of the force-displacement plots for all the tested specimens per given fibre orientation*

238 areas under the appropriate curve segment. The areas were calculated numerically using the
 239 trapezoidal rule. For the three-point bending test, the ζ_d parameter represents a quantitative
 240 measure the energy dissipation following successive cyclic loading of the the material[29].

241 Consider the cyclic flexural response of $[(\pm 45)_5]_s$ PlytronTM laminate, the test specimen
 242 was subjected to 5 cycles of flexural loading as shown in *Figure 7(a)*. The short cycle
 243 flexural response the PlytronTM laminates shown in *Figure 7(b)* indicates that at the end
 244 of the unloading curve, the laminate does not return to zero deflection but there exists a
 245 residual deflection. For the first cycle of the $[(\pm 45)_5]_s$ PlytronTM laminate, the residual
 246 deflection was 2.1 mm and by the fifth cycle, this has reduced 0.32 mm. Similarly, the area
 247 under the load-unload curves evolves from 504 N-mm to 147 N-mm.

248 It is customary to represent ζ_d as a percentage, thus providing a measure of energy
 249 dissipation between successive cycles. Using the cyclic plots of *Figure 5* and *Equation 5*,
 250 the plots of energy dissipation factor, ζ_d against number of cycles, N for both angle-ply and
 251 cross-ply laminates are shown in *Figure 8*. The plot of *Figure 8* shows that for increasing
 252 angle, θ between the fibre and the main axis of bending, the percentage energy dissipation
 253 factor, ζ_d increases. Cross-ply laminates show very small energy dissipation compared with
 254 the angle-ply laminates.

255 The data in *Figure 8* shows the most significant energy loss occurred between the first
 256 two cycles. For example, for the $[(\pm 75)_5]_s$ laminates, the percentage change in energy
 257 dissipation $\Delta\zeta_d = 70\% - 46\% = 24\%$ whilst for the $[(0/90)_5]_s$ laminates, $\Delta\zeta_d = 3\%$. The
 258 cyclic flexural response of the test composite indicates significant energy dissipation in those
 259 laminate arrangements that experience significant plastic deformation. Therefore, structural
 260 applications requiring significant energy absorption in the first loading cycle can be made
 261 based on the test material with the $[(\pm 45)_5]_s$ laminate arrangement.

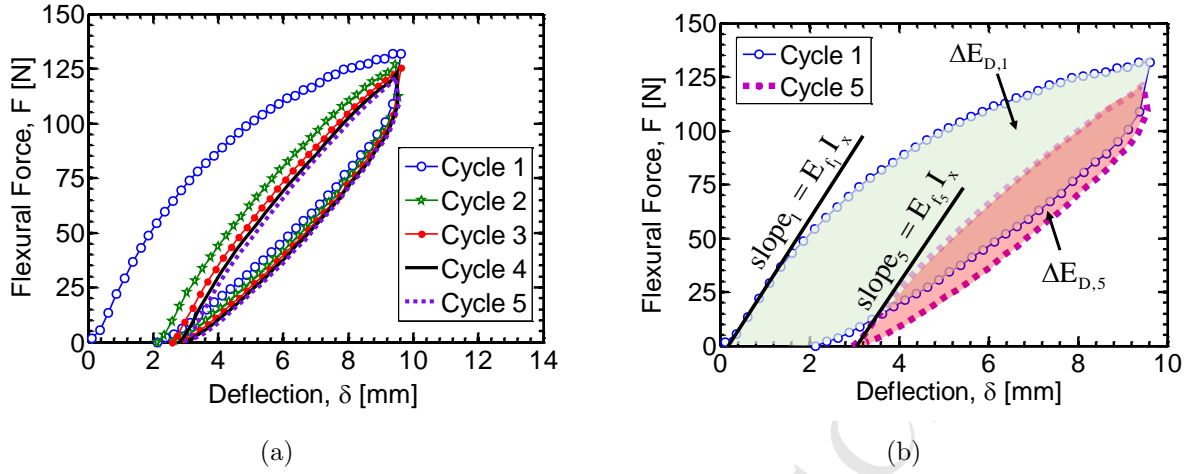


Figure 7: Short cycle flexural response of the $[(\pm 45)_5]_s$ laminates for: (a) all 5 cycles and : (b) first and fifth cycles. The energy dissipation for an n – th cycle ΔE_n is shown as well as the slopes for the first and fifth cycles where the tested specimen flexural modulus of n – th cycle is $E_{f,n}$ and the area moment of inertia about the x-axis is I_x .

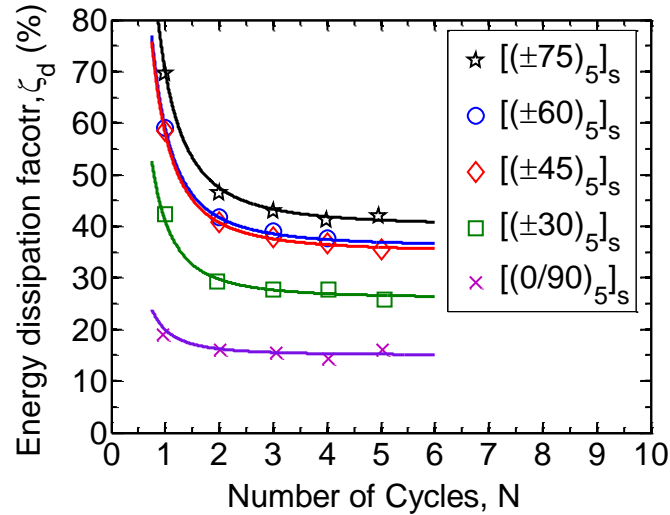


Figure 8: Plots of energy dissipation for all the tested composites. Note: Solid trend lines indicate model predictions while markers are experimental data.

262 To quantify mathematically the dependence of percentage energy dissipation factor, ζ_d
 263 with number of cycles, N , this study established that the numerical fit of the experimental
 264 data of *Figure 8* can be expressed by the power law:

$$\zeta_d = AN^b + \zeta_{d,0} \quad (6)$$

265 where ζ_d = energy dissipation factor, N = number of cycles, $\zeta_{d,0}$ = threshold energy
 266 dissipation factor and finally A, b are material constants. Typical values of A, b and $\zeta_{d,0}$
 267 for the laminates arrangements tested here are shown in *Table 3*.

268 *Equation 3* represents a phenomenological model for characterizing the short cycle flexu-
 269 ral response of PlytronTM laminates. The A -parameter is a measure of *first-cycle percentage*

Table 3: Typical values of material constants for the energy dissipation function associated with short cycle flexural behaviour of PlytronTM laminates.

Stacking Sequence	A	b	$\zeta_{d,0}$
$[(0/90)_5]_s$	5	-2.0	15
$[(\pm 30)_5]_s$	15	-2.0	26
$[(\pm 45)_5]_s$	23	-2.0	35
$[(\pm 60)_5]_s$	23	-2.0	36
$[(\pm 75)_5]_s$	30	-2.0	40

270 *change in energy dissipation.* The A -parameter is high for laminate arrangements that show
 271 dominant plastic deformation (e.g. the $[(\pm 75)_5]_s$ and $[(\pm 60)_5]_n$ angle-ply laminates). The
 272 $\zeta_{d,0}$ -parameter is a measure of the energy dissipation of the test material that remains con-
 273 stant with increasing number of cycles. Kar and co-workers [31] reported similar observation
 274 for the bending fatigue of hybrid composites.

275 5.4. Damage accumulation in cyclic flexural response

276 With increasing number of cyclic loadings, the mechanical properties of composite ma-
 277 terials degrade progressively. The global damage caused by the cyclic flexural loading of
 278 the test composite can be measured effectively by measuring the stiffness degradation of the
 279 material [31–33]. An objective measure of this degradation is the global damage index, D
 280 which can be defined as:

$$D = \left[1 - \frac{(EI)_n}{(EI)_1} \right], \quad (7)$$

281 where $(EI)_n$ is the cyclic flexural rigidity after the n -th cycle, and $(EI)_1$ is the flexural
 282 rigidity after the first cycle for the tested cyclic deflection level, δ_{limit} . These flexural moduli
 283 are related to a chosen δ_{limit} and should be differentiated from statically determined flexural
 284 modulus, $(EI)_0$ which is a reference value for monotonic flexural test deformed till failure
 285 of the test material. Several microscopic damage mechanisms are assumed to contribute to
 286 the D . For the thermoplastic matrix composites under investigation, these damage/failure
 287 mechanisms are usually matrix-dominated and can include: matrix cracking (or more gen-
 288 erally inter-fibre failure), fibre-matrix debonding, ply delamination, fibre rupture and finally
 289 fibre kinking[34]. Kar and co-workers [31] observed that with subsequent cycles, the micro-
 290 scopic damage continues to localize especially in the matrix and matrix-interphase phases
 291 until eventual failure occurs. The final flexural rigidity at failure, $(EI)_f$ is calculated at the
 292 last loading cycle before eventual failure. This will yield a GDI value of D_f defined as global
 293 damage index at failure of the test material.

294 Combining Equation 3 and the cyclic flexural data of Figure 5, the flexural rigidity, EI
 295 for each loading cycle was determined. These EI 's were then used with Equation 7 to
 296 calculate the associated *global damage index*, D for the last n -th cycle (representative of
 297 accumulated damage). The plot of the percentage *global damage index*, D against the tested
 298 laminate arrangements is shown in Figure 9.

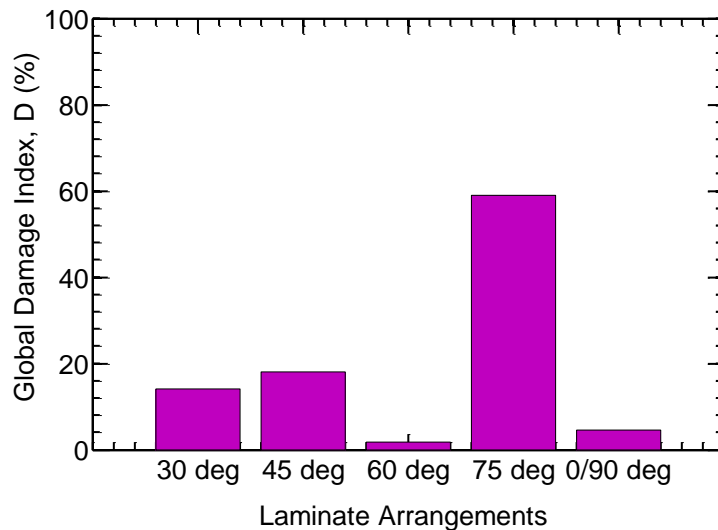


Figure 9: Histogram of global damage index, D for all laminate arrangements. Note: D is expressed here as a percentage, for easy interpretation of the extent of damage.

299 The result shown in *Figure 9* shows that damage accumulation was the most in the
 300 $[(\pm 75)_5]_s$ laminates and the least in the $[(\pm 60)_5]_s$ laminates. The high damage accumula-
 301 tion seen in the $[(\pm 75)_5]_s$ indicates that for same cyclic deflection level (CDL), δ_{limit} , more
 302 microscopic damage occur than in say the $[(\pm 60)_5]_s$ laminates. Gamstedt and Talreja [35]
 303 investigated the fatigue damage mechanisms in unidirectional carbon-reinforced composites
 304 with thermoset and thermoplastic matrices. The study showed the fatigue damage is a
 305 cumulative effect of microscopic damage of the matrix and was particularly pronounced
 306 in the thermoplastic (PEEK) matrix composites. These microscopic damages comprise of
 307 widespread propagating debonds and matrix cracks. The authors did not assess the effect
 308 of changing fibre orientation on the evolution of these microscopic damage. This evidence,
 309 amongst several others, support the claim in this work that the observed damage accumula-
 310 tion as shown in *Figure 9* is a consequence of progressively accumulating microscopic failure
 311 mechanisms

312 6. Conclusions

313 The paper reports on the monotonic and cyclic flexural response of continuous E-glass
 314 fibre reinforced polypropylene composites (marketed as PlytronTM). By changing the ori-
 315 entation, θ of the fibre to the main bending axis, the flexural response of the laminates
 316 changed from a linear elasticity dominant response until damage to a plasticity-dominant
 317 response. Therefore, the effect of the matrix is significant in understanding the flexural
 318 response of this type of thermoplastic matrix composite. Experimentally-derived flexural
 319 modulus of all tested laminates were found to agree with predictions of the same based on
 320 classical lamination theories.

321 Similarly, short cycle flexural response of PlytronTM was investigated for five stacking
 322 sequences of the laminates. The study also developed a phenomenological model that de-
 323 scribes the dependence of energy dissipation with number of cycles. The study observed

324 that for all tested laminates, the largest energy dissipation occurred after the first cycle
325 and thereafter, the energy dissipation converges quickly to a cycle-independent energy dis-
326 sipation value which is a material constant. This cycle-independent energy dissipation is
327 thought to be a saturated damaged state (or maximum energy-dissipation) within the test
328 material. The size of initial and saturated energy dissipation increases as the angle of fibre
329 orientation relative to the main bending axis increases too.

330 In conclusion, the flexural response of continuous E-glass fibre reinforced polypropylene
331 composites has been studied experimentally. It is concluded here that the plastic deforma-
332 tion of the matrix contributes significantly to the flexural response. To encourage wider
333 application of this type of material in structural designs, it is imperative that a constitutive
334 model need to be developed. On evidence from this work, such model development must
335 consider the contribution of a robust matrix model towards the accurate prediction of the
336 mechanical response of PlytronTM. The experimental data presented here and the associ-
337 ated phenomenological models will help in the micro-mechanical modelling for the class of
338 thermoplastic matrix composites as PlytronTM with a comparatively higher matrix volume
339 fraction.

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