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Effects of thermoplastic preforming binder on the properties of S2-glass fabric reinforced epoxy composites

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Abstract

The effect of a thermoplastic polyester binder on the thermophysical and mechanical properties of an S2-glass/epoxy–amine system was investigated. The purpose of the polymeric binder is to bond the individual fabric layers together during preforming prior to composite fabrication. This paper will address the significance of the binder chemistry, i.e., the compatibility of the binder with the matrix polymer, on the composite properties. The peel strength of preforms consolidated with various concentrations of binder was evaluated using the T-peel test. The highest peel resistance was obtained from preforms that have full coverage of the binder on the glass fabric. Further increase of the concentration of the binder does not change the peel strength. Scanning electron microscopy (SEM) on peel test fracture surfaces revealed mostly adhesive-type failure between binder and fiber. Double cantilever beam (DCB) and short beam shear (SBS) test results of the composite showed that the presence of about 2.6 wt% of the polyester binder reduces the Mode I interlaminar fracture toughness and apparent interlaminar shear strength of the S2-glass/SC-15 epoxy-amine system by about 60% and 25%, respectively. Moreover, the T_g of the matrix polymer within the interlaminar region decreases about 6°C due to the presence of the binder. The dissolution of the polyester binder within the reacting matrix resin is limited for the standard cure cycle. © 2001 Elsevier Science Ltd. All rights reserved.

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1. Introduction

Resin transfer molding (RTM) and vacuum assisted resin transfer molding (VARTM) techniques have recently been used to fabricate polymeric composite structures for automotive, aerospace and civil engineering applications. A typical polymeric composite is made of reinforcing fabric containing glass and carbon fibers (i.e., preforms) and either a thermoplastic or a thermoset polymeric matrix. RTM and VARTM processes involve several stages: loading the preform into a mold cavity or a flat tool followed by resin infusion, curing and demolding. Preform placement into complex shapes is time consuming and is an important cost driver. Thermoset-based tackifiers have been developed to aid in compaction and placement of preforms prior to composite fabrication. Most recently, thermoplastic binders have been introduced, enabling thermoforming of preforms into net shape. In this approach thermoplastic binders are deposited onto the fabric surface prior to preform consolidation. The application of heat and pressure during thermoforming softens the binder and upon cooling re-solidifies to bond the adjacent plies together. During the consolidation stage, a desired shape and thickness can be obtained from the preform using appropriate tooling and applying pressure and heat on the lay-up. Thermoplastic bound preforms can also be machined to final shape and offer significant potential for cost saving. Binders that bond well to the fibers and do not dissolve in the resin offer the potential to maintain preform compaction, thickness (i.e., higher fiber volume fraction) and overall superior dimensional tolerances. Binders that are compatible with the resin offer the potential to tailor properties at the

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interlaminar region between layers. One must also consider the effects of binders on preform permeability that may affect resin infusion.

The presence of the binder material within the preform may significantly affect the processing of the composite, (i.e., resin flow). Rohatgi et al. [1] showed that the thermoplastic binder might flow along the capillaries during the preform consolidation and block the large gaps that govern the mold filling. If the binder used stays outside the fiber tows and does not dissolve extensively during the infusion of the matrix resin, it lowers the permeability of the preform [1,2]. It was also reported that the processing variables such as modulus of the binder or the area of binder coverage on the fibers controls the feasibility of fabricating net-shaped preforms with reduced springback [1]. Furthermore, Kittelson et al. [3] showed that concentration gradients of the thermoplastic binder within the composite might form during the infusion process due to the migration of the binder aided by the resin flow depending on the dissolution rate of the binder. In the case of a binder that is soluble in the matrix resin, this phenomenon would cause property gradients within the composite part. Moreover, the viscosity of the matrix resin may increase due to the binder dissolution, especially if the binder dissolution time is much less than the resin gelation time [4]. It has been addressed by several researchers [2,5–7] that the chemical compatibility of the binder with the matrix resin is an essential issue for the mechanical (interlaminar shear strength and fracture toughness, etc.) and physical properties (T_g, etc.) of the polymeric composites. However, to our knowledge, limited work has been reported on the effects of the thermoplastic binder on the properties of thermosetting composites.

The present paper reports an investigation of the effects of a thermoplastic polyester binder on the properties of composites made of S2-glass fibers and epoxy–amine resin systems. Scanning electron microscopy (SEM) was conducted to determine the distribution and particle attachment of the polyester binder on the glass fabric. The T-peel test technique (ASTM D-1876-95) [8] was used to determine the strength of preforms having various concentrations of binder and consolidated under vacuum at elevated temperatures. The loci of failure of the peel specimens were inspected with SEM to determine the adhesion mechanism between fiber and the binder. The residual peel strength of the preforms after exposure to the matrix resin was also measured to evaluate the durability of the binder/fiber bonding. Double cantilever beam (ASTM D-5528 94-a) [8] and short beam shear (ASTM D 2344-84) [8] test techniques were used to evaluate the effect of binder on the composite mechanical properties, i.e., fracture toughness and interlaminar shear strength, respectively. Dynamic mechanical analysis (DMA) was employed to evaluate the effect of the polyester binder on the thermophysical properties, i.e., T_g and flexural modulus of the composite. Furthermore, model specimens were fabricated to simulate the interactions between the binder and the matrix polymer and to monitor the extent of dissolution of the binder in matrix resin. The results from model specimens were correlated with the results from composite specimens processed under standard conditions.

2. Experimental

2.1. Materials

S2-glass fabrics with various weight fractions of preforming binder were provided by Solectria. The glass fibers (OCF 463), manufactured by Owens Corning Fiberglass Corporation (Toledo, OH), contained epoxy–amine compatible coupling agent on the surface. Silane coupling agent, already reacted with the glass fiber surface, can react with the epoxy–amine matrix resin. The preforming binder was a powdered (average powder size is about 250 μm) thermoplastic polyester that was applied to the fabric by Solectria. The binder was also obtained separately to study the interactions between the binder and the reacting resin system. The resin system used was an epoxy/amine thermoset system designated as SC-15 and purchased from Applied Poleramic. The epoxy resin was diglycidyl ether of bisphenol-A (DGEBA) and the amine was an aromatic amine. Stoichiometric amount of amine (100–30 weight ratio of epoxy and amine) was used for all samples made in this study.

2.2. Thermophysical and microstructural characterization of the preforming binder

Glass transition temperature (T_g) and elastic flexural modulus (E) of the polyester binder were characterized by means of differential scanning calorimetry (DSC) and dynamic mechanical analysis (DMA). The DSC experiment was carried out by ramping the temperature at 5°C/min from 10°C to 140°C in a Perkin Elmer DSC instrument. The DMA samples were prepared by heating the binder to 60°C (above its T_g) under vacuum in a rubber mold for 5 hours. Vacuum was applied to remove air bubbles from the sample. After the cooling process, the samples were carefully machined to approximately $3 \times 10 \times 52$ mm. DMA was carried out on a Rheometrics Solid Analyzer, model RSA-2, in a 3-point bend-loading configuration by ramping the temperature at 5°C/min from -100°C to 70°C with a liquid-nitrogen-cooling accessory. The specimens were loaded within the linear range of the stress-strain response in flexure, and the loading frequency and maximum strain values were 1 Hz and 2×10^{-4}, respectively.

Degree of tow impregnation was determined with scanning electron microscopy (SEM). The areal coverage of the binder on as-received fabric was determined by analysis of SEM micrographs using an image analyzer. The amount of binder on the glass fabric was measured with burnout experiments. For the burnout test, S2-glass fabrics with binder were dried in a vacuum oven at -75 cm Hg at room temperature until they reached a constant mass, i.e., until all moisture was removed from the fabric. The mass loss was negligible. The fibers were placed in a dry crucible and left in a furnace at 620°C for 35-40 min. The samples were then cooled to room temperature under vacuum and weighed. The amount of the fiber sizing was also measured by means of the same burnout procedure used on the S2–glass fabric without binder. The weight fraction of the binder was calculated by subtracting the weight fraction of the sizing from the weight fraction of the binder including the sizing.

2.3. Preform consolidation and composite fabrication

For preparation of peel test specimens, glass preforms were consolidated from the glass fabrics containing 2.67, 4.83 and 6.77 wt% of the polyester binder on one side. Preforms for the peel test were made by laminating two layers (binder sides up) of glass fabrics 25 mm in width and 305 mm in length. A Kapton® film was inserted in the midplane and on one end of each specimen such that the initial unbound peel length was 76 mm. Preform consolidation took place at 80°C for 30 min under vacuum in a vacuum bag.

S2-glass/SC-15 composites were manufactured with preformed glass fabric with 2.6 wt% and without preform binder (for baseline value) by means of the vacuum assisted resin transfer molding (VARTM) technique. These composites were used for SBS and DCB testing. Prior to infusion, fabric with binder was consolidated at 80°C for 30 min under vacuum. For DCB specimens, a 70 mm long Kapton® film was inserted in the midplane of the composites as a crack initiator for interlaminar fracture toughness test specimens. The composites (about 4.8 mm thick) were cured at room temperature for 12 h and then post-cured at 121°C for 2 h.

2.4. Mechanical property measurements

Peel strength of the consolidated preforms was determined by means of the T-peel test method. The locus of failure of the peel specimens was inspected with SEM to determine the adhesion mechanism (chemical/mechanical) between the fiber and the binder. The peel test was also used to determine the effect of preform consolidation temperature and the exposure of each of the matrix resin components on the peel resistance of preforms. The peel specimens were loaded with the peel fixture shown in Fig. 1 in an Instron universal test machine at a constant cross-head speed of 25.4 mm/min. T-peel strength of the samples was calculated based on the average peeling load per unit width of the bond line. At least five specimens from each material were tested to obtain an average peel strength value.

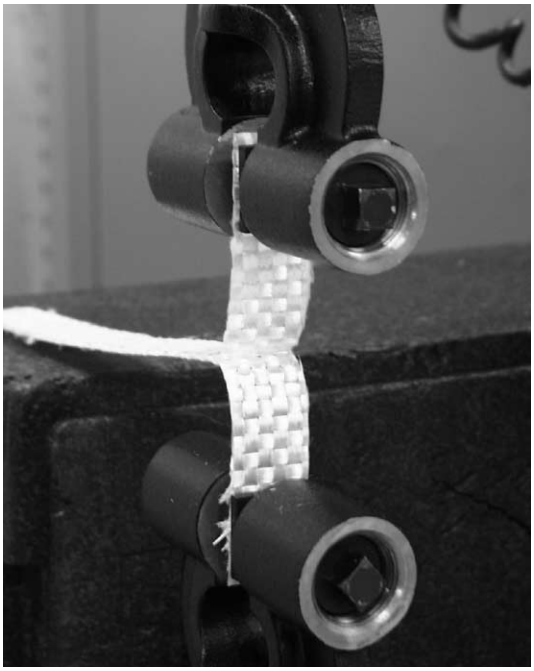
Fig. 1. T-peel test specimen under load.

Mode I interlaminar fracture toughness of S2-glass/SC-15 composites with (2.6 wt%) and without preform binder was measured by means of the double cantilever beam (DCB) method. The DCB specimens were fabricated by sectioning composite laminates to lengths of about 125 mm and widths of 25 mm. The loading blocks were bonded at the insert side of the specimens such that the initial delamination length, a_0 was about 50 mm. At least five specimens from each set were tested on universal Instron machine at a cross-head speed of 0.5 mm/min. The Mode I interlaminar fracture toughness ($G_{\rm Ic}$) values were calculated from Equ. (1), which is based on modified beam theory (MBT):

$$G_{\rm Ic} = \frac{3P_c\delta_c}{2b(a+|\Delta|)}, \tag{1}$$

where $P_{\rm c}$ and $\delta_{\rm c}$, are the applied load and load point displacement at crack initiation at delamination length a and specimen width b, respectively. Since the DCB specimens are not perfect cantilevers (beam cross section deviates from beam theory at the crack tip) the delamination length needs to be corrected by considering $a + |\Delta|$, where Δ is determined experimentally by generating a least square plot of the cube root of compliance $C^{1/3}$ as a function of delamination length, a [8]. The compliance C is given as $C = \delta/P$.

Apparent interlaminar shear strength of S2-glass/SC-15 composites with (2.6 wt%) and without binder was determined by means of the short beam shear (SBS) method. The SBS specimens were sectioned from composite laminates to lengths of 35 mm and widths of 20 mm. The length-to-thickness and span-to-thickness ratios were kept at 7 and 5, respectively. At least 10 specimens from each set were tested on an Instron machine, and the load to break was recorded. The apparent shear strength ($\tau_{\max}$) was calculated as follows:

$$\tau_{\max} = \frac{0.75P_B}{bd}, \qquad (2)$$

where P_{B}, b and d are the breaking load, width of the specimen and thickness of the specimen, respectively.

2.5. Analysis of the interactions between the binder and the matrix resin

Dynamic mechanical analysis (DMA) was conducted on S2-glass/SC-15 composite specimens with and without binder to evaluate the T_g and flexural storage and loss modulus values and to determine the effects of binder on those properties. In addition to composite specimens, model test specimens were synthesized to simulate the interactions between the binder and the reacting resin system. The T_g and flexural modulus values of the model specimens were also characterized using DMA.

Three types of model specimens were synthesized. Table 1 summarizes the procedure used to fabricate the model specimens. The first model specimen is SC-15 at stoichiometric composition representing the neat matrix polymer (baseline). The second specimen was made by simply mixing the binder (3%) and the SC-15 resin. This specimen closely simulates the interaction in real composites. The third specimen was made to simulate the extensive dissolution of the binder in the matrix resin by melting and dissolving the binder in the resin. All of the model specimens were cured at RT for 12 h and post-cured at 121°C for 2 h. The DMA specimens were prepared from slabs of cured material (from composite laminates or model specimens). The specimens were carefully sanded to approximately 2 × 15 × 50 mm.

The second type of model specimen (3 wt% binder mixed at room temperature for 10 min) had a translucent appearance before curing. This indicates limited dissolution of the binder and suspension of the binder particles in the mixture. The more transparent appearance of the third sample (3 wt% of binder mixed at 55°C for 4 hours) indicated the extended dissolution process.

Dynamic mechanical characterization was carried out by ramping the temperature at 5°C/min. from 25°C to 140°C with a Rheometrics Solid Analyzer, model RSA-2, and a 3-point bend-loading configuration. The loading frequency and strain values were 1 Hz and 2×10^{-4}, respectively. In this analysis, T_g is defined as the temperature at which a maximum on loss modulus occurs.

3. Results and discussion

3.1. Preforming binder and binder modified preform fabric characterization

Fig. 2 shows the DSC profile for binder after two heating ramps. The first heating ramp exhibits a peak at about 55°C. This peak is associated with the melting of the binder, and the T_g of the binder is hidden by the melting transition. In the second heating ramp, another transition with lower intensity appears at about 46°C and the T_g is visible as a step change in the baseline. In

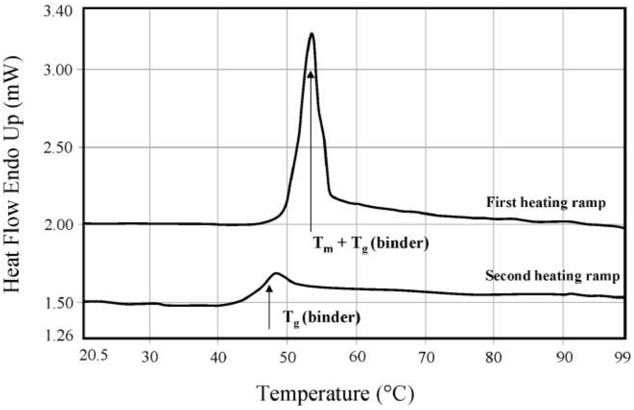

Fig. 2. DSC profile for binder.

Table 1
Experimental procedure to prepare the model and S2-glass/SC-15 epoxy–amine composites (Part A: epoxy, Part B: amine curing agent)

Specimen	Mass % of binder	Preparation procedure	Curing procedure
Neat SC-15	—	Mix for 10 min at room temperature	Cure at RT for 12 h Post-cure at 121°C for 2 h
Blend (physical mix)	3	Mix for 10 min at room temperature	
Blend (extensive dissolution)	3	Stir in Part B at 55°C for 4 h Add Part A Mix all of the polymers for 10 min Mix for an extra 10 min	
Composite without binder	—	VARTM Infusion time: 30 min	
Composite with binder	2.6	VARTM Infusion time: 30 min	

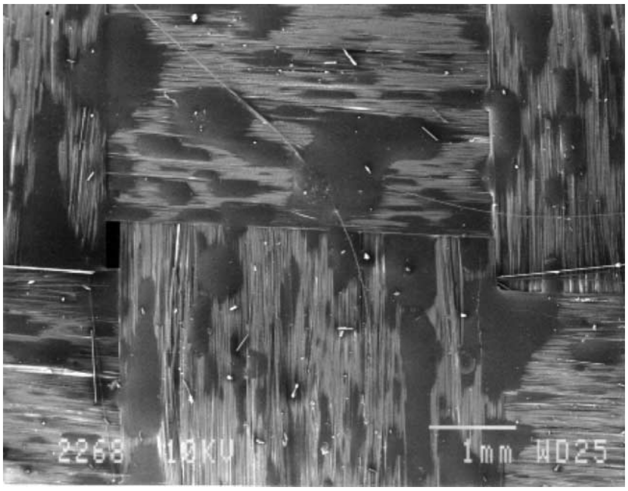

Fig. 3. Scanning electron micrographs showing the as received glass fabric with 2.6 wt% binder at low magnification.

Table 2
Mass fraction of the polyester binder on the preforming S2-glass fabric after burnout experiments. Mass fraction of the fiber sizing is also shown

Preforming S2-glass fabric	Average wt% of the binder and sizing (Std. Dev.)	Average wt% of the binder
1 (no binder, baseline)	1.0 (0.1)	0
2	3.6 (0.1)	2.6
3	5.8 (0.4)	4.8
4	7.7 (0.6)	6.7

the first ramp, the T_g is shifted to a higher temperature due to the crystallinity of the material; therefore, it overlaps with the melting peak. Since there is no time for re-crystallization in the second ramp, the second heating profile exhibits only the T_g of the binder. Also, a T_g of about 55°C and the storage and loss modulus values (at the room temperature) of 1.1 and 0.023 GPa, respectively, were measured with the DMA for the polyester binder.

Fig. 3 is a typical example showing the as-received glass fabric with 2.6 wt% binder material. It was observed that the fabric surface is covered with randomly distributed aggregates of the former binder particles. The particles seem to have infiltrated into the tows to some degree due to melting during the binder application stage done by Solectria. It was determined that about 50% of the fabric surface area was covered by the binder for this concentration level.

The amount of binder and the sizing on the glass fabric was determined from the burnout experiments. The burnout results are summarized in Table 2. Binder concentration ranged from a low of 2.6 to a maximum of 6.7 wt%. Fig. 4 shows the area of binder coverage on the glass fabric as a function of binder concentration. Full coverage of the surface is possible with 5–6% binder concentration.

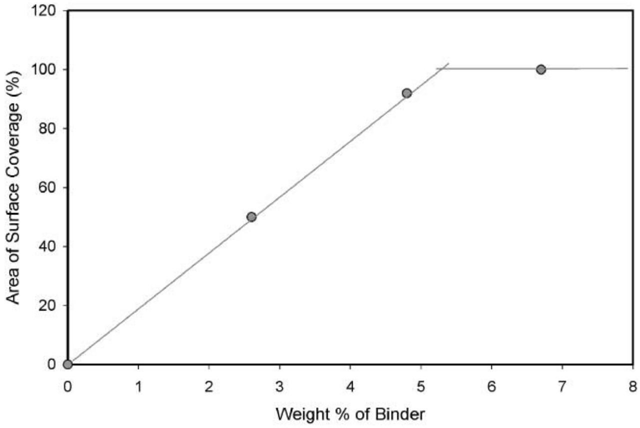

Fig. 4. Area of binder coverage on the glass fabric surface as a function of binder concentration.

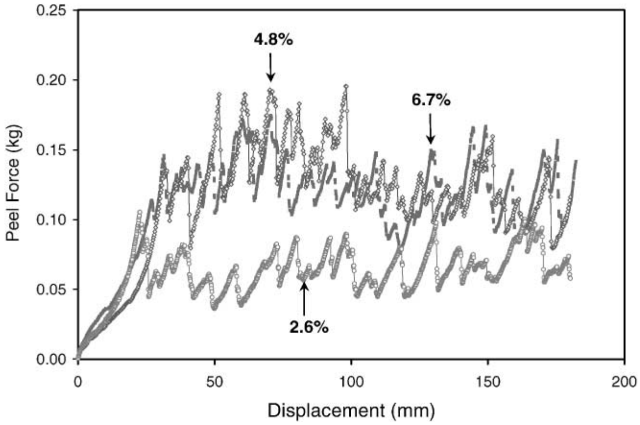

Fig. 5. Peel force vs. displacement response of peel specimens consolidated at 80°C with various binder concentrations.

3.2. Peel strength of glass preforms

Fig. 5 shows typical peel force vs. cross-head displacement plots for preforms consolidated at 80°C with various weight fraction of the binder. In this test, load is increased to a critical value where crack growth initiates. The specimens exhibit stick/slip type crack growth in the plateau region. Fig. 6 shows the average peel strength as a function of weight fraction of the binder for the same preforms. As can be seen from the plots, the preforms consolidated with 4.8 wt % of the binder exhibited higher peel strength compared to that of samples made with 2.6 wt %. The initial increase of peel strength with increased binder is due to the larger surface area of the fabric bonded together by the binder. At about 4.8%, the binder covers almost the entire surface, and further

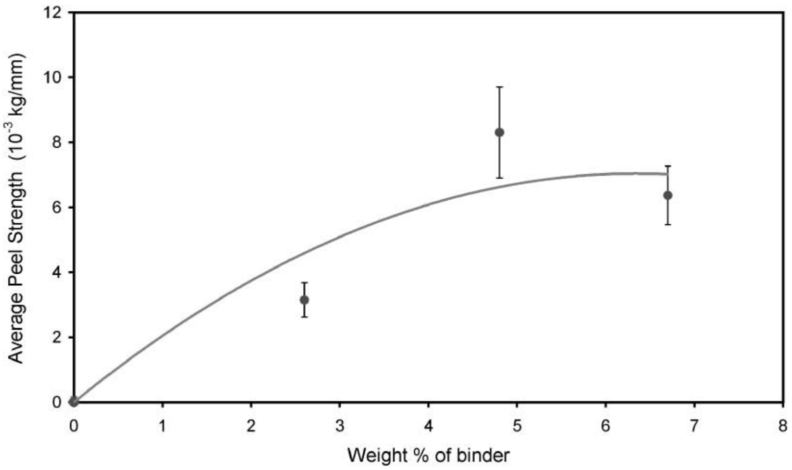

Fig. 6. Average peel strength of preform specimens consolidated at 80°C as a function of binder concentration.

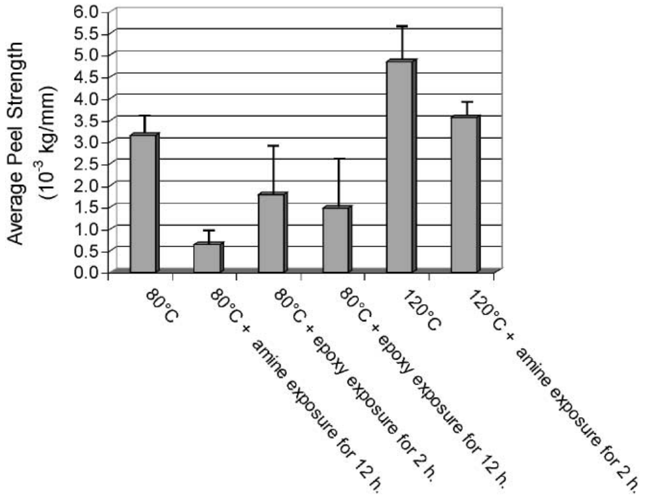

Fig. 7. Average peel strength of peel specimens consolidated at 80°C and 120°C (with 2.6 wt% binder) before and after epoxy or amine exposure.

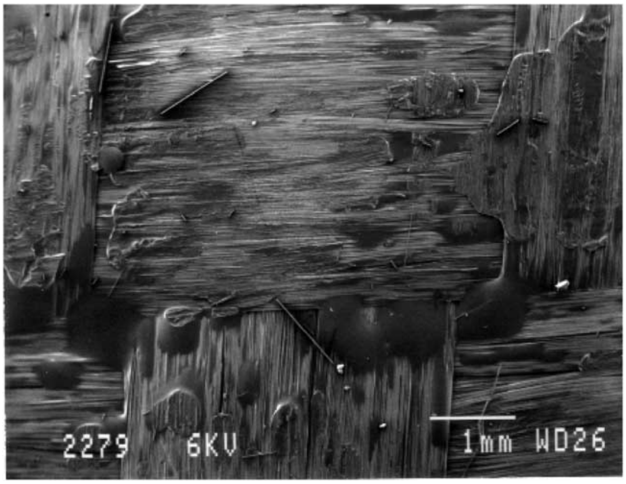

Fig. 8. Fracture surface scanning electron micrographs of the preform (consolidated at 80°C) after peel test at low magnification (binder side of the fabric).

increase in the amount of the binder does not increase the preform peel resistance.

The results presented in Fig. 7 for 2.6% binder concentration show that the preforms consolidated at 120°C exhibited higher peel force compared to that of samples made at 80°C. It was found that the average peel strength could be increased approximately 50% by consolidating the preforms at 120°C instead of 80°C. This may be attributed to the higher degree of mechanical interlocking between the layers of fiber tows due to higher tow impregnation caused by the lower viscosity of the binder at higher temperature. Knight et al. [4] reported a greater compaction of the glass preforms made from a thermoformable polyester binder at relatively higher temperatures. This was attributed to higher flow of the binder and thus better penetration and spreading of the binder into the gaps between the individual filaments due to the lower viscosity of the binder at relatively higher temperatures. Investigation of this mechanism will be continued in future work.

Fig. 7 also summarizes the average peel strength values for specimens consolidated at 80°C and 120°C before and after exposure to epoxy or amine for various time ranges. As can be seen from the figure, the peel strength of the specimens decreases due to the epoxy or amine exposure. As an example, the peel strength of the specimens (80°C) decreases by about 80% and 50% due to the 12 h of exposure to amine and epoxy, respectively. The reduction in strength was attributed to the dissolution of the binder and the detachment of the binder from the fiber. The results also show that the degree of dissolution is higher in the amine compared with that of the epoxy.

SEM was used to determine the locus of failure of peeled specimens. Figs. 8 and 9 show the fracture surface SEM micrograph of the preforms consolidated at 80°C

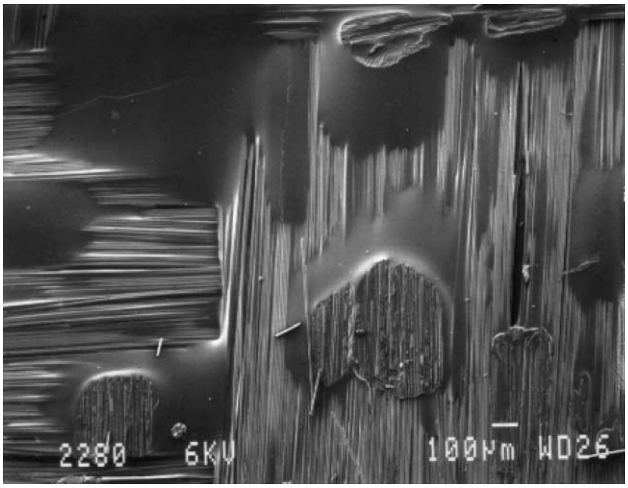

Fig. 9. Fracture surface scanning electron micrographs of the preform (consolidated at 80°C) after peel test at high magnification (binder side of the fabric).

with 2.6 wt% of the binder after peel test. In general, specimens exhibited adhesive-type failure between binder and glass fibers. Some cohesive type failure was also observed on the untreated side of the fabric.

3.3. Effects of the binder on the mechanical properties of the composite

Typical load vs. displacement traces obtained from DCB specimens with (2.6 wt%) and without binder are shown in Fig. 10. Delamination initiates and propagates at relatively lower displacement and load values in laminates with binder. Fig. 11 shows the delaminated edge of a typical DCB specimen and the crack position for composites with binder. Mode I interlaminar fracture toughness (G_{IC}) as a function of delamination crack extension (Δa) is shown in Fig. 12 for S2-glass/SC-15 composites with 2.6 wt% and without binder. In these resistance curves (R-curve), the experimental and least square polynomial fits are represented by circles and solid line, respectively. Composites both with and without binder exhibit rising R-curve behavior. The delamination initiates at G_{IC} values of 800 and 300 J/m^2 without and with binder, respectively. The average G_{IC} values increase and reach a constant propagation value. The average propagation G_{IC} values are 1600 and 617 J/m^2 for specimens without and with binder, respectively (see Table 3). These results reveal that the fracture toughness of the S2-glass/SC-15 system is reduced about 60% due to the presence of 2.6 wt% of the binder. The SBS test results are summarized in Table 3. The interlaminar shear strength of the S2-glass/SC-15 composite is reduced about 25% due to the presence of 2.6 wt % preforming binder. The DCB and SBS results are consistent with the limited compatibility/dissolution observed between binder and SC-15. Furthermore, results indicate that properties at the interlaminar region between layers can be tailored using appropriate binder/resin combinations.

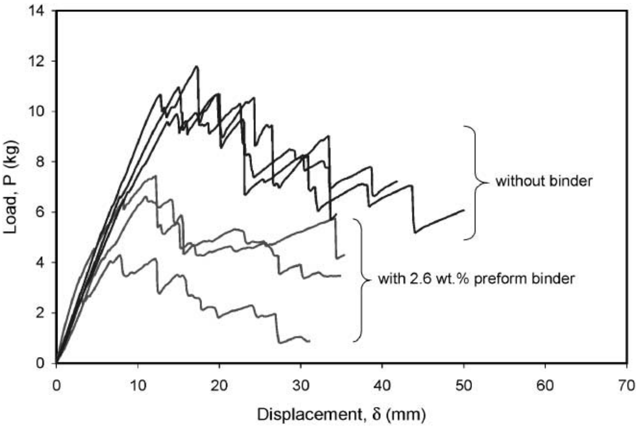

Fig. 10. Load vs. displacement from DCB test with 2.6% preform binder and without binder.

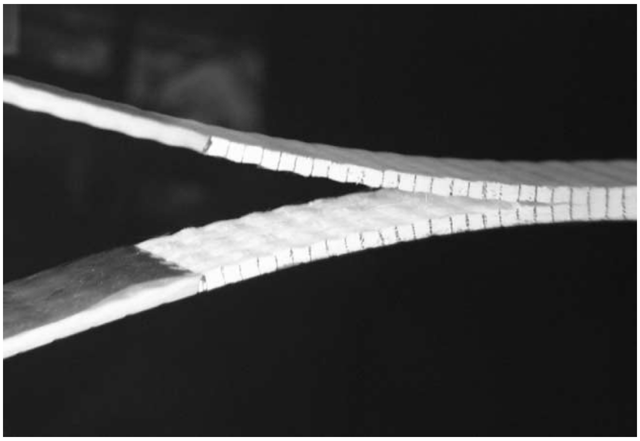

Fig. 11. S2-glass/SC-15 (with 2.6% binder) DCB test specimen under load showing delamination region and crack extension.

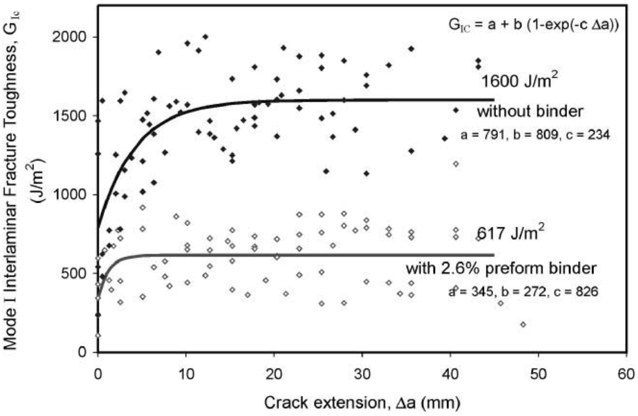

Fig. 12. Delamination resistance curve (R-curve) from DCB test for woven S2-glass/SC-15 epoxy–amine composites with 2.6 wt% preform binder and without binder.

Table 3
Mode I interlaminar fracture toughness and apparent interlaminar shear strength of S2-glass/SC-15 epoxy–amine composite systems with and without preform binder

Specimen	Mode I interlaminar fracture toughness, G_{IC} (J/m^2)	Interlaminar shear strength, τ_{max} (MPa)
Without binder	1600 (± 176)	41.13 (± 1.82)
With 2.6 wt% preform binder	617 (± 193)	30.95 (± 4.35)

3.4. Interactions between the polyester binder and the matrix resin

The flexural storage modulus, E' and the loss modulus, E'' values are plotted as a function of temperature in Fig. 13 for S2-glass/SC-15 composite system with and without binder. It was measured that the composite sample without binder exhibits a T_g of about 101°C. On the other hand, the composite with binder exhibits two transitions: matrix T_g of about 95°C and binder T_g of about 55°C. The value of the binder T_g is consistent with the DSC and DMA findings obtained for neat binder as mentioned earlier. Since the DMA profile in Fig. 13 shows the binder peak, it is expected that the binder will not dissolve completely in the matrix resin during the infusion and gelation stages. However, the presence of the 2.6 wt % of the binder lowers the T_g of the matrix approximately 6°C. This indicates the possibility of limited dissolution of the binder in the resin. Different degrees of dissolution might be possible by modifying the cure cycle.

Fig. 14 shows the storage modulus, E' and the loss modulus, E'' values as a function of temperature for model specimens. The neat SC-15 exhibits a T_g (α-transition) of about 89°C. The model specimen (physical mix, 3% binder mixed at room temperature for 10 min) that has limited dissolution exhibits two transitions: T_g of the matrix at 82°C and T_g of the binder at 55°C. In contrast to the physical mix, the model specimen that has extensive dissolution of the binder (3% binder mixed at 55°C for 4 h) exhibits only one transition: T_g of matrix at 64°C. This specimen does not show any peak associated with the T_g of the binder. This confirms that extensive dissolution of the binder in the matrix resin has occurred. The results indicate that the results from the model specimen (physical mix) correlate well with the results from S2-glass/SC-15 composite with binder. Based on this result, limited dissolution of the binder in the matrix resin is expected in composite panels processed under standard conditions. The results show that the T_g of the polymer matrix is affected significantly by the presence of the polyester binder. As an example, T_g of the matrix is reduced about 7°C due to limited dissolution of the binder in the reacting resin (room temperature cure for 12 h). On the other hand, elevated-temperature cure at 55°C for 12 h (3 wt%) further reduces T_g by 24°C due to the extensive dissolution of the binder. The flexural modulus values were found to be affected only slightly by the presence of the binder whether the dissolution was extensive or not.

4. Conclusions

The present study shows that the thermoplastic polyester binder has a significant effect on the properties

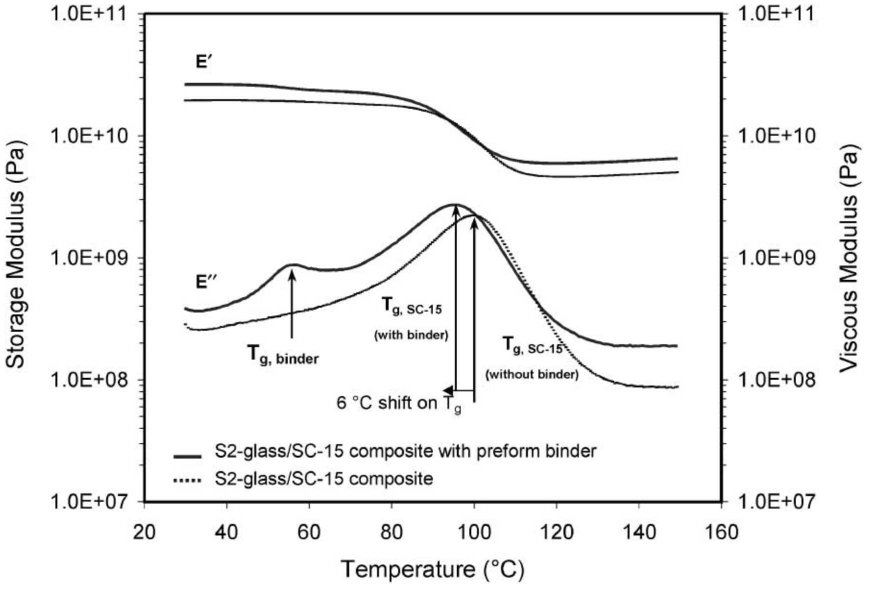

Fig. 13. DMA results for S2-glass/SC-15 composite w/wo. polyester binder.

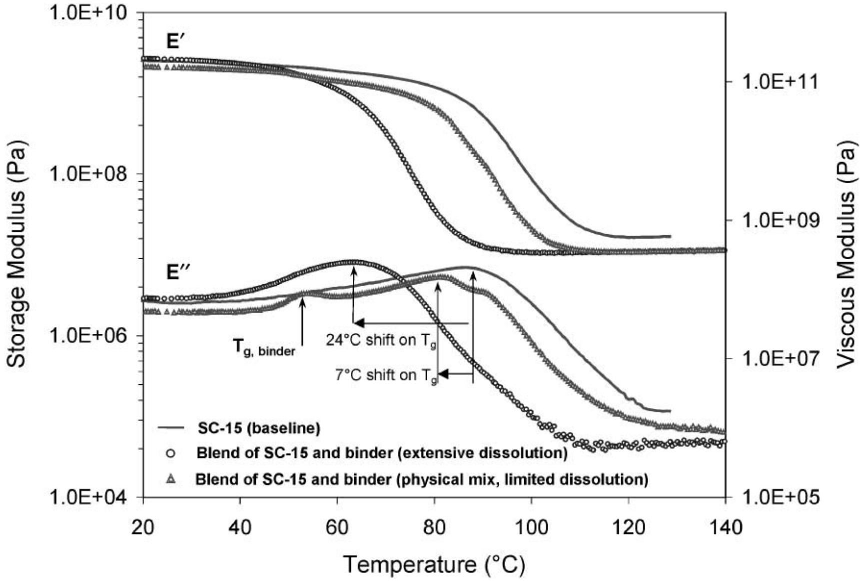

Fig. 14. DMA results for model specimens.

(T_g, interlaminar shear strength, mode I interlaminar fracture toughness) of an S2-glass/SC-15 composite system. It was determined that approximately 5–6 wt% of the binder provides full coverage of the glass fabric surface. The highest peel strength was measured from the preform that had almost full binder coverage on the fabric. The increase in the binder concentration did not significantly change the peel strength values. It was found that preform consolidation temperature can significantly change the peel strength of preforms, i.e., peel strength increased about 50% with an increase in temperature from 80°C to 120°C. The peel strength of the preforms decreases due to the exposure of the peel specimens to the SC-15 resin components (epoxy and amine). This was attributed to some extent of dissolution of the binder in the resin. The interlaminar shear strength and Mode I interlaminar fracture toughness of the S2-glass/SC-15 composite system were significantly affected by the thermoplastic binder, i.e., 25% and 60% reduction, respectively, by the presence of 2.6 wt% of the binder. Moreover, the T_g of the polymer matrix was reduced about 6°C by the same amount of binder. It was found that the polyester binder had limited dissolution in the composite panels processed under standard cure cycle. The results indicate that thermoplastic binder offers the potential to tailor the properties of the interlaminar region between fiber layers (strength, fracture toughness, modulus, T_g).

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