

Mechanical properties of *in situ* composites based on polycarbonate and a liquid crystalline polymer

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The mechanical properties of *in situ* composites based on blends of polycarbonate and a liquid crystalline copoly(ester amide) (Vectra B950) have been studied as functions of the liquid crystalline polymer (LCP) concentration and the draw ratio, a processing parameter. It is shown that both the elastic modulus and the tensile strength of the *in situ* composites increase steadily with the LCP concentration and the draw ratio. However, the ultimate tensile strain decreases with these two parameters. A model is proposed for the longitudinal elastic modulus of the *in situ* composites, which is based on the Halpin-Tsai equation and Northolt's model for the LCP phase. The experimental elastic moduli of the *in situ* composites are found to conform fairly well with the theoretical values derived from the model.

(Keywords: *in situ* composites; liquid crystalline polymer; polycarbonate)

INTRODUCTION

Blending two or more polymers is an effective way to combine the useful properties of each component into one system. Recently, polymer blends of liquid crystalline polymers (LCPs) with isotropic polymers have attracted considerable attention because of the potential for producing high performance polymeric materials at reduced costs. Addition of a LCP to an isotropic polymer not only enhances its mechanical properties, due to the easy molecular orientation of the LCP, but also improves its processing properties, as a result of the low melt viscosity of the LCP. On the other hand, combining a LCP with a conventional isotropic polymer is expected to improve the weak lateral strength of the LCP and also to enhance the compressive resistance of the oriented polymer. Moreover, addition of conventional, isotropic thermoplastics also lowers the cost of the final products.

Another important advantage of polymer blends containing LCPs is that addition of the latter makes it possible to control blend morphology through the processing. Under the appropriate processing conditions, the dispersed LCP phase can be deformed into a fibrillar phase, which acts as the reinforcing phase. The resultant product is hence called an *in situ* composite¹. The deformation of the LCP phase into the fibrillar form not only increases the aspect ratio of the reinforcing phase, but also improves the molecular orientation of the LCP chains. As a consequence, significant reinforcement can be achieved with the addition of only a small amount of LCP¹⁻¹³. The *in situ* composite concept has emerged as a new route for the development of advanced polymeric materials.

The key to successful fabrication of *in situ* composites is the formation of a fibrillar LCP morphology. To generate this type of morphology, three major methods have been used, namely fibre spinning, extrusion, and injection molding. It has been found that fibre spinning is particularly advantageous because the elongational force field in the spinning process is more effective for producing the LCP fibrillar morphology than the shear force field that is present in other processing methods.

As in other multiphase polymer systems, the mechanical properties of *in situ* composites depend on the composition, the properties of each component, the macro- and microstructures of the composites, and on the interfacial adhesion and compatibilization. In particular, the shape and dispersion of the LCP phase and the orientation of the LCP chains play primary roles in determining the mechanical properties. Furthermore, the overall properties of the composites are dominated by the properties of the reinforcing LCP phase, since the mechanical properties of the LCP fibrils are typical of those of high performance fibres. It has been well established that a fibrillar LCP morphology translates into higher moduli and strengths. The reinforcement effect arises from the larger aspect ratio of the fibrillar LCP phase and the better molecular orientation of the LCP chains^{5,7,9}.

Previous studies of the mechanical properties of *in situ* composites focused mainly on the effects of LCP concentration, certain processing conditions, such as drawing, and compatibilization. Addition of a LCP to an isotropic polymer generally leads to an increase in modulus and strength, and a decrease in ultimate tensile strain. Improvements in modulus and strength, which range from tens of percent to an order of magnitude,

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have been found in blends of LCPs and isotropic thermoplastics, such as polycarbonate^{1,4-9,11,14-22}, polystyrene^{3,13,17,23}, polyamides (nylon)^{1,2,14,24,25}, polyethylene terephthalate^{17,26-30}, poly(butylene terephthalate)^{1,14}, polypropylene^{12,27,31}, poly(ether imide)^{1,8,10,32}, poly(ether ether ketone)^{1,33}, poly(ether sulfone)¹, polysulfone³⁴, poly(phenyl sulfide)³⁵, poly(arylate)^{1,36}, and the blends of some of these polymers^{13,22,37,38}. Drawing leads to additional improvements in mechanical properties, e.g. both modulus and strength have been found to increase with drawing. The improvements in the mechanical properties are generally related to a fibrillar LCP morphology which is produced by the elongational stress field during the drawing process. Recently, Bretas and Baird³⁸ demonstrated that compatibilization of the immiscible *in situ* composites of polypropylene and a liquid crystalline copoly(ester amide) (Vectra) also results in an impressive improvement in mechanical properties. However, there are instances in which no reinforcement was achieved with addition of a LCP to either polypropylene or poly(butylene terephthalate)^{31,39}.

We have been interested in developing these novel self-reinforced *in situ* composites, using a liquid crystalline copoly(ester amide) as the discrete LCP phase and polycarbonate or polystyrene as the continuous phase³⁻⁷. We used the fibre spinning method to generate the self-reinforced composite structure. Our previous results showed that the addition of only 30 wt% LCP to polycarbonate leads to a 7.3 times increase in elastic modulus and a 1.4 times increase in tensile strength, relative to the corresponding properties of pure polycarbonate⁷.

The purpose of this paper is to present a systematic study of the mechanical properties of the *in situ* composites based on polycarbonate and a thermotropic copoly(ester amide) (Vectra B950). Specifically, the effects of the LCP concentration and the blend draw ratio on tensile modulus, tensile strength and ultimate tensile strain are investigated. In addition, a model is proposed for the elastic modulus of the *in situ* composites, which is based on the Halpin-Tsai equation and Northolt's model for the LCP phase. Theoretical predictions of the model are then compared with the experimental results obtained for the longitudinal elastic modulus of the *in situ* composites.

EXPERIMENTAL

Materials

The LCP used in this research was Vectra B950 (Hoechst Celanese). Vectra B950 is a random liquid crystalline copoly(ester amide) of three components: 60 mol% 2,6-hydroxynaphthoic acid, 20 mol% terephthalic acid and 20 mol% *p*-aminophenol. The isotropic thermoplastic matrix that was used was a bisphenol-A polycarbonate (PC) (Lexan 151, GE Plastics).

Processing

In situ composite fibres were fabricated by melt extrusion of blends of the LCP and polycarbonate with a ZSK-30 twin-screw extruder (Werner and Pfleiderer). The extrusion was followed by drawing the extruded melt directly with a take-up machine, with the draw ratio being varied by changing the speed of the take-up reels. The detailed procedures are described elsewhere⁷. Three series of samples were fabricated: one varied in LCP content but had a constant draw ratio of ~ 11.5 , while

the other two varied in draw ratio but had constant LCP concentrations of 20 and 26.9 vol%, respectively. The draw ratio was determined experimentally from the ratio of the cross-sectional area of the die to that of the extruded composite fibre.

Mechanical testing

The longitudinal elastic moduli of the extruded fibres were measured on an Instron Mechanical Tester (Model 4502), while the elongation was determined with the aid of an extensometer attached directly to the specimen. To reduce stress concentration at the grips, the specimens were held in place with rubber pads. Since the strain was measured directly over a predetermined gauge section, the displacement in the grips did not affect the accuracy of the results. All tests were run at a constant crosshead speed of 0.5 mm min⁻¹. All data were acquired and processed by an IBM computer, with each data point reported here being the average of five measurements.

RESULTS AND DISCUSSION

Stress-strain behaviour

Typical stress-strain curves of the *in situ* composite are shown in Figure 1. We note that, in contrast to pure polycarbonate, most drawn composite fibres do not exhibit any yielding during tensile testing. Only those with low LCP contents and low draw ratios show limited yielding and stress whitening.

Effect of melt drawing

One of the advantages of *in situ* composites is the possibility of improving the mechanical properties by controlling the blend morphology through processing. Previous studies demonstrated that melt drawing, which usually generates fibrillar LCP morphologies, led to enhancements in modulus and strength in many of the *in situ* composite systems^{7-11,18,24,25,36,39,40}. However, other researchers^{13,20,32,37} have reported that drawing did not appreciably change either the modulus or the strength. In some instances, drawing was even reported to have a detrimental effect on the mechanical properties^{13,37}.

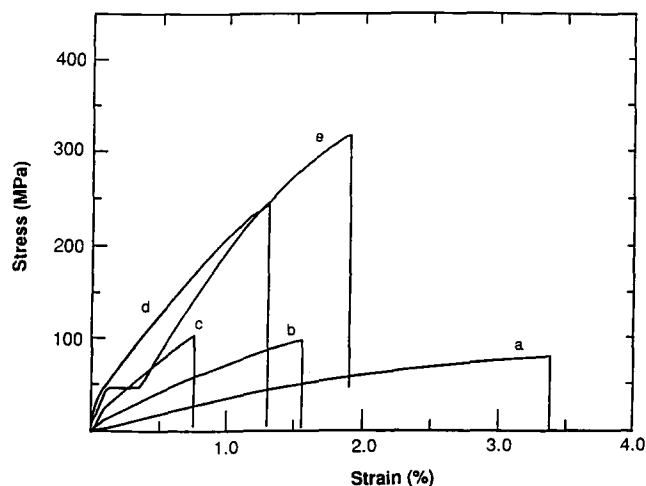


Figure 1 Typical stress-strain curves of PC/LCP *in situ* composites with different LCP contents: (a) 10 vol%; (b) 20 vol%; (c) 26.9 vol%; (d) 57.3 vol%; (e) 77.4 vol%. The draw ratio for all samples was 11.5

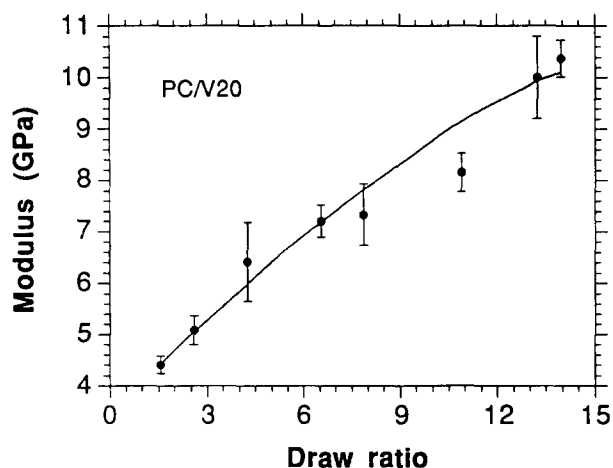


Figure 2 Effect of draw ratio on elastic modulus of PC/LCP *in situ* composites with 20 vol% LCP

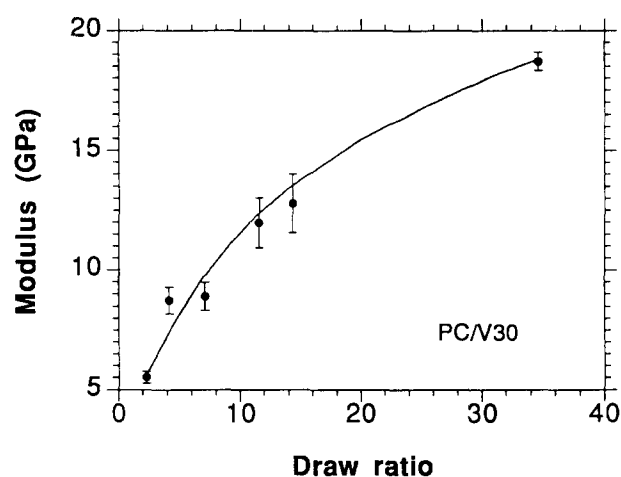


Figure 3 Effect of draw ratio on elastic modulus of PC/LCP *in situ* composites with 26.9 vol% LCP

We have found that drawing the extruded strands results in marked improvements in the mechanical properties. Figures 2–5 show the effect of drawing on the elastic modulus and tensile strength of the *in situ* composites containing 20 and 26.9 vol% LCP. It is seen that both the elastic modulus and the tensile strength increase monotonically with the draw ratio. Increasing the draw ratio from 1.6 to 14 leads to a 136% improvement in the modulus and a 67% improvement in the tensile strength for the composite with 20 vol% LCP. With higher draw ratios, the improvement for the composites with 26.9 vol% LCP is even more significant. For example, the elastic modulus of the composites with a draw ratio of 35 is more than three times that of an undrawn specimen. This finding is in agreement with results previously reported for most of the LCP/isotropic polymer systems. The apparent drop in the tensile strength of the composite with 26.9 vol% LCP at the highest draw ratio is due to a change in the gripping method that was used during the mechanical testing (see Figure 5)⁷.

Our previous studies have demonstrated that the application of a stress field through drawing brings about two major structural changes in the blends⁷:

(1) With increasing draw ratio, the dispersed LCP phase is deformed from a spherical form, first into an

ellipsoidal, and then into a fibrillar form. In other words, the aspect ratio of the dispersed LCP phase is increased during drawing.

(2) The (semi)rigid LCP chains are oriented along the fibre axial direction during drawing. X-ray scattering measurements of the blends indeed showed a steady increase in the order parameter of the LCP chains with draw ratio⁵. Both of these structural changes could lead to improvements in the elastic modulus and tensile strength of the composites.

During the drawing process, some molecular orientation of the polycarbonate phase may result. However, mechanical measurements of a number of isotropic polymers showed that melt drawing does not appreciably alter the modulus and the strength of the isotropic polymers^{3,18,20,32,36,37,40}. This may be due to the flexible nature of these isotropic polymer chains and/or the rapid relaxation of their molecular orientation. It is much more difficult to extend and orient flexible chains than rigid chains, because the former involves a huge penalty in the entropy. Therefore, the structural change of the polycarbonate during drawing will not have an appreciable influence on the overall mechanical properties of the *in situ* composites. This is also true because of the fact that the mechanical properties of the blends are dominated by the reinforcing LCP phase.

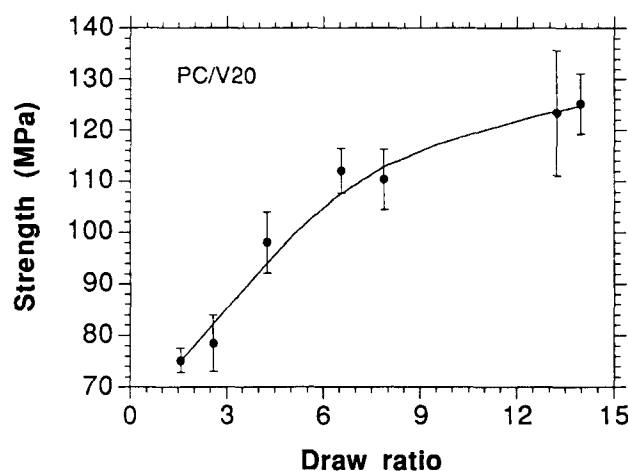


Figure 4 Effect of draw ratio on tensile strength of PC/LCP *in situ* composites with 20 vol% LCP

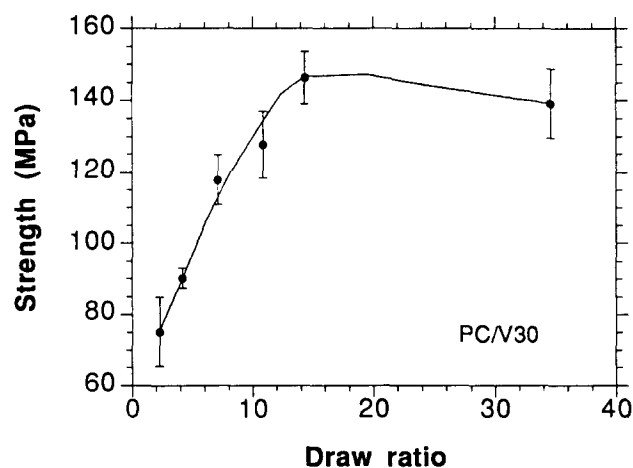


Figure 5 Effect of draw ratio on tensile strength of PC/LCP *in situ* composites with 26.9 vol% LCP

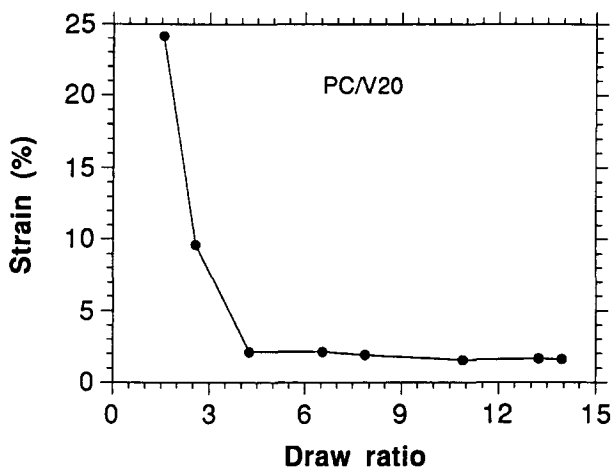


Figure 6 Effect of draw ratio on ultimate tensile strain of PC/LCP *in situ* composites with 20 vol% LCP

Figure 6 gives the variation of ultimate tensile strain with the draw ratio for the *in situ* composites containing 20 vol% LCP. It is clearly shown that the ultimate tensile strain decreases dramatically with drawing, with this behaviour beginning at the lower draw ratios. It then levels off at a draw ratio of ~ 4 . Similar behaviour has previously been observed in other *in situ* composite systems, but no explanation, however, has been offered in these cases^{10,13,25}.

Effect of LCP concentration

According to the published literature, the effect of LCP concentration on the mechanical properties of *in situ* composites is more complicated than the effect of melt drawing. Most of the published papers have shown reinforcement with the addition of a LCP to an isotropic polymer, although the magnitude of the reinforcement and the dependence on LCP content vary^{1,2,4,6,8,11-16,18,20,21,23-28,30,35,36,38}. We and many other groups^{1,4,7-14} have found steady improvements in modulus and strength with increasing LCP concentration. However, Crevecoeur and Groeninckx¹³ and others^{19,23} found that the reinforcement effect resulted only after the LCP concentration exceeded certain values in injection moulded samples. Li *et al.*²⁶ observed that the dependence of elastic modulus and tensile strength on the LCP content followed two distinct paths as the LCP changed from the discrete phase to the continuous phase in drawn composite fibres. Isayev and Subramanian³³ noted maxima and minima in modulus and strength as the LCP concentration was increased. Finally, Ajji *et al.*³⁹ reported that the mechanical properties did not change with the addition of up to 30 wt% LCP to poly(butylene terephthalate).

We found steady increases in elastic modulus and tensile strength with the LCP content in the polycarbonate/Vectra *in situ* composites. The results are depicted in Figures 7 and 8 for samples with an average draw ratio of 11.5. It is seen that reinforcement can be achieved throughout the whole concentration range studied, from 10 up to ~ 80 vol% LCP. This is in good agreement with most of the previous studies on other LCP/isotropic polymer systems. The reason appears to be our ability to consistently generate fibrillar morphologies as long as the LCP remains as the discrete phase. Aside from the LCP volume effect, X-ray measurements showed that the

molecular orientation of the LCP chains increases slightly with increasing LCP concentration⁴¹. It is interesting to note that both the elastic modulus and the tensile strength follow a similar trend with increasing LCP content without any abrupt deviation, even though phase inversion occurs in the vicinity of 60 vol% LCP in the LCP/polycarbonate system.

Figure 9 gives the variation of ultimate tensile strain versus LCP concentration at an average draw ratio of 11.5. It is noted that the ultimate tensile strain decreases with the LCP content at lower LCP concentrations, but increases slightly at higher LCP concentrations, with the minimum strain being located at a LCP concentration of ~ 30 vol%. The reason for the existence of the minimum ultimate tensile strain has not yet been completely understood at this point, although it appears to be related to the change in morphology of the blends. A continuous polycarbonate phase was observed for blends with LCP concentrations below 30 vol%, while a continuous LCP phase was seen for blends with LCP concentrations above 60 vol%.

A model for the elastic modulus of *in situ* composites

The extensive experimental work on the mechanical properties of *in situ* composites has prompted researchers to model these properties. While modelling tensile

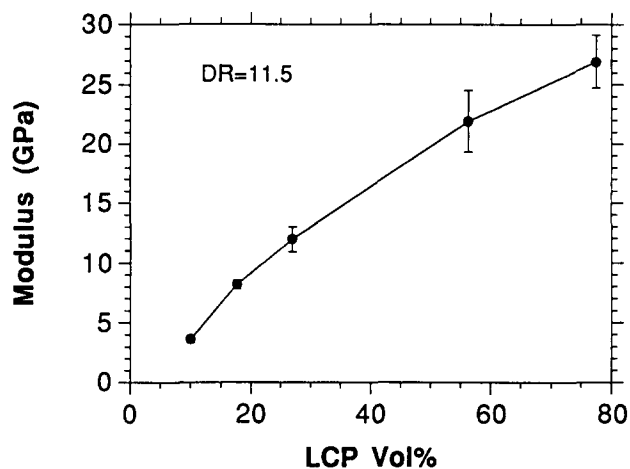


Figure 7 Effect of LCP concentration on elastic modulus of PC/LCP *in situ* composites with a draw ratio of 11.5

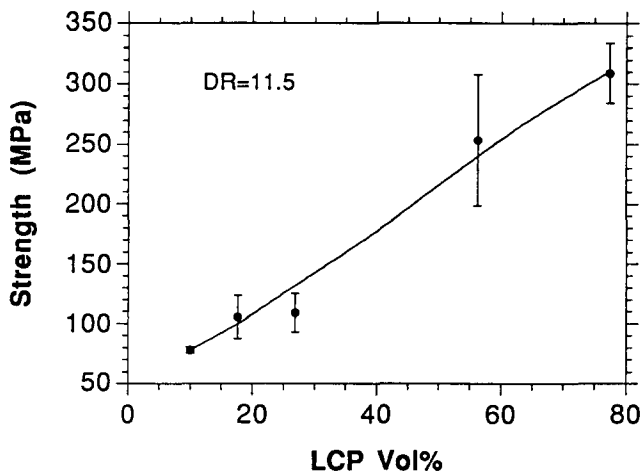


Figure 8 Effect of LCP concentration on tensile strength of PC/LCP *in situ* composites with a draw ratio of 11.5

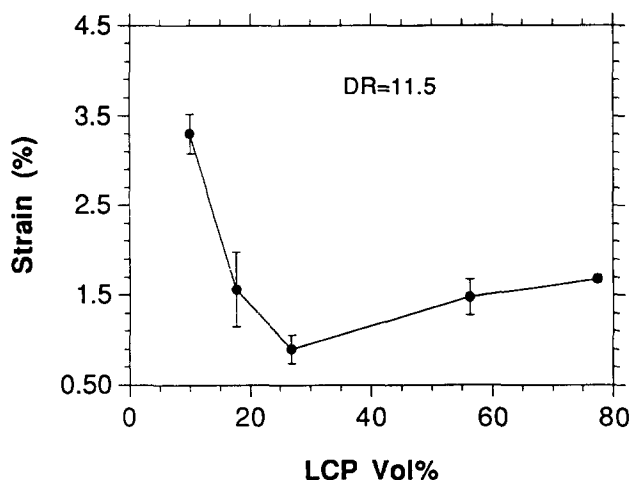


Figure 9 Effect of LCP concentration on ultimate tensile strain of PC/LCP *in situ* composites with a draw ratio of 11.5

strength is difficult, attempts have been made to apply the simple rule of mixtures of continuous fibre-reinforced composites to modelling the modulus of the *in situ* composites, even though in most cases the morphologies of the *in situ* composites were far from that of continuous fibre-reinforced composites^{3,4,8,13,15,16,22,26,32,33,35,37}. Recently, some researchers^{13,26,29} have indicated that the aspect ratio of the LCP reinforcing phase could seriously affect the elastic modulus of the *in situ* composites. However, even in these cases, constant elastic moduli were assumed for the reinforcing LCP microfibrils. This assumption is not reasonable because the molecular orientation of the LCP reinforcing phase increases with drawing, and the elastic modulus of the LCP microfibrils is expected to change accordingly. In fact, a careful examination of the literature revealed that the values used for the elastic modulus of the reinforcing LCP fibrils (Vectra A) ranged from 2.2 GPa²² to 75 GPa¹³ with many intermediate values, e.g. 10 GPa³⁵, 24.6 GPa⁸, 26.4 GPa¹⁵, 38 GPa¹⁶, and 70.7 GPa²⁶. This wide range of LCP modulus values reflects the fact that the elastic modulus of an oriented LCP depends heavily on the processing conditions. Indeed, many researchers⁴²⁻⁴⁵ have found that the elastic modulus of extruded Vectra A and B rods increased dramatically with increasing draw ratio.

We have proposed a model to predict the elastic modulus of *in situ* composite strands with fibrillar LCP morphologies^{6,46}. Our approach took into account both the change in aspect ratio of the reinforcing LCP phase and the variation of the orientation of the LCP chains during processing. We related the elastic moduli of the *in situ* composites to the LCP concentration and draw ratio through the well known Halpin-Tsai equation and a composite model of the reinforcing LCP phase. We envisaged the LCP phase itself as a composite of a perfectly oriented chain aggregate and a randomly oriented chain aggregate which are connected in series (see Figure 10). With increasing draw ratio, the fraction of oriented chains increases at the expense of random chains. Our results showed that the theoretical predictions agreed fairly well with the experimental results of the longitudinal elastic modulus of polycarbonate/Vectra *in situ* composite strands.

In this paper, we adopt the aggregate model first proposed by Crawford and Kolsky⁴⁷, and further

developed by Ward⁴⁸ and Northolt and van der Hout⁴⁹ to account for the change of the elastic modulus of the reinforcing LCP fibrils with drawing. Both of the models proposed by Ward⁴⁸, and Northolt and van der Hout⁴⁹ considered an oriented fibre as an assembly of polymer chain aggregates which are connected in series. They further assumed a uniform stress distribution throughout the oriented fibre. An external stress causes stretching of the aggregates as well as rotation of the aggregates due to misalignment (Figure 11). Here, we use the modified aggregate model proposed by Northolt and van der Hout⁴⁹ for the oriented LCP phase.

Northolt and van der Hout⁴⁹ considered an oriented LCP fibre as being built up of a parallel array of identical fibrils. Each fibril consists of a series of end-to-end connected aggregates of parallel aligned polymer chains. The orientation of the polymer chains (and the aggregates) is characterized by the angle between the chain axis (or the aggregate symmetry axis) and the fibre axis (θ). An external tensile stress, σ , applied along the fibre axis causes both an elongation of the aggregates and a rotation of the aggregates towards the fibre axis, as a result of shear deformation. Therefore, the total strain of the fibre is the sum of the strains due to the elongation and the shear deformation. If the polymer has a chain modulus E_c , and a shear modulus G along the chain direction, the tensile strain of the fibre (ϵ_f) can be derived as follows:

$$\epsilon_f = \frac{\sigma}{E_c} + \frac{\langle \sin^2 \theta \rangle}{2} [1 - \exp(-\sigma/G)] \quad (1)$$

where $\langle \sin^2 \theta \rangle$ represents the average orientation of the polymer chains with respect to the fibre axis. At small

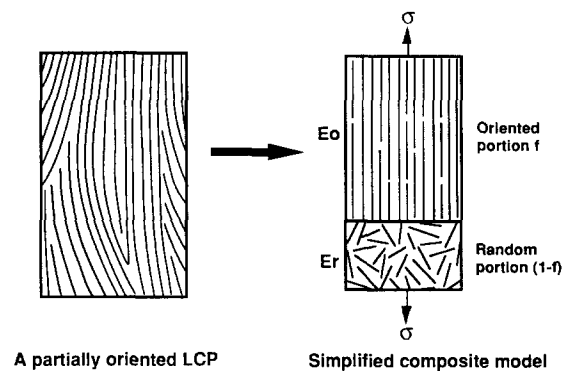


Figure 10 Composite model of an oriented LCP

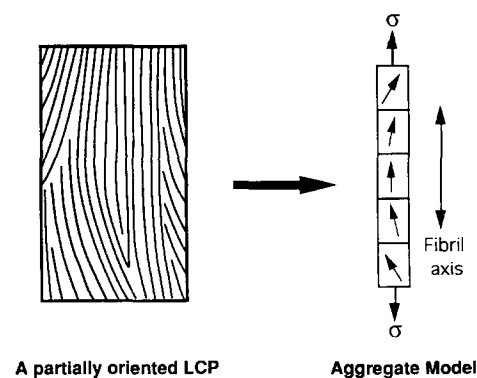


Figure 11 Aggregate model of an oriented LCP (after Northolt and Sikkema⁵⁰)

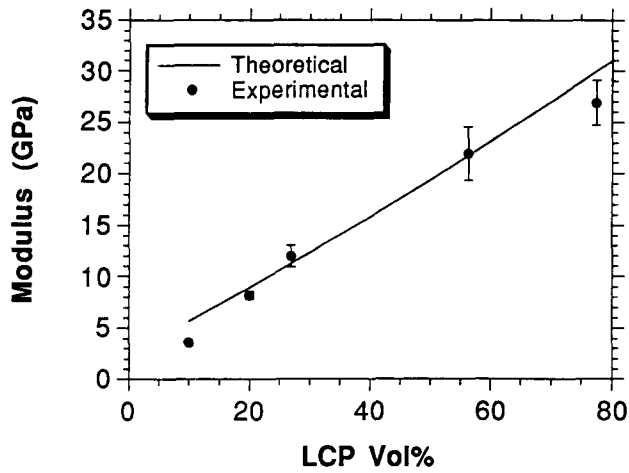


Figure 12 Comparison of theoretical predictions and experimental results for the variation of elastic modulus with LCP content of PC/LCP *in situ* composites with a draw ratio of 11.5

deformations, a simple relationship for the elastic modulus of the fibre, E_f , can be obtained from equation (1):

$$\frac{1}{E_f} = \frac{1}{E_c} + \frac{\langle \sin^2 \theta \rangle}{2G} \quad (2)$$

The equations derived by Ward and others differ from the above equation in the pre-factor of the second term^{51,52}, which results from the difference in the proposed shape of the aggregates. Ward^{51,53} adopted cube-shaped aggregates, while Northolt and van der Hout⁴⁹ used oblong-shaped aggregates. The aggregate models have been successfully applied to many oriented polymer systems^{49,54}, including liquid crystalline polymers^{49-51,54-58}. However, in most cases, $\langle \sin^2 \theta \rangle$ was used as a fitting parameter, but no correlation was obtained between $\langle \sin^2 \theta \rangle$ and the processing parameters. In this paper, we have attempted to relate the $\langle \sin^2 \theta \rangle$ parameter to the processing parameters, specifically the draw ratio λ , by using the so called floating rods model^{5,46}.

According to the floating rods model, the order parameter (f) of LCP chains with a draw ratio of λ can be approximated as follows⁴⁶:

$$f = \frac{1}{2}(3\langle \cos^2 \theta \rangle - 1) = \frac{2\lambda^3 + 1}{2(\lambda^3 - 1)} - \frac{3}{2} \left[\frac{\lambda^3}{(\lambda^3 - 1)^{3/2}} \right] \sin^{-1} \left(\frac{\lambda^3 - 1}{\lambda^3} \right)^{1/2} \quad (3)$$

We therefore have:

$$\langle \sin^2 \theta \rangle = -\frac{1}{\lambda^3 - 1} + \frac{\lambda^3}{(\lambda^3 - 1)^{3/2}} \sin^{-1} \left(\frac{\lambda^3 - 1}{\lambda^3} \right)^{1/2} \quad (4)$$

By inserting equation (4) into equation (2), we can relate the elastic modulus of the reinforcing LCP fibrils to the draw ratio of the composite by the following:

$$\frac{1}{E_f} = \frac{1}{E_c} + \frac{1}{2G} \left[-\frac{1}{\lambda^3 - 1} + \frac{\lambda^3}{(\lambda^3 - 1)^{3/2}} \sin^{-1} \left(\frac{\lambda^3 - 1}{\lambda^3} \right)^{1/2} \right] \quad (5)$$

In a previous paper⁴⁶, we showed that the longitudinal elastic modulus (E_1) of *in situ* composite strands can be

written as follows:

$$E_1 = E_m \left(\frac{1 + 2\lambda^{3/2}\eta V_f}{1 - \eta V_f} \right) \quad \text{with } \eta = \frac{E_f/E_m - 1}{E_f/E_m + 2\lambda^{3/2}} \quad (6)$$

where E_f and E_m are the elastic moduli of the reinforcing fibrous LCP phase and the matrix, respectively. λ is the draw ratio of the composite fibre and V_f is the volume fraction of the fibrous LCP phase.

As in our previous paper, we again use $E_m = 2.5$ GPa for the modulus of the polycarbonate matrix and $E_c = 110$ GPa for the chain modulus of the Vectra liquid crystalline polymer⁴⁶. The shear modulus (G) of the Vectra liquid crystalline polymer was measured by Zhang *et al.*⁵⁷ to be 1.2 GPa at room temperature. This value falls well within the reported shear moduli values for Vectra liquid crystalline polymers, such as 0.9 GPa⁵⁶, 2.0 GPa⁵¹, 1.7 GPa⁵⁸, and 1.47 GPa⁵⁹. With these values, the longitudinal elastic modulus of the *in situ* composites can be calculated as a function of the LCP content and the draw ratio by using equations (5) and (6).

Comparisons of the theoretical predictions with the experimental results are shown in Figures 12–14. Again, we see a fairly good correspondence between the theoretical values and the experimental ones. This good

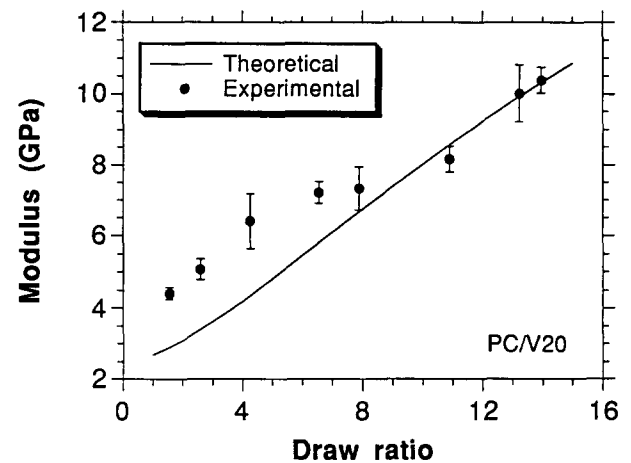


Figure 13 Comparison of theoretical predictions and experimental results for the variation of elastic modulus with draw ratio of PC/LCP *in situ* composites with 20 vol% LCP

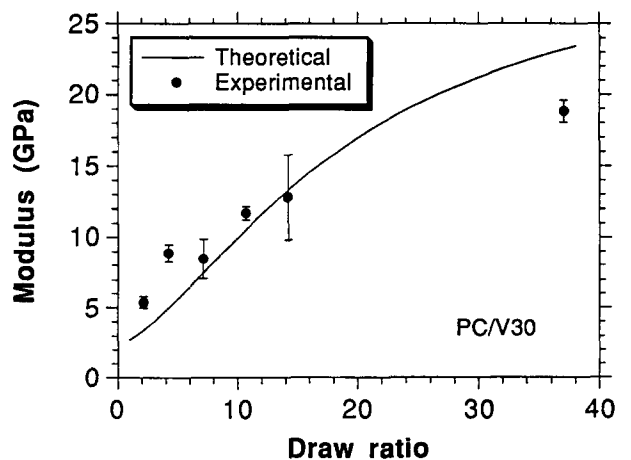


Figure 14 Comparison of theoretical predictions and experimental results for the variation of elastic modulus with draw ratio of PC/LCP *in situ* composites with 26.9 vol% LCP

agreement between theory and experiment once again corroborates our previous conclusion that the elastic modulus of *in situ* composites is dominated by the aspect ratio of the reinforcing LCP phase and the molecular orientation of the LCP chains^{5,7}.

We can see that the performance of the present model is almost as good as the previous one^{4,6}. This is not surprising in view of the many similarities between the two. First, both models are based on the Halpin–Tsai equation, and use the affine deformation scheme to derive the aspect ratio for the LCP phase. Furthermore, both models conceive the oriented LCP phase as aggregates of LCP chains, and propose a series connection of the chain aggregates. Thirdly, in both models, the orientation of the LCP chains is approximated by the affine deformation of the rigid-rod molecules. In fact, equation (6) in this paper is exactly the same as equation (5) in ref. 46. This present model differs from the previous one in that Northolt's aggregate model is used for the reinforcing LCP phase instead of the composite model. According to Northolt's model, an external tensile stress causes both the stretching, and the rotation of the chain aggregates due to misorientation, while in the composite model, the external stress causes only the stretching of both the perfectly oriented chain aggregate and the randomly oriented chain aggregate. The LCP chain misorientation is taken into account by changing the volume ratio of the two aggregates.

CONCLUSIONS

The mechanical properties of *in situ* composites based on blends of polycarbonate and a liquid crystalline copoly(ester amide) (Vectra B950) have been studied as functions of LCP concentration and draw ratio. It is found that both the elastic modulus and the tensile strength of the *in situ* composites increase steadily with the LCP concentration and the draw ratio, while the ultimate tensile strain, however, decreases with these two parameters. A model is proposed for the elastic modulus of the *in situ* composites, in which the elastic modulus of the LCP phase is related to the draw ratio via Northolt's model for the LCP phase. Equations for the longitudinal elastic modulus of the *in situ* composites are then derived based on the Halpin–Tsai equation. The experimental elastic moduli of the *in situ* composites are found to agree fairly well with the theoretical values.

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