polymer

Structure and property studies of poly(ethylene terephthalate)/ poly(ethylene-2,6-naphthalate) melt-blended fibres

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Abstract

The effects of blend composition and take-up velocity on structural variations and physical properties of poly(ethylene terephthalate) [PET]/poly(ethylene-2,6-naphthalate) [PEN] blend fibres were studied using WAXS, n.m.r., d.s.c., density, tensile and thermal shrinkage tests. Over the ranges studied, the structure and properties of the blend fibres were significantly influenced by the composition and melt spinning velocity. With the increase in content of any second component (PEN or PET), the crystallization process in the as-spun fibres becomes difficult, because of the formation of random copolymers via a transesterification reaction occurring between PET and PEN molecules. This reaction also produces a miscible amorphous phase with the result that all blend samples exhibit a single glass transition temperature intermediate to those of the two homopolymers. The WAXS results of annealed samples, however, indicate separate crystals for the two components rather than a co-crystallization. Room temperature mechanical properties, for fibres spun at constant take-up velocity, improved gradually with increasing PEN content. The thermal shrinkage, however, is relatively high for blend fibres and reaches a maximum for the 50/50 PET/PEN blend, indicating that the copolymer chains are oriented fairly well, but exist in the amorphous state. © 1998 Elsevier Science Ltd. All rights reserved.

Keywords: Poly(ethylene terephthalate); Poly(ethylene-2,6-naphthalate); Blend

1. Introduction

Blends containing polyester have been studied by many workers for both industrial application and academic interest. Most of the work carried out in this field was devoted to the miscibility, phase structure and physical properties, for certain polymer pairs chosen from poly(ethylene terephthalate) [PET], poly(butylene terephthalate) [PBT], polycarbonate (PC), polyarylate (PAr), Vectra, poly(ethylene-2,6-naphthalate) [PEN], etc. [1–16]. It has been found that the transesterification (or ester exchange) reaction may take place during high temperature melt blending of two polyesters [2–4,7,8,10,12–15]. As the transesterification occurs, the chemical structure of the resultant product will vary. The reaction may play a decisive role in miscibilizing two immiscible polymers. In the extreme case of a very high extent of reaction, the product would be a random copolymer of the original components.

With the recent increased interest in the study of PEN, the miscibility and phase structure of PET/PEN have been studied by several researchers [9,10,13,14]. It was reported that the physical blend of PET/PEN is immiscible. At a temperature near or above their melting points, however, transesterification occurs in the blend. The entire system was gradually homogenized due to the miscibility enhancement by the resultant PET/PEN copolymer. At this stage, blends are not crystallizable and show a single glass transition temperature between those of the starting polymers. Also, it was indicated that the transesterification is significantly influenced by temperature and time, but only negligibly by blend composition [13].

In our laboratory, we have continuously studied the structural features and mechanical properties of both PET and PEN homopolymer products, especially for uniaxially oriented filaments, made under various processing conditions [17–26]. As a continuous development of these studies, we are now involved in the study of a blend in which PET and PEN are included. In this article, melt spinning of PET/PEN blends with various compositions was done in a wide velocity range between 1 and 10 km·min^{-1}. The effect of blend composition and spinning

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Table 1
Preparation conditions and identification of the PET/PEN blend fibre samples

Sample series	Composition PET/PEN (wt%)	Take-up velocity [a] (km·min^{-1})	Sample i.d.
PET100	100/0	1,2,...10	PET100-1, PET100-2,...PET100-10
PET90	90/10	1,2,...10	PET90-1, PET90-2,...PET90-10
PET75	75/25	1,2,...10	PET75-1, PET75-2...PET75-10
PET50	50/50	1,2,...10	PET50-1, PET50-2,...PET50-10
PET35	35/65	1,2,...5	PET35-1, PET35-2,...PET35-5
PET10	10/90	1,2,...4	PET10-1, PET10-2,...PET10-4
PET0	0/100	1,2,...10	PET0-1, PET0-2,...PET0-10

[a] For each sample series, take-up velocity starts from 1 km·min^{-1}, and is increased in 1 km·min^{-1} intervals.

conditions on the structure and physical properties of as-spun and post-treated fibres were studied.

2. Experimental

2.1. Materials

Commercial poly(ethylene terephthalate) [PET] and poly(ethylene-2,6-naphthalate) [PEN] chips used in this study were kindly provided by Goodyear Tire and Rubber, and Allied-Signal, respectively. The intrinsic viscosities of the two homopolymers as measured in a 60/40 (wt%) mixture of phenol and tetrachloroethane are 0.60 dl·g^{-1} for PET and 0.89 dl·g^{-1} for PEN.

Physical mixing of the PET and PEN was carried out prior to the chip drying process. The respective weight percentages in various mixtures were PET/PEN = 100/0, 90/10, 75/25, 50/50, 35/65, 10/90 and 0/100, which formed seven sample series: PET100, PET90, PET75, PET50, PET35, PET10 and PET0 (see Table 1). Mixed chips were dried in a vacuum oven at 140°C for *ca.* 16 h and subsequently used.

2.2. Blend fibre spinning

Mixed polymer chips were melt spun in an extruder with a screw diameter of 25 mm and a round single-hole spinneret having a 0.6 mm diameter at a denier of 4.5. Melt spinning temperatures were adjusted between 300 and 310°C, depending on the composition variation in the blends. Table 1 identifies each blend and shows the conditions under which each was prepared. For each sample series, the take-up velocity started at 1 km·min^{-1}, and was increased in 1 km·min^{-1} intervals. The sample i.d. indicates both the blend composition and take-up velocity. For example, sample PET100-1 means that the content of PET in the mixed chips is 100, the content of PEN is 0% and the fibre was wound at 1 km·min^{-1}; and sample PET90-10 signifies that the content of PET is 90, the content of PEN is 10% and the blend fibre was wound at 10 km·min^{-1}. For most series (series PET100, PET90, PET75, PET50 and PET0), the spinnability appeared excellent, resulting in the attainment of superhigh speed spinning with take-up

velocity reaching 10 km·min^{-1}. In contrast, series PET35 and PET10 showed poor spinnability, and the velocity was limited in a narrow range below *ca.* 5 km·min^{-1}. The reason for this phenomenon is not yet understood.

2.3. Measurements

Wide-angle X-ray scattering (WAXS) intensity curves for equatorial scan were obtained with a Siemens type-F X-ray diffractometer using nickel-filtered CuKα radiation. The scattering intensities were recorded every 0.05° from 2θ scans in the range 5–40°.

For ^{1}H n.m.r., various samples were dissolved in a 30/70 (wt%) mixture of deuterated trifluoroacetic acid and chloroform. All spectra were obtained at room temperature on a Bruker ARX400 n.m.r. spectrometer operating at 400.13 MHz. Tetramethylsilane (TMS) was used as internal standard.

A Perkin-Elmer DSC-7 system was used for the thermal analysis of the blend fibres. The d.s.c. experiment was performed at a heating rate of 20°C·min^{-1}, and the thermogram was recorded from 20 to 320°C. The calibration of temperature was made with a standard sample of indium.

The dimensional change of the heated fibres was measured as a function of the temperature using a home-made thermal mechanical analyzer (t.m.a.). The heating rate during the t.m.a. scan was 5°C min^{-1} and the pretension applied to the fibre bundle was *ca.* 0.2 MPa. Shrinkage was calculated according to the following formula

Shrinkage = [(initial length − final length)/(initial length)]

$$\times 100\% \tag{1}$$

Details of other structural characterizations and property measurements including fibre denier, density values and more routine mechanical tests followed the same methods reported previously [17,22].

3. Results and discussion

3.1. Structural characterization

In order to study the effect of composition variation on the structural features, analysis was first performed on
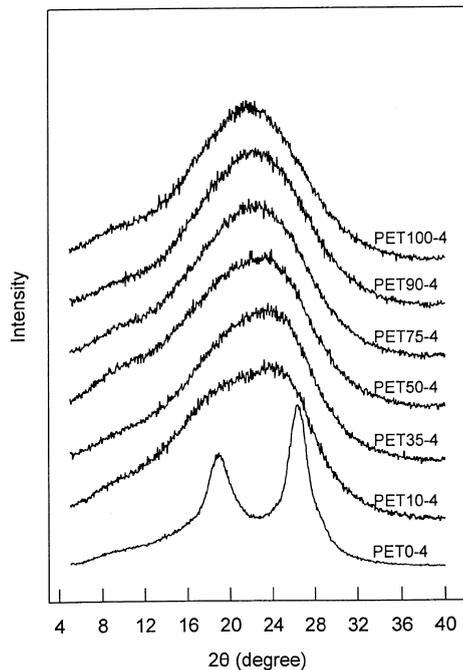
Fig. 1. Wide-angle X-ray intensity curves of various fibre samples PET100-4 to PET0-4.
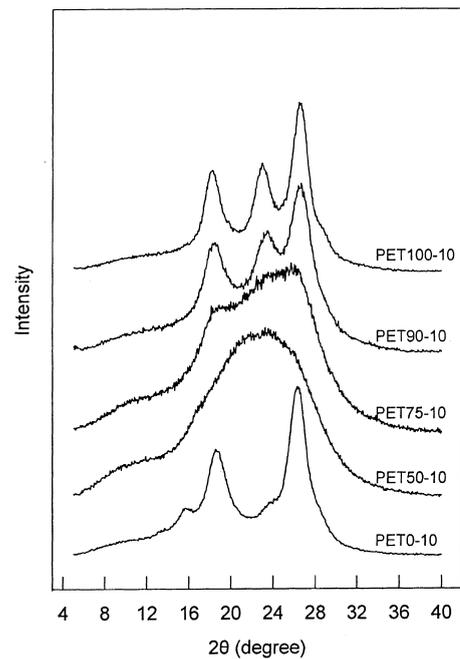

Fig. 2. Wide-angle X-ray intensity curves of various fibre samples PET100-10 to PET50-10 and PET0-10.

samples PET100-4 to PET0-4 with WAXS measurements. These samples were prepared with an increasing component of PEN in the blend starting from 0 to 100%, and all prepared at the same take-up velocity of 4 km·min^{-1}. The intensities of equatorial X-ray scattering *versus* 2θ angle for these samples are shown in Fig. 1. At the top of the figure, the pure PET fibre reveals only a broad unresolved pattern. Its peak centres at *ca*. 21° and appears to be almost symmetric. This feature indicates that sample PET100-4 is essentially amorphous [27,28]. The intensity curve of pure PEN fibre, as seen at the bottom of the figure, shows well-resolved diffraction peaks at *ca*. $2\theta = 19°$ and 27°, which may be assigned to (020) and (200) reflections of the β form crystals, and an overlapped ($\bar{1}10$) reflection of the α form crystal [29–34], indicating that a crystalline structure has developed in this sample. The WAXS curve of any blend fibre shows only one characteristic peak. This pattern does not change significantly, even though the content of PEN was increased from 10 to 90%, proving that the crystallization process is restricted in these melt-blending fibres. By careful observation, it is evident that the diffuse scattering curves of PEN-rich blends (PET35-4 and PET10-4) appear to be slightly asymmetric in their shape. For comparison, observe the intensity curve of sample PET0-4, where two diffraction peaks with different intensity are seen. Such asymmetric profiles of PET35-4 and PET10-4 may be considered to be a summation of a symmetric amorphous scattering and a few peaks diffracted from PEN crystals, although these crystals may have small sizes or imperfect structure.

The dominant amorphous structure existing in the five

blended fibres may be caused by either relatively low take-up velocity or a second component being added which may disturb the structural regularity. For clarifying the point at issue, the equatorial WAXS traces of some fibres spun at a superhigh velocity of 10 km·min^{-1} are shown in Fig. 2. By comparison with Fig. 1, an evident effect of take-up velocity on fibre structure is observed. Intensity curves of the homopolymer fibres (PET100-10 and PET0-10) and fibre with high content of PET (PET90-10) show a developed crystalline structure with strong peaks from various crystalline planes. As seen in the bottom curve, sample PET0-10, two new peaks have appeared at 15.7 and 23.3° which overlap with the (020) and (200) peaks reflected from β form crystals. These new peaks which appeared in the superhigh speed spun PEN fibre are (010) and (100) reflections from the α modification [31–34], indicating that the content of the α form crystal is promoted by increased take-up velocity. On the other hand, when the second component is increased from a minor level to a relatively high one, e.g. from PET90-10 to PET50-10, the crystallization process seems to become difficult. In the case of PET75-10, the crystalline diffractions are not resolved. Its WAXS curve reveals several strong, but diffuse reflections on the equator which seem to contain three PET diffraction (010, $\bar{1}10$ and 100) rather than two PEN peaks, suggesting that small PET crystals are formed preferable to PEN crystals in this blend. Further, no crystalline trace can be observed in the scattering profile of PET50-10. This means that neither PET nor PEN can form a crystalline structure, even at this superhigh take-up velocity.

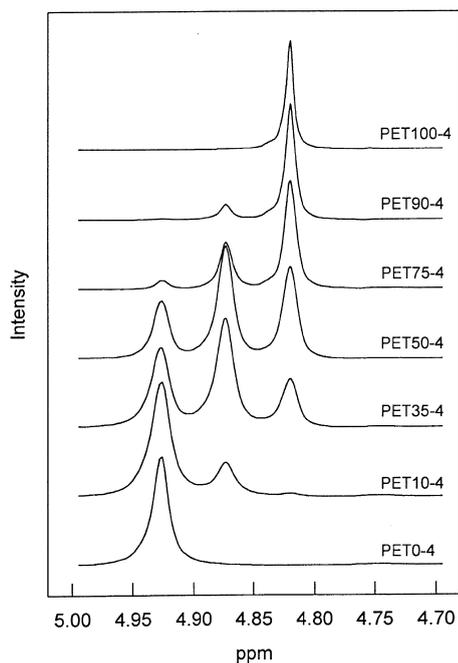As previously mentioned, another reason which might

Fig. 3. ^{1}H n.m.r. spectra of the ethylene region of various fibre samples PET100-4 to PET0-4.

Fig. 4. d.s.c. scans of various fibre samples PET100-4 to PET0-4.

restrain the development of crystalline structure in the blends may be the formation of block or random copolymers which are introduced by transesterification. It was reported that evidence of this reaction can be detected for a PET/PEN blend which was heat-treated at 280°C for only 5 min [13]. In this study, because an extrusion temperature of *ca*. 300°C was chosen and polymer would remain in the molten state longer than 5 min, the transesterification, therefore, would very likely occur. To verify this inference, n.m.r. measurements were performed. ^{1}H n.m.r. spectra in the region corresponding to the ethylene unit for samples PET100-4 to PET0-4 are presented in Fig. 3. According to Refs [10,13,14], PET and PEN homopolymers, respectively, exhibit one characteristic n.m.r. peak associated with the ethylene proton. For PET, the resonance peak representing the ethylene unit between two terephthalate groups (TET) appears at 4.82 ppm, while the resonance peak representing the ethylene unit between two naphthalate groups (NEN) appears at 4.92 ppm. The reaction extent of transesterification in the PET/PEN blends can be recognized from a third resonance peak which represents the ethylene unit between one terephthalate and one naphthalate group in the backbone (TEN). This new peak located at 4.87 ppm is evident in melt blends of PET and PEN which have undergone transesterification [10]. As shown in Fig. 3, in the cases of PET and PEN fibre samples, only one resonance peak was, respectively, observed at their characteristic location. This is a reasonable result for these homopolymers. In contrast with the homopolymers, a resonance peak at 4.87 ppm clearly appears for all blend fibres, indicating that transesterification indeed did occur during the melt spinning
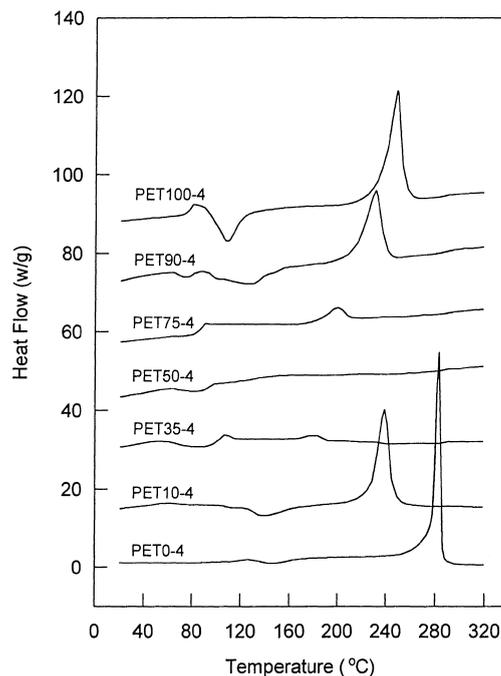
process. The intensity of the 4.87 ppm peak changes significantly with blend composition. It is relatively weak when the second component is in a minor level (PET90-4 and PET10-4). The resonance peaks corresponding to the second component, e.g. 4.92 ppm peak appearing in sample PET90-4 and 4.82 ppm peak appearing in sample PET10-4, have essentially disappeared, because most molecules of the second component have been consumed in producing PET/ PEN copolymer chains. The blend fibres PET90-4 and PET10-4, therefore, may be considered to be a mixture of a vast amount of the first component and a small amount of copolymer. This means that crystallization in such samples becomes possible when the blends were spun at superhigh take-up velocity or subjected to heat treatment, as confirmed by WAXS and d.s.c. data (Figs 2, 4 and 5). On the other hand, intensity of the resonance peak at 4.87 ppm increases significantly when the content of the second component is increased to be close to that of the first component, as represented in spectra of PET75-4 and PET35-4, and especially in the case of sample PET50-4 with equal content of two components. The extremely intense peak appearing at 4.87 ppm means that the molecular backbone contains a massive concentration of TEN groups, resulting in noncrystallizability in sample PET50-4, even though it is spun at 10 km·min^{-1} or subjected to heat treatment, as will be discussed later. Following Yamadera's method, it is possible to evaluate the actual level of transesterification in a sample from the relative area of the above-discussed three resonance peaks [35]. The details of quantitative analysis for these blend fibres will be reported elsewhere.

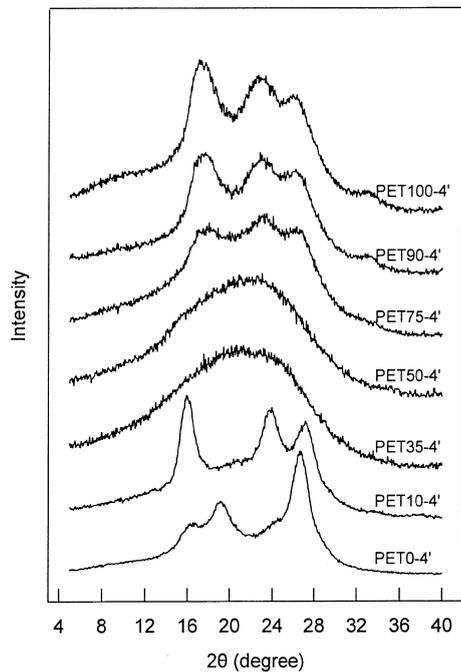Fig. 4 gives d.s.c. curves of the seven samples PET100-4

Fig. 5. Wide-angle X-ray intensity curves of heat-treated fibre samples PET100-4′ to PET0-4′.

Fig. 6. The density of various fibre samples as a function of composition and take-up velocity.

to PET0-4. For PET fibre PET100-4, a well-known thermogram was observed where the glass transition can be observed at *ca.* 78°C, followed by the cold crystallization exotherm with a peak temperature near 108°C, and after that a melting endotherm at *ca.* 247°C. In contrast to this trace is PEN fibre PET0-4. A minor increase of the heat flow appearing near 120°C is attributed to the glass transition, although it becomes somewhat difficult to recognize. Next to this heat flow change, the curve shows a small cold crystallization peak near 145°C, and a sharp melting endotherm at 283°C. Such a small cold crystallization peak and a relatively intense melt peak indicate that the PEN as-spun fibre is already partially crystalline. This conclusion is consistent with the X-ray data shown in Fig. 1. The respective situations between these two homopolymers exhibit a gradual change. A single glass transition was found for all blend fibres. Because a physical PET/PEN blend has been reported to be immiscible [10,13], the appearance of only one glass transition indicates, again, that the components have been transesterified and then are compatible in the amorphous phase. These T_g values mediate between those of the two homopolymers, and are seen to shift to higher temperatures with an increased PEN component in the blend fibres. This may be simply attributed to a much higher T_g in PEN than that in PET. It is also clear that if the content of the second component is quite low, e.g. in the cases of samples PET90-4 and PET10-4, the obvious crystallization exotherm and melt endotherm peaks are observed. With an increase in the content of the second component (fibres PET75-4 and PET35-4), however, the trace of the cold crystallization disappears, whereas only a small endotherm peak appears
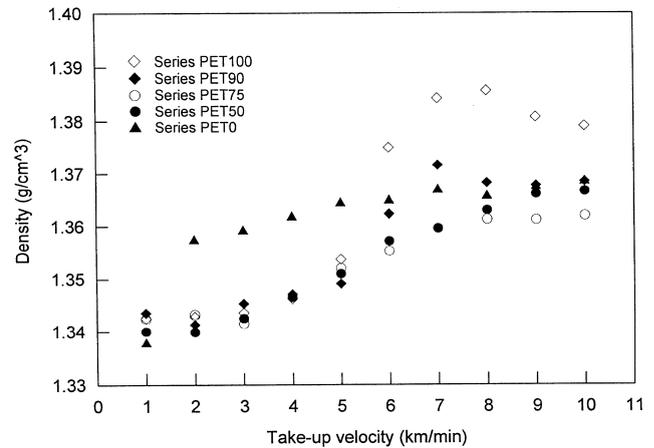
at a relatively low temperature. Further, in the case of sample PET50-4 with an equal content of PET and PEN, neither cold crystallization nor the melting peak is observed any longer in its d.s.c. diagram. The exception, the glass transition appears, indicating that the original blend fibre with this special composition is essentially amorphous and not crystallizable. The above d.s.c. observations are closely consistent with the conclusions drawn from WAXS and n.m.r. data. By comparing Fig. 4 with Fig. 1, one may also deduce that some as-spun fibres (PET90-4, PET75-4, PET35-4 and PET10-4) with essentially amorphous structure (affirmed by WAXS results) may be transferred to a semi-crystalline structure during the heating process of the d.s.c. testing. Such newly formed crystals, confirmed by large or small endothermic peaks, show a lower melting point than that of either of the two homopolymers. In order to identify the crystalline structure existing in such fibres, annealing treatment was done on these samples to promote a further developed crystalline structure.

Conditions for the annealing treatment were determined experimentally. To encourage the cold crystallization and avoid melting, the annealing temperature should be slightly higher than the cold crystallization temperatures of samples PET90-4 and PET10-4, which exhibit a higher cold crystallization temperature, and slightly lower than the melt temperature of samples PET75-4 and PET35-4, which have a lower melt temperature. For this temperature, 145°C was chosen, and correspondingly, a treatment period of 8 min was used. A new series of heat-treated samples, respectively, prepared from samples PET100-4 to PET0-4 was labelled PET100-4′ to PET0-4′ and examined by WAXS measurements. Fig. 5 shows the WAXS scans performed at room temperature for the seven annealed samples. As revealed by the 2θ scan of annealed fibres PET50-4′ and PET35-4′, a broad pattern of as-spun fibres PET50-4 and PET35-4 observed in Fig. 1 was retained after the annealing process. By comparison with WAXS, n.m.r. and d.s.c. data,
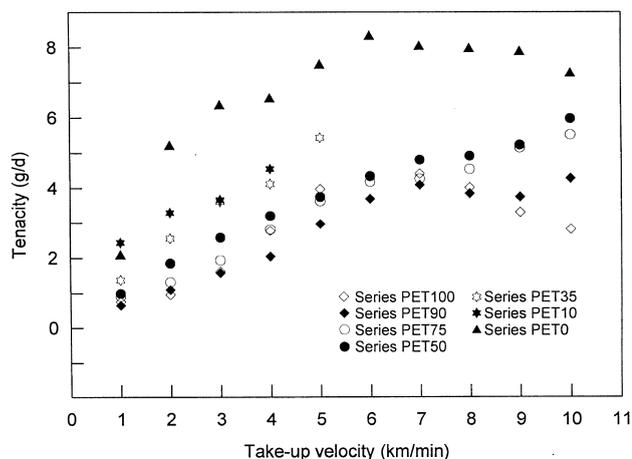
Fig. 7. The tenacity of various fibre samples as a function of composition and take-up velocity.

Fig. 8. The initial modulus of various fibre samples as a function of composition and take-up velocity.

represented in Figs 2–4, these observations indicate again that the blend fibres containing an equal (50/50) or relatively closer content (35/65) are non-crystallizable, even though they are annealed under the stated conditions or spun at superhigh velocity. On the other hand, individual diffraction peaks are fairly well presented in blend fibres of PET90-4′ and PET75-4′ which are rich in PET, as well as in the pure PET sample PET100-4′. Such peaks in the three profiles seem to be located at the same angles which centre near $2\theta = 17.4°$, 22.6° and 25.7°, respectively. Because the three Bragg's angles mentioned just correspond to the (010), ($\bar{1}$10) and (100) planes of PET crystal, the developed crystalline structure in blend fibres PET90-4′, PET75-4′ and of course in PET100-4′ are considered to be PET triclinic crystals. From this result, it may be concluded that the PET crystals will be formed separately in PET-rich blends rather than forming PET/PEN co-crystals, although it is not clear whether weak PEN reflection peaks overlap those of PET or not. A similar phenomenon was observed in the cases of PEN-rich blend. For sample PET10-4′ and the pure PEN fibre PET0-4′, only reflections of PEN crystal were observed. The situation, however, becomes slightly complicated, because two crystal modifications appeared. These modifications are the same as those reported by Buchner, and Cakmak and co-workers [30,33,34], where characteristic reflections of the β form crystal are (020) and (200) at $2\theta = 18.6°$ and 26.9°, whereas the α form shows three reflections (010), (100) and ($\bar{1}$10) at $2\theta = 15.7°$, 23.3° and 27.0°, respectively. As compared with the WAXS scan of as-spun fibre PET0-4 shown in Fig. 1, the annealed sample PET0-4′ is still crystallized mainly in the β form. However, related to the newly appearing (010) reflection just mentioned, a small amount of α modification seems to coexist in this sample. In the case of sample PET10-4′, it may be seen that three new reflection peaks appear and are assigned to the α modification of PEN. The profile seems to contain a small amount of β form from indications of a
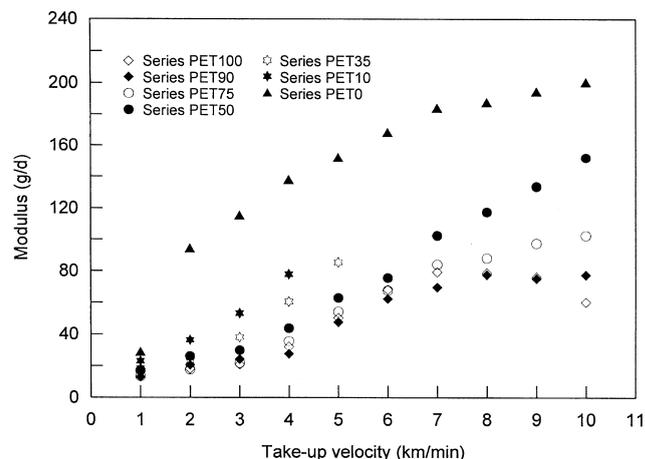
broad portion between the (010) and (100) diffractions, and may conceal small peaks of PET crystals. However, similar to the conclusion drawn for PET-rich fibres PET90-4′ and PET75-4′, instead of PET/PEN co-crystallization, only separate PEN crystals were observed.

Densities of the blend fibres with varying composition *versus* take-up velocity are compared with those of the two homopolymers in Fig. 6. Samples prepared from PET homopolymer exhibit a sudden increase in the density at *ca.* $5\ \mathrm{km·min}^{-1}$ because of a well-known crystallization phenomenon induced by an increased spinline stress at this speed. A density decrease is also observed in the velocity range beyond $8\ \mathrm{km·min}^{-1}$, due to the formation of micro-voids and skin-core structures. These observations are consistent with other reported results [27,36]. On the other hand, a monotonic increase in density as a function of take-up velocity is found in the case of the PEN homopolymer. The density rises rapidly at the lower velocity range and then levels off at *ca.* $4\ \mathrm{km·min}^{-1}$. For each blend fibre series, an obvious increase in density at higher speed is observed. In comparison with the density data of the two homopolymers, however, their values are relatively low. This is considered to be due to these samples generally retaining an amorphous structure and hence possessing low crystallinity, as mentioned earlier.

3.2. Physical properties

Figs 7–9 show the dependence of tenacity, initial modulus and elongation at break of various as-spun fibres, on their composition as well as on the take-up velocity. In the case of pure PET fibres (series PET100), a well-known tendency is observed. It appears that tenacity and modulus first increase with increasing take-up velocity, reach maximum values at *ca.* $7\ \mathrm{km·min}^{-1}$ and then decrease beyond this velocity. This decrease in mechanical properties has been attributed to the formation of voids and skin-core structure
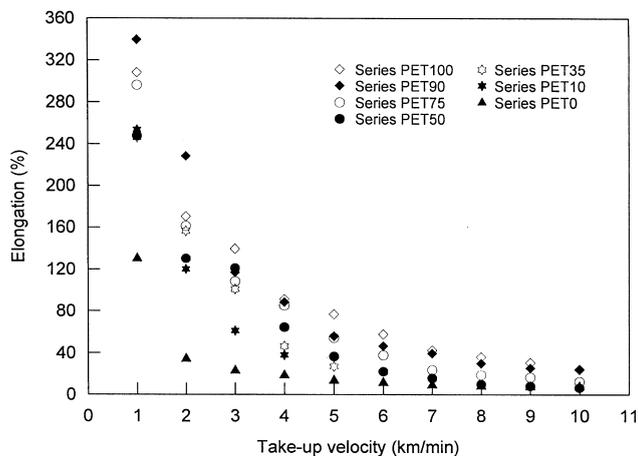
Fig. 9. The elongation of various fibre samples as a function of composition and take-up velocity.
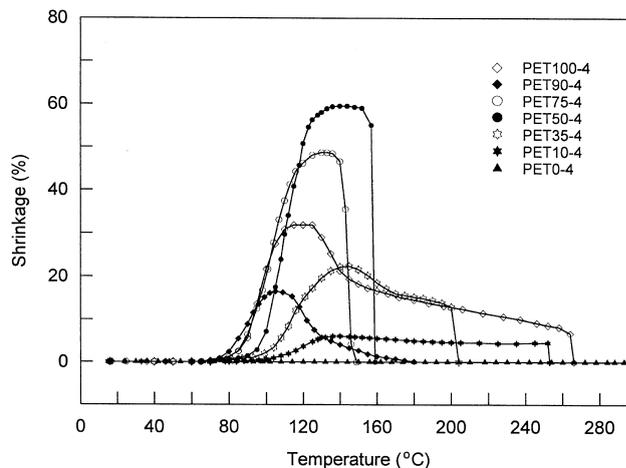

Fig. 10. Temperature dependence of the shrinkage of various fibre samples PET100-4 to PET0-4.

at superhigh spinning speed [36]. However, such a decrease is not observed in other cases. For most series, tenacity and initial modulus increase, and elongation decreases with increased take-up velocity, indicating that an improved molecular orientation is achieved when as-spun fibres were wound at higher speed. In the superhigh speed range from 7 to 10 km·min^{-1}, mechanical properties of these samples still increase or only level off. At the present time, we have no further data to explain this phenomenon. On the other hand, as expected, PEN homopolymer exhibits both the highest tenacity and initial modulus, and the lowest elongation compared to other samples, over the entire velocity range, with a few exceptions which may be caused by experimental error. Generally, at the same take-up velocity, fibre samples with high PEN content possess better mechanical properties.

The percent thermal shrinkage of the two homopolymers and the five blend fibres wound at a take-up velocity of 4 km·min^{-1} are plotted as a function of temperature in Fig. 10. The dimensional change in all samples is low in the range between room temperature and *ca.* 70°C, due to restricted molecular motion. At *ca.* 70°C, which is close to the glass transition temperature of PET, shrinkage begins to occur for several samples. The temperature at which a fibre starts to shrink quite obviously seems to depend on the PEN content. The higher the PEN content, the higher this temperature, indicating good heat resistance in PEN. In the case of pure PET, observed shrinkage behaviour strongly confirms results reported by others [27]. A significantly increased shrinkage exhibited in the temperature range from its T_g to *ca.* 110°C is caused by a disorientation process occurring in the oriented amorphous phase. The decrease with further increase in temperature is associated with the diminished disorientation process and the onset of crystallization during heating [37]. In comparison with this relatively high shrinkage peak, pure PEN fibre exhibits an undetectable dimensional change when it is subjected to

high temperature. This can be attributed to its more rigid backbone in which the benzene ring in PET is replaced by the naphthalene ring, and a high crystallinity which is achieved in the melt spinning process (see Fig. 1) and promoted further during the shrinkage test. The traces of the various blend fibres reveal obvious differences among themselves. In general, high temperature shrinkage behaviour is mainly considered to depend on the fraction of oriented amorphous phase, and such a fraction in an as-spun fibre is related principally to the take-up velocity in the melt spinning process. For the blend fibre samples, furthermore, high temperature shrinkage also seems to be related to the composition. It may be assumed that similar amounts of oriented amorphous chains exist in the five as-spun blend fibres, because the same take-up velocity was used for preparing these samples, but such a fraction may change, respectively, during the heating process and during the shrinkage test. As stated earlier, the blend fibres with a greater disparity between the two contents can be crystallized easier than fibres containing equal or similar contents of the two components when they were heated. In other words, more oriented amorphous chains are retained in sample PET50-4 than others, resulting in the highest level of shrinkage in this sample. Compared to PET50-4, relatively lower shrinkages are found in other samples, because molecular thermal motion is partially restricted by newly developed crystals during the test.

4. Conclusions

In this paper, the effects of blend composition and take-up velocity on structural features and physical properties of PET/PEN melt-blended fibres were studied using WAXS, n.m.r., d.s.c., density, tensile and thermal shrinkage. n.m.r. experimental results suggested that the transesterification proceeds during the melt extrusion process and PET/PEN

copolymers were formed. Over the ranges studied, the structure and properties of the blend fibres were significantly influenced by the composition and spinning velocity. All blend samples exhibit a single glass transition temperature intermediate between those of the respective two homopolymers, indicating that a miscible amorphous phase has been formed through transesterification. The WAXS and d.s.c. results on both as-spun and annealed samples indicated separate crystals for the two components rather than co-crystallization, and crystallization in either PET or PEN was inhibited in the blend by the presence of the other component. Transesterification also results in a reduction in the melting point, crystallization temperature and crystallinity. In general, at a certain take-up velocity, mechanical properties including tenacity and modulus of the various blended fibres were improved gradually with the increasing PEN content, while their breaking elongation is lower than that of the PET homopolymer. The thermal shrinkage, however, is quite pronounced for the various blended fibres and reaches a maximum in the case of 50/50 PET/PEN blend, indicating that the copolymer chain is fairly well oriented but exists in the amorphous state.

Acknowledgements

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