

Structure and Properties of Polyethylene–Polypropylene Blend

J. W. TEH, *School of Applied Sciences, University Science Malaysia, Penang, Malaysia*

Synopsis

The structure and mechanical properties of blends of low-density polyethylene and isotactic polypropylene were studied. The blends behaved like a simple composite obeying the rule of mixture for the modulus dependence on composition. Tensile strength of the LDPE was enhanced with the addition of PP but elongation at break was drastically reduced for all blend compositions. Studies with WAXD, hot-stage microscope, and DSC indicated lack of interaction between the LDPE and the PP. The crystallographic structure of LDPE and PP remained unchanged. However, the spherulite size of the PP was found to be reduced in the presence of LDPE, possibly due to an increase in nucleation density.

INTRODUCTION

When two or more different polymers are intimately mixed to form a single continuous solid product, the resultant composition is referred to as a polymer blend or polyblend. Polymer blends have demonstrated some superior quality either in cost consideration or special performance requirement or both. The complexity in the structure of polyblends, controlled mainly by the compatibility and interaction between the components, often leads to difficulty in predicting the properties as well as structure-property relation.

Polymer blends have recently become major new areas of macromolecular science research.^{1–6} Depending mainly on the compatibility between the composite polymers, the resultant blend may exhibit properties ranging from those of a homogeneous mass similar to a copolymer to those of a simple composite mixture or reinforced composite. Polymer–polymer compatibility has been well documented by Krause.⁴ Compatibility between polyolefins was shown to be limited depending on the type of polyolefins. For example, polyethylene (PE) and atactic polypropylene (PP) were found to be conditionally compatible and miscible, but polyethylene and isotactic polypropylene were completely incompatible especially in bulk.⁷ However, blends of polyolefins have become commercially important where certain properties of the blends were shown to be better than the parent homopolymers. Although toughened polypropylene has been manufactured by copolymerization of ethylene and propylene monomers,⁸ mechanical blending of PE and PP, if resulting in an improved product, may prove to be a more economical and easier method to produce new polymeric materials for specific applications.

Studies on PE–PP blends are relatively few.^{9–15} From structural consideration, though both exhibit the same spherulitic morphology, the two polymers may not be compactible in bulk since polyethylene crystallizes in the more stable orthorhombic form whereas isotactic polypropylene crystallizes in the monoclinic

α -form.¹⁶ The following is a report on some work done on PE-PP blends, which may help us in the understanding of the mechanical behavior and structure of this polyblend.

EXPERIMENTAL

Materials and Specimens Preparation

The polyethylene used was Lupolen BASF 2425K by Bayer with a density of 0.924–0.926 and an MFI of 3.4–4.6. The polypropylene used was Novalen BASF 1120 HX by Bayer with density of 0.907 and an MFI of 1.5–2.2. The raw materials were in the form of granules. Samples were prepared either in the form of sheets or square bars. Initially, sheets of 1 mm thickness were prepared by compression molding directly from the mixture of granules of LDPE and PP. These sample sheets were found to be macroscopically inhomogeneous and gave very low tensile strength. Consequently, a premixing process as described below was carried out before compression moulding.

PP granules were added slowly to a two-roll mill (8 cm diameter and 15 cm length) maintained at 175°C until a continuous band was obtained. LDPE granules were then added. The mixture was allowed to melt blend for about 30 min to achieve homogeneity, removed, and air-cooled. The resultant crepe was then compression-molded in a cavity mould of dimensions 15 cm × 25 cm × 0.1 cm between Melinex sheet backed with steel plates at a temperature of 175°C and a pressure of 20 tons (4-in. diameter ram). The molded sheets were air-cooled and stored for more than a week before testing.

Rectangular bars of dimensions 21.8 cm × 1.0 cm × 1.0 cm were obtained by injection moulding using a Fox Oxford U 1469 Unimoulder with a barrel temperature of 175 ± 2°C and a mould temperature of 98 ± 2°C.

The blend compositions are given in Table I.

Mechanical Testings

Tensile test was carried out using the Instron 1114 with a 2- to 100-kg load cell and a crosshead speed of 2 cm·min⁻¹. Dumbbell specimens of thickness 1.0 mm, width 7 mm, and gauge length 10 cm, were stamped out using a Wallace cutter (ASTM D412 Die F). Flexural modulus was obtained by cantilever deflection method using the injection moulded bar. Shear modulus was determined using the apparatus described by Ladizesky and Ward,¹⁷ using parallel strip

TABLE I
Composition of PE-PP Blend

Sample	Proportion (%) of LDPE by weight	Proportion (%) of PP by weight
A	0	100
B	20	80
C	40	60
D	60	40
E	80	20
F	100	0

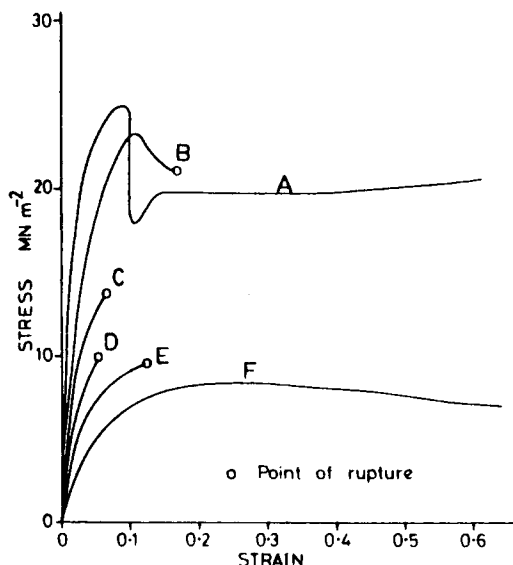


Fig. 1. Uniaxial stress-strain curves of: (A) polypropylene; (B) 80 PP/20 PE; (C) 60 PP/40 PE; (D) 40 PP/60 PE; (E) 20 PP/80 PE; (F) polyethylene.

specimens of dimensions 4 cm \times 0.3 cm \times 0.1 cm. Polyblend specimens of dimensions 5.5 cm \times 1 cm \times 1 cm were used in the Charpy Impact Test with Zwick Model 5101. A standard 45° V-groove of 0.25-mm R was introduced using a Blacks Equipment Groove Cutter. Specimens were tested at 25°C, 0°C, and -196°C (liquid nitrogen temperature).

Structure and Morphological and Thermal Studies

A specimen in the sheet form of 1 mm thickness was mounted in a Phillips X-Ray Diffractometer. A powder diffraction pattern was obtained with Cu $K\alpha$ (Ni filter) radiation (40 kV and 20 mA) and registered on a flat film placed at 5.0 cm from the specimen exposed for 30 min.

The morphology of the homopolymers and the blends were observed under cross-polarized light and the melting behavior was studied with a Mettler FP 5 hot-stage microscope and with a Perkin-Elmer Differential Scanning Calorimeter. For optical microscopic observation, approximately 5 mg of the sample were melted on a glass slide over a hot plate at 180°C, and a thin film of approximately 100 μ thickness was obtained by pressing the molten sample with a thin cover slide. For the DSC study, approximately 10 mg of sample were used with a heating rate of 20°C \cdot min $^{-1}$. The instrument was calibrated with indium.

RESULTS AND DISCUSSION

Uniaxial Stress-Strain Behavior

The stress-strain curves are as shown in Figure 1. The tensile strength and the elongation at break are plotted against the composition as shown in Figure

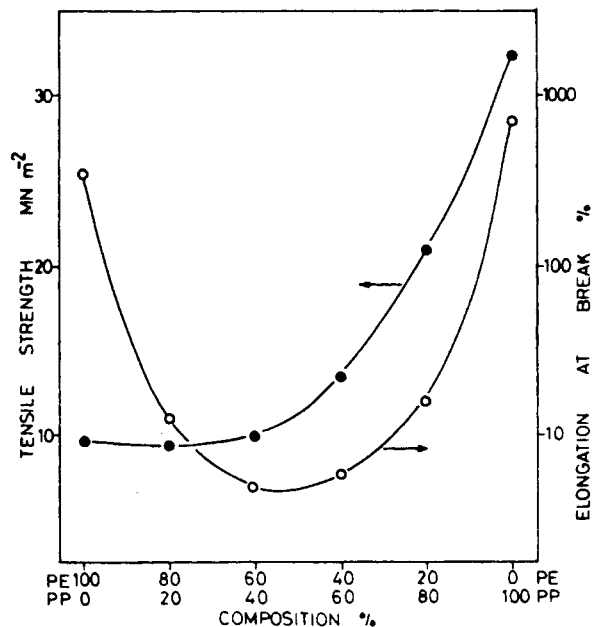


Fig. 2. Tensile strength and elongation at break of PE-PP blends.

2. For the homopolymers of LDPE and PP, yielding was observed to occur at approximately 30% and 9% strain, respectively. Necking and cold drawing ensued up to several hundred percent strain after which strain hardening took place when necking had spread throughout the gauge length. This change was followed by ultimate rupture. This deformation behavior was typical of semicrystalline polymers capable of undergoing chain orientation. However, for the polyblends, a drastic reduction in elongation at break was observed, as shown in Figure 2. Similar observation was made by Deanin and Sansone¹¹ for LDPE-PP blend. This was attributed to the two-phase character of the blends due to the incompatibility of LDPE and PP. In the case of HDPE-PP blend, the same trend was observed.¹⁰⁻¹² Yielding was observed for all compositions in the case of Deanin¹¹ and Lovinger.¹² However, Noel and Carley¹⁰ observed that for HDPE-PP blend, yielding was not prominent for the blends, similar to the present finding for LDPE-PP blend, where yielding was observed for homopolymers and for 80% PP only. For other compositions, rupture occurred prior to any yielding.

In the case of LDPE-HDPE blend, Shishesaz and Donatelli¹⁵ found that elongation at break decreased rapidly with increase in HDPE content, and approached the value of homopolymer HDPE. This indicated compatibility between LDPE and HDPE. Deanin and Sansone,¹¹ on the other hand, found that LDPE-HDPE blend also exhibited incompatibility by the lower values of elongation at break for the blend compared with that of the homopolymers.

The inability of the LDPE-PP polyblend to undergo yielding and cold drawing clearly points to the fact that LDPE and PP are incompatible and the resultant blends resemble a composite mixture of individual LDPE and PP spherulites as shown later by X-ray and thermal-optical analysis. The presence of a second phase, though of similar spherulitic morphology, introduces weak interspherulitic

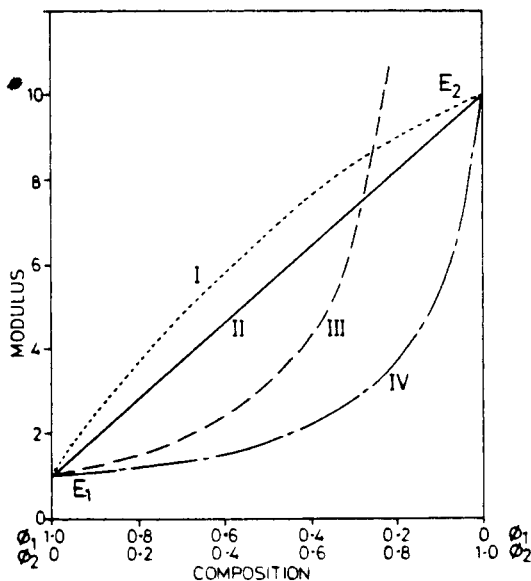


Fig. 3. Theoretical curves of modulus of two-component composite: (I) simplex equation; (II) rule of mixture, Voigt model; (III) Nielsen and Halpin modified Kerner equation; (IV) rule of mixture, Reuss equation.

boundaries resulting in the rupture of the blends without any cold drawing occurring.

The nominal tensile strength (rupture load divided by original cross-sectional area) increased with increase in PP content. The result differs from that obtained by Deanin and Sansone,¹¹ who found that the nominal tensile strength showed a complex pattern showing a deterioration in strength at low PP content and a synergism at high PP content. For HDPE-PP blend, synergism was observed by Noel,¹⁰ Deanin,¹¹ and Lovinger.¹² The observed tensile strength variation for LDPE-PP was similar to the trend exhibited by the LDPE-HDPE blend.¹⁵ On the other hand, the true tensile strength of the polyblends is much lower than that of the homopolymers.

Modulus of Polyethylene-Polypropylene Blends

In the case of a particulate-filled or two-phase composite system, the resultant modulus of the composite is a function of the modulus of the individual pure components, the volume or weight fraction, the geometry and packing of the disperse phase, and the Poisson ratio of the matrix. Four possible ways for the dependence of modulus on composition are, as shown in Figure 3, the "rule of mixture" given by curve II for the Voigt model and Curve IV by the Reuss equation, Nielsen and Halpin modified Kerner equation¹⁸⁻²¹ given by Curve III, and the simplex equation^{22,23} given by Curve I. The proposed form of simplex equation was

$$E = \phi_1 E_1 + \phi_2 E_2 + \beta \phi_1 \phi_2$$

where E is the modulus, ϕ is the weight fraction or volume fraction, and the

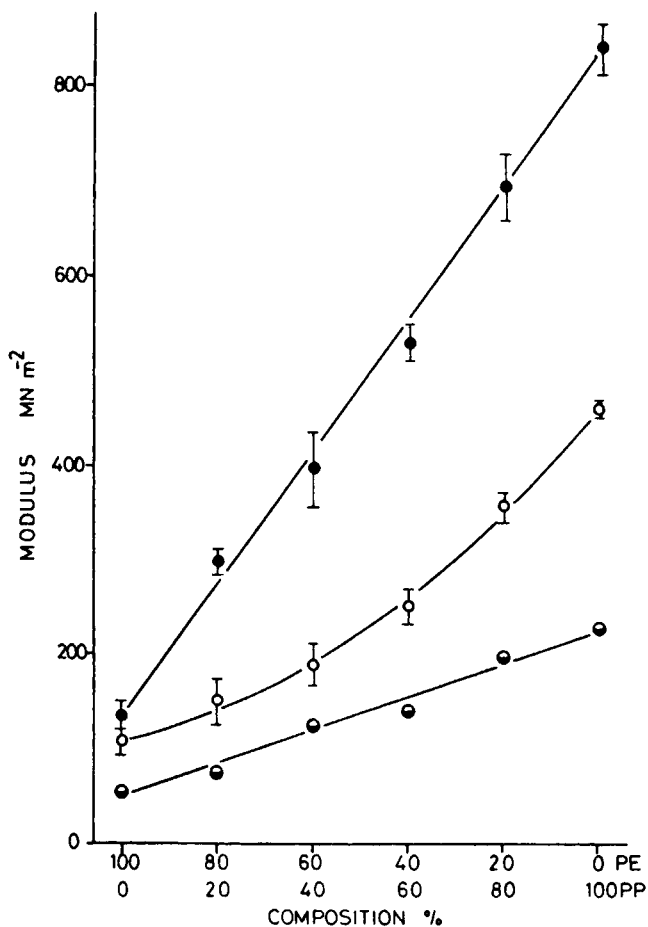


Fig. 4. Flexural (●), shear (○), and tensile (⊙) modulus of PE-PP blends.

subscript 1 and 2 refer to the component 1 and 2 of the composite system, respectively. Positive and negative values of β represent nonlinear synergism and antagonism, respectively.

The dependence of modulus on blend composition is as shown in Figure 4. The modulus at 5% elongation was obtained from the uniaxial tensile test. For the flexural and shear modulus, the data conform to those predicted by the rule of mixture equation. In the case of modulus at 5% elongation, the simplex equation gave a good fit when β was assumed a value of $-240 \text{ MN}\cdot\text{m}^{-2}$. The tensile modulus data conformed to those obtained by Deanin and Sansone¹¹ for LPPE-PP blend, and also to those of Shishesaz and Donatelli¹⁵ for LDPE-HDPE blend. The modulus behavior for LDPE-PP was, however, different from that of HDPE-PP, which showed a definite positive synergism.¹⁰⁻¹² The positive synergistic effect was attributed to the fact that the spherulite size of PP was reduced in the presence of PE.

The enhancement of modulus could also be due to an increase in crystallinity. Kleiner et al.²² suggested that modulus may be related to the density. As density is closely related to the crystallinity of the polymers, the density of the polyblends

was measured by the density gradient column method (ASTM D1505) using a water-methanol mixture. The dependence of density on composition was found to obey the rule of mixture, indicating there was no interaction between LDPE and PP or any enhancement in crystallinity.

Hence for LDPE and PP blends where no interaction had taken place between the homopolymers, antagonism observed was due possibly to the weak boundary between the two incompatible phases, as shown by the present findings. On the other hand, when HDPE interacted with PP to alter the crystalline as well as the morphological state, an enhancement in modulus was observed.¹⁰⁻¹²

Modulus of crystalline polymer with spherulitic morphology was shown to depend both on spherulite size as well as lamellar thickness.²⁴ By annealing high-density polyethylene at various temperatures, spherulite size will remain constant while the lamellar thickness was varied. At constant spherulite size, modulus was found to increase linearly with lamellar size. If a spherulitic polymer is considered to be a composite system where the amorphous portion is reinforced with randomly arranged fibers,²⁵ then the modulus may be related to the aspect ratio (defined as the ratio of the spherulite's diameter to the lamellar thickness) and the volume fraction of the crystalline material.²⁴ However, with the present polyblend systems (where we have a real composite system with a possibly bimodal distribution of spherulite size and unknown interacting forces which may alter the lamellar thickness, spherulite size, and crystallinity of the individual component), prediction of modulus-composition-structure interrelationship would have to involve holding various parameters constant. Such prediction would need more elaborate study.

Impact Strength

Blends of PE and PP have been found to give high-impact strength and low-temperature toughness. The toughening mechanism could be the same as that of toughening glassy polymers by rubbery fillers.³ The reported hardness for LDPE and PP are Rockwell hardness R10 and R95, respectively.²¹ For LDPE-PP blends where LDPE is the disperse phase and PP the matrix, we may consider LDPE as the soft rubbery phase improving the impact resistance by inducing crazing of PP or absorbing more energy during impact fracture. For a ethylene-propylene copolymer, impact strength was found to increase with ethylene content.²⁶ The result given for an ICI Propathene homopolymer GWM 201 indicates that the higher ethylene content copolymer gave the best impact performance at room temperature, but improvement at low temperature was only slight.

The impact strength results were plotted as shown in Figure 5. At room temperature, the impact strength was seen to increase sigmoidally with increase in LDPE content. This was similar to the observation made by Deanin and Sansone.¹¹ At 0°C, impact strength increased slightly with increasing PE content to a maximum for LDPE content between 40% and 60%, and then decreased again. The pure homopolymers of LDPE and PP gave similar impact strengths. At liquid nitrogen temperature which is well below the glass transition temperature of LDPE and PP, impact strength remained the same for all compositions. Comparing impact strength of copolymer and polyblend with that of homopolymer at 25°C, the impact strength of high ethylene copolymer GWM

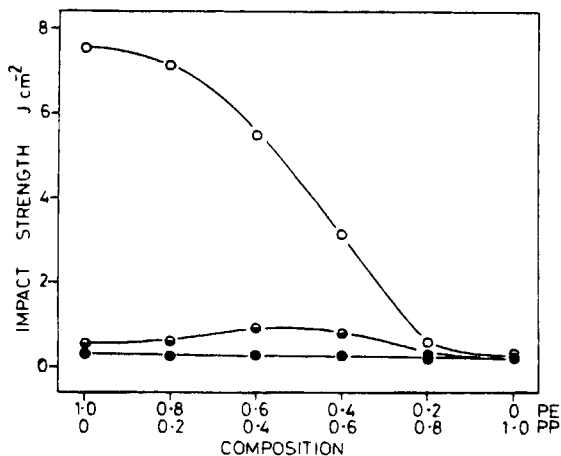


Fig. 5. Impact strength of PE-PP blends: (○) 25°C; (◐) 0°C; (●) 196°C.

201 ($1.68 \text{ J}\cdot\text{cm}^{-2}$) was approximately 13 times that of homopolymer GWM 22 ($0.126 \text{ J}\cdot\text{cm}^{-2}$)²⁶; a 50-50 polyblend gave an impact strength of $4.5 \text{ J}\cdot\text{cm}^{-2}$, which is also 13 times that of homopolymer ($0.35 \text{ J}\cdot\text{cm}^{-2}$). Hence a polyblend was shown to exhibit comparable enhancement of impact resistance.

At low temperature, addition of LDPE showed no significant improvement in the impact behavior. This lack of improvement was due to the fact that LDPE became brittle at low temperature and hence no longer acted as rubbery reinforcing fillers. The slight improvement in impact resistance at 0°C for 50-50 blend might have been due to the fact that the crystallinity of PP was reduced with the presence of LDPE. Bhateja et al.²⁷ have shown that the impact fatigue of ultrahigh molecular weight linear polyethylene was improved with decrease in crystallinity. The rate of crack propagation of the slowly cooled samples (high crystallinity) was much higher than that of the quenched samples (low crystallinity), showing that a reduction in crystallinity enhanced impact performance. A quenched sample normally yielded smaller spherulites than an annealed or slowly cooled sample. Hence we may deduce that impact strength increases with decrease in spherulite size.

Thermal Behavior

Figure 6 shows the thermogram scan of the homopolymers and the blends. Two peaks were observed for the blends. Although the use of peak temperatures to define T_m of polymers is almost universal, it must be realized that the melting peak temperatures are sensitive to the rate of heating, sample size, and shape whereas the initial melting point (obtained from the intersection of the maximum slope with the base line) has been shown to be less dependent.²⁸ The peak temperatures of LDPE varied from 110°C to 115°C and that of PP varied from 160°C to 165°C. This variation was found to be due to the effect of sample size. A determination of initial melting point gave $105.5 \pm 0.5^\circ\text{C}$ and $152.5 \pm 0.5^\circ\text{C}$ for LDPE and PP, respectively. Hence no detectable shift in melting point was observed for all the blends investigated, suggesting further that there was no interaction between the blends. Ke²⁹ has shown that addition of diluent sup-

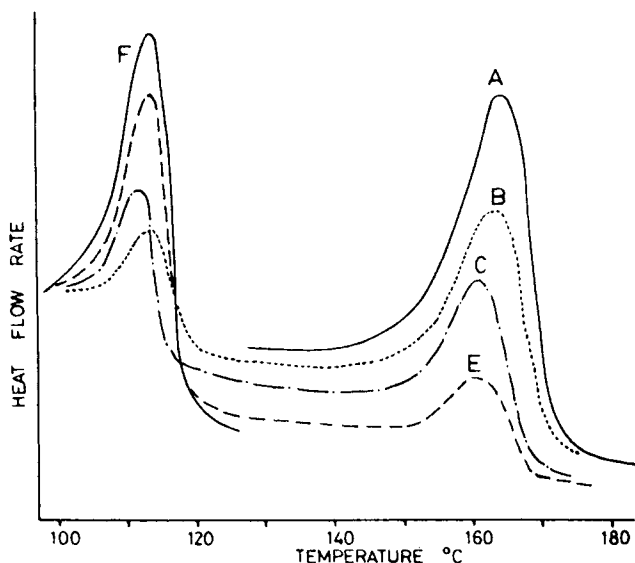


Fig. 6. Thermogram scan of homopolymers and polyblends of PE and PP: (A) PP; (B) 80 PP/20 PE; (C) 60 PP/40 PE; (E) 20 PP/80 PE; (F) PE.

pressed the melting point of PE with increasing diluent concentrations because of an interaction between the diluent and PE. Hence a zero shift in melting points indicates complete incompatibility between LDPE and PP. Ke³⁰ also found that an ethylene-propylene block copolymer gave two crystallization peaks but a physical blend of high-density polyethylene and isotactic polypropylene yielded only a single peak. The present findings gave two distinct peaks for LDPE and PP.

Wide-Angle X-Ray Diffraction

X-ray diffraction has been employed for the identification of polymers. In the case of copolymers, besides superimposing patterns of homopolymers, a new diffraction pattern, distinct from the homopolymer of either component, may appear. Copolymerization may also result in the destruction of crystallinity and hence alter the diffraction pattern to just amorphous halos.³¹ In the case of polyblends, if interaction is absent, we should observe superimposed patterns of the homopolymer. On the other hand, if interaction occurs, a new pattern may occur or the crystallized pattern may be destroyed or reduced in intensity.

The X-ray diffraction patterns of isotropic samples of LDPE, PP, and polyblends are shown in Figure 7. The corresponding interplanar spacing d_{hkl} determined experimentally and the calculated d_{hkl} value based on published crystallographic data¹⁶ are given in Table II.

If we compare the interplanar spacing of PE (110), α -PP (111), and β -PP (301), these planes will give rise to diffraction rings superimposed upon and indistinguishable from one another. Similarly, for PE (200) and α -PP (131) the rings will superimpose whereas for α -PP (040) and β -PP (300) the rings are quite close to each other. For a noninteracting blend, the ratio of the components should

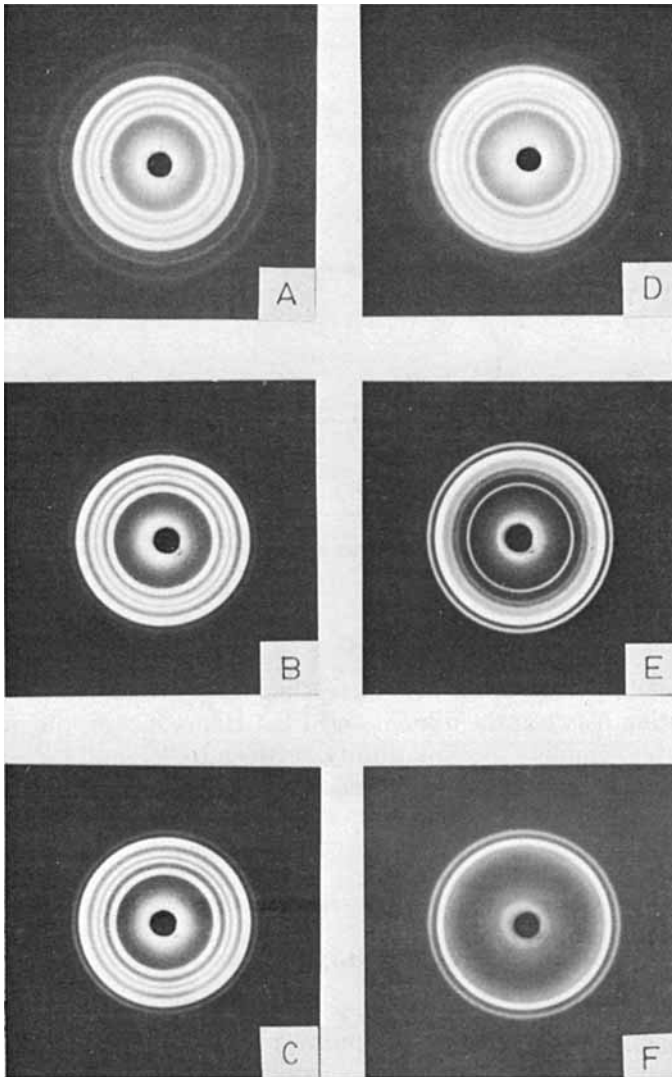


Fig. 7. Wide-angle X-ray diffraction pattern of: (A) polypropylene; (B) 80 PP/20 PE; (C) 60 PP/40 PE; (D) 40 PP/60 PE; (E) 20 PP/80 PE; (F) polyethylene.

be proportional to the intensities of the crystalline diffraction patterns of the components and could be used for assessing the ratio of the components. For the PE-PP blend, however, accurate assessment is not possible. If we study Figure 7, the intensity due to PP diffraction is observed to decrease from A to E, while that of PE is seen to increase from B to F, giving us a rough picture of the composition of the blend. There is no detectable change in the interplanar spacing and no new diffraction pattern, further suggesting that there is no interaction between PE and PP.

PP has been found to crystallize in two major types of spherulites,^{32,33} a monoclinic crystal modification whose melting point was 168°C, and a hexagonal modification whose growth was enhanced by rapid cooling rates and whose

TABLE II
Diffraction Data of PE and it-PP

Polymer	Crystal system	Interplanar spacing Å		Indices <i>hkl</i>	Observed intensity
		Exptl	Calcd		
PE	Orthorhombic	4.10	4.114	(110)	Strong
	<i>a</i> = 7.417 Å	3.70	3.709	(200)	Strong
	<i>b</i> = 4.945 Å	3.00	2.967	(210)	Weak
	<i>c</i> = 2.547 Å	2.46	2.472	(300)	Weak
		2.24	2.211	(310)	Weak
it-PP	α - monoclinic	6.24	6.262	(110)	Strong
	<i>a</i> = 6.65 Å	5.23	5.240	(040)	Medium
	<i>b</i> = 20.96 Å	4.76	4.783	(130)	Medium
	<i>c</i> = 6.50 Å	4.10	4.170	(111)	Strong
	β = 99° 20'	3.52	3.634	(131)	Weak
		3.10	3.131	(220)	Weak
		2.67	2.656	(221)	Weak
	2.10	2.085	(222)	Weak	
it-PP	β - hexagonal	—	5.508	(300)	—
	<i>a</i> = 19.08 Å	—	4.199	(301)	—
	<i>c</i> = 6.49 Å				

melting point was 150°C. When rapidly quenched PP was annealed at 150–168°C, the hexagonal β -form was converted to the monoclinic α -form. Oriented β -form PP has been obtained by crystallizing the polymer melt in a temperature gradient.^{34,35} For extruded PP, it was found that the interior unoriented portion which experienced slow cooling was composed of virtually the α -form while the exterior surface layer which experienced high shear as well as rapid cooling was made up predominantly of the β -form.³⁶ From the X-ray diffraction patterns of Figure 7, β -PP (300) diffraction ring was not observed, hence indicating the absence of β -form PP in the compression-molded sheet.

Morphology

Figures 8(A) and 8(B) show the morphology of pure LDPE and pure PP respectively. The size of the spherulites of the homopolymer PP was much larger than that of the spherulites of LDPE. Figure 8(C) shows a typical micrograph of the LDPE-PP blend. A sea of small spherulites similar in size to those of LDPE, with isolated islands of large and distinct spherulites was observed. The distribution as well as the concentration of the large spherulites were random for all blend composition. In order to identify the nature of the large spherulites, a sample was heated in a hot-stage microscope at the rate of 10°C·min⁻¹. At 140°C, the large spherulites were observed to start melting. At 155°C, complete melting of the large spherulites was observed. At 166°C, all traces of spherulites had disappeared. The above observation indicated that the large distinct spherulites observed in the blends were β -spherulites similar to those reported by Padden and Keith.³²

In order to show that the formation of these large β -spherulites was due to the presence of LDPE, a small sample of pure LDPE was melted side by side with a small sample of PP and the molten polymers were allowed to flow over one another to achieve mixing at the boundary. Figure 9(A) depicts the boundary

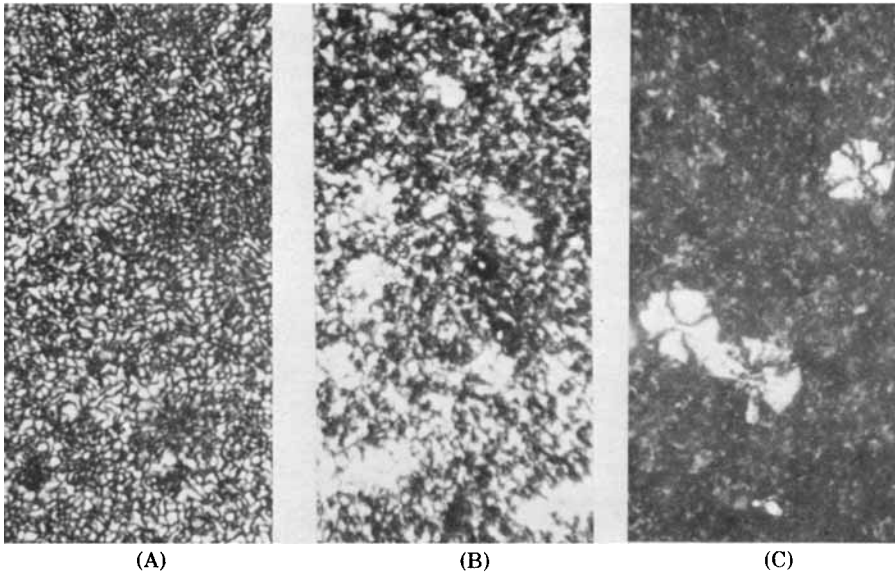


Fig. 8. Spherulitic morphology of: (A) LDPE; (B) PP; (C) 20 PP/80 PE blend ($\times 200$).

morphology. The molten LDPE being the less viscous had flown into the molten PP. At the boundary, on the side of the PP, large distinct bright spherulites were observed. Small LDPE spherulites were also seen to overlap with the larger PP spherulites. Far away from the boundary into the PP region, no large distinct bright spherulite was observed. When the sample was heated at $10^{\circ}\text{C}\cdot\text{min}^{-1}$, at 113°C , the LDPE spherulites melted, and the region originally occupied by the pure LDPE showed complete darkness, as indicated in Figure 9(B). The PP spherulites could now be observed more clearly without the interference of the small LDPE spherulites. At 152°C , the large distinct bright spherulites had melted partially, and become dark as indicated in Figure 9(C). At 165°C , the other PP spherulites were observed to start melting as indicated in Figure 9(D). The above observation showed that the large distinct bright spherulites which

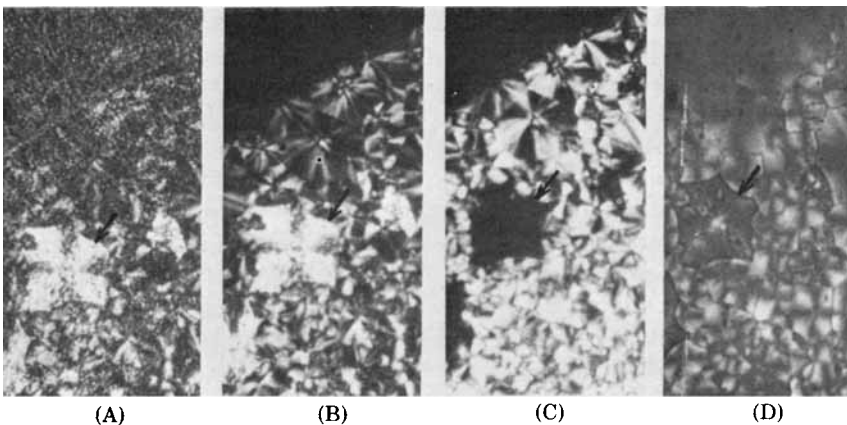


Fig. 9. Boundary morphology of LDPE-PP mix: (A) as crystallized at room temperature; (B) at 113°C ; (C) at 152°C ; (D) at 165°C . Arrow indicates the large distinct bright spherulite identified as β -spherulite. ($\times 1000$.)

were formed only at the LDPE and PP boundary were the β -form of PP spherulites.

Hence the addition of LDPE seemed to have two effects on the morphology of PP. Firstly, the LDPE acted as nucleating centers and hence increased the number of nuclei for PP and reduced the spherulite size. Secondly, the presence of LDPE enhanced the formation of large β -form PP spherulites to several times the size of the other spherulites of LDPE and PP.

However, β -spherulites were not detected in the WAXD and DSC work. This was probably because the samples used for WAXD and DSC were bulk crystallized whereas the observation for the hot-stage microscope was made on the re-crystallized thin film where the kinetic of crystallization may be different.

Reduction in spherulite size was observed by Lovinger and Williams¹² for HDPE-PP blends on microtomed sections of the bulk sample. In the case of PP and polybutene-1 blend, Siegmann³⁷ also made the same observation, where the thin film crystallized blends gave a mixture of small crystalline aggregates. When there is a large difference in the melting points of the two homopolymers, crystallization of the higher melting polymer will occur first. During crystallization, the lower melting polymer may be segregated, remaining as a molten form. This rejected material may have an effect on the nucleation rate, rate of crystallization, and/or morphological structure of the higher melting polymer. In the case of LDPE-PP blend, the effect of the lower melting polymer (LDPE) on the crystallization behavior of the higher melting polymer (PP) was reflected in a reduction of the spherulite size and the enhancement on the formation of β -spherulites.

CONCLUSIONS

Interest in polyblends or polymer alloys has been increasing recently. The important factor deciding whether properties are enhanced through blending is the miscibility and interaction between the two components. This consideration determines the resultant structure of the blend and hence control their properties.

Compression melt blending of LDPE and PP gave a blend consisting of a mixture of spherulites of LDPE and PP. The crystallographic structure of LDPE and PP as determined by WAXD remained unchanged. However, the spherulitic morphology of the blends differed from that of the homopolymer, where the nucleation density of α -form PP increased, resulting in smaller spherulites; and the rate of growth of β -form PP was enhanced, resulting in larger and distinct spherulites. Crystallinity estimation of density determination, WAXD, and DSC indicated each individual component retaining its own crystallinity where the overall crystallinity of the blends was given by the sum of the crystallinity of the individual components. Lack of interaction was substantiated by the melting points from DSC measurements where no shift in melting points was observed. The mechanical properties obtained showed the behavior as expected from the above findings of the structure. The blends behaved like a simple composite obeying the rule of mixture. However, departure away from the ideal behavior was observed. This departure was probably due to the modification in spherulitic morphology of the blends.

Blending of PE and PP gives properties in between those of the homopolymers.

The main shortcoming of the blends is the drastic reduction in elongation at break. Hence, for successful application of PE-PP blends, improvement in elongation at break is important. This improvement may be achieved by introducing a suitable compatibilizer.

The author thanks Janizar Nonchik for his assistance in preparation of figures and photographs.

References

1. J. A. Manson and L. H. Sperling, *Polymer Blends and Composites*, Plenum, New York, 1976.
2. M. O. W. Richardson, *Polymer Engineering Composites*, Applied Science, 1977.
3. C. B. Bucknall, *Toughened Plastics*, Applied Science, 1977.
4. D. R. Paul and S. Newman, *Polymer Blends*, Academic, New York, 1978, Vols. 1 and 2.
5. S. L. Cooper and G. M. Estes, *Multiphase Polymers*, American Chemical Society, Washington, D.C., 1979.
6. D. R. Paul and J. W. Barlow, *J. Macromol. Sci., Rev. Macromol. Chem.*, **C18**, 109 (1980).
7. R. Koningsveld, H. A. C. Chermin, and M. Gordon, *Proc. Roy. Soc. London A*, **319**, 331 (1970).
8. D. C. Allport and W. H. Janes, *Block Copolymers*, Wiley-Interscience, New York, 1964.
9. K. Yamaguchi, T. Yagi, S. Machi, and M. Takehisa, *J. Appl. Polym. Sci.*, **19**, 1959 (1975).
10. O. F. Noel and J. F. Carley, *Polym. Eng. Sci.*, **15**, 117 (1975).
11. R. D. Deanin and M. F. Sansone, *Polym. Prepr.*, **19**, 211 (1978).
12. A. J. Lovinger and M. L. Williams, *J. Appl. Polym. Sci.*, **25**, 1703 (1980).
13. P. Robson, G. J. Sandilands, and J. R. White, *J. Appl. Polym. Sci.*, **26**, 3515 (1981).
14. W. J. Ho and R. Salovey, *Polym. Eng. Sci.*, **21**, 839 (1981).
15. M. R. Shishesaz and A. A. Donatelli, *Polym. Eng. Sci.*, **21**, 869 (1981).
16. H. Tadokoro, *Structure of Crystalline Polymers*, Wiley, New York, 1979.
17. N. H. Ladizesky and I. M. Ward, *J. Macromol. Sci., Phys.*, **B5**, 745 (1971).
18. L. E. Nielsen, *J. Appl. Phys.*, **41**, 4626 (1970).
19. J. C. Halpin, *J. Composite Mater.*, **3**, 732 (1969).
20. E. H. Kerner, *Proc. Phys. Soc.*, **B69**, 808 (1956).
21. L. E. Nielsen, *Mechanical Properties of Polymers and Composites*, Marcel Dekker, New York, 1974.
22. L. W. Kleiner, F. E. Karasz, and W. J. MacKnight, *Polym. Eng. Sci.*, **19**, 519 (1979).
23. L. W. Kleiner, Ph.D. dissertation, University of Massachusetts, 1978.
24. J. Patel and P. J. Phillips, *J. Polym. Sci., Polym. Lett. Ed.*, **11**, 771 (1973).
25. J. C. Halpin and J. L. Kardos, *J. Appl. Phys.*, **43**, 2235 (1972).
26. R. M. Ogorkiewicz, *Engineering Properties of Thermoplastics*, Wiley-Interscience, New York, 1970.
27. S. Bhateja, J. Rieke, and E. Andrews, *Ind. Eng. Chem. Prod. Res. Dev.*, **19**, 607 (1980).
28. M. J. Richardson, *Dev. Polym. Charact.*, **1**, 205 (1978).
29. B. Ke, *J. Polym. Sci.*, **50**, 79 (1961).
30. B. Ke, *J. Polym. Sci.*, **61**, 47 (1962).
31. M. Kakudo and N. Kasai, *X-Ray Diffraction by Polymers*, Elsevier, New York, 1972.
32. F. J. Padden and H. D. Keith, *J. Appl. Phys.*, **30**, 1479 (1959).
33. H. D. Keith, F. J. Padden, N. M. Walter, and H. W. Wyckoff, *J. Appl. Phys.*, **30**, 1485 (1959).
34. A. J. Lovinger, J. O. Chua, and C. C. Gryte, *J. Polym. Sci., Polym. Phys. Ed.*, **15**, 641 (1977).
35. T. Asano and Y. Fujiwara, *Polymer*, **19**, 99 (1978).
36. H. Dragaun, H. Hubery, and H. Muschik, *J. Polym. Sci., Polym. Phys. Ed.*, **15**, 1779 (1977).
37. A. Siegmann, *J. Appl. Polym. Sci.*, **24**, 2333 (1979).

Received September 10, 1981

Accepted August 23, 1982