Effect of Flexibility of Ethylene Vinyl Acetate and Crystallization of Polypropylene on the Mechanical Properties of i-PP/EVA Blends

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ABSTRACT: Polymer blend technology has been widely used for the past several years for the modification or enhancement of mechanical properties of polymers to obtain an overall balance of properties over those of the constituents. Despite its interesting mechanical and thermal properties, the impact strength of polypropylene leaves wide scope for improvement. A series of blends of ethylene vinyl acetate (EVA) copolymer with an impact grade of isotactic polypropylene (i-PP) were prepared by single screw extrusion at 0–0.32 volume fraction of the dispersed phase. The mechanical properties such as tensile behavior, elongation-at-break, and impact strength of these blends systems as well as crystallinity were evaluated. Crystallinity data have been used in greater depth to support the mechanical properties. Differential scanning calorimetry studies conducted to study the modification in crystallinity of the crystalline component, i-PP, of the blend revealed that the rubber component of the blend enhanced the crystallinity of i-PP phase by providing sites for nucleation. Tensile modulus and strength decreased while the impact strength and breaking elongation enhanced with blending elastomer concentration. The improved properties of these PP/EVA blends are encouraging for carrying out further work on this system (composites) and suggest potential high impact strength applications for PP. © 2011 Wiley Periodicals, Inc. J Appl Polym Sci 123: 1905–1912, 2012

Key words: polypropylene; ethylene vinyl acetate; mechanical properties; impact strength; crystallinity

INTRODUCTION

Blending of polymers is an economically viable and technologically potential method that has been increasingly utilized for tailor making new materials with a specific balanced combination of physical properties.[1–5] A unique spectrum of material properties can be achieved through blending by a convenient selection of the components. The performance parameters most frequently targeted for improvement via blending are the mechanical properties, impact strength, processability, heat deflection temperature, and cost : performance ratio. Elastomeric rubber–plastic blends have been technologically interesting for use as thermoplastic elastomers.[6,7] They not only exhibit some of the physical properties of the conventional elastomers at the service temperatures but also at the same time are processable at elevated temperatures.

Isotactic polypropylene (i-PP) is the lightest major thermoplastic with desirable properties, such as high melting temperature, low density, and high chemical inertness, which make it a versatile material, particularly suitable for structural and long-life applications.[8,9] In addition, its cost places it in an advantageous position in comparison to most of the other plastic materials. Although i-PP is extensively used in many fields of applications owing to its interesting mechanical and thermal properties, its insufficient low temperature impact strength, particularly its notched toughness, hinders its wider utilization. Even at room temperature, the impact strength of some grades of i-PP leaves wide scope for improvement. Hence, enhancement of impact strength of i-PP has become the subject of interest and research. The melt blending of i-PP with different elastomers and rubbers such as butyl rubber, ethylene–propylene copolymers, ethylene–propylene–diene terpolymers, polyisobutylene, and styrene–butadiene block copolymers has been carried out for improving the impact strength of PP especially at low temperatures.[3–7]

Earlier studies on PP/EVA blends have proved EVA as an impact modifier for i-PP.[10,11] While linear viscoelastic and transient behaviors were studied by Tyagi and Ghosh,[12] the morphological, mechanical, and rheological properties of these blends using homopolymer polypropylene were reported by several researchers in separate studies.[13–16]

In this article, we report the role of the thermoplastic elastomer EVA in modifying the mechanical

properties of an impact grade of i-PP. This grade of polypropylene is one in which, EPR, the known impact modifier for PP has been incorporated during its polymerization process and thus becomes an integral part of PP. A comprehensive study on all the mechanical properties of the blend of EVA elastomer with an impact modified grade of polypropylene has been carried out and thoroughly correlated with the crystallinity data. Tensile properties such as the modulus, strength, and elongation-at-break and impact strength have been evaluated as functions of the blend composition. The tensile data have been analyzed following simple predictive models to understand phase interactions. The role played by the crystallinity of the impact modified polypropylene on the mechanical properties of these blends has been extensively evaluated and analyzed unlike other works.

EXPERIMENTAL

Materials

Injection moldable, medium impact grade of polypropylene (hitherto i-PP) with 18.5% EPR content (REPOL B030 mg, MFI-2.84 g/10 min at 230°C, 21.6 N load) was obtained from Reliance Industries, Mumbai, India.[17] Ethylene vinyl acetate (EVA) copolymer with vinyl acetate content of 18% (PILENE 1802, MFI = 2.0 g/10 min at 150°C, 21.6 N load) manufactured by National Organic Chemicals India was used for the study.[18]

Preparation of blends

Binary blends of i-PP with varying concentrations of EVA (0–33 vol %) were prepared by melt mixing on a Windsor twin screw extruder (KTS 140e, L/D ratio 25 : 1). Before use, EVA was dried in a vacuum oven at 65°C for 24 h to ensure the removal of moisture. The granules of i-PP and EVA were then tumble mixed in appropriate ratios and extruded at a temperature profile of 175/190/200/200°C in the feed zone, compression zone, metering zone, and the die zone, respectively. The screw speed was adjusted to 30 rev min^{−1}. The extruded strands were immediately quenched in a water bath at room temperature and chopped into granules, which were dried at 70°C for 4 h in vacuum before molding. The i-PP granules were also extruded under the same conditions to ensure identical thermal and shear history.

Molding of test specimens

Test specimens for the evaluation of mechanical properties were molded on a R.H. Windsor (Model SP-1) screw-type injection-molding machine using screw speed of 80 rev min^{−1} and temperatures of 190

and 200°C in the feed and die zones, respectively. Injection pressure and mold-locking pressures were 58.8 MPa and 1.9 GPa, respectively.

Measurements

Tensile properties

Tensile properties were measured on an Instron Universal Tester (Model 1121) using injection-molded dumb-bell-shaped test specimens with a gauge length of 50 mm according to ASTM D 638 (type I) test procedures[19] at a cross-head speed of 50 mm min^{−1}. Izod impact strengths of the notched specimens were measured on a pendulum hammer-type impact tester (FIE instrument Model IT-0.42). A notch of 2.5 mm depth with an angle of 45° was made on the impact specimens conforming to ASTM D-256 specifications.[19] At least five samples were tested at each blend composition, and their average values are reported. All the tests were performed at ambient temperature 30 ± 2 °C.

Crystallization behavior

Crystallization behavior of i-PP and that of the blends was studied by differential scanning calorimetry (DSC) on a Dupont 2100 Thermal Analyzer having a 910 DSC module, in an atmosphere of liquid nitrogen.[20] The samples obtained by cutting small chips from the injection-molded samples were dried in a vacuum oven before starting the experiment. Thermograms were recorded during the heating cycle at a constant heating rate of 10 °C min^{−1}, using identical settings of the instrument for all the samples. The heat of fusion values were used to calculate the percent crystallinity of i-PP in the different blend compositions. The enthalpy for 100% crystalline i-PP used was 209 J g^{−1}.[21]

SEM studies

Scanning electron microscopy (SEM) studies of the cryofractured surfaces of the tensile specimens were carried out on a Cambridge Stereoscan (Model S4-10), to determine the phase morphology of the blends. The fractured surfaces were etched in toluene for 12 h at 85°C to dissolve out the EVA phase. The surfaces were sputter coated with silver before scanning.

RESULTS AND DISCUSSION

Degree of crystallinity

In a semicrystalline polymer matrix, the crystallinity decreases on incorporation of elastomers/additives, the decrease in crystallinity being proportional to the increase in the concentration of the inclusions.[22,23] This has been attributed to the interference

TABLE I
Crystallinity (%) of i-PP/EVA Blends (from DSC)

Φ_d (%)	T_m (°C)	T_{onset} (°C)	Enthalpy of fusion, ΔH_f (J g^{-1})	Normalized enthalpy	% X_c
0	164.92	153.39	56.16	56.16	26.9
1.9	164.40	149.99	60.24	61.44	29.5
4.6	162.73	151.39	61.39	64.44	30.8
8.8	163.69	151.86	60.19	66.21	31.7
16.1	164.29	150.02	54.22	65.06	31.1
22.4	163.37	151.56	55.10	71.63	34.3
32.4	161.82	150.14	58.40	87.59	41.9

of the dispersed phase to the matrix's polymer chain mobility through imposition of physical barrier. However, incorporation of EVA copolymer into i-PP enhanced the crystallinity of the latter, and the crystallinity increases with the concentration of EVA copolymer as shown by the DSC measurements given in Table I. This may be attributed to the nucleating ability of EVA copolymer which provides a surface for the crystallization of i-PP.[24] The crystallinity of the matrix i-PP plays an important role on the properties of the i-PP/EVA blend system, which will be presented in the subsequent sections.

Tensile properties

The tensile properties, that is, tensile strength and modulus, and elongation-at-break of the blends were estimated from the tensile stress–strain curves (not shown). These data are then presented in Figures 1–7 as plots of variations in the ratios of the properties (subscript b) to that of the matrix i-PP(subscript p) against the volume fraction, Φ_d, of the blending polymer EVA.

Tensile modulus

The plot of the relative tensile modulus, E_b/E_p, of the i-PP/EVA copolymer blends is shown in Figure
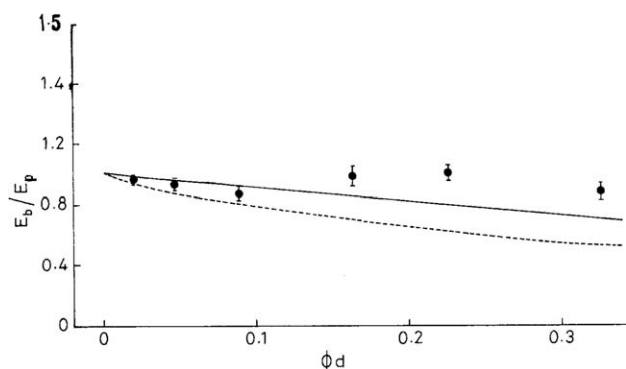
Figure 2 Dependence of normalized relative moduli, $[(E_b/X_{c(b)})/E_p/X_{c(p)}]$ of i-PP/EVA blends (●) on Φ_d.

1 as functions of Φ_d. The parameter decreased marginally at first upto $\Phi_d = 0.09$, whereas, with further increase in Φ_d, the modulus enhanced, inappreciably, the value varied between 0.89 and 1.05 depending on Φ_d. It implied that the EVA copolymer marginally softened the i-PP upto EVA content of 9 vol %, while at $\Phi_d > 0.09$, the modulus of i-PP is unaffected by the presence of EVA polymer. To understand the blend structure, the moduli data were analyzed and compared with predictive models according to "rule of mixture"[25,26] as in polymer blends and composites, eq. (1), and the "foam model" of Cohen and Ishai[27] eq. (2):

$$\frac{E_b}{E_p} = \left(\frac{E_d}{E_p} - 1\right)\phi_{d+1} \tag{1}$$

$$\frac{E_b}{E_p} = 1 - \phi_d^{2/3} \tag{2}$$

The tensile moduli data for the i-PP ($E_p = 324$ MPa), the EVA copolymer (E_d), and the blends (E_b) were determined from the initial slopes of the stress–strain curves.

The relative moduli data were higher than the "foam model," Figure 1. Initially, upto $\Phi_d = 0.08$, the moduli values were marginally higher, whereas,

Figure 1 Relative tensile modulus, E_b/E_p, of i-PP/EVA copolymer blends(●) and predicted behavior according to Rule of Mixtures (___) eq. (1) and Foam Model (- - -) eq. (2) versus Φ_d.
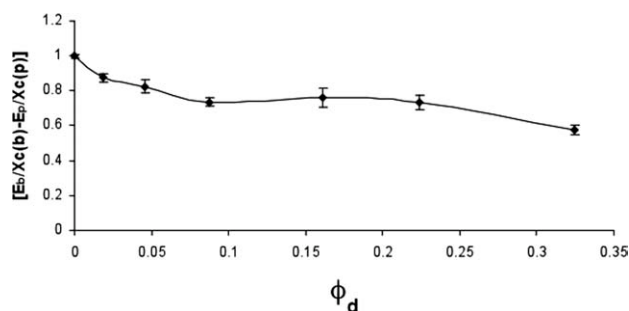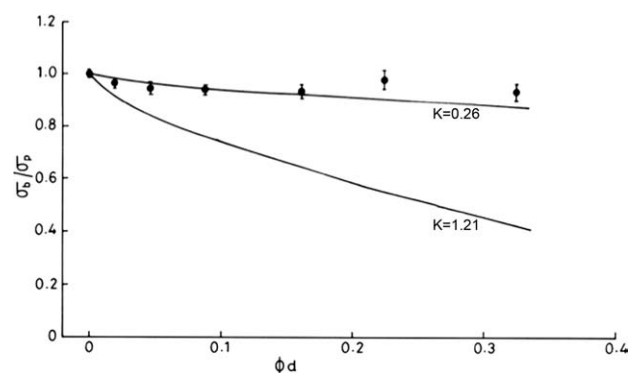
Figure 3 Relative tensile stress, σ_b/σ_p, of i-PP/EVA copolymer blends (●) and predicted behavior according to Nicolais-Narkis model (___) eq. (3) with k values as indicated, against Φ_d.
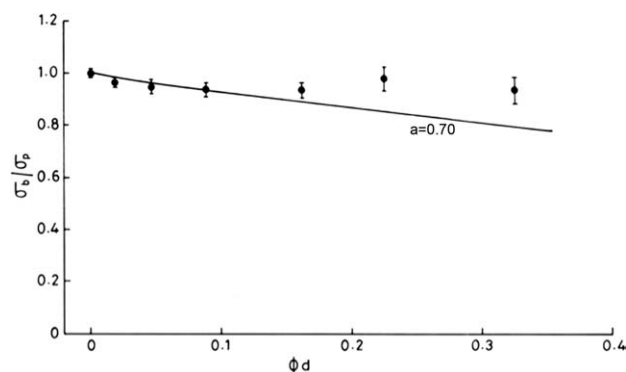
Figure 4 Plot of relative tensile stress, σ_b/σ_p, i-PP/EVA copolymer blends (●) and predicted behavior according to Porosity Model [eq. (4)] with $a = 0.70$(___), against Φ_d.

beyond $\Phi_d = 0.08$, the values were much higher. This indicates that the EVA copolymer was not exactly equivalent to a void or a pore, rather it may be effective, through any kind of interaction, or by any other means, as a stiffening blending phase.

The "rule of mixture" curve exhibited marginally higher values upto $\Phi_d = 0.08$ and at $\Phi_d > 0.08$, the moduli data were higher, and closer to unity, Figure 1. This implies that the blending polymer, EVA, although dilutes the matrix initially; the tensile modulus has not decreased significantly, contrary to the expected decrease on account of rubber toughening. It has been observed in other works that EVA copolymer provides a surface for crystallization of i-PP.[10] The crystallinity data also show an overall increase with Φ_d, Table I. Thus, the EVA copolymer may soften the i-PP matrix initially upto $\Phi_d = 0.08$; however, on further increase in Φ_d, it resists the fall in the modulus by assisting in the crystallization of i-PP so that the relative moduli remain close to unity, Figure 1. In other systems where the blending polymers are elastomeric, the decrease in modulus is to much higher extents.[27–29]

The EVA copolymer enhanced the crystallinity of the matrix i-PP, also, it is a softer copolymer than i-PP. These effects required to be considered to deter-
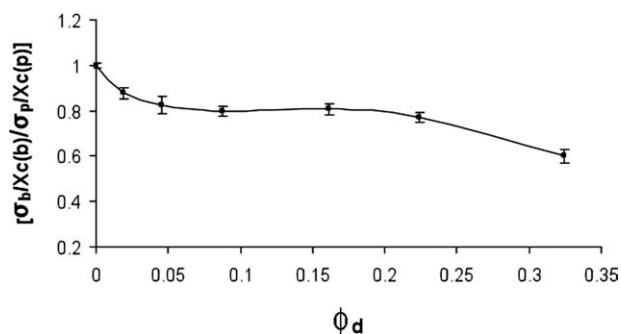
mine the properties of these blends. The tensile moduli were, therefore, normalized by dividing the properties by the crystallinity of i-PP in the blend and in the matrix, respectively, to eliminate the effect of crystallinity in these systems. Figure 2 shows the variation of normalized relative moduli, $[(E_b/X_{c(b)})/E_p/X_{c(p)}]$, vs. Φ_d. The value decreased upto $\Phi_d = 8\%$ and remained almost unaltered upto $\Phi_d = 23\%$ finally falling to ~ 0.6 at $\Phi_d = 33\%$. Thus, the data show nonspecific phase interaction of i-PP with EVA copolymer, the latter effective only as a softening discontinuous phase.

Tensile strength

The relative tensile strength of the i-PP/EVA blends (ratio of tensile strength of i-PP/EVA blend to that of i-PP, σ_b/σ_p) is shown as functions of Φ_d in Figures 3 and 4. Addition of EVA copolymer decreased the tensile strength of i-PP marginally, the data varied between 0.96 and 0.98 depending on Φ_d. This indicated that the blending polymer weakens the i-PP matrix polymer structure, the weakening may be due to a decrease in the effective load-bearing cross-sectional area of the matrix, similar to other elastomer containing polymer blends.[28,29]

The tensile strength data decreased despite increase in the crystallinity of i-PP in the presence of

Figure 6 Plot of relative elongation-at-break, $\varepsilon_b/\varepsilon_p$ of i-PP/EVA copolymer blends(x) and predicted behavior according to Mitsuishi Model [eq. (5)] (___), as a function of Φ_d.

Figure 5 Dependence of normalized relative tensile strength, $[(\sigma_b/X_{c(b)})/(\sigma_p/X_{c(p)})]$ of i-PP/EVA copolymer blends (●) on Φ_d.
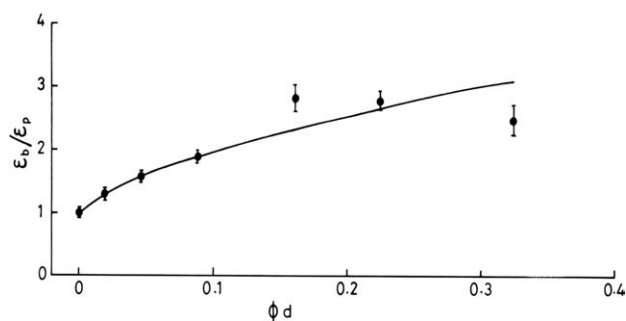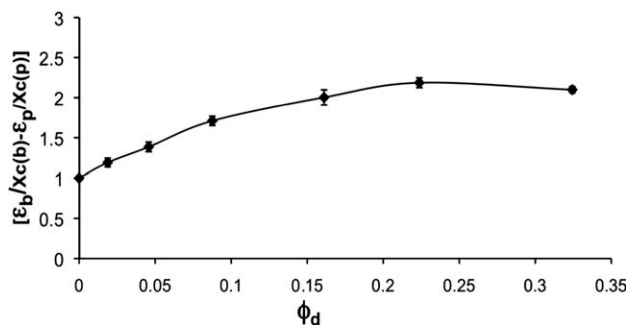

Figure 7 Variations of normalized elongation-at-break, $[(\varepsilon_b/X_{c(b)})/(\varepsilon_p/X_{c(p)})]$ of i-PP/EVA copolymer blends (●) against Φ_d.

EVA copolymer, Table I. The analysis of these tensile strength data was carried out using some predictive theories to estimate the discontinuity/weakness in the blend structure introduced by the blending polymer EVA:

$$\frac{\sigma_b}{\sigma_p} = 1 - k\phi_d^{2/3} \qquad (3)$$

$$\frac{\sigma_b}{\sigma_p} = \exp(-a\phi_d) \qquad (4)$$

These models have also been used in other two-phase blend/composite systems.[30,31] Here, σ_b and σ_p represent the tensile strength of the blend and the i-PP matrix, respectively. These equations are based on the relationship of area fraction or the volume fraction of the discontinuous phase[32,33] in no-adhesion types of blend structure. Realistic features, such as stress concentrations at the narrow portions of the matrix at the inclusion-matrix interphase, were incorporated in the two-thirds power law by Nicolais–Narkis, eq. (3), where area fraction of the discontinuous phase is considered effective.[34,35] The parameter, k, is a weightage factor, is also denoted as the phase interaction constant, which describes the adhesion quality between the two phases. The value of k depends on the blend structure. Dense hexagonal packing of the dispersed phase in the plane of highest density is described by a value of $K = 1.1$. In the extreme case of poor adhesion of the matrix with the spherical inclusions, $K = 1.21$ when minimum cross section is generated between adjacent spherical particles.[33–36] Values of K less than 1.21 indicate adhesion between phases, the lower the value of k the better the adhesion.[37] While $K = 1$ stands for no stress concentration,[38] $k = 0$ represents the upper bound for the phase adhesion so that the strength of the matrix polymer does not decrease.

According to the porosity model, eq. (4), the dispersed phase appears to be analogous to pores/voids in the metals and ceramics[39] and polymer blends/composites.[40] The pores/voids are assumed not to influence the mechanical properties of the two-phase systems owing to non-adhesion in the phase boundary. The parameter a is a measure of the stress concentration (weakness) in the structure.[33] Its relationship to the stress concentration effect is suggested from the similarity of the effects of tensile deformation and stress concentration on its value. The higher the value of a the higher is the extent of stress concentration in the structure.[33]

Table II presents the values of the parameters K and a from eqs. (3) and (4), respectively, obtained by comparing the models with the experimental tensile stress data at each individual blend composition. According to eq. (3), the Nicolais–Narkis model, the

TABLE II
Values of Adhesion Parameter K [eq. (3)], and Stress Concentration Parameter a [eq. (4)] in i-PP/EVA Blends

Φ_d (%)	K	a
0	–	–
1.88	0.44	1.66
4.58	0.48	1.17
8.76	0.29	0.68
16.11	0.22	0.42
22.37	0.05	0.09
32.44	0.14	0.21
Mean value	0.26	0.70

values of K are less than unity with an average of $K = 0.26$, which indicates a high degree of phase interaction and a lesser extent of defects in the blend structure. Similar results were observed by Gupta and Purwar[31,41] and Maiti and Das.[26] The porosity model, eq. (4), showed a significant weakness in the structure, the average value of the stress concentration parameter, a, being 0.70 (Table II). This value is, however, much lesser than the value 1.35 of Ref. 18. This also points toward a good extent of phase interaction in the i-PP/EVA polymer blends.

Comparison of the tensile stress data with the Nicolais–Narkis model, eq. (3) and the porosity model, eq. (4), are shown in Figures 3 and 4, respectively. With the Nicolais–Narkis model and with the average value of $K = 0.26$, the data showed good agreement upto $\Phi_d = 0.16$, Figure 3. However, at $\Phi_d > 0.16$, the data showed some scattering, the deviations are all on the positive side. Nevertheless, the individual data points imply a significant extent of phase interaction. The data also showed quite good agreement with the porosity model, eq. (4), with an average $a = 0.70$, Figure 4, upto $\Phi_d = 0.16$ while the data at $\Phi_d > 0.16$ were higher than the model. This also implies an extent of phase interaction.

It may be mentioned that the polymer polypropylene is an olefin polymer with very low degree of interchain attraction forces. The mechanical properties of such a polymer are derived from its crystallinity.[8,40] Hindrance to fitting of the chains into a crystal structure by the dispersed phase may decrease its strength properties. However, the crystallinity of i-PP in the i-PP/EVA blends was found to increase with Φ_d Table I. Therefore, to clearly understand the effects of crystallinity and phase interaction, if any, the normalized relative tensile strength, $[(\sigma_b/X_{c(b)})/(\sigma_p/X_{c(p)})]$, was plotted vs. Φ_d, Figure 5. Here, $X_{c(b)}$ denotes crystallinity of i-PP in the blend while $X_{c(p)}$ denotes that of i-PP in the matrix polymer. The data remained less than unity varying from 0.8 to 0.5 as Φ_d enhanced from 0.02 to 0.32. This plot seemingly negates the possibility of
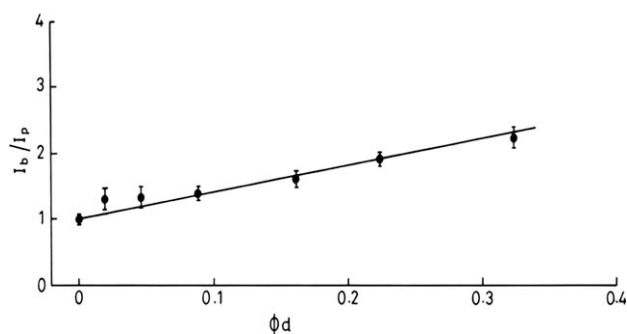
Figure 8 Plot of relative impact strength (I_b/I_p) of i-PP/EVA copolymer blends as a function of Φ_d.

any significant interphase adhesion between the elastomer particles and the polymeric matrix. It may be concluded that the EVA copolymer, although provides a surface facilitating the crystallization of i-PP, is a softer elastomeric polymer which offers no phase adhesion. The polymer's flexibility overrides the enhancement of crystallinity of i-PP so that the tensile strength of the blend eventually decreases in the presence of the blending EVA copolymer.

Elongation-at-break

The relative elongation-at-break, $\varepsilon_{b}/\varepsilon_{p}$, of the i-PP/EVA blends showed an overall increase with ϕd, indicating matrix softening by the rubbery dispersed phase, Figure 6. The increase in the relative elongation-at-break of the blend is quite significant (<180%) up to $\phi d = 0.16$, compared to the impact i-PP ($\Phi_d = 0$), whereas with further increase in Φ_d the value shows a slight decrease. Moduli data also showed matrix softening in the presence of the blending polymer EVA.

The elongation data were compared with a modified Mitsuishi model:[42]

$$\frac{\varepsilon_b}{\varepsilon_p} = (1 + K\phi_d^{2/3}) \qquad (5)$$

where ε_b is the elongation-at-break of the blends, ε_p the value of the matrix polymer, and K is the flexibility coefficient. The data showed a close agreement with the model upto $\Phi_d = 0.08$ and at $\Phi_d > 0.08$ the experimental data points were uniformly scattered around a plot with a mean K value of 4.49.

It may be mentioned here that the crystallinity of the matrix i-PP enhanced with Φ_d which should have raised its rigidity. Instead, there was enhancement in the ductility which may be due to the overriding effect of the flexibility of the elastomer phase compared to its function as a crystallinity facilitator. This matrix softening indicates further toughening of the impact i-PP, which will absorb additional energy before impact failure, similar to other systems.[26,43,44]

To evaluate the role of the elastomeric blending polymer EVA on flexibility of impact i-PP the effect of crystallinity by EVA was eliminated by plotting the relative normalized elongation-at-break data, $[(\varepsilon_b/X_{c(b)})/(\varepsilon_p/X_{c(p)})]$, vs. Φ_d, Figure 7. The value enhanced upto $\Phi_d = 0.16$, showing a maximum value of 2.4 here. The data decrease with further increase in Φ_d remaining, however, higher than unity. The variation of the data strengthens the assumptions of the modified Mitsuishi et al. model[42] and clearly demonstrates that the EVA copolymer makes the impact i-PP more ductile implying an impact toughening.

Impact behavior

Figure 8 exhibits variation of the relative Izod impact strength (I_b/I_p) of the i-PP/EVA blends vs. Φ_d. The impact strength enhances with Φ_d and goes upto 1.24 at $\Phi_d = 0.32$. The impact behavior showed similar variation of the ductility of the blends. The enhanced impact strength may be attributed to the flexibility effect of EVA which, along with the amorphous chains cushions the spherulites of i-PP, enabling the absorption of high impact energy.

As crystallinity of the i-PP/EVA blends increased with EVA copolymer content, the impact toughening contribution of the polymer was evaluated from a plot of normalized relative Izod impact strength vs. Φ_d, Figure 9. The relative impact strength values increased with Φ_d, to 1.2–1.5, depending on Φ_d. This clearly demonstrates that the EVA polymer further increases the impact strength of impact i-PP. Such an impact modification increases the softness and decreases the modulus of the polymer further and can also pave the way for further enhancement of properties by incorporation of fillers. Similar impact toughening of i-PP was obtained by our group in i-PP/chlorosulphonated polyethylene (CSM) blends also[26] which were later modified with filler.
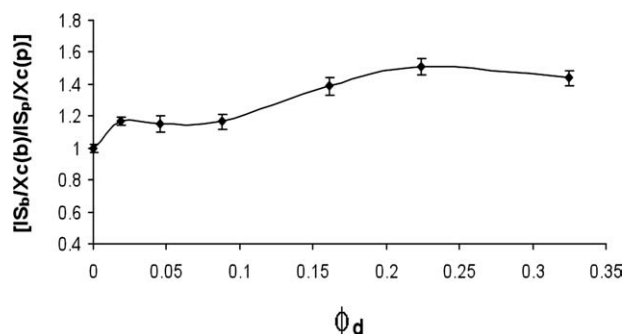

Figure 9 Dependence of normalized relative impact strength $[(I_b/X_{c(b)})/(I_p/X_{c(p)})]$ of i-PP/EVA copolymer blends on Φ_d.
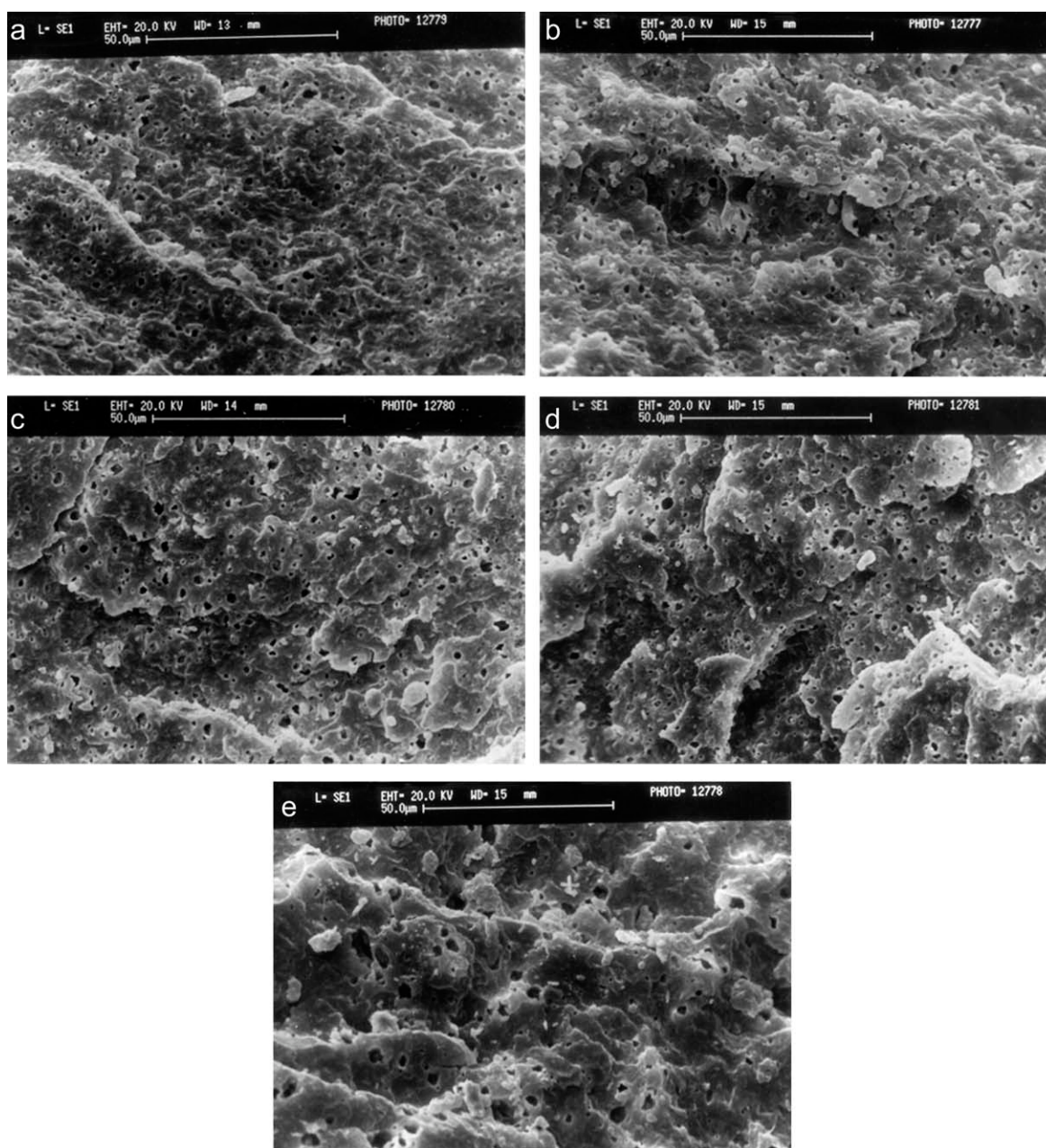
Figure 10 Scanning electron micrographs of (a) i-PP and i-PP/EVA blends at varying Φ_d: (b) 0.08, (c) 0.16, (d) 0.22, and (e) 0.32.

The impact strength enhancement may be due to shear yielding initiated at the region of stress concentration which in turn leads to local strain inhomogeneities.[45,46] The formation of stress concentration points by the EVA polymer was indicated in the analysis of tensile stress data of these blends. This shear yielding may also be observed in the SEM of the impact fractured surfaces in the form of whitening in the interphases (shown later).

State-of-dispersion

SEM photomicrographs of cryogenically impact fractured and etched surfaces of the studied five blend compositions are presented in Figure 10(a–e). In general, two-phase dispersed morphology is clearly seen in the blends with the EVA droplets homogeneously dispersed in the i-PP matrix. Preferential etching of the minor component was carried out in toluene and the etched component can be seen as black regions on these micrographs, representing droplets uniformly dispersed in the i-PP continuous phase. The shape of these droplets on an average is spherical in all the blends and their sizes vary from 0.5 to 2.2 μm with the blending ratio as the EVA polymer content varies from 0.19–0.32 volume fraction. The average size of the droplets increases with the EVA content. From the micrographs, it is observed that

with increasing percentage of the elastomer, the average domain size increases which may be due to increasing dynamic coalescence. This indicates a significant degree of interfacial tension between i-PP and EVA. Some whitening in the interphases of the impact fractured surfaces appears to be due to the occurrence of shear yielding in these blends.

CONCLUSIONS

Incorporation of EVA copolymer into an impact grade of i-PP enhances its flexibility. The tensile modulus and strength decrease while ductility and toughness increases. The EVA polymer provides a surface for the heterogeneous crystallization of i-PP enhancing the overall crystallinity. In addition to crystallinity, other features such as lamella thickness, spherulite size and number of tie molecules may also influence the mechanical properties. Though these are not covered in this article, they may form part of our future work on this topic.

However, the flexibility of the elastomeric EVA copolymer overrides the enhancement of crystallinity facilitating the molecular chain mobility of i-PP. This explains the decrease in modulus and strength as well as increase in the elongation-at-break of the i-PP. Because of this enhanced ductility coupled with strain inhomogeneities at the interphase, shear yielding is facilitated which enhances the impact strength of i-PP. A two-phase structure is seen by the SEM studies. Spherical-shaped EVA polymer particles are uniformly dispersed in the impact i-PP matrix. The size of the dispersed droplets increases with the rubber content. Though the models discussed in this article do not explain the blend properties completely, this elastomer modified blend can be a potential candidate for several applications by addition of particulate fillers which will widen the applicability of this existing polypropylene. This will be taken up as part of our future studies on this system.

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