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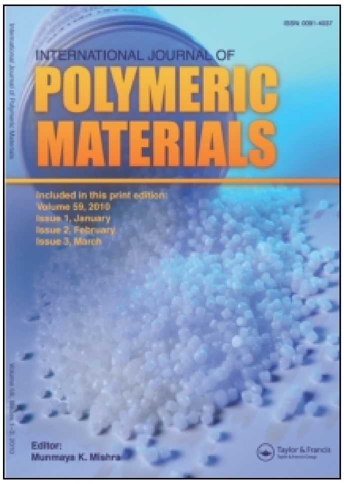

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Mechanical Properties of Talc Filled i-PP/CSM Rubber Composites

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Mechanical Properties of Talc Filled i-PP/CSM Rubber Composites

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Tensile and impact properties of talc-filled i-PP/CSM rubber (20 phr, 0.13 vol fraction) blends were studied in the talc concentration range 0–0.149 vol fraction (0–50 phr). The tensile modulus increased whereas the tensile breaking strength and the strain-at-break decreased with increase in the talc concentration. The modulus increase and the strain decrease were due to enhanced mechanical restraint imposed by the talc particles on the polymer blend decreasing its deformability. Formation of stress concentration points explained the decrease in the tensile strength. The Izod impact strength showed a significant decrease with increase in the filler content. Surface treatment of the talc particles with a titanate coupling agent LICA 12 increased the wetting of the talc by the polymer blend, further modifying the strength properties. Scanning electron microscopic studies showed enhanced dispersion of the filler particles sequential to the surface treatment, effecting modifications of the composite strength properties.

Keywords: i-PP/CSM rubber/Talc composites, mechanical restraint, interphase interaction, stress concentration, phase morphology

INTRODUCTION

Particulate-filled polymer composites are a group of useful materials to suit a wide range of applications at economical costs [1–9]. The property modifications achieved are enhanced stiffness, dimensional stability, creep resistance, reduced mold shrinkage, and the like. A large

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number of these particulate materials are inorganic compounds such as talc, $CaCO_3$, clay, mica, $CaSiO_3$, glass beads, $BaSO_4$, and so on.

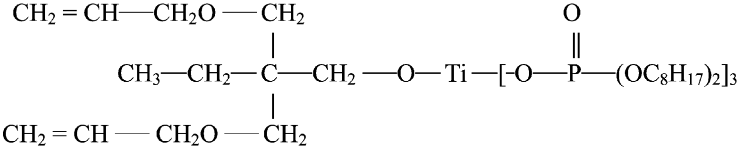
To obtain the best property combinations from these composites good adhesion between the dispersed phase and the polymer has to be achieved. Surface treatment of the filler with a suitable coupling agent is often recommended, which also brings about efficient dispersion of the powdery fillers into the matrix polymer [9–10].

In this article the effects of talc particles on the mechanical properties of a blend of an impact grade of isotactic polypropylene (i-PP) and chlorosulfonated polyethylene (CSM) rubber (0.13 vol fraction, 20 phr) have been studied. The tensile modulus, breaking strength, and strain have been compared with the theories for two-phase systems. The Izod impact strength values were also evaluated. Scanning electron microscopic studies have been made to determine the composite structure. The effects of a titanate coupling agent, LICA 12, on the aforementioned properties have also been analyzed.

EXPERIMENTAL

Materials

The isotactic polypropylene used was a reactor grade impact polymer (hitherto i-PP) B 030 MG (MFI 3, density $0.9\,g\,cm^{-3}$, rubber content 18.5%) manufactured by M/S Reliance Industries, India, Ltd. [11]. The chlorsulfonated polyethylene (CSM) rubber used was Hypalon 40 with 1% and 35% sulfur and chlorine contents, respectively, and density $1.18\,g\,cm^{-3}$ [12]. The viscosity average molecular weight of the CSM rubber was 78,000 in toluene at 313 K [13–14]. Tribasic lead sulfate and lead stearate were obtained from M/S Fine Chemical Company Ltd., Mumbai. The talc, supplied by M/S Pioneer Chemical company Ltd. Mumbai, was of average particle diameter 27 μm, density $2.67\,g\,cm^{-3}$. The surface treatment agent to modify the talc surface was neopentyl(diallyl)oxytri(dicotyl)phosphatotitanate, LICA 12, of the following chemical formula (6,10):



$$CH_3{-}CH_2{-}\underset{\underset{CH_2{=}CH{-}CH_2O{-}CH_2}{|}}{\overset{\overset{CH_2{=}CH{-}CH_2O{-}CH_2}{|}}{C}}{-}CH_2{-}O{-}Ti{-}[{-}O{-}\overset{\overset{O}{\|}}{P}{-}(OC_8H_{17})_2]_3$$

obtained from M/S Kenrich Petrochemicals Inc., Bayonne, New Jersey, USA.

Surface Treatment of Talc

2 g LICA 12 were diluted with 40 ml toluene and the solution was added to 500 g talc in a high-speed laboratory mixer (Phillips, Model-2L) at 2400 rpm for 10 min. The treated filler was then dried in a hot air oven at 373 K for 2 h followed by vacuum drying at 363 K for 4 h. The surface treatment was confirmed by water floatation technique: the treated talc powders spread over water surface whereas the untreated particles readily sank. Similar technique of filler surface treatment was employed in other works too [15–16].

Preparation of the Composites

A blend of the i-PP and 0.13 vol fraction (20 phr) CSM rubber together with 0.4 phr each (based on the CSM rubber) of both the lead sulfate and lead stearate heat stabilizers, was first made by melt mixing the polymers following a procedure described elsewhere [17]. A 0.149 vol fraction (50 phr) talc (both surface treated and untreated) filled composite masterbatch was then made, using the polymer blend in a co-rotating twin screw extruder Berstoff, Model E0440/83, with an L/D = 28/1 and screw diameter 25 mm. Separate hoppers were used to feed the polymer blend and the filler. A reverse temperature profile, for example, 473 K and 533 K in the feed and 413–433 K in the barrel and the die zones, respectively, were maintained while the screw speed was 500 rpm.

Composites of the lower filler contents (10–35 phr, 0.034–0.109 vol fraction, ϕ_F) were made by subsequent dilution of the 50 phr masterbatch in a Windsor Klockner single screw extruder (Model–S × 30, L/D ratio 20:1), at a screw speed 20 rpm, using temperature profiles of 453–473 K from the feed zone to the die zone. The extruded strands were water cooled, chopped into granules, and dried in vacuum for 3 h at 353 K. The 50 phr masterbatch and the neat blend were also subjected to identical extrusion processes in order to maintain the thermal and mechanical history similar to that of the i-PP/CSM component in the i-PP/CSM rubber/Talc composites.

Preparation of Test Specimens

Dumb-bell shaped tensile specimens and rectangular bar impact samples were made by injection molding of the composites on a Windsor machine Model Sp1, using a melt temperature 473 K, injection pressure 1.2 Pa, and mold temperature 301 K. Small pieces cut from the tensile specimens were used for X-ray crystallization studies.

ANALYSIS AND MEASUREMENTS

X-Ray Diffraction Studies

Wide-angle X-ray diffraction measurements were performed on a Phillips Norelco X-ray diffractometer equipped with a scintillation counter and recorder. Radial scans of intensity (I) versus diffraction angle (2θ) were recorded in the 2θ range of 10°–35° using Ni-filtered CuK_α radiation of wavelength 1.5418 Å. Diffractograms of all the samples were recorded at identical settings of the instrument (operating voltage 40 kV, scan rate 2°/min, filament current 30 mA).

The experimental I versus 2θ diffractograms were converted to Is^2 versus curves [18]. Amorphous scattering curve and the baseline were drawn following the method described elsewhere [19–21]. The degree of crystallinity, X_c, was calculated according to Eq. 1:

$$X_c = \frac{\int_0^\infty s^2 I_{cr}(s) ds}{\int_0^\infty s^2 I(s) ds} \cdot K \tag{1}$$

where $I_{cr}(s)$ is the coherent intensity in the crystalline peaks, I (s) the total coherent intensity scattered, s the scattering vector (expressed as $s = (2/\lambda)\sin\theta$), K the correction factor that is governed by atomic scattering factors and the disorder function [20–21]. Due to the uncertainty about the value of the disorder function the value of K is taken as unity in these calculations. The estimated degree of crystallinity, X_c, is thus an apparent value, which may be used for comparisons in similar materials.

Scanning Electron Microscopy

Cryo-fractured tensile specimens were sputter-coated with silver and scanned on a Cambridge Stereoscan, Model-360, using an accelerating voltage of 20 kV, to investigate the state of dispersion of the filler in the polymer blend.

Mechanical Tests

A Zwick Universal Tester, Model-Z010, was used to evaluate tensile properties following ASTM D638 test procedure at initial cross head separation of 6 cm and cross-head speed of 5 cm min^{-1}. The Izod impact strength data were determined on notched specimens on an FIE instrument, Model IT-0.42, according to ASTM D256 test procedure. At least five samples were tested at each talc concentration and the average values (within ±3%) reported. The tests were performed at ambient temperature of $301 \pm 2\,K$.

RESULTS AND DISCUSSIONS

X-Ray Crystallization

The X-ray diffractograms of the i-PP/CSM rubber blend and the talc-filled composites showed peaks at 2θ of 14.0, 16.8, 18.5, and 21.4° corresponding to the (110), (040), (130), and (111) diffraction peaks (Figures 1 and 2). Similar crystalline peaks for the i-PP have been reported earlier by Purwar and Gupta [19]. The presence of these characteristic peaks of i-PP in the i-PP/CSM rubber blend and the talc-filled composites indicate that the crystalline structure of i-PP remains unchanged upon addition of the rubber and the filler. However, the intensities of the peaks decreased indicating that by incorporation of the filler the proportion of the monoclinic (α form) and hexagonal (β form) is affected. This variation in the peak heights could be due to the variation of the mean spherulite size or their distribution, deformation at the spherulite boundaries, or any long-range order induced in the structure by dispersion of the filler in the i-PP/CSM rubber blend.

The crystallinity data estimated from the diffraction patterns following the procedure described in the experimental section are presented in Table 1. The values in the last column were estimated by multiplying the volume fraction, v, of the i-PP component in the blend and the composites with the crystallinity (%) of the neat i-PP (68.9%) in the impact grade of i-PP, which contained 18.5% of EPR [11, 18]. The experimental crystallinity values decreased compared to those corresponding to the proportion of the i-PP, column 4, Table 1. This may be due to the presence of the foreign materials, that is, the CSM rubber and the talc filler, which may restrict the molecular mobility of the i-PP chains for fitting into the crystal structure. In the i-PP/CSM rubber/Talc (treated) composites the crystallinity values decreased somewhat with increase in the filler concentration, compared to the data in the untreated talc-filled systems, indicating a degree of restricted i-PP chain mobility, possibly through increased interphase interactions.

Tensile Properties

Table 2 exhibits the tensile properties, such as the tensile modulus, the breaking strength, and the strain-at-break (%) of the i-PP/CSM

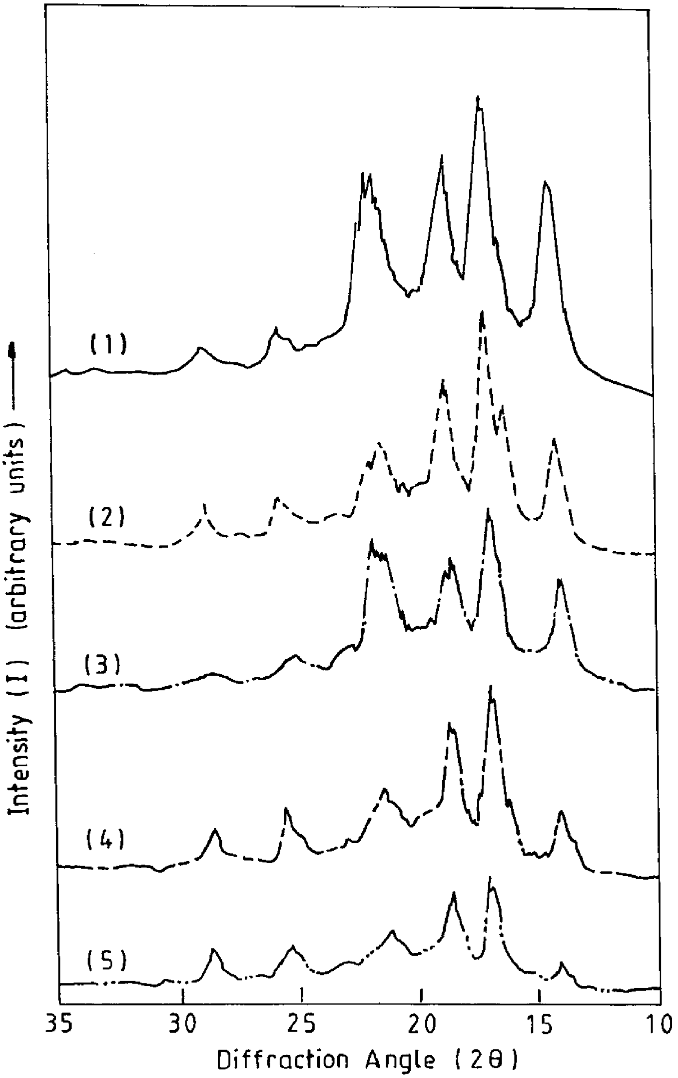


FIGURE 1 Variations of X-ray scattering intensity (I) vs. diffraction angle (2θ) for the i-PP/CSM rubber (0.13 vol fraction) (1) and the i-PP/CSM rubber/Talc (untreated) composites at varying talc concentrations, ϕ_F: (2) 0.034; (3) 0.065; (4) 0.109; (5) 0.149.

rubber/Talc composites, determined from the stress strain curves (not shown). These tensile parameters are presented in Figures 3–6 as the variations of the ratio of the property (subscript c) to

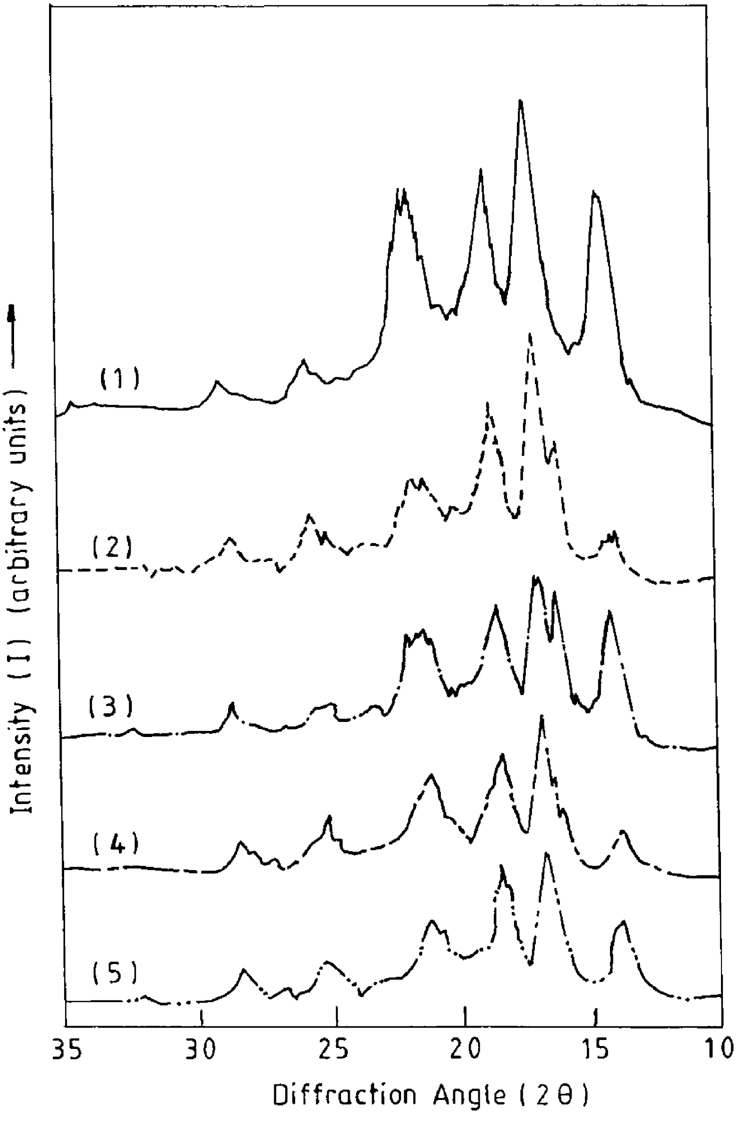


FIGURE 2 Plots of I vs. 2θ for i-PP/CSM rubber (0.13 vol fraction) (1) and the i-PP/CSM rubber Talc (treated) composites at varying ϕ_F: (2) 0.034; (3) 0.065; (4) 0.109; (5) 0.149.

that of the matrix polymer blend (i-PP/CSM rubber, 0.13 vol fraction, 20 phr) (subscript p) versus volume fraction, ϕ_F, of the filler, talc.

TABLE 1 X-Ray Crystallinity Data for the i-PP/CSM Rubber/Talc (Untreated and Treated) Composites

ϕ_F	Volume fraction of i-PP (v)	Crystallinity (%)	$(X_c)_{i\text{-}PP}$
0	0.707	43.6	48.7
0.034	0.687	41.8 (40.6)	47.3
0.065	0.668	39.2 (37.4)	46.0
0.109	0.641	36.8 (35.1)	44.2
0.149	0.617	33.9 (31.8)	42.5

Values in parentheses are for the i-PP/CSM rubber/Talc (treated) composites.

Tensile Modulus

The plot of the relative tensile moduli (E_c/E_p) against ϕ_F is presented in Figure 3. The moduli of the untreated talc-filled composites show a continuous enhancement with talc concentration. At the maximum ϕ_F (0.149 vol fraction), the value is ~1.25 times that of the i-PP/CSM rubber. Filler surface modification brings about a relatively lesser degree of modulus increase at corresponding ϕ_F values, compared to the unmodified talc-filled composites. The modulus values were compared with some predictive models for two-phase composite systems, which take into account the shapes, packing fraction, and interphase adhesion between the matrix polymer and the discrete phase (Figure 3). Curve I describes Einstein's equation without adhesion [22–24] (Eq. 2), curve II represents Einstein's equation with adhesion [22–24] (Eq. 3) for filled polymer systems:

$$E_c/E_p = 1 + \phi_F \tag{2}$$

$$E_c/E_p = 1 + 2.5\,\phi_F \tag{3}$$

TABLE 2 Mechanical Properties of i-PP/CSM Rubber/Talc (Untreated and Treated) Composites

ϕ_F	Tensile modulus (MPa)	Tensile breaking strength (MPa)	Strain-at-break (%)	Izod impact strength (J/M)
0	235.1	12.9	74.8	146.6
0.034	264.3 (253.1)	12.8 (12.3)	74.7 (56.1)	109.8 (114.8)
0.065	275.4 (255.4)	12.3 (12.2)	69.4 (50.1)	83.5 (75.0)
0.109	288.8 (267.5)	12.2 (11.8)	36.9 (27.6)	66.8 (46.7)
0.149	293.2 (293.7)	11.9 (11.7)	17.4 (18.7)	43.3 (19.9)

Values in the parentheses are for the i-PP/CSM rubber/Talc (treated) composites.

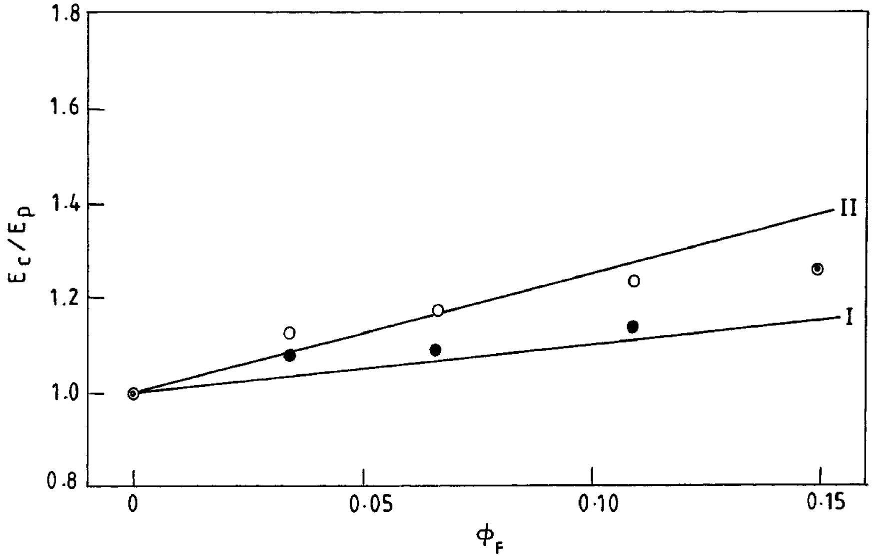


FIGURE 3 Plot of the relative tensile modulus (E_c/ E_p) of the i-PP/CSM rubber/Talc (O) and the i-PP/CSM rubber/Talc (treated) (●) composites against the talc volume fraction ϕ_F. Curves I and II describe Einstein's equation without adhesion (Eq. 2) and with adhesion (Eq. 3), respectively.

The experimental modulus data show a reasonably good agreement with the Einstein model with adhesion, except for the values at $\phi_F = 0.149$, which lie between the two curves. The i-PP/CSM rubber/Talc composites seem to be of the adhesion type, which may be attributed to an extent of physical interaction in the form of surface adsorption of the polymer phase by the solid phase, along with surface interlocking of the phases arising out of the differential thermal contraction of the two phases [22]. The softer polymer matrix is stiffened by the talc particles, which impede the molecular mobility/deformability of the polymer chains through the introduction of a mechanical restraint, the extent of which is a function of the particulate spacing as well as the properties of the component phases [25].

It may be noted that the crystallinity of the i-PP component, which is the major phase in these systems, decreased with increase in the talc concentration (Table 1). This should have led to a decrease in the moduli of the blend in presence of the filler. It appears that the effect of the filler as mechanical restraint to the polymer matrix predominates, giving rise to the enhancement. Other polymer systems also

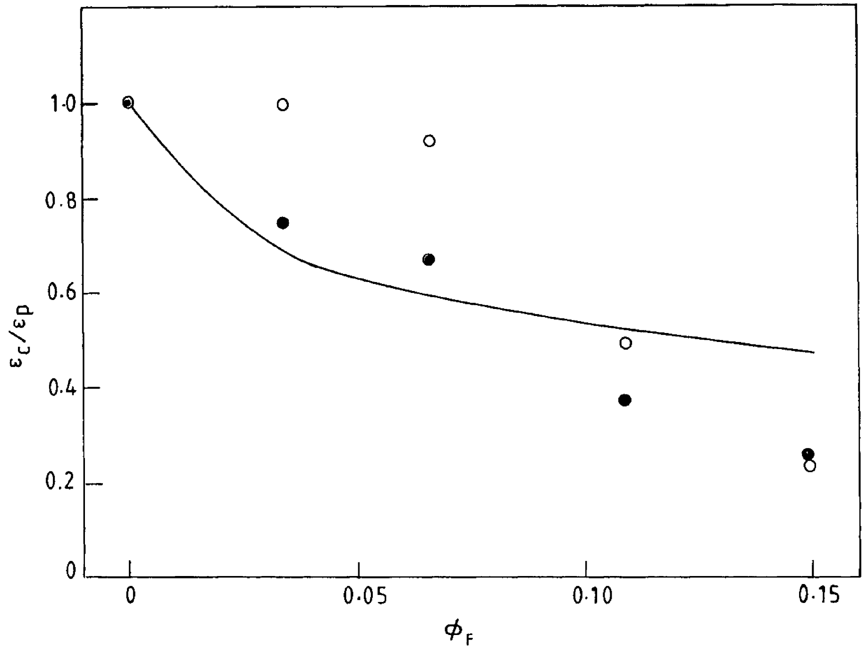


FIGURE 4 Relative strain-at-break (ϵ_c/ϵ_P) of the i-PP/CSM rubber/Talc (O) and the i-PP/CSM rubber/Talc treated (•) composites vs. ϕ_F. The curve represent predicated behavior according to Eq. 4.

showed modulus increase in the presence of rigid particulate fillers [24, 26–27]. Polymer modulus increase in the presence of solid particles was also attributed to an enhancement of the attraction potential between the segment of the polymer chain as a result of the repulsive potential that the polymer is subject to when it approaches the solid particles [28].

Surface modifications of the talc with LICA 12 also increases the modulus over that of the polymer blend, the values lie around curve I (Figure 3). The data are, however, lower than those of the untreated talc-filled systems at the corresponding ϕ_F values, except at $\phi_F = 0.149$ where the two values are similar. This may be due to a plasticizing/lubricating function by the coupling agent similar to other reports [24, 29]. The modulus decrease may also be partly due to an extent of increased interphase interaction, which marginally decreases the crystallinity of the i-PP component of the blend (Table 1).

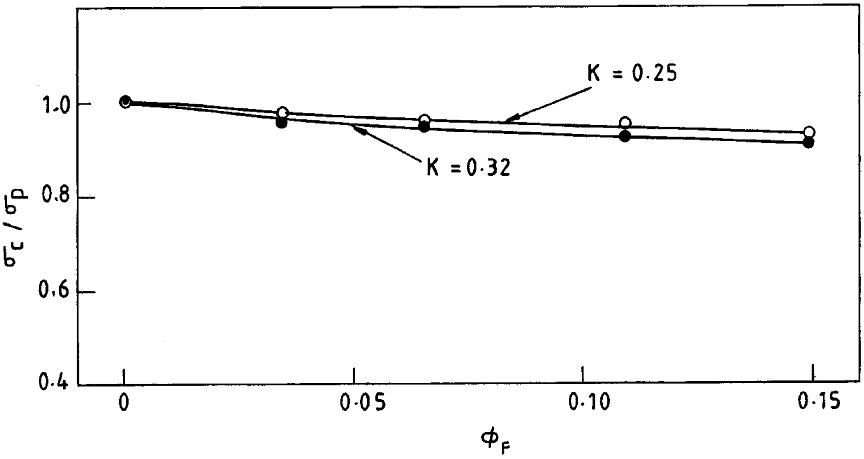


FIGURE 5 Variation of the relative tensile stress (σ_c/σ_p) of the i-PP/CSM rubber/Talc (O) and i-PP/CSM rubber/Talc (treated) (•) composites versus ϕ_F. The curves represent the Nicolais-Narkis model (Eq. 5) with $K = 0.25$ and 0.32 for the composites, respectively.

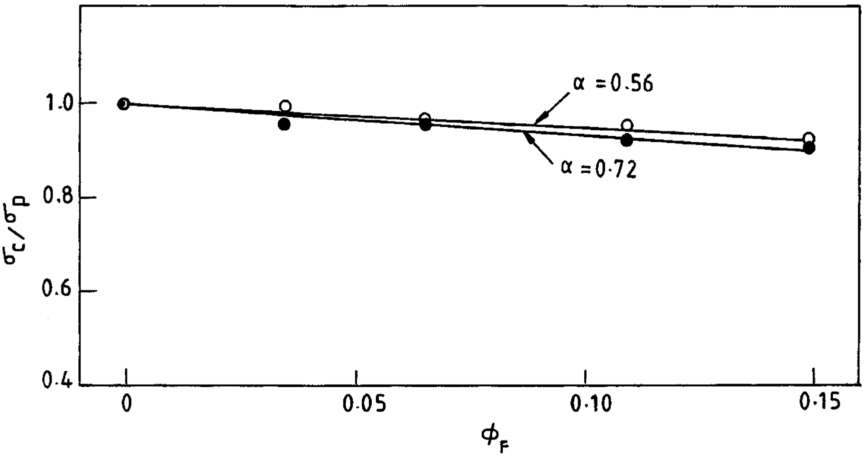


FIGURE 6 Plot of the relative tensile stress (σ_c/σ_p) of the i-PP/CSM rubber/Talc (O) and the i-PP/CSM rubber/Talc (treated) (•) composites versus ϕ_F. The curves describe the porosity model (Eq. 6) with $\alpha = 0.56$ and 0.72 for the composites, respectively.

Strain-at-Break

Breaking strain of the i-PP/CSM rubber/Talc composites decreases with increase in ϕ_F, (Figure 4), similar to other works [24,27]. The decrease is rather marginal up to $\phi_F = 0.065$, whereas beyond this point the data decrease quite drastically. With surface modification of the talc, the decreasing trend of the strain is maintained, the reduction of the data is rapid right form low ϕ_F values, however, except at $\phi_F = 0.149$ where the data for both types of the composites become similar. The strain data for the i-PP/CSM rubber/Talc (untreated) composites were significantly higher than the Nielsen's model for perfect adhesion [22] (Eq. 4) up to $\phi_F = 0.065$ whereas the

$$\in_c / \in_p = 1 - \phi_F^{1/3} \tag{4}$$

data at $\phi_F > .065$ were less than the curve (Figure 4). The data of the treated talc-filled systems lied marginally above and significantly below the predicted model up to and beyond $\phi_F = .065$, respectively. The decrease in the strain of the i-PP/CSM rubber blends in the presence of the talc implies a reduction of molecular deformability through immobilization of the polymer matrix by imposition of mechanical restraint as shown in the modulus data. In the treated talc-filled system the strain is decreased to a degree higher than the previous composites, which may be due to a degree of enhanced interphase interaction, which decreased the i-PP crystallinity (Table 1). Although the coupling agent would also be expected to function as a plasticizing/lubricating agent, at least in part [10, 24, 29], it seems that the dominating effect is the interphase interaction and imposition of mechanical restraint on the deformability of the matrix blend.

An extent of interphase interaction was indicated in the modulus data of the i-PP/CSM rubber/Talc composites. Surface treatment of the talc enhanced the interaction (decreasing crystallinity of i-PP to a small extent). Breaking strain values also indicated hindrance to molecular deformability in the untreated talc-filled composites where this deformability is further decreased upon surface treatment of the talc. In particulate-filled polymer systems the interphase plays a significant part in controlling the overall composite properties [22, 25]. If the dispersed phase does not adhere to the polymer, discontinuity will be created for the stress transfer and stress concentration points will be generated around the particles. The effect will be aggravated in the case of breakdown of the particle agglomerates. To achieve enhanced mechanical and other properties in these two phase systems, two important critieria are (i) maintenance of continuity in the stress-transfer and (ii) interphase adhesion.

Tensile Strength

To evaluate weakness in the composite structure, the tensile breaking strength data were analyzed according to theories describe elsewhere [24, 30]. The following models were used:

$$\sigma_c/\sigma_p = 1 - K\,\phi_F^{2/3} \tag{5}$$

$$\sigma_c/\sigma_p = \exp\,(-\alpha\,\phi_F) \tag{6}$$

where σ_c and σ_p are the tensile strength of the composite and the matrix, respectively. These models denote no-adhesion type structure and are based on the area fraction or volume fraction of the inclusions [31–32]. In Eq. 5 the parameter K describes adhesion quality between the polymer and the discrete phase [33]; the lower the value, the better the adhesion. With spherical fillers and poor adhesion, $K = 1.21$ [31]. In the porosity model, Eq. 6, the inclusion is assumed to be equivalent to pores/voids in metals/ceramies [34] and polymer blends and composites [31, 35]. The pores do not influence the composite mechanical properties due to lack of interphase adhesion. The parameter α accounts for stress concentration: the higher the value the higher the stress concentration [31].

Comparing the experimental tensile strength data with the theories, the values of adhesion parameter K and stress concentration factor α (Eqs. 5 and 6), can be evaluated (Table 3). Comparisons of the data and the theories are shown in Figures 5 and 6. The data with untreated talc show good agreement with the Nicolais-Narkis model (Eq. 5) with $K = 0.25$, which indicates a good degree of two-phase

TABLE 3 Values of the Adhesion Property K (Eq. 5) and the Stress Concentration Parameter α (Eq. 6) in the i-PP/CSM rubber/Talc and the i-PP/CSM rubber/Talc (treated) composites

ϕ_F	K	α
0	—	—
0.034	0.02 (0.44)	0.07 (1.38)
0.065	0.27 (0.30)	0.68 (0.77)
0.109	0.22 (0.34)	0.47 (0.74)
0.149	0.27 (0.25)	0.53 (0.56)
Mean value	0.25 (0.32)	0.56 (0.72)

Due to data scatter, mean values were taken excluding some data points, for example, for K and α at $\phi_F = .034$. Values in the parentheses are for the i-PP/CSM rubber/Talc (treated) composites; the mean was taken at $\phi_F > .034$.

adhesion resulting in quite a strong interphase. The value of the adhesion parameter K is lower than that in other systems [2, 24]. Possibly the adhesion is promoted through an interaction between the polar CSM rubber and the talc. The porosity model (Eq. 6) exhibits some degree of stress concentration with $\alpha = 0.56$ which is, however, significantly lower than the i-PP/Talc system where the value of α was 6.18 [27].

In the presence of the coupling agent LICA 12 the interphase adhesion is reduced marginally, K being equal to 0.32. The stress concentration parameter also increases to an extent with $\alpha = 0.72$ (Table 3). Enhancement of stress concentration sequential to the filler treatment was observed in other works too [2, 24].

The matrix polymer used in these systems was a blend of i-PP and CSM rubber. The i-PP component is the major phase and crystallizable, which contributes to the strength properties [4, 24], whereas the elastomeric components tend to dilute the i-PP and decrease its crystallinity. In the presence of the particulate talc the overall effect of the crystallinity of the i-PP and the filler-polymer interaction would determine the strength properties.

In the i-PP/CSM rubber/Talc composites, the tensile strength and the breaking strain decreased whereas the tensile modulus increased with the talc content. Some degree of interaction of the hydrophillic talc [9] is possible with the polar CSM rubber of the blend. In the presence of the talc the crystallinity of the i-PP component decreased (Table 1). However, the filler particles also caused mechanical restraints, which played a dominant role to increase the modulus of the polymer blend. The tensile strength decreased marginally with increase in the talc concentration, which may be attributed to the decrease in the i-PP crystallinity coupled with the formation of stress concentration points around the filler particles. The latter effect is quite expected becasue any flaw in the composite structure is likely to be magnified in large deformations in which the breaking strength is measured [22, 24]. The breaking strain decrease is due to the enhanced mechanical restraint by the discrete phase, which restricts the deformation of the polymer matrix [25–27].

Upon surface treatment of the talc with LICA 12, the tensile strength decreased from the values of the previous composites, which may be attributed to the enhanced adhesion of the treated talc with the i-PP component impeding its crystallinity, (Table 1), in preference to the CSM rubber phase. Enhancement of interphase interaction between the LICA 12 treated talc and the i-PP is expected from the chemical structure of the two phases [10]. The resultant effect may be a degree of decreased interphase adhesion giving rise to a marginally increased value of K (0.32) and the stress concentration parameter α (0.72) (Table 3). The tensile strength decrease may also be due partly to the plasticizing/lubricating effect of function by the coupling agent, similar to other works [24, 29]. Decrease in the strain at break may be due to a degree of predominant interphase interaction between the treated talc and the i-PP component in the polymer blend.

Impact Properties

Figure 7 presents the variation of the relative Izod impact strength values (I_c/I_p) of the composites as function of ϕ_F. In the untreated talc-filled composites, the Izod impact strength decreases quite sharply with increasing talc concentrations. This may be due to matrix immobilization and eventual stiffening of the blend so that the stress concentrations around the particles give rise to cracks and the impact load is unable to be transferred. Formation of stress concentration was shown in the analysis of the tensile strength data in the previous

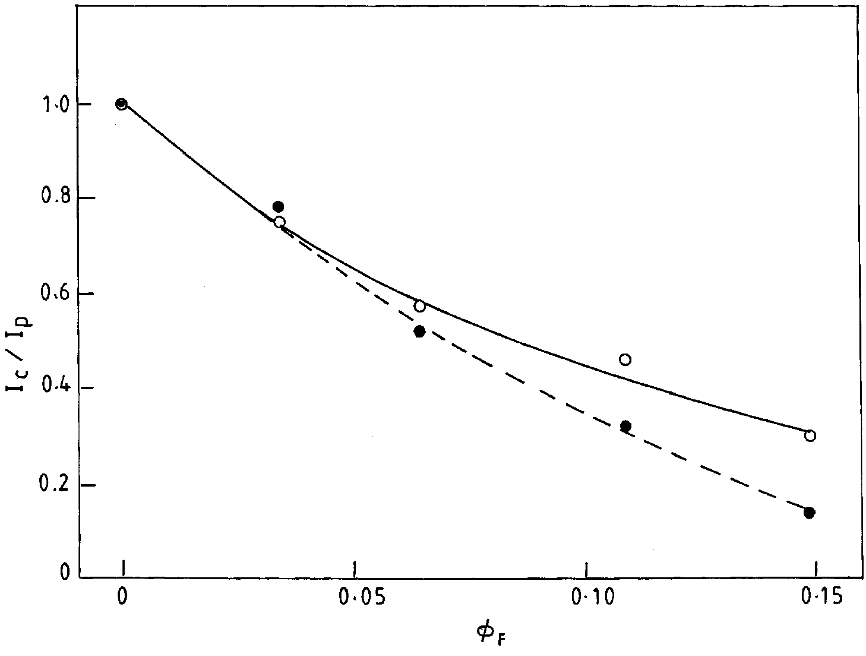


FIGURE 7 Dependence of the relative Izod impact strength (I_c/I_p) of the i-PP/CSM rubber/Talc (O) and the i-PP/CSM rubber/Talc (treated) (●) composites against ϕ_F.

section. With surface modification of the filler, the impact strength decreased further, in particular beyond $\phi_F = .034$, which may be ascribed to the interphase adhesion of the talc that with the i-PP leading to a further decrease in the deformability of the polymer blend. It appears that the plasticizing/lubricating function of the coupling agent is less effective than the enhanced interphase adhesion under impact mode of load application.

Fracture Surface Morphology

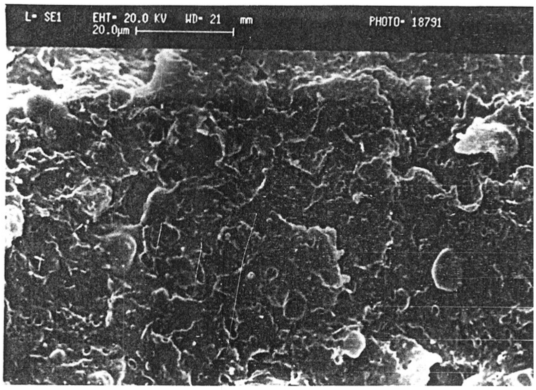
The scanning electron microscopic (SEM) studies of the i-PP/CSM rubber blend and its talc-filled composites are shown in Figure 8 (a)–(g). The composites with the untreated talc show quite good dispersion of the talc in the polymer matrix, in particular up to $\phi_F = 0.034$, Figure 8(b). However, beyond this ϕ_F value, the filler dispersion becomes poor with increased extent of bare and unwetted talc surfaces (Figures 8(d), (f), (h)). The flaky nature of the talc also becomes evident. Upon surface treatment of the talc the fracture surface becomes smoother on account of enhanced adhesion of the polymer to the filler surface with decreased extent of bare filler particles (Figures 8(e), (g), (i)). This indicates an enhanced surface wetting of the talc by the polymer blend.

CONCLUSIONS

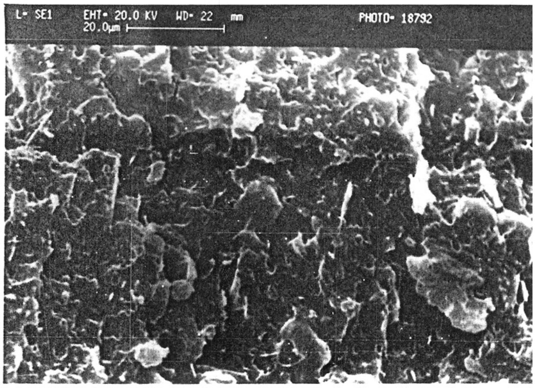
This study reveals that the addition of talc significantly modifies the mechanical properties of the i-PP/CSM rubber (20 phr, 13 vol fraction) blend. The tensile strength and breaking strain decreased whereas the modulus increased with the talc concentration. Izod impact strength decreased significantly with increase in the talc content, at talc concentration of 0.149 vol fraction the value decreased to 0.3 times that of the blend. The composites exhibit good interphase adhesion and a small extent of discontinuity and stress concentration in the composite structure.

Surface treatment of the talc with a titanate coupling agent LICA 12 modified the composite structure further. The tensile modulus

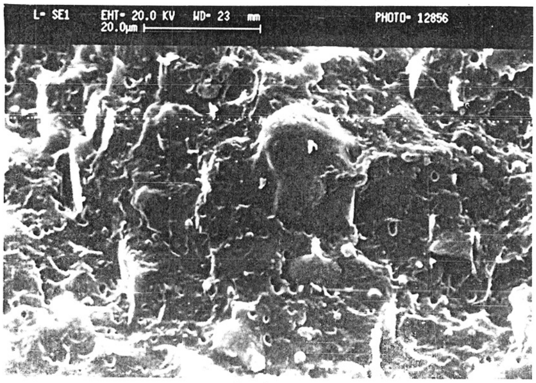
FIGURE 8 SEM photomicrographs of the i-PP/CSM rubber blend (a) and fractured surface of the i-PP/CSM rubber/Talc composites at varying ϕ_F: (b) 0.034; (d) 0 .065; (f) 0.109; (h) 0.149. The micrographs of the i-PP/CSM rubber/Talc (treated) composites at the corresponding ϕ_F values are presented in (c), (e), (g), and (i), respectively.


L= SE1
EHT= 20.0 KV
WD= 21 mm
PHOTO= 18791
20.0µm

(a)


L= SE1
EHT= 20.0 KV
WD= 22 mm
PHOTO= 18792
20.0µm

(b)


L= SE1
EHT= 20.0 KV
WD= 23 mm
PHOTO= 12856
20.0µm

(c)

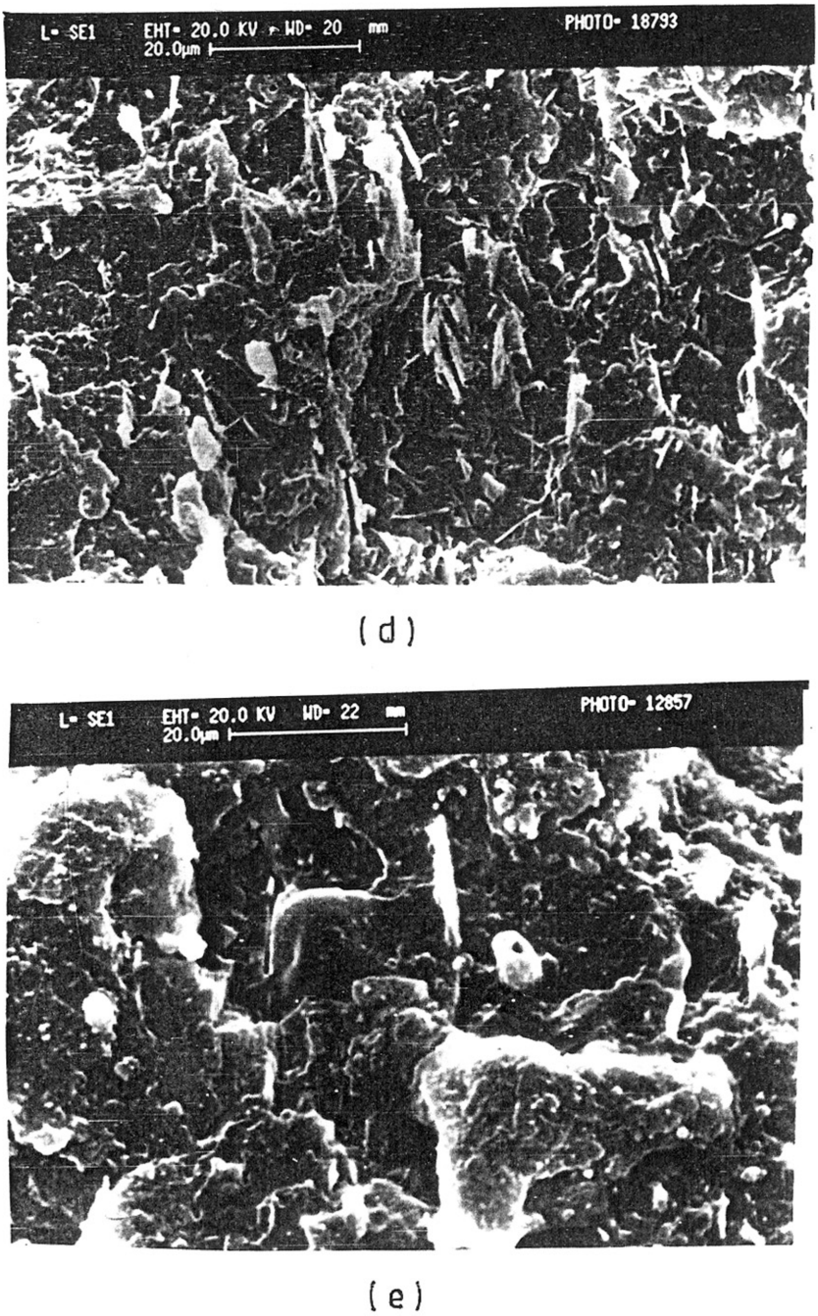


FIGURE 8 Continued.

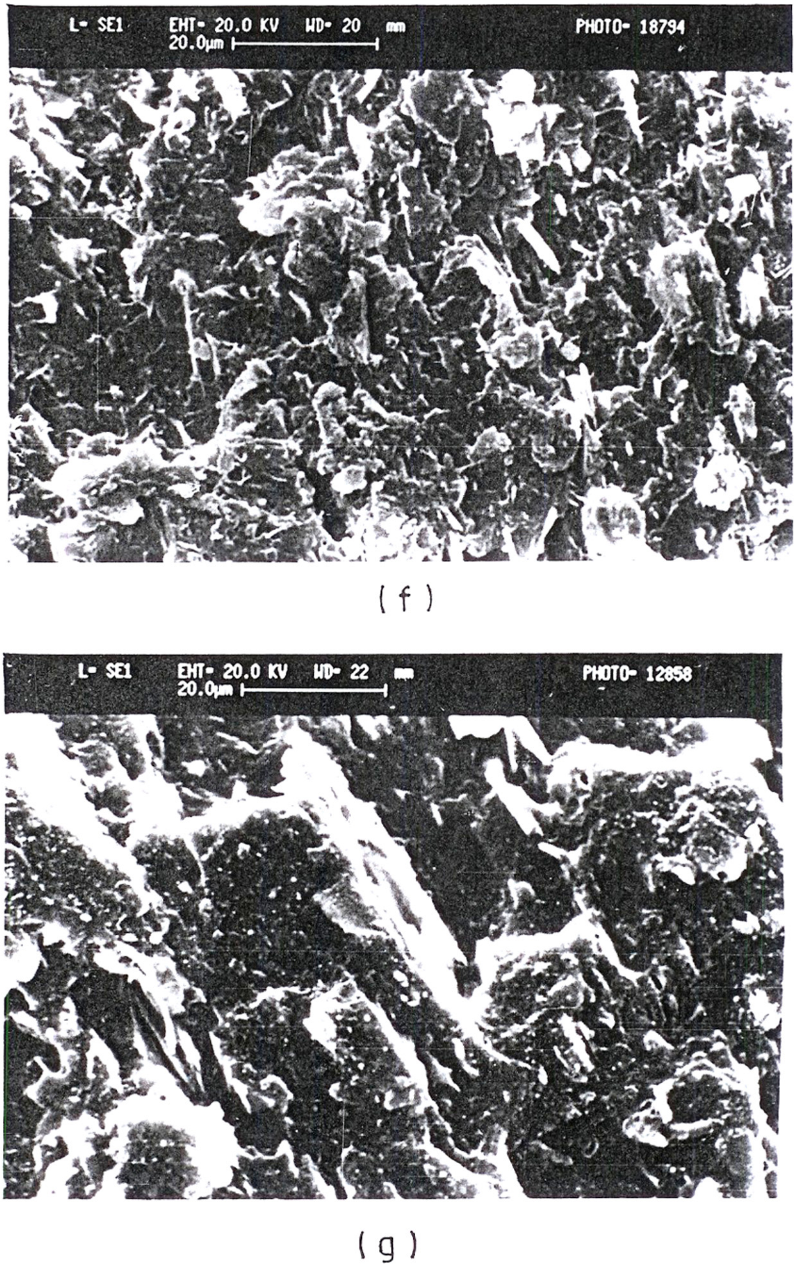


FIGURE 8 Continued.

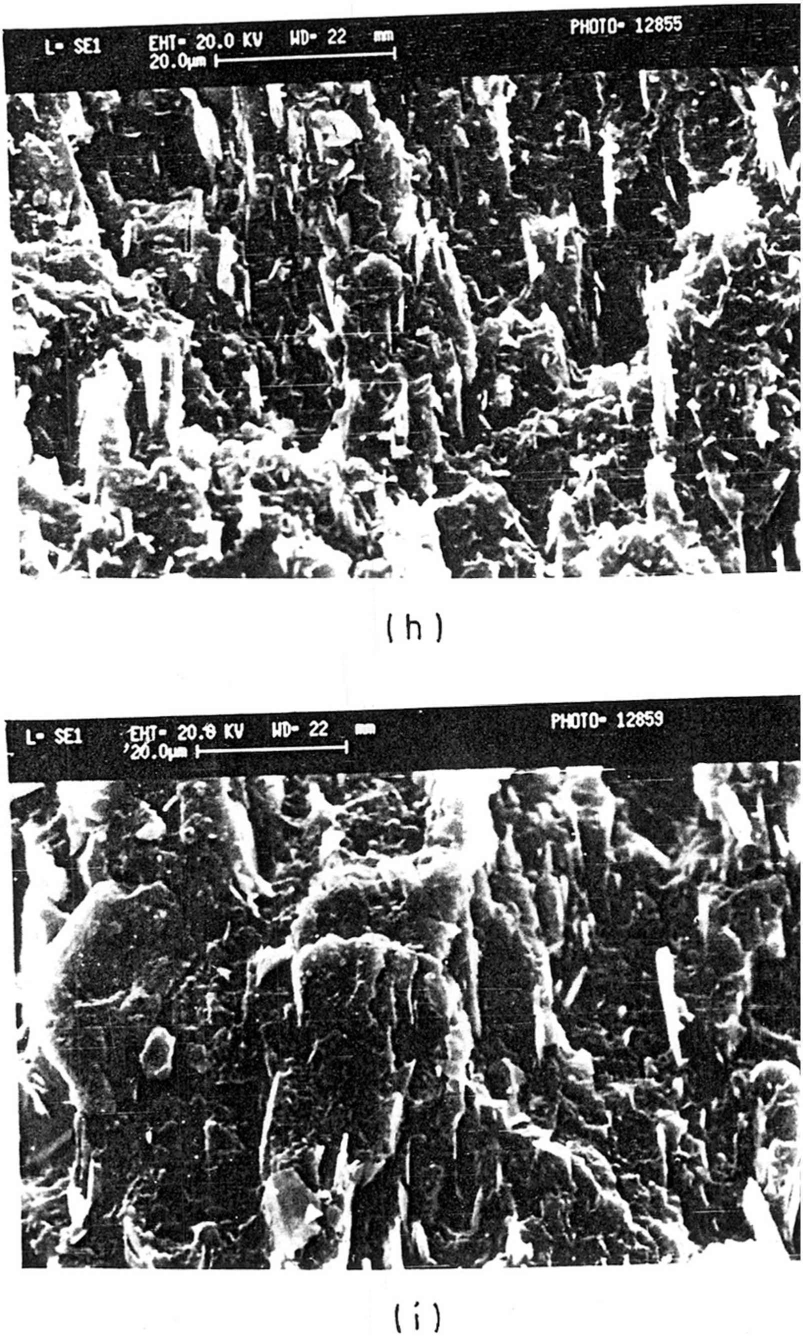


FIGURE 8 Continued.

registered a lesser increase. The tensile strength and strain also decreased, which was attributed to an enhanced interaction of the talc and the i-PP and imposition of mechanical restraint, which impedes molecular deformation of the blend. Izod impact strength also decreased due to matrix immobilization resulting from an enhanced interphase interaction by the surface treated talc. However, the plasticizing/lubricating effect of LICA 12 is superceded by enhanced interphase adhesion of the talc particles with the i-PP.

SEM studies indicated increased surface wetting of the filler leading to its better dispersion upon surface treatment with LICA 12.

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