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A SMALL-ANGLE X-RAY SCATTERING STUDY OF BLENDS
OF ISOTACTIC AND ATACTIC POLYSTYRENE^{*}

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ABSTRACT

The small angle x-ray scattering from blends of isotactic and atactic polystyrene has been studied. Results have been interpreted and compared using the Tsvankin, Vonk and Hosemann techniques. The studies suggest that segregation of the atactic component occurs during crystallization within the growing spherulite of the isotactic component. However, since the interlamellar distance does not increase with atactic content, segregation is believed to occur with the formation of domains larger than interlamellar but smaller than spherulite size.

INTRODUCTION

A considerable research effort is being devoted to commercially important blends of both compatible and incompatible polymers. The structures of several crystalline blend systems have been examined in detail by small angle x-ray scattering (SAXS), notably the compatible blends of isotactic polystyrene/poly 2,6-dimethyl phenylene oxide (i-PS/PPO)¹ and polycaprolactone/polyvinyl chloride (PCL/PVC)². In both these blends the non-crystalline component was found to reside between the individual lamellae as opposed to an interfibrillar or interspherulitic position. In order to obtain this structural information theoretical models based on lamellar stacking were fitted to the experimentally observed SAXS. The Hosemann and Tsvankin/Buchanan models were used in the i-PS/PPO and PCL/PVC blends respectively. These results are in contrast with the optical microscopy of Keith and Padden³ on blends of the isotactic and atactic forms of polystyrene and polypropylene. Qualitatively they found segregation of the atactic component to the interfibrillar regions as opposed to the interlamellar regions. The scale of segregation was placed on a somewhat quantitative basis by the use of the expression $\delta = D/G$ where δ is the dimensional order of segregation, D is the diffusion coefficient of the non-crystalline component and G is the spherulitic radial growth rate. Qualitatively, in the case of i-PS/PPO and PCL/PVC both non-crystallizing components have considerably higher T_g 's than the crystallizing polymer which, therefore, may cause the diffusion term to dominate and thus cause segre-

gation at the lamellar level. However, in the case of mixtures of isotactic and atactic polymers of similar molecular weight the decreased growth rate may induce segregation into regions with larger dimensions.

The purpose of this paper is twofold; firstly, to examine the detailed morphology of an "ideal" polymer blend, in this case blends of i-PS/a-PS, and to determine the level of rejection of the non-crystallizing component whether at the fibrillar or lamellar level. The second objective is to examine the applicability of three popular theoretical models used in the analysis of SAXS from semi-crystalline polymers to i-PS and its blends with a-PS. Previous workers^{1,4} have applied Tsvankin/Buchanan⁵ and Hosemann⁶ models to pure i-PS with conflicting results with respect to the degree of crystallinity and lamellar thickness.

EXPERIMENTAL

Isotactic polystyrene was obtained from Polysciences Inc. in the form of a crystalline powder. After being quenched from the molten state (260°C) it was dissolved in boiling toluene, filtered and then precipitated into vigorously stirred methanol. Subsequent soxhlet extraction of the precipitated polymer for 48 hours with methyl ethyl ketone selectively extracted any a-PS present. The dried extracted i-PS was used for the morphological studies on the i-PS and for blending with a-PS. The molecular weight was found from gel permeation chromatography, $\bar{M}_n = 133,000$, $\bar{M}_w = 724,000$.

Atactic polystyrene (HH101) was obtained from the Monsanto Co. ($\bar{M}_n = 92,000$, $\bar{M}_w = 270,000$). It was dissolved in toluene, filtered and precipitated in excess methanol prior to blending with the i-PS.

The i-PS/a-PS blends were prepared by dissolving the weight fractions of each polymer in boiling toluene at a 2% concentration then coprecipitating into a large excess of vigorously stirred methanol by dropwise addition. The dried blends and pure i-PS were compression molded at 6000 psi and 270°C in a 1 mm thick mold between aluminum foil and then quenched in water at 25°C. Crystallization from the quenched state was carried out on these samples in the mold (sealed from the atmosphere) at 200°C and 210°C for 76 hours to ensure complete crystallization, i.e. volume filled with spherulites.

SAXS studies were carried out using a Kratky camera assuming infinite slit height geometry. Ni filtered CuK α radiation was employed with a step

scanning NaI scintillation detector and electronic pulse height analysis.

Samples were of uniform thickness and approximately 1 mm thick.

Melting and heats of fusion measurements were carried out using a Perkin Elmer DSC 1B. The weight fractions of crystallinity, W_c , were derived from an averaged literature value^{7,8} of 20.0 cal g^{-1} for the melting enthalpy of the crystal. The weight fractions of crystallinity were converted to volume fractions using the crystal and amorphous densities of 1.112 and 1.05 g cc^{-3} respectively. Melting points were measured at $40^\circ\text{C min}^{-1}$, at this speed only one peak is observed and characterizes the material as crystallized. Peak melting points are quoted since this is more representative of the average lamellar thickness as measured by SAXS.

TREATMENT OF SAXS DATA

The SAXS was measured up to $2\theta = 4^\circ$ for the sample both in the crystalline and quenched states. After subtracting the constant liquid scattering⁹ from each scan, the residual quenched sample scattering due to foreign material and parasitic scattering was then subtracted from the crystalline scattering.

Deviations from Porod's law caused by the presence of a diffuse amorphous-crystalline interphase boundary were examined in the high angle region of the SAXS curve. Ruland¹⁰ and Vonk¹¹ have shown that the intensity observed $\tilde{I}(s)_d$ is related to the intensity for the same system with an idealized sharp phase boundary $\lim_{s \rightarrow 0} \tilde{I}(s)_o = K/s^3$ by the expression

$$\lim_{s \rightarrow 0} \tilde{I}(s)_d = (1 - 2\pi^2 E^2 s^2 / 3) K/s^3 \quad (1)$$

where $s = (2\sin\theta)/\lambda$ and E is the width of the transition region. E may be obtained from the plot $\tilde{I}(s)s^3$ vs s^2 and the data may be corrected over the whole SAXS range to the idealized case. Idealized intensities were employed to calculate the correlation function for comparison with the theoretical correlation function derived from the simple two phase lamellar model given by Vonk et al.¹²

The total integral, $\tilde{Q}(s)$, was calculated from $\tilde{I}(s)_o$ and can be related to the electron density difference between phases, $\Delta\rho_e$, by employing a Kratky standard sample.

$$\tilde{Q}(s) = \int_0^\infty \tilde{I}(s)_o s ds = \frac{P_s I_e \lambda^3 N^2 d \phi(1 - \phi) \Delta\rho_e^2}{2\pi a} \quad (2)$$

where i_e is Thomson's factor, P_s is the sample attenuated intensity of the main beam, d is the sample thickness, ϕ is the volume fraction of crystalline phase and a is the sample to plane of registration distance.

Desmearing of the experimental SAXS was carried out as described by Dijkstra et al.¹³ Prior to calculating the correlation functions or comparing the experimental SAXS to theoretical curves the desmeared curve was "Lorentz" corrected by multiplying by s^2 as is discussed in the paper by Crist and Morosoff¹⁴

Since the long period, L , of the Lorentz corrected SAXS curve is only characteristic of the crystal thickness, C , plus the amorphous thickness A , it is necessary to use model calculations to determine C and A individually. The three frequently used models of Vonk¹⁵, Tsvankin/Buchanan⁵ and Hosemann⁶ will be compared here in their application to the i-PS/a-PS system. Other workers have applied the latter two models to i-PS as mentioned earlier. It is not the intention here to review these models but only to indicate pertinent differences and similarities. All three models are based on a stack of parallel alternating layers of crystalline and amorphous regions. In the Hosemann model the number of repeat units, N , in the stack is not taken to be infinity as in the Vonk and Tsvankin/Buchanan models.

The correlation function method involves calculation of the Fourier transform of the scattering curve and then fitting this with a theoretical correlation function (TCF) from the layer model. The TCF is given by equation 3

$$\gamma(r) = \frac{\phi}{1-\phi} \left[\frac{1}{\phi} \int_0^\infty (r_c - r) P_c(r_c) dr_c + P_{cac} + P_{cacac} + \dots = 1 \right] \quad (3)$$

where ϕ is the linear crystallinity and equals $C/(C+A)$, P_c and P_a are the thickness distribution functions of the crystalline and amorphous regions, P_{cac} and following terms are functions of convolution products of P_c and P_a . In this work both log normal and Gaussian distributions have been used for P_c and P_a . Since negative thicknesses are not permitted, a normalisation procedure was necessary to redistribute the Gaussian function, this is detailed elsewhere¹⁶ together with the curve fitting procedure.

The Tsvankin model¹⁶ as modified by Buchanan uses an exponential amorphous thickness distribution and a symmetric rectangular crystal thickness distribution. A trapezoidal electron density profile along the lattice is included to represent a linear density change from the crystalline to amorphous regions characterized by a diffuse boundary width E . Due to the type of distributions chosen the crystallinity is uniquely and directly related to the position and width of the scattering curve for a given dispersion of thicknesses and E . Buchanan established calibration curves for various dispersions of crystal thicknesses. In order to determine the linear crystallinity, and hence crystal and amorphous layer thicknesses, it is necessary to assume values for these parameters. However, of course, the theoretical scattering curve should be calculated for comparison with the experimental curve.

The model developed by Hosemann in its general form considers the scattering from a stack of N parallel lamellae. Unlike the Vonk and Tsvankin models this introduces additional scattering due to zero order scattering of the whole stack. This has been observed and generally increased with increasing content of a-PS. Limiting the size of the lamellar stack produces

some broadening of the whole scattering curve in addition to the zero order scattering at very low angles. The SAXS is given by equation 4

$$I(s) = \frac{1}{2\pi s^2} \left\{ \text{Re} \left(N \frac{(1-f_x)(1-f_y)}{1-f_x f_y} \right) + \text{Re} \left[f_y \left(\frac{1-f_x}{1-f_x f_y} \right)^2 \left(1-(f_x f_y)^N \right) \right] \right\} Z(s) \quad (4)$$

where f_x and f_y are the Fourier transforms of the thickness distributions of the crystalline and amorphous layers. A transition zone term $Z(s)$ introduced by Tsvankin¹⁷ is given by

$$Z(s) = \frac{1}{2\pi i s^2 E} |1 - \exp(-2\pi i s E)|^2 \quad (5)$$

In this work Gaussian distributions for the crystalline and amorphous thickness distributions have been used as given by

$$P(r) = \left(2 g^2 \bar{r}^2 \right)^{-1/2} \exp \left(-(r-\bar{r})^2 / 2 g^2 \bar{r}^2 \right) \quad (6)$$

where $\bar{r}$ is the average amorphous or crystalline layer thickness. Theoretical SAXS curves with these distributions have been shown to represent the shape of the SAXS from i-PS and i-PS/PPO blends.

The SAXS curves were calculated in this work for different crystallinities with a wide range of g_c , g_a , N and the experimentally determined value for E . After normalizing the theoretical curve to the experimental

curve on a total integral basis the deviation Δ was calculated and minimized.

$$\Delta = \int_{s_{\max}}^{s_{\min}} \frac{|I(s)_T - I(s)_E|}{I(s)_E} \quad (7)$$

RESULTS AND DISCUSSION

Crystallization of i-PS and its blends with a-PS was carried out from the quenched state to deliberately aid nucleation¹⁸ and thus complete crystallization. Optical microscopy showed the system to be completely volume filled with spherulites within 24 hours at 200°C. No significant change in the spherulite texture was observed on addition of 30% a-PS although higher concentrations have been shown in this work and elsewhere³ to cause coarsening of spherulite texture.

In this study i-PS was crystallized at 200°C primarily to avoid any possible misinterpretation due to the multiple melting peaks observed in DSC experiments. Only one peak is observed for i-PS and the i-PS/a-PS blends at a heating rate of 40°C min⁻¹ and this is due to the melting of the originally formed lamellae. Within the reproductibility of the instrument the melting curves and melting temperatures were identical throughout the blend composition studied as given in Table 1. This may be compared with similar observations by Keith and Padden³ and Boon and Azcue¹⁹ but differed from results by Yeh and Lambert²⁰ who observed a depression of optical melting points using lower molecular weights. For comparison purposes data on i-PS crystallized at 210°C is included. This in itself suggests that the size, perfection and thickness distribution of the i-PS lamellae are identical in the blends examined. A melting point depression would probably be observed if the lamellae were more imperfect, smaller or if there were specific mixing interactions in the blends. Although unlikely, there is, of course,

the possibility of these factors being exactly counteracted by a lowering of the lamellar surface free energy in the blends. Crystallinities were found to decrease on addition of a-PS by an amount corresponding almost exactly to the dilution of the i-PS with a-PS. Optical microscopy has shown the system to be completely volume filled with spherulites, therefore, the amorphous material may be between the lamellae or on a larger scale between the crystalline fibrils.

The SAXS from the blends was analyzed to examine the details of the lamellar structure. The desmeared and Lorentz corrected SAXS curves are illustrated in Figure 1. It is most striking that the peak position does not shift with addition of a-PS, this was also observed in the i-PS/PPO system¹. Increasing the amount of a-PS does not significantly alter the general shape of the curve, however, some broadening occurs indicating an increase in disorder in the lamellar stacking. Attempts were made to fit each of the three models described earlier to the scattering curves.

Application of the Tsvankin/Buchanan model through the calibration curves was attempted for pure i-PS. Employing the notation of Buchanan; the dispersion of crystallite lengths about their mean, β/α , was varied from 0.1 to 0.3 and the ratio of transition zone thickness to crystal thickness, ϵ , was set to 0.2. For the pure i-PS crystallized at 200°C the crystallinity from calibration curve analysis varied from 49 to 53% in the range $\beta/\alpha = 0.1$ to 0.3 which is in approximate agreement with the results of Berghmann's et al.⁴ using the same technique. While it might be thought that variations of crystallinity in this range could not be differentiated in view of Babinet's principle, this distinction is actually possible using

the Tsvankin technique (as can be seen from the calibration curves) because of the difference in the nature of the assumed distribution functions for the crystalline and amorphous regions. However, the complete theoretical scattering curves were subsequently calculated and a very poor fit to the experimental curve was obtained, the best fit being at $\beta/\alpha = 0.1$ as shown in Figure 2. Further experiments have shown that this model does not fit the SAXS from i-PS in the tested crystallization range of 170°C to 210°C and the calibration curve method gives completely erroneous values for the crystallinity.

The correlation function of the scattering curve of pure i-PS was broad and broadened further on addition of a-PS as shown in Figure 3. Theoretical correlation functions with Gaussian thickness distributions could only be fitted with very broad and unrealistic crystal distributions. Correlation functions based on log normal distributions for both the crystalline and amorphous thicknesses gave only a reasonable fit to the experimental data as illustrated in Figure 3. However, in the pure i-PS the crystallinity from this model (0.24) was much lower than the volume fraction of crystallinity from DSC (0.30) and thus probably incorrect. This may be simply understood as follows, the depth of the minimum in the correlation function is strongly decreased by three factors (i) decreasing crystallinity (below $\phi = 0.50$) as shown in Figure 4; and (ii) broad phase thickness distributions and; (iii) when there is a stacking disorder such as angular branching of lamellae or lamellar stacks of only limited size. The latter effect is illustrated in Figure 5 where the correlation function of scattering curves

from the Hosemann model are given as a function of N . Therefore, if there is a disordered stacking of lamellae, then fitting a theoretical correlation function based on Vonk's model will give an artificially lower crystallinity than reality.

In a previous publication from this laboratory² it was found for PCL/PVC blends that reasonable agreement was found between parameters determined using the Tsvankin and the Vonk models. We believe that this agreement was fortuitous and might not generally be expected. It should be pointed out that while predictions of lamellar spacing and crystallinity did agree, the calculated scattering curves did not, indicating the inadequacy of the Tsvankin approach.

We realize that there may be some uncertainty in the DSC values of crystallinity arising from uncertainties in the melting enthalpy and in drawing base lines which may cast doubt upon our conclusions based upon comparisons of DSC and SAXS values of crystallinity. However, our rejection of the infinite stack model is also based upon the following considerations:

a. The experimental scattering curves are very broad and dissimilar to any normally observed for semicrystalline polymers with known well stacked lamellae. Crist¹⁴ indicates that the ratio of peak position to peak width for most polymers is around 0.5. Larger values than this cannot be accounted for by the infinite layer model of Vonk, et al.

b. The very broad experimental correlation function can only be fitted by very broad and unrealistic log normal distributions of crystal thicknesses. For example, for the 70/30 blend a distribution characterized by $\beta_c = 0.7$

and $\phi = 0.20$ could be required. This specified a high population of lamellae thinner than $15\text{\AA}$ along with a high proportion of thicker lamellae ($>50\text{\AA}$). Such a distribution would give rise to an extremely broad melting curve which would further broaden with increasing concentration of a-PS if the results derived from the Vonk model are accepted (i.e. β_c increases from 0.33 to 0.7). However, the melting point range observed from the DSC peak width remains unchanged with increasing a-PS content (see Table I).

The results from these model studies, therefore, indicated that the Hosemann model would more closely represent the real system. Excellent agreement ($\Delta < 4\%$) was found between theoretical scattering curves based on Hosemann's model with Gaussian thickness distributions and the experimental curves as shown in Figures 6 and 7. If a value for N greater than 2 had been adopted, then Δ would have increased to 20-40% rather than the observed agreement of ($<4\%$). Thus the fit is very sensitive to the choice of N. The structural information obtained is given in Table 2.

The superstructure of the lamellar system in the i-PS/a-PPO blends appears to change very little throughout this blend composition range studied. Lamellar thickness has remained constant which, coupled with the constancy of the melting points indicates little or no difference in the individual lamella in i-PS and the blends. The interlamellar thickness is also unchanged. However, the value of N decreased from 2.2 in pure i-PS to 1.6 in the 70/30 blend indicating increasing disorder. This parameter N in the Hosemann model is strictly speaking the number of lamellae oriented parallel to each other. In the i-PS system N is very small and thus

indicative of a very disordered stacking probably due to small angle lamellar branching within a fibril. This will decrease the order in the stack without necessarily reducing the number of lamellae; N should at present be viewed only as a disorder parameter.

In our calculations using the Hosemann model, we have arbitrarily assumed that the electron density of the medium surrounding the stack of lamellae had a value characteristic of amorphous PS. Alternatively, one could have followed the procedure of Schultz²¹ who assumed a matrix density equal to that of the stacks. He demonstrated an appreciable effect of the assumed value of the matrix density on the scattering curves. We feel that the true value lies somewhere between these two extremes of electron density depending upon how much non-crystalline material is interfibrillar. If we had adopted the Schultz assumption, it would have required that we adopted even lower values of N than the almost unreasonably low values that we propose in order to fit the experimental data. However, we feel that the role of stacking disorder is of more significance than that of matrix density in leading to the broad scattering curves and the very low values of N .

Examining the crystallinities from DSC and the SAXS model analysis we observe an increasing difference between them as a function of blend composition. This is direct evidence that the atactic material is being rejected from the growing fibril and is forced between the fibrils which themselves contain the alternating layers of amorphous and crystalline regions. The internal morphology of the fibrils has been found not to change on addition of a-PS except for a slight increase in disorder. The agreement between the DSC and SAXS crystallinities for the pure i-PS crystallized

at 200°C suggest that the amorphous i-PS is interlamellae in this case. The lower SAXS crystallinity for the 210°C crystallized sample may indicate that some of the i-PS is interfibrillae at this temperature. It might be also noted that $\langle \eta^2 \rangle$ is also lower for the i-PS crystallized at 210°C than for the 200°C crystallized sample.

Additional confirmatory evidence for this is gained from the SAXS total integral (Eqn. 2, Table 2) which depends only on the total volume fraction of crystalline material present in the sample and the electron density difference between phases. This is in contrast to the shape of the scattering curve which is characteristic of localized structure. The total integral decreases with addition of a-PS in approximate proportion to the DSC crystallinity which is consistent with the proposed morphology.

Similar conclusions were reached by Keith and Padden in their study of isotactic and atactic mixtures of polypropylene and polystyrene using optical microscopy. In the case of i-PS/a-PS they found the a-PS to be dispersed on a finer scale than is the polypropylene in i-PP/a-PP blends and resulted in a fine textured spherulite. However, at high a-PS concentrations a coarsening of the spherulite could be observed optically at the spherulite growth front thus being indicative of large scale rejection of a-PS from the growing fibrils. Our quantitative results agree with their conclusions and we have also found that despite the slower growth rate in the blends the internal structure of the fibrils remains unchanged on addition of up to 30% atactic material.

These conclusions are essentially in agreement with those of Yeh and Lambert²⁰ based upon kinetic studies. They observed variations in spherulite growth rate with the molecular weight of narrow distribution a-PS in i-PS/a-PS blends and noted an increase in growth rate with increasing molecular weight occurring in the range of $20-50 \times 10^3$. Their interpretation was that above this molecular weight range "significantly greater amounts of non-crystallizable rejected impurity are trapped within the growing spherulites". They further conclude that "the trapped impurity probably resides between the spherulite fibrils and not within the crystalline domains". The implication is that below this molecular weight range the a-PS is excluded from the spherulite and must concentrate in interspherulitic rather than interfibrillar regions. It is noted that the a-PS molecular weight studied here was above the range of this transition region. It is expected that this would result in a) departure from constancy of spherulite growth rates, b) cessation of crystallization before spherulites become volume filling and c) a decrease in zero-order SAXS due to the reduction of the interfibrillar a-PS contribution. It is also expected that another high molecular weight transition region might occur above which the a-PS may reside in interlamellar regions due to a decrease in the diffusion coefficient, D , in the Keith δ parameter. This should lead to a break in a curve of interlamellar spacing vs. molecular weight (at constant a-PS/i-PS ratio). Such studies are currently in progress²². It is possible that another break may occur in the spherulite growth rate vs. molecular weight plot in this region.

The morphology of the blends will, of course, be temperature, composition and molecular weight dependent, but the scale of rejection of non-crystallizable material may be predicted by the equation $\delta = D/G$. In this respect it is interesting to briefly mention the i-PS/PPO system where the non-crystallizing PPO is distributed between individual lamellae and not isolated on a larger scale. This suggests qualitatively that addition of PPO ($T_g \approx 210^\circ\text{C}$) to i-PS predominantly decreases the diffusivity term such that segregation occurs at the lamellar level. It is difficult to predict the level of segregation in blends of different compatible polymers and such generalizations from system to system should be made only with caution.

CONCLUSIONS

The constancy of the long period and melting point indicates that the fibrillar structural unit is unchanged on addition of up to 30% a-PS to i-PS. In addition the decrease in overall crystallinity, therefore, shows segregation of the a-PS to the interfibrillar region of the spherulite. These results agree well with the qualitative findings of the Kieth and Padden with respect to the increasing coarseness of texture on addition of atactic material to the crystallizable isotactic polymer.

In the i-PS/a-PS blends the SAXS could not be represented by the Vonk model using Gaussian or log normal thickness distributions due to the large degree of disorder in the lamellar stacking. The Tsvankin/Buchanan model did not fit the shape of the experimental SAXS curves and gave erroneous crystallinities from the calibration curve technique. However, the Hosemann model using Gaussian distributions could represent the SAXS curves within experimental error. The results showed an increase in disorder of the lamellar stacking on increasing the a-PS content despite the lamellar and interlamellar thicknesses remaining unchanged.

TABLE 1

MELTING POINT DATA FOR I-PS/A-PS BLENDS

BLEND I-PS/A-PS	CRYST'N TEMP. °C	T _M PEAK	1. ΔT °C	W _c	φ _c
100/0	200	223.3	8.8	0.307	0.295
90/10	200	223.7	9.5	0.278	0.266
80/20	200	223.5	8.5	0.252	0.241
70/30	200	223.2	9.2	0.199	0.190
100/0	210	229.2	8.0	0.274	0.263

1. Width of the melting peak.

TABLE 2

STRUCTURAL INFORMATION GAINED FROM THE HOSEMANN ANALYSIS
FOR I-PS/A-PS BLENDS

BLEND I-PS/A-PS	CRYST'N TEMP. °C	ϕ	$C(\text{\AA})$	$A(\text{\AA})$	N	g_c	g_a	$E(\text{\AA})$	$\langle \eta^2 \rangle^{1/2}_{\text{gcm}^{-3}}^*$
100/0	200.0	0.30	48.0	112.0	2.2	0.11	0.30	21.0	0.0372
90/10	200.0	0.32	47.5	100.9	1.7	0.22	0.24	21.0	0.0366
80/20	200.0	0.31	47.5	105.7	1.7	0.18	0.26	21.2	0.0338
70/30	200.0	0.30	46.0	110.2	1.6	0.14	0.26	21.2	0.0329
100/0	210.0	0.30	51.5	120.2	2.4	0.3	0.3	14.4	0.0328

*These values are expressed in this table in units of fluctuation in actual density rather than electron density.

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CAPTIONS FOR FIGURES

1. Desmeared and Lorentz corrected SAXS from i-PS/a-PS blends of various compositions.
2. A Comparison of the experimental desmeared and Lorentz corrected SAXS data on an isotactic polystyrene sample with the best fit obtained using Tsvankin theory.
3. Experimentally obtained correlation functions (line) for i-PS and a 70/30 i/a PS blend with the best theoretically calculated fit (points) using log normal distributions.
4. The effect of crystallinity (ϕ) on the correlation function with no dispersion of phase thickness.
5. Correlation functions obtained using the Hosemann model obtained using $g_c = g_a = 0.15$ and $N = 100$ (—) and $N = 1.6$ (- - -).
6. A comparison of the experimentally determined desmeared and Lorentz corrected SAXS from isotactic polystyrene (—) with the best Hosemann model fit (- - -).
7. A comparison of the experimentally determined desmeared and Lorentz corrected SAXS from a 70/30 i/a PS blend (—) with the best Hosemann model fit (- - -).

Fig 1

Fig 2

Fig 3

Fig 4

Fig 5,

Fig 6