Rheological and Mechanical Study of Polypropylene Ternary Blends for Foam Application

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ABSTRACT

Melt strength of polymer is one of the main primaries for the success of fabricating low-density extrusion foam and thermoforming process. Polypropylene (PP) as a linear polyolefin has low melt strength for bubble stabilization and sagging. Large molecules and long chain branches of PP cause higher molecular entanglements and increase elongational strength of PP melt. In this paper the rheological behaviour, melt elasticity, mechanical properties, and crystallinity of linear PP blends are studied and compared with blends containing a long chain branched PP resins. It is found that ternary blends of linear PP improve the melt elasticity in a certain composition, comparing to binary blends. Branched PP resins increase molecular entanglements, which it leads to higher melt elasticity. The results of this study help to understand the effects of chain size and chain architecture in increasing the melt strength and melt drawability of PP blends. These are the most important factors for producing low density PP foam and high quality thermoformed products.

Key Words:
melt strength; branched polypropylene; melt drawability; molecular entanglements; melt creep compliance.

INTRODUCTION

Polypropylene (PP) is one of the most widely used thermoplastics. High melting point, excellent chemical resistance, and acceptable range of mechanical properties together with low price are the main advantages that make it useful for many applications [1, 2].

The high degree of crystallinity of standard PP contributes to poor melt strength and low melt drawability which are not adequate for a number of applications, for instance, foaming and thermoforming [3, 4].
There are two different aspects in melt strength study of resins subjected to deformation. The first aspect is the resistance (strength) of the melt to the applied force, that is a measure of melt strength. The second one is the amount of induced strain in the melt because of the applied stress, which is called melt drawability. Melt drawability is a function of melt strain under tensile force. When melt is subjected to tensile forces, the melt strength is usually taken for melt elongational strength and melt extensibility as well.

Due to its Ziegler-Natta catalysis system, PP is a totally linear molecule with about 60% crystallinity [5]. The common linear semi-crystalline PP resins have low shear and extensional viscosity above their melting temperature.

If the melt strength and melt drawability are very weak, the thin wall of PP melt, in the foam cells walls, will not be strong enough to bear the applied extensional forces due to expansion, and it ruptures easily.

There are some known pathways, to increase the melt strength of PP as follows [6]:
- Increasing average molecular weight
- Broadening molecular weight distribution
- Introducing long chain branched structure on PP
- Cross-linking

Molecular entanglements, because of side branches, give rise to an improved strength and drawability of the melt as known for low-density polyethylene. Various research works have been conducted to compensate for PP weak melt strength.

Alteepping and Nebe [7] used a binary blend of two various grades of PP with different weight average molecular weights (MW) of 1,800,000 and 280,000 and polydispersity of 2.5 and 5, respectively. The composition of blend comprises a major proportion of the low viscosity PP (MW = 280,000) component having a melt viscosity of less than 2 \(10^3\) Poise (at 190 C and shear rate 1000 s\(^{-1}\)) and a minor proportion of the high viscosity PP component having a melt viscosity greater than 2.5 \(10^3\) Poise, at the same condition.

Using resins with bimodal [8] molecular weight distribution yield higher melt strength and also higher melt extensibility. The major component of these resins is largely linear while the minor component with higher molecular weight is highly branched.

The main known procedure for having higher melt strength and melt extensibility is to introduce long chain branched on PP molecule [3,5]. It is found that, high melt strength PP resins and long chain branched ones, have improved foams and their thermoform abilities [8, 9].

In the past few years different companies have made a lot of effort to evaluate the possibilities and potentials of polypropylene's post reactor modification [10]. Recently, various grades of long chain branched PP (lcb PP) have been presented to the market [11].

The effect of molecular size, molecular weight distribution and long chain branched PP on melt strength and extensional behaviour is quite different. Broadening the molecular weight distribution, leads to an improvement in melt strength but has not any significant effect on melt drawability. Having almost the same width of molecular weight distribution, the higher the average molecular weight (low melt flow rate), the higher the melt strength.

By introducing or creating long chain branched on polypropylene backbone, entanglement of branched chains give rise to improvement of melt extensibility. Because of processing limitation, pure lcb PP cannot be used.

In recent years, some efforts have been made to blend long chain branched and linear injection grade PP, in binary blends [12]. High molecular weight linear PP resins are more preferred to increase the molecular entanglements. On the other hand, it leads to increase the viscosity of blends. In this study, the ternary blends of different grades of PP were investigated.

As the mechanical properties of resins are important for final strength of products, the tensile properties and impact strength of blends are measured.

Viscosity as well as melt strength, is the response of polymer melts to applied stress. Shear viscosity is the main factor on mixing and processing of resins. Elongational viscosity and elastic behaviour of polymer melt are the key properties in foaming process. To understand the above mentioned behaviour, the rheological properties of the blends are investigated.

**EXPERIMENTAL**

**Materials**

The materials used in this study have four different grades of standard polypropylene impact copolymer, which are shown in Table 1. The PP1 is a standard injec-
tion grade of PP with high melt flow index (MFI). The MFI (ISO 1133, 230°C, and 2.16 kg) of this linear PP is 7 g/10 min. Two low MFI grades (PP2 and PP3) are used as high molecular weight resins with different distributions [13]. The lcb PP used in this study is Profax PF814 which is supplied by Basel Polyoolefin Co.

To study the effect of blend composition on melt strength and melt extensibility, two sets of blends were prepared. In the first set of experiments, three different grades of PP with various MFIs and molecular weight distributions (PP1, PP2, and PP3) were blended in melt state in a twin screw extruder. The recipes of this set of blends and their MFIs are presented in Table 2.

To study the effect of long chain branched effect on rheological properties, a linear PP is replaced by Profax PF814 in ternary blends. The recipes of these blends together with their MFIs are shown in Table 3.

Procedure
The polypropylene pellets with 0.8% talc as nucleating agent were the first dry blended according to recipes of Table 2 and 3. No antioxidant additives are used in the formulations. The physically blended mixtures were fed into a Dr. Collin twin screw extruder (120 rpm). The temperature profile of 195 to 220°C is set from barrel to die exit. The die used was a 3 mm round die and the strand was changed to cylindrical shape granules in a strand pelletizer.

The strength of a layer of polymer after solidification is mainly determined by the tensile strength, modulus, impact strength, and crystalline structure of polymer matrix. In order to understand the role of compositions on the final properties of blends the mechanical properties and crystallinity of blends were studied.

The samples for tensile and Izod notched impact test were prepared by injection moulding process in a standard mould. The temperature set on injection moulding machine was 190°C to 230°C from hopper to nozzle.

The tensile test is carried on a Zwick Z020 universal testing machine, on 5 sample bars (type I) at 50 mm/min according to ASTM D638 test method. The impact strength of blends were measured, at three various temperatures of 23°C, 0°C and -20°C, according to ASTM D 256A, by a Zwick/Roell B5102 pendulum impact tester, on five samples.

To study the flow behaviour of blends, the MFIs of linear PP blends were measured by Ceast Junior melt flow indexer, at 230°C, 210°C, and 190°C with 2.16 kg load.

An Instron capillary testing machine at 220°C and different shear rates was used to measure the shear rheological properties of the blends. The capillary barrel

### Table 1. Experimental raw materials.

<table>
<thead>
<tr>
<th>Grade name</th>
<th>Code</th>
<th>Description</th>
<th>MFI (g/10min) (230°C, 2.16 kg)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Moplen EPC40 R</td>
<td>PP1</td>
<td>Standard injection grade</td>
<td>7</td>
</tr>
<tr>
<td>Moplen EPD60 R</td>
<td>PP2</td>
<td>Low MFI pipe grade</td>
<td>0.3</td>
</tr>
<tr>
<td>Moplen EPO30 RF</td>
<td>PP3</td>
<td>High melt strength for blow moulding</td>
<td>0.8</td>
</tr>
<tr>
<td>Profax PF814</td>
<td>lcb PP</td>
<td>Long chain branched PP</td>
<td>2.5</td>
</tr>
</tbody>
</table>

### Table 2. Melt flow index of linear PP blends at 190°C, 210°C, 230°C.

<table>
<thead>
<tr>
<th>Blend No.</th>
<th>Blend desig.</th>
<th>PP1 (%)</th>
<th>PP2 (%)</th>
<th>PP3 (%)</th>
<th>MFI (g/10min)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>190°C</td>
</tr>
<tr>
<td>B</td>
<td>R-2</td>
<td>50</td>
<td>50</td>
<td>-</td>
<td>0.7</td>
</tr>
<tr>
<td>1</td>
<td>R-H-2</td>
<td>50</td>
<td>35</td>
<td>15</td>
<td>0.8</td>
</tr>
<tr>
<td>2</td>
<td>H-R-1</td>
<td>50</td>
<td>25</td>
<td>25</td>
<td>1.02</td>
</tr>
<tr>
<td>3</td>
<td>H-R-2</td>
<td>50</td>
<td>15</td>
<td>35</td>
<td>1</td>
</tr>
<tr>
<td>E</td>
<td>H-2</td>
<td>50</td>
<td>-</td>
<td>50</td>
<td>0.93</td>
</tr>
</tbody>
</table>

### Table 3. Melt flow index of blends with lcb PP at 190°C, 210°C, 230°C.

<table>
<thead>
<tr>
<th>Blend No.</th>
<th>Blend desig.</th>
<th>PP1 (%)</th>
<th>PP2 (%)</th>
<th>lcb PP (%)</th>
<th>MFI (g/10min)</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td>(230°C, 2.16 kg)</td>
</tr>
<tr>
<td>4</td>
<td>60-B-2</td>
<td>50</td>
<td>35</td>
<td>15</td>
<td>2</td>
</tr>
<tr>
<td>5</td>
<td>B-60-1</td>
<td>50</td>
<td>25</td>
<td>25</td>
<td>2.4</td>
</tr>
<tr>
<td>6</td>
<td>B-60-2</td>
<td>50</td>
<td>15</td>
<td>35</td>
<td>3.5</td>
</tr>
</tbody>
</table>
diameter was 9.52 mm, capillary die of 1.4 mm and capillary length of 53.3 mm.

Since the deformation of thin layer of melts stops just at the moment of crystallization, the crystallization behaviour of blends is important to understand. The crystallization behaviour of the blends was investigated using a PL differential scanning calorimetry (DSC) at 10 °C/min heating and cooling rate.

The viscoelastic characterization of PP blend’s melt was conducted with a Paar Physica Modular Compact Rheometer MCR300, with TC30 temperature control unit. To prepare the sample for this test, resin pellets were compression moulded into sheets of 1.2 mm in thickness. Samples were stamped out with a 25 mm diameter circular die.

Two different approaches are used to characterize the viscoelastic responses of polymer melts by this equipment. In the first approach, the steady state (equilibrium) compliance of melts was measured using a rotational rheometer at a constant stress of 100 Pa at 210 °C. The creep data were obtained in a time period of 0-150 s.

The next approach is based on measuring the loss tangent (tan δ) of material resulting from an induced oscillatory strain on a molten polymer. Once the sample is heated to desired temperature, one of the parallel plates rotates relative to the other one with oscillation frequency of 1 rad/s. Maximum strain on the sample was limited to 10%, to reduce strain-induced effects on the results.

The loss tangent of the blends was calculated by measured values of G’ and G” at 190 °C.

**RESULTS AND DISCUSSION**

The tensile properties and impact strength of blends are summarized in Tables 4 and 5.

Figure 1 shows the impact strengths of linear PP blends at 23 °C, 0 °C, and -20 °C. It is observed that, the impact strength of linear PP blends is a direct function of PP2 content. The impact strength increases by increasing the amount of PP2 in the blends. Nevertheless, the impact strength data of blends at 23 °C are greater than impact strength at 0 °C and -20 °C, but the trend of changes shows similar behaviour at all temperatures.

H-R-1 (Blend with 50% PP1, 25% of PP2, and 25% of PP3) showed higher impact strength than what was expected (Figure 1). This might be due to lower crystallinity of the blend. The molecular interaction in this certain percentages of composition affects crystallinity of the resins. The effect of various polymer phases on morphology should be similar to known effect of low molecular weight additives on the crystalline structure of the blends [14]. Blending affects the spherulite sizes,
intercrystalline links between lamellas, and the interactions between polymer phases in the boundaries. It is expressed in the experimental reports that, addition of a minor amount of another polymer greatly affects the morphologies, surface characteristics [15], and mechanical properties [16].

To investigate the crystallinity of the blends, all the samples (Table 2) are tested by using differential scanning calorimetry at the heating rate of 10°C/min. The samples were heated to 250°C, treated 5 min in this temperature then cooled down with the same rate to room temperature, under nitrogen blanket. Normally lower cooling rate of 5°C/min is used to study the crystallinity of resins. Referring to fast cooling rate in real processing condition in thin walls of foam or films, and as a basis for comparison, the cooling rate of 10°C/min is selected for all samples. The crystallinity test results are summarized in Table 6. The heat of crystallization of blend No. 2 (H-R-1) is -11 mcal/mg which is lower than the other linear PP blends. The crystallization peak of blend No 2 (Figure 3) is shorter and starts to rise up at lower temperature. This is due to interaction between polymer phases in the ternary blend of H-R-1, which suppresses the crystallization, at these percentages of compositions. These results interpret the higher impact strength of H-R-1.

By changing the ratio of each PP in the ternary blends, the rate and the specific heat of crystallization will also increase, as we have observed in blend H-R-2 (No. 3). By increasing the crystallinity of resins, the impact strength will decrease, and the elastic modulus has to change accordingly. The tensile properties of blends were summarized in the Table 5. However, there is an evidence of lower elastic modulus for blend No. 2 (H-R-1) that can interpret the impact data, but the difference is not big enough to cover the test error limits.

Using long chain branched PP, showed a distinct effect on impact strength of blends. It is observed that lcb PP leads to a sharp decline in the impact strength curve (Figure 2), which is more obvious at room temperature. Blend No. 2 (H-R-1) and blend No. 5 (B-60-1) have the same composition, where linear PP2 in H-R-1 is replaced by lcb PP in B-60-1. The impact strength of H-R-1 and B-60-1 are 22.5 and 16 (kJ/m²), respectively which means 30% lower impact strength for blends containing lcb PP. This is due to lcb PP molecular structure. The lcb PP is substantially an isotactic homopolymer that long chain branches are created on the backbone via a post reactor modification. The impact strength of PP homopolymers is lower than PP copolymers. The long chain branches have no major effect on impact strength of lcb PP resins.

### Table 6. Crystallization study of blends, DSC,10°C/min.

<table>
<thead>
<tr>
<th>Blend No.</th>
<th>Blend design</th>
<th>Start of crystallization (°C)</th>
<th>Heat of crystallization (mcal/mg)</th>
<th>Melting point (°C)</th>
</tr>
</thead>
<tbody>
<tr>
<td>B 1</td>
<td>R-2</td>
<td>119</td>
<td>-14</td>
<td>166.5</td>
</tr>
<tr>
<td>1</td>
<td>R-H-2</td>
<td>124</td>
<td>-13</td>
<td>167.5</td>
</tr>
<tr>
<td>2</td>
<td>H-R-1</td>
<td>118.5</td>
<td>-11</td>
<td>167.5</td>
</tr>
<tr>
<td>3</td>
<td>H-R-2</td>
<td>126</td>
<td>-20</td>
<td>167</td>
</tr>
<tr>
<td>E</td>
<td>H-2</td>
<td>118</td>
<td>-15.6</td>
<td>168</td>
</tr>
</tbody>
</table>

### Figure 2. Impact strength of ternary blends of lcb PP.

### Figure 3. DSC Crystallization study of linear PP ternary blends.
Rheology of Blends

Linear polymers have low shear and extensional viscosities. The shear viscosity is more important in mixing. The MFI as an indication of shear viscosity is measured. The MFI (230°C and 2.16 kg) of both sets of blends (linear PP and lcb PP containing blends) are shown in the Table 2 and 3. The PP1 content is 50% for all blends. The variation of MFIs at different percentages of composition is shown in Figures 4 and 5. The melt flow index increases by increasing the percentage of component having higher MFI. To follow the effect of temperature on melt flow behaviour, the MFIs of linear PP blends (Table 2) are measured at 190°C, 210°C, 230°C, and 2.16 kg load.

The inverses of MFIs as indications of viscosities were calculated and plotted versus temperature (Figure 6). It is observed that the slope of the line (log MFIs versus temperature) for H-R-1 ternary blend is lower than the slope of the lines of the other two binary blends. This is due to lower sensitivity of viscosity of this blend to temperature comparing to binary blends.

For detail characterization and analyzing the melt behaviour of the blends under shear, all of the formulations are tested with capillary rheometer at 220°C and various shear rates. The apparent shear viscosities of No. 1, 2, and 3 formulations of linear PP blends, versus shear rate are shown in Figure 7.

At higher shear rates the viscosity of all blends are nearly the same, whereas at low shear rate (lower than 100 s⁻¹). The formula No. 2, displayed distinct behav-

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**Figure 4.** Melt flow index trend of PP ternary blends with HMS 230°C.

**Figure 5.** Melt flow indices' variation of PP ternary blends with lcb PP at 230°C.

**Figure 6.** Effect of chain branch length on viscosity.

**Figure 7.** Capillary shear viscosities of PP ternary blends with different percentages of HMS PP.
The MFI s of blends No. 1, 2, 3 at 230°C and 2.16 kg are 2.1, 2.3, and 2.5 (g/10min), respectively.

The viscosity curve of formula No. 2 (H-R-1), versus temperature, at lower shear rates, has to have sharper slope than the viscosity curve of formula No. 3. It has lower apparent viscosity at lower shear rate comparing to the blend with higher MFI. This is due to different molecular interactions of various PP molecules, which changes the flow behaviour of the melt in shear mode.

The apparent shear viscosity of lcb blends are shown in Figure 8. There is no evidence of unusual behaviour of shear viscosities and no distinct effects of long chain branches.

Mutual solubility of polyolefin turns out to be highly dependent on the details of the chemical architecture of polymer chains, especially the degree, length, and location of side groups [17].

Polypropylenes have the same molecular structure, and can be blended at higher viscosity ratio of blend components.

Viscoelastic Study of the Blends

The melt (tensile) strength and melt extensibility of polymers are important in processes where the nature of deformations is substantially elongational. High molecular weight polypropylene resins are frequently characterized as high melt strength (HMS) resins, however, unexpectedly, it has been found that this characterization is inadequate to interpret the processing behaviour and foaming of PP resins [8].

The viscoelastic behaviour can be estimated by different methods [16, 17]. In this study, it is measured by dynamic mechanical spectroscopy.

The damping factors of melts, are summarized in Table 7. The change of tan δ at various percentages of PP2 and lcb PP are plotted in Figure 9. The tan δ curve shows a minimum for blend No. 2 (H-R-1) of linear PP blends and also blend No. 6 (B-60-2) of the ternary blends containing lcb PP.

Since it is desirable to have a resin with relatively greater melt elasticity, a resin with lower tan δ is preferred. This study reveals that PP ternary blends, improved the tan δ compared with binary blends.

The different size of molecules at a certain percentage of components enforces specific molecular interactions that suppress the molecular movement. This is the reason of lower heat of crystallization and higher impact strength of H-R-1 ternary blend.

<table>
<thead>
<tr>
<th>Blend No.</th>
<th>Blend desig.</th>
<th>Damping factor-tan δ</th>
</tr>
</thead>
<tbody>
<tr>
<td>B</td>
<td>R-2</td>
<td>2.35</td>
</tr>
<tr>
<td>1</td>
<td>R-H-2</td>
<td>2.33</td>
</tr>
<tr>
<td>2</td>
<td>H-R-1</td>
<td>2.31</td>
</tr>
<tr>
<td>3</td>
<td>H-R-2</td>
<td>2.32</td>
</tr>
<tr>
<td>E</td>
<td>H-2</td>
<td>2.32</td>
</tr>
<tr>
<td>4</td>
<td>60-B-2</td>
<td>2.08</td>
</tr>
<tr>
<td>5</td>
<td>B-60-1</td>
<td>1.96</td>
</tr>
<tr>
<td>6</td>
<td>B-60-2</td>
<td>1.87</td>
</tr>
<tr>
<td>7</td>
<td>B-60-3</td>
<td>1.9</td>
</tr>
<tr>
<td>-</td>
<td>PROFAX PF814</td>
<td>1.4</td>
</tr>
</tbody>
</table>

| Figure 8. Capillary shear viscosities of PP ternary blends with different percentages of lcb PP. |
| Figure 9. Damping factor of PP ternary blends at 190°C and 1 rad/s. |
It is found that (Figure 9) ternary blends containing a lcb PP, show a minimum tan δ at a certain percentage of components. High molecular weight PP together with lcb PP enhances many friction sites that can resist against small deformation. The short arm entanglements of linear PP cannot withstand higher deformation any more. In this regard long chain branches will play an efficient role. Due to this, PP resins with strain hardening characteristic is preferred for large deformations [20]. From another point of view, the lcb PPs show lower resistance in small deformation and need to be strained before hardening. Therefore, blending with high molecular weight linear PP improves this weak point of branched PPs.

As an evidence to check the damping behaviour of blends, steady-state compliance of blends are measured by creep test. The creep compliance $J(t)$ is given by [21, 22]:

$$J(t) = \frac{Y(t)}{\delta_0} = J_{e0} + \left(\frac{t}{\eta_0}\right)$$

where:

$Y(t) = \text{strain}$; $\delta_0 = \text{stress}$; $J_{e0} = \text{equilibrium creep compliance}$; and $\eta_0 = \text{zero shear viscosity}$.

The strain as a function of time is divided by the stress to give $J(t)$. $J_{e0}$ is the intercept of the $J(t)$-time curve. Figure 10 and Figure 11 show the $J(t)$ versus time for blend H-R-1 (ternary blend of Linear PP) and blend No. 6 (B-60-2), containing long chain branched PP polymers.

Equilibrium creep compliance $J_{e0}$ is a function of melt elasticity. Higher $J_{e0}$ means higher elasticity of polymer melts. The results of this test are surveyed in Table 8. The equilibrium creep compliances $J_{e0}$ of both sets of blends versus blend compositions are plotted in Figure 12, which confirm the test results of tan δ.

Test results show that, lcb PP improves the melt elasticity of blends. Blend No.6 (B-60-2), displayed higher compliance than that was expected. This is due to mutual interaction of high molecular weight PP and lcb PP. It is observed that, the long chain branched PP is sensitive to melt fracture. In the capillary test of lcb PP the uneven surface of extrudate visibly addressed us to the melt fracture phenomena in nearly all ranges of

![Figure 10. Creep compliance versus time for blend H-R-1 (No. 2).](image1)

![Figure 11. Creep compliance versus time for blend B-60-2 (No. 6).](image2)

<table>
<thead>
<tr>
<th>Blend No.</th>
<th>Blend designation</th>
<th>Equilibrium creep compliance $J_{e0} \times 10^{-4}$ (1/Pa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>B</td>
<td>R-2</td>
<td>1.5</td>
</tr>
<tr>
<td>1</td>
<td>R-H-2</td>
<td>2</td>
</tr>
<tr>
<td>2</td>
<td>H-R-1</td>
<td>3</td>
</tr>
<tr>
<td>3</td>
<td>H-R-2</td>
<td>2</td>
</tr>
<tr>
<td>E</td>
<td>H-2</td>
<td>2</td>
</tr>
<tr>
<td>4</td>
<td>60-B-2</td>
<td>6</td>
</tr>
<tr>
<td>5</td>
<td>B-60-1</td>
<td>7</td>
</tr>
<tr>
<td>6</td>
<td>B-60-2</td>
<td>10</td>
</tr>
<tr>
<td>7</td>
<td>B-60-3</td>
<td>9</td>
</tr>
<tr>
<td>-</td>
<td>PROFAX PF814</td>
<td>14</td>
</tr>
</tbody>
</table>
shear rates. Baik and Tzoganakis[23] have been observed from their study on controlled rheology of PP that, the extrudate distortion has decreased with increasing the shear rate of a given material and processing condition.

Sammler et al. [24], have studied melt fracture for high melt strength PP. The presence of this melt fracture is expected from the strong elastic components observed in the extensional rheology and the long chain branching is the most probable molecular origin of spurtting melt fracture in the branched PP resin.

The rheological studies of lcb PP ternary blends have shown that, higher lcb PP content improves the melt elasticity indexes and decrease tan $\delta$. On the other side, high level of lcb PP intensifies the melt fracture. On the other hand, the higher molecular weight PP resins are preferred as a couple for lcb PP in the blends to increase the molecular entanglement. The low MFI of high molecular weight linear PPs limits the usage of this type of PP because of processing limitations. To be able to tune the MFI of blend in certain limits, a linear PP with higher MFI (PP1) was used as a third member of the blends. The preferred MFI for the extrusion of PP foam is considered 2.5 to 3.5 g/10 min.

We believe that blends of high molecular weight linear PP together with lcb PP produce molecular entanglements and many friction sites that resist against slippage of molecules on each other and it leads to improved melt elasticity.

CONCLUSION

The crystallinity, mechanical, rheological, and viscoelastic properties of binary and ternary blends of PP resins containing a broad and high molecular weight PP and a long chain branched PP are studied and reported. The followings are concluded:

- The ternary linear PP blends have shown better melt elasticity and are preferred to binary blends.
- An optimum percentage of components for linear PP ternary blends is found which has a distinct effect on mechanical properties, crystallinity, and melt elasticity.
- The composition of ternary blends, containing a long chain branched PP can be formulated according to processing and properties requirements.
- Blending of standard PP having high melt strength, together with long chain branched PP leads to many sites for entanglements, and need much energy to be extended. High molecular weight PP have low MFI and is not easy to use in some processes.
- To tailor made the formulation for a certain application and to balance the processing requirements, a linear PP with lower MW (higher MFI) is used as the third member of the blends.
- Long chain branched PP is sensitive to melt fracture and because of processing limitations it cannot be used as pure form.
- Blending of long chain branched PP with linear PP, facilitates and broaden the processing windows.
- An engineering basis is developed to balance the composition percentages of PP ternary blends with improved melt elastic properties.

REFERENCES