CHAPTER 3

LINSEED VINYL ESTER FATTY AMIDE MODIFIED EPOXY LAYERED SILICATE NANOCOMPOSITES

3.1 INTRODUCTION

Epoxy resins are an important class of thermosetting materials with unique combination of properties that can be tailored according to the type of curing agent, modifier utilized. The cured networks show evidence of high modulus, low creep, good adhesion, physico-mechanical and corrosion/chemical resistance properties as well as good performance at high temperatures for other applications such as adhesives, lubricants, industrial tooling, reinforced plastics, paints and coatings. Beyond all these good properties, the major drawback of epoxy resins that they contract and develop internal stress on curing. These internal stresses affect the weather resistance, chemical resistance; mortify their fracture energy, hydrophobicity and impact strength.

The modification of epoxies with polystyrene (PS), polymethyl methacrylate (PMMA), polyether ether ketone (PEEK), Polyether sulfone (PES), polyurethane (PU) (Guo Yang et al 2007, Yamanaka et al 1989 and Hourston 1992) liquid rubber, siloxane based polymer, hydroxyl terminated polybutadiene, polyesters, and liquid elastomers dates back to few years. In earlier work, an improvement in thermo-mechanical, dielectric and aging characteristics have been achieved by the introduction of bismaleimide into
siliconized epoxy (Ashok Kumar et al 2001 and Ashok Kumar. A et al 2002), siloxane modified epoxy (Suresh Kumar. R et al 2006) and unsaturated polyester modified epoxy resin (Dinakaran K et al 2002). In the present study an attempt has been made to improve the thermomechanical and insulating behavior by partial replacement of commercial epoxy resin with bio based linseed vinyl ester fatty amide and organically modified organoclay.

In recent years, polymers developed from renewable sources have gained the centre of attraction for its economic, environmental, and societal reward. Bio-based resin systems can be reinforced with nano clays to obtain novel, value added applications for natural polymers. Bio-based resins are defined as a combination of primary constituent namely petroleum based resin and a secondary constituent namely natural bio-resin. These resins exhibit improvement in toughness which is due to the reduction in cross–link density in the system, paving way for increased plastic deformation. However, this increase in toughness gravely impinges the modulus, thermal and barrier properties of the resulting polymer. This implies that stiffness and toughness are contrasting performance parameters and hence an apt weighing scale is required to develop an efficient biocomposites. Moreover, research has evidenced that the vegetable oil modified polymeric materials do not show passable properties of rigidity and strength for load-bearing applications by themselves and hence require modification. One of the approaches to recover the vanished properties of bio-resin is by the addition of layered silicate or nanoclay.

In the present work, an attempt has been made to develop DGEBA epoxy modified vinyl ester by blending, in different formulations by weight (DGEBA/linseed vinyl ester fatty amide). Linseed oil vinyl ester fatty amide
is obtained by condensation of acrylic acid and N, N-bis (2-hydroxyethyl) linseed amide (HELA) obtained from linseed oil. In linseed vinyl ester reactive sites are positioned only at the end of the molecular chain, therefore cross linking can take place only at the chain ends. The long flexible aliphatic fatty amide chain absorbs shock loading, subjecting toughness and resilience to the resulting composites. The formation of linseed vinyl ester is confirmed by IR studies. DGEBA/linseed vinyl ester fatty amide blends were cured by diaminodiphenylmethane (DDM) in appropriate stoichiometry and castings of the composites were prepared. These composite sheets were investigated by physico-mechanical, thermal (TGA, DSC) and X-ray diffraction. The Data resulted from different studies indicated that HELA can be used as sustainable resource based environment friendly reactive modifier for epoxy resin. The varying percentages of organo modified clay were incorporated into the polymer matrix in order to enhance the thermo mechanical and dielectric properties.

3.2 FABRICATION OF BIO BASED EPOXY MATRICES

DGEBA epoxy and linseed vinyl ester fatty amide were mixed in predetermined ratio 90/10, 80/20, and 70/30 by weight to obtain their blends and are referred to as E$_{90}$L$_{10}$, E$_{80}$L$_{20}$ and E$_{70}$L$_{30}$ respectively. Each of these samples was mixed by continuous agitation over magnetic stirrer for 24 h and a stoichiometric amount of curative 4,4'-diaminodiphenylmethane (9.72g) corresponding to epoxy equivalents was also added. The product was subjected to vacuum to remove the trapped air and then cast and cured at 120°C for 3 h. The castings were then post-cured at 180 ºC for 2 h and finally removed from the mould and characterized.
Scheme 3.1 Formation of linseed vinyl ester fatty amide toughened epoxy interpenetrating network

3.3 PREPARATION OF OMMT CLAY FILLED HYBRID LINSEED VINYL ESTER TOUGHENED EPOXY NANOCOMPOSITES

The epoxy resin was mixed with the desired amount of organo clay (1, 3 and 5%) at 70°C for 24 h. To the organo clay filled epoxy resin, the fixed amounts of linseed vinyl ester (20 wt%) were added and heated at 70°C in an oil bath for 24 h with constant stirring. A stoichiometric amount
of curative 4,4’-diaminodiphenylmethane (9.72 g) corresponding to epoxy equivalent was also added. The product was subjected to vacuum to remove the trapped air and then cast and cured at 140 °C for 3 h. The castings were then post-cured at 200 °C for 2 h and finally removed from the mould and characterized.

3.4 RESULTS AND DISCUSSION

3.4.1 FT-IR Spectroscopy

The FT-IR spectra of N, N-bis (2-hydroxyethyl) linseed amide (HELA) and linseed vinyl ester fatty amide are presented in Figure 3.1. The strong absorption band (1463-1466 cm\(^{-1}\)) in the IR spectrum of HELA and linseed vinyl ester fatty amide can be attributed to a tertiary amide. A strong absorption band at 2854 cm\(^{-1}\) - 3010 cm\(^{-1}\) may be due to –CH stretching of linseed oil fatty amides. The strong intensive band at 1740 cm\(^{-1}\) indicate that the presence of α, β unsaturated carbonyl group present in the linseed oil fatty amides. A strong absorption band at 3392 cm\(^{-1}\) is due to free –OH group of HELA. The free –OH group of HELA undergoes esterification with acrylic acid to form linseed vinyl ester fatty amide (LVEFA), which was identified by using FT-IR spectra.

Figure 3.1b shows the IR spectrum of LVEFA. The formation of LVEFA observed by the disappearance of band at 3392 cm\(^{-1}\) (free –OH group of HELA) and the appearance of a new absorption band at 1408 cm\(^{-1}\) is attributed to the terminal vinyl (HC=CH\(_2\)) group. The bands at 2854 cm\(^{-1}\) and 2927 cm\(^{-1}\) are attributed to (C-H stretching) alkane chains of linseed oil. A very strong absorption band at 1740 cm\(^{-1}\) is attributed to ester stretching vibration of LVEFA (Figure 3.1b). The absorption peak at 917 cm\(^{-1}\) for the oxirane ring of epoxy resin disappeared in the cured product of hybrid vinyl ester fatty amide toughened epoxy resin.
Figure 3.1  (a) FT-IR spectra of N, N-bis (2-hydroxyethyl) linseed amide (HELA) and (b) linseed vinyl ester fatty amide
Figure 3.2  a) Epoxy cured by DDM, b) epoxy linseed vinyl ester fatty amide blend 100:20, epoxy linseed vinyl ester fatty amide blend and clay with the composition c)100:20:01 d) 100:20:03 and e)100:20:05
The disappearance of absorption band at 917 cm$^{-1}$ confirm the Michael addition reaction between DDM and terminal vinyl group of linseed vinyl ester fatty amide (Figures 3. 2 b, c and d). The appearance of OH peak at 3392 cm$^{-1}$ is due to the opening of the oxirane ring of the epoxy in the cured product of hybrid vinyl ester fatty amide toughened epoxy resin.

Table 3.1 TGA results of linseed vinyl ester fatty amide toughened epoxy clay nanocomposites

<table>
<thead>
<tr>
<th>Epoxy(E)/LVEFA(L)/clay(C)</th>
<th>Initial decomposition temperature (°C)</th>
<th>Temperature characteristic wt loss (°C)</th>
<th>Char yield at 750° C wt%</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td>20%</td>
<td>40%</td>
</tr>
<tr>
<td>E$_{100}$</td>
<td></td>
<td>355.0</td>
<td>372</td>
</tr>
<tr>
<td>E$<em>{80}$L$</em>{20}$</td>
<td></td>
<td>358.0</td>
<td>360</td>
</tr>
<tr>
<td>E$<em>{80}$L$</em>{20}$C$_{1}$</td>
<td></td>
<td>347.8</td>
<td>390</td>
</tr>
<tr>
<td>E$<em>{80}$L$</em>{20}$C$_{3}$</td>
<td></td>
<td>366.8</td>
<td>400</td>
</tr>
<tr>
<td>E$<em>{80}$L$</em>{20}$C$_{5}$</td>
<td></td>
<td>376.8</td>
<td>410</td>
</tr>
</tbody>
</table>

3.4.2 Thermal Properties

3.4.2.1 Thermo gravimetric analysis

TGA results (Table 3.1) indicate that the incorporation of linseed vinyl ester fatty amide to the epoxy decreases the thermal stability. The 20% weight loss temperature for unmodified epoxy and 20 wt% linseed vinyl ester fatty amide modified epoxy is 372 °C and 360° C respectively. The exact reason for this behaviour may be due to the thermoplastic nature imparted by linseed vinyl ester fatty amide. The initial degradation temperature of organoclay filled, lined vinyl ester fatty amide modified epoxy system is slightly higher than that of unmodified epoxy system. An incorporation of
organoclay of 1 wt%, 3 wt%, and 5 wt% into 20 wt% linseed vinyl ester-modified epoxy resin, the char yield increases with respect to organoclay content.

Figure 3.4 DSC curves of a) epoxy cured by DDM, b) epoxy linseed vinyl ester fatty amide blend with the composition b) 100:10 (c)100:20, (d)100:30, epoxy linseed vinyl ester fatty amide blend and clay with the composition e)100:20:01 f) 100:20:03 and g)100:20:05

The reason for enhancement in the degradation temperature is due to the dispersed clay layers which act as a barrier for both the incoming gas and also the gaseous by-products, which increases both the degradation temperature and also widens the degradation process. The addition of clay enhances the performance of the char formed by acting as superior insulator
and mass transport barriers to the volatile products generated during decomposition (Leszczynska et al 2007).

Table 3.2  Thermo mechanical and mechanical properties of linseed vinyl ester fatty amide toughened epoxy clay nanocomposites

<table>
<thead>
<tr>
<th>Epoxy(E)/LVEFA (L)/clay (C)</th>
<th>Glass transition temperature (°C)</th>
<th>Tensile Strength (MPa)</th>
<th>Tensile modulus (GPa)</th>
<th>Flexural Strength (MPa)</th>
<th>Flexural Modulus (GPa)</th>
<th>Impact Strength kJ/m²</th>
</tr>
</thead>
<tbody>
<tr>
<td>E₁₀₀</td>
<td>165.0</td>
<td>61.3</td>
<td>2.71</td>
<td>104.2</td>
<td>2.37</td>
<td>70.0</td>
</tr>
<tr>
<td>E₉₀L₁₀</td>
<td>146.8</td>
<td>43.0</td>
<td>2.41</td>
<td>92.91</td>
<td>2.25</td>
<td>76.66</td>
</tr>
<tr>
<td>E₈₀L₂₀</td>
<td>136.7</td>
<td>40.6</td>
<td>2.25</td>
<td>90.16</td>
<td>2.02</td>
<td>83.33</td>
</tr>
<tr>
<td>E₇₀L₃₀</td>
<td>128.3</td>
<td>35.8</td>
<td>2.19</td>
<td>83.16</td>
<td>1.75</td>
<td>86.66</td>
</tr>
<tr>
<td>E₈₀L₂₀C₁</td>
<td>153.3</td>
<td>67.5</td>
<td>2.50</td>
<td>105.84</td>
<td>2.48</td>
<td>90.0</td>
</tr>
<tr>
<td>E₈₀L₂₀C₃</td>
<td>158.8</td>
<td>71.6</td>
<td>2.90</td>
<td>116.29</td>
<td>2.57</td>
<td>93.33</td>
</tr>
<tr>
<td>E₈₀L₂₀C₅</td>
<td>165.9</td>
<td>75.8</td>
<td>3.10</td>
<td>120.21</td>
<td>2.59</td>
<td>98.33</td>
</tr>
</tbody>
</table>

Table 3.2 indicates that the Tg of modified epoxy system decreased from 165 °C to 136.7 °C with increasing concentration of linseed vinyl ester. The reason behind this is due to the effective cross linking density which decreases with increasing weight percentage of flexible linseed vinyl ester fatty amide. The addition of an aliphatic chain to the polymer would also be responsible for reducing the Tg due to a “comonomer” effect. The decreasing trend of Tg in the case of linseed vinyl ester fatty amide modified epoxy is due to the thermoplastic nature attributed by the flexible long aliphatic fatty amide (Raymond et al 1998 and Sharmin et al 2010). It was observed that the Tg increases from 136.7 °C to 165.9 °C with increasing percentage of organoclay incorporation (Table 3.2). The value of Tg for amine cured epoxy/organoclay nanocomposites increases due to the restricted segmental motions.
in the neighbourhood of organic-inorganic interface of the intercalated nanocomposites (Miyagawa et al 2004).

3.4.3 Mechanical Properties

3.4.3.1 Tensile properties

The values for tensile strength of linseed vinyl ester fatty amide modified DGEBA and organo-clay filled linseed vinyl ester modified epoxies are presented in Table 2. An introduction of 10 wt%, 20 wt% and 30 wt% linseed vinyl ester into epoxy decreased the tensile strength to 43 MPa, 40 MPa and 35.78 MPa respectively when compared with those of unmodified epoxy matrix which is 61.3 MPa. An incorporation of 20wt% linseed vinyl ester amide modifies the brittle nature of epoxy to the flexibility imparted by long pendant aliphatic fatty amide chains of linseed vinyl ester. An incorporation of 1 wt%, 3wt% and 5 wt % organoclay to 20wt% linseed vinyl ester resin epoxy increased the tensile strength from 40 MPa to 67.5 MPa, 71.6 MPa and 75.8 MPa respectively. Effective stress transfer is the vital factor which contributes to the strength of the composite material. In this case, there are well bonded particles, which when incorporated to the polymer will lead to an increase in strength especially for nanoparticles with high surface area. This is further confirmed from the SEM micrographs of the tensile fractured surfaces of epoxy/linseed vinyl ester composites. The tensile modulus of nanocomposites is systematically increased with increasing organo-clay loading. The MMT has a high modulus and hence the resulting composites would be expected to have modulus (Fornes et al 2003). The enhancement of modulus was reasonably ascribed to the high resistance exerted by the MMT clay platelets against the plastic deformation and the stretching resistance of oriented polymer backbone in the galleries. From this it is inferred that the layered silicates of OMMT act as reinforcement for polymer chains (Shia et al 1998).
3.4.3.2 Flexural properties

As expected, the addition of linseed vinyl ester having soft segments into the DGEBA resulted in decrease in the flexural properties and this is due to the reduction in the rigidity of the epoxy network. For instance the incorporation of 20wt % LVEFA the flexural strength decreased from 106.2 MPa to 90.16 MPa when compared with that of the neat epoxy matrix. This reduction in flexural behavior could be due to the plasticization effect which increased the flexibility resulting in decreasing the elastic modulus of DGEBA/ linseed oil vinyl ester hybrids (Hiroaki Miyagawa et al 2005). However, the reinforcing effect of organo-clay is reflected in enhancing the flexural strength. For instance the incorporation of 5 wt % of organoclay in 20wt% LVEFA increased the flexural strength from 90.16 MPa to 120.21 MPa and flexural modulus from 2024 MPa to 2590 MPa respectively (Table 3.2).

3.4.3.3 Impact properties

From Figure 6a, it is observed that the fractured surface of 20 wt% LVEFA modified epoxy fracture surface is ductile. According to Mehta G et al (2004) the rubbery phase imparted by the linseed oil modified epoxy act as an impact modifier which absorbs more impact energy in addition to high energy dissipation during the crack propagation resulting in delayed catastrophic failure resulting in high impact strength from 70 kJm$^{-2}$ to 83.3 kJm$^{-2}$.

3.4.4 Fractography

In the case of neat epoxy matrix there is no evidence of any ductile fracture process (matrix shear yielding) which further substantiates the brittle nature of the fracture process. The poor fracture property of the sample can
also be explained by considering the morphology as observed in SEM. Figure 3.5a shows SEM micrographs of tensile fractured surfaces of linseed vinyl ester modified epoxy resin. It is obvious that the surface (Figure 3.5b) is rougher than the surface of neat epoxy matrix (Figure 5a). This infers that the material experiences enormous amount of plastic deformation before fracture indicating that the increase in the ductility of the material.

![SEM images of tensile fractured surface](image-url)

**Figure 3.5** SEM images of tensile fractured surface of a) neat epoxy matrix, (b) LVEFA epoxy composites in the ratio (100:20), and (c) epoxy LVEFA clay nanocomposites in the ratio (100:20:05)
Figure 3.5c shows the SEM micrographs of the nanocomposites with the higher volume fraction of intercalated clay stacks which indicate that plastic stretching still occurs. The SEM micrograph of 5wt% clay nanocomposites clearly shows the presence of some remaining clay aggregates. These aggregates in turn explain the observed tensile behaviour of the nanocomposites as compared to that of the neat epoxy matrix. In a soft polymer matrix, the intercalated OMMT aggregates provide a stiffening effect; even though such aggregates of silicate layers act as stress concentrators initiating multiple voids they nevertheless seldom detrimentally affect the over ductility of the nanocomposites. Hence, the critical energy release rate of 5wt% OMMT clay reinforced LVEFA/epoxy system (Figure 3.5c) is larger and so more strain energy is used in the failure of the system leading to an improvement in the tensile strength and tensile modulus (Mehta G et al 2004) and it is evidenced from larger fractured surface area.

The fractured surface of organoclay modified composites at different magnification is shown in Figure 3.6 (b, c, d). The larger and rougher strain area is proportional to the higher failure energy and high impact strength of nanocomposites on adding organoclay up to 5wt% (Mehta et al 2004). There are several views to account for the increase in impact strength of the nanocomposites. Since the interfacial adhesion between the organoclay and the polymer matrix is strong, these clay layers are not easily de-bonded from the matrix instead the aggregates acts as obstacle for the crack growth thus diverting from its path and hence the energy dissipated in deflecting the crack resulted in high impact energy (Mohan et al 2006 and Roulin Moloney et al 1988).
Figure 3.6 SEM images of impact fractured surface of a) neat epoxy matrix, (b) LVEFA epoxy composite in the ratio (100:20), and (c) epoxy LVEFA clay nanocomposites in the ratio (100:20:05)

For intercalated organoclay nanocomposites, the crack tends to avoid reaching the aggregations of intercalated organo-clay nanoplatelets, since the adhesion between DGEBA/LVEFA and organoclay interface was excellent and the strength of clay aggregation prevents cracks from propagating. Therefore, the crack tends to deflect on the micrometer scale in the vicinity of the intercalated clay nanoplatelets, and this result in the higher critical energy release rate with the rougher fracture surface (Mehta et al 2004). One more reason is that the clay layers possess excellent strength which is not easily broken by the impact forces resulting in nanocomposites.
with impact strength increased by 19% and 40.47% by incorporating LVEFA and organoclay respectively (Table 3.2).

3.4.5 Dielectric Properties

Commercially available epoxy resin has a dielectric constant of about 4 which inhibits the efforts to increase the effective dielectric constant of the composites at a low ceramic loading level. Organoclay has been used to tailor the capacitive and conductive properties and enhance its performance as a dielectric. 20 wt% linseed vinyl ester modified epoxy was taken as reference and 1 wt%, 3 wt% and 5 wt% organoclay were incorporated and hence their dielectric properties were studied.

![Figure 3.7 Dielectric constant of linseed vinyl ester fatty amide epoxy clay nanocomposites](image-url)
From the above results it is clear that the dielectric constant increases from 4 to 5.9 with the addition of organically modified nano filler (Figure 3.7). The reason for the improved dielectric constant is attributed to the Polarization of $\Pi$ electrons of the double bond and the polar carbonyl groups of the pendant chain segments of linseed vinyl epoxy matrix. The external charges accumulated at both the clay layers and the polymer molecules can change the electrical properties of the composites, and the effectiveness of the bulk permittivity enhancement was related to the structure of organoclay in the polymer matrix defined by the interplanar d-spacing of the silicate layers. It was shown that there is an optimum interplanar d spacing of about 1.7 nm at which charges can accumulate and the dielectric permittivity of the composites will be maximized.

In the present study it is inferred that the diffracted peaks of OMMT clay are shifted from $2\theta=4.7$ nm towards lower angle namely $2\theta=4.19$ nm and 4.22 nm for 1wt% and 5wt% clay in 20 wt% LVEFA epoxy respectively thereby increasing the d spacing for the accumulation of more charges. This polarization mechanism promotes the filler in enhancing the dielectric constant of the modified epoxy resulting in dielectric constant greater than that of neat epoxy matrix. First this polarization can be explained by the presence of intrinsic surface negative charges on the surface of silicate platelets due to positive counter ions placed in the interlayer, which is the cause for ionic polarization in the nanocomposites.

Hence, the interlayer organic surfactant in a modified nano-clay can move along the surface of clay platelets, when it is mixed with a polymer and produce dipoles on the layers and tactoids (Miyagawa et al 2004). Secondly the dielectric constant of the organoclay loaded epoxy/LVEFA nanocomposites is monotonously increased with the increase of organoclay concentration due to enhancement in number of free $\text{Na}^+$ cations and their mobility in the increased spacing of intercalated clay galleries (Karikal
Chozhan et al 2007). These results suggest that clays with high layer charge density and high population density of onium ions limit intragallery diffusion of epoxy and amine and tend to form intercalated nanocomposites rather than exfoliated nanocomposites (Razzaghi-Kashani et al 2008). As a result of all these additional polarization mechanisms, the dielectric permittivity of organoclay composites is higher than that of neat DGEBA matrix. So when filler with higher dielectric constant is used, other additional polarization mechanisms also will come into play. These include interfacial polarization or the accumulation of charges at the interface between the polymer matrix and the heterogeneous inclusions with higher permittivity than the matrix in addition to the distortion and amplification of electric fields around the filler particles especially flat platelets. As a consequence the dielectric permittivity of organoclay filled composites is higher (k=5.9) than that of neat epoxy matrix (k=4).

### 3.4.6 XRD Studies

XRD is a common tool used to probe the structure of nanocomposites. The state of dispersion and exfoliation of silicate layers has been typically established using X-ray diffraction analysis. By monitoring the position, shape and intensity of the basal reflections from the distributed silicate layers, the nanocomposites structure may be identified. For exfoliated nanocomposites, the extensive layer separation and delamination of the original silicate layers in the polymer matrix results in the total disappearance of the coherent diffraction peaks from the silicate layers. While there is a finite layer expansion associated with intercalation of the polymer chain in between the clay layers resulting in the appearance of new basal reflection. Montmorillonite clay consists of number of individual layers and the interplanar distance between the two adjacent layers is called basal spacing or d-spacing calculated using the Braggs equation $2d \sin \theta = n \lambda$ where $\theta$ is the Bragg angle, $\lambda$ is the wavelength used, $n$ is the order of the plane.
The d spacing corresponding to the (001) peak position of OMMT used in the study is 1.85 nm. The XRD is given for OMMT, 1wt% clay and 2wt% clay modified nanocomposites (Figure 3.8). When \( \theta = 0^\circ - 10^\circ \), the peak at \( 2\theta = 4.7^\circ \) corresponds to a basal spacing of 1.85 nm for OMMT. For 1wt% and 5wt% clay nanocomposites the \( 2\theta \) values are 4.19nm and 4.22nm and their corresponding d spacing values are 2.11nm and 2.12nm respectively. The presence of alkyl ammonium ions at gallery region has increased the interplanar spacing.

![XRD spectra showing d-spacing for OMMT, 1wt% and 2wt% clay nanocomposites.](image)

Figure 3.8  Toughened epoxy linseed vinyl ester fatty amide / clay nanocomposites in the composition \( C1= E_{80}L_{20}C_1 \) and \( C5 = E_{80}L_{20}C_5 \)

As a result the diffracted peaks are shifted to a lower angle thus increasing the interplanar spacing into which the polymer chains are intercalated. XRD results confirm that the formation of intercalated clay nanocomposites (Shabeer et al 2007).
3.4.7 TEM Analysis

This can be further confirmed by the TEM observation. All three nanocomposites containing 1, 3 and 5 wt% of OMMT clay is represented respectively show nonhomogeneous distribution of OMMT clay platelets. The light, white area is the epoxy matrix, and the black area is made up of clay layers. The TEM photographs for the LVEFA/epoxy/OMMT nanocompoite indicates the tactoids or bundles of clay platelets do exist. Thus an analysis of the TEM images and XRD patterns indicate, OMMT clay forms intercalated tactoids with LVEFA toughened epoxy matrices.

Figure 3.9 TEM images showing the distribution of a) 1wt% OMMT clay b) 3wt% OMMT clay and c) 5wt% OMMT clay in 20wt% LVEFA toughened epoxy matrices