Hydrophilic fumed silica/clay nanocomposites: Effect of silica/clay on performance

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Abstract

Room temperature cured epoxy (LY-556/HY-951) system filled individually with fumed silica (FS) and modified clay (MC) were synthesized by mechanical shear mixing with the addition of tri-ethylene-tetra-amine (TETA) hardener. The effect of the fumed silica can be negated by the application of a shear force (e.g. mixing, brushing, spraying etc), allowing the liquid to flow, level out and permit the escape of entrapped air. The reinforcement effects of FS and MC in the epoxy polymer on thermal, mechanical and vibration properties were studied. Curing study shows that the addition of FS in epoxy resin aids the polymerization by catalytic effect, and MC addition does not show any effect in the curing behavior of epoxy polymer. Thermogravimetry analysis (TGA) shows enhanced thermal stability for epoxy with FS fillers than that of epoxy with MC fillers. The epoxy with FS fillers shows considerable improvement on tensile and impact properties over pure epoxy polymer and epoxy with MC fillers. SEM studies shows that addition clay significantly turns the epoxy system from brittle to ductile nature was played instrumental in scaling performance. The improvement in tensile and impact properties of nanocomposites is supported with the fracture surface studies. Epoxy with FS fillers shows enhanced vibration characteristics than that of the pure epoxy polymer and epoxy with MC fillers. FTIR studies indicated the formation of C-H bonds on the surface of the nanocomposites.

Key words: Fumed silica, Hydrophilic nano clay, Epoxy nanocomposite, Mechanical properties

1. Introduction

The engineering benefits of smaller systems can justify the cost inherent in designing and building them. Shrinking a thing means that lesser material is required to build it. Material is like an excess baggage. It costs money, adds weight and takes up space. These considerations weigh heavily on all engineering decisions and are of utmost importance for certain applications-for example, satellite and space craft systems, which must be as small as possible. Moreover, it is expensive and inefficient to take heavy things into space: the cost of launching the space shuttle in 2004 is about $10,000 per pound of weight. Smaller devices are imperative in the medical field and have enabled unprecedented surgical and imaging techniques. Bulky tools and big cameras simply do not fit inside the delicate path ways of the human body. Smaller systems perform quicker because they have less mass and therefore lesser inertia (the tendency of mass to resist acceleration). This improved speed leads to products that perform tasks faster, just as a fly can flap its wings much faster than the bird. Another example cited is an assembly robot in a factory. It might perform ten welds in a second, while an enzyme in our body performs as
many as millions chemical operations in the small amount of time. To design tailor-made high technology materials by focusing on molecular structure/processability/properties relationships of high performance polymers such as heterocyclic polymers known for their outstanding thermal and mechanical behaviour on comparison with functionalized polymers or inorganic polymers. Nowadays there is a strong demand, for such nanocomposite materials. The addition of nanoparticles has proven to exhibit a high potential for significantly improving mechanical properties of polymers. This behaviour justifies the great interest of the scientific community and industry. Nevertheless, most of the available data refer to static properties while the behaviour under cycling and dynamic loading is rarely investigated. A brief overview on some of the most important work is presented below. Siegel et al. [1] obtained an increase of 15% of the strain to failure filling an epoxy resin with 10 wt.% of nanometric TiO2 particles. Evora et al. [2] found that adding only 1 vol.% of TiO2 nanoparticles within unsaturated polyester resin increased the fracture toughness of about 57% was due to the uniform and fine dispersion of the filler within the resin at low volume contents. More significant enhancements in fracture toughness (almost 100% at 4.5 vol.% of Al2O3 nanoparticles in unsaturated polyester) were achieved improving the particle-matrix adhesion through a silane surface treatment [3]. Wetzel et al. [4] studied the effects of nano (alumina) and micro-spherical (calcium silicate) particle addition to epoxy resin and found increases in flexural modulus (up to 35%), strength (up to 20%) and Charpy impact energy (up to 35%). In a following, interesting work [5], neat epoxy reinforced with Al2O3 nanoparticles at different volume contents was investigated. The 10 vol. %epoxy/ Al2O3 nanocomposite exhibited significant improvements in flexural modulus (around 40%), strength (15%) and fracture toughness (120%). Furthermore, the crack propagation threshold and resistance turned out to be improved dramatically, with the crack propagation rates for nanocomposites being orders of magnitude slower than neat resin for the same range of SIF. Adebahr et al. [6] proposed a novel route to prepare nanocomposites consisting of monodispersed SiO2 nanoparticles and reactive resin. The addition of 23 wt.% of particles subjected to thermal anhydride curing induced a 66% increase in $K_{IC}$, while UV curing led to an improvement of 82% at 50 wt.%. Lin et al. [7] reported that tensile and impact strength of titanium dioxide and montmorillonite filled epoxy resin reached a maximum for a filler content of 5–8 vol. % and decreases at higher filler contents, sometimes even below the neat resin values. Ragosta et al. [8] improved the mechanical properties of epoxy resin adding 10 wt. % of silica particles with a diameter of 10–15 nm. The normalized elastic modulus reached the value of 1.5, while the normalized yield strength increased up to 1.3. The addition of silica raised the fracture energy of the epoxy matrix by a factor of about 4, whereas the increase of $K_{IC}$ was twofold. Zheng et al. [9] found that the addition of 3 wt. % of silica nanoparticles within epoxy matrix leads to an increase in tensile strength of 115%, while the impact strength increases by 56%. In the literature, the toughening effect due to the addition of particles to polymers has been studied for a long time [10–12]. Different toughening mechanisms have been mentioned, such as the localized inelastic matrix deformation and void nucleation, particle debonding, crack deflection, crack pinning, crack tip blunting, particle deformation or breaking at the crack tip. However, it is still an open question which is the effective mechanisms responsible for toughening on nanocomposites [13]. Furthermore, experimental techniques and descriptive models are based on macro-mechanical concepts. Thus, their application to nanocomposites is not straightforward and indeed questionable. Particle-matrix debonding and localized deformations in the process zone ahead of the crack tip are probably responsible of the considerable toughening effect brought by nanomodification. Recent experimental investigations by Johnsen et al. on silica nanoparticle reinforced epoxy polymers confirm these assumptions [14]. Because of the very high specific surface area, even very low filler contents can significantly contribute to matrix reinforcement. Especially interface related effects, such as debonding mechanisms and void nucleation could play a significant role even at low volume contents. Although classical mechanical theories concerning particle toughening sometimes even predict a decrease of toughening contribution with decreasing particle size, the increasing amount of interfacial area and absolute number of particles in the process zone can be reasons for the experimentally observed increases in $K_{IC}$ [15]. Xie et al. [16] reported the improvement of the mechanical properties of PVC with the addition of CaCO3. At 5 vol. %, optimal performances were achieved in Young’s modulus, tensile yield strength, and strain to failure and Charpy impact energy. The filler enabled ductile fracture caused by elevated triaxial stresses at the neck region and consequently debonding at the particle-matrix interface. Increasing the load, the ligaments between the voids were stretched increasing the energy consumption. The unmodified MMT clay addition leads to the conventional composites. The exfoliated structure possesses superior properties among the three existing structures. The high surface contact area of matrix polymer to nanolayers, uniform distribution of nano layers, etc. of exfoliated structure enhance the properties than other two structures [17–20]. Thermoset epoxy-clay nanocomposites (ECN) are studied under different curing conditions, synthetic routes, organoclays, etc. The report suggests that good exfoliation can be achieved in ECN when amine based curing agents are used [21–24]. The main objective of the researcher is to attempt a lighter material which promises to have high performance applications, as nanocomposites form a good platform for
Fig. 1. DSC scans of (a) epoxy with FS series (b) epoxy with MC series.

generating lighter materials. Although several types of amine curing agents are available in the literature, limited reports are available on TETA curing agent [25-31]. The TETA curing agent is widely used for making glass fibre reinforced epoxy composites. The main purpose of nanoclay filler is to increase the matrix properties of fibre reinforced polymer composites. The addition of nanoclay in epoxy resin with addition of TETA curing agent is an important phenomenon to consider as large amount of glass fibre reinforced composites are used in several applications. In this work, the effect of modified MMT clay and fumed nano silica addition in the epoxy polymer matrix, under TETA curing is studied. The curing behaviour, structure, tensile, thermal and vibrational, SEM and FTIR properties are studied as a function of silica and clay concentrations.

2. Experimental details

2.1 Materials

The matrix material used in this present study is a commercially available epoxy resin (Aradur LY-556) and hardener (TETA, Araldite HY-951) supplied by Huntsman, Switzerland. Two nano fillers were used in this study among first up Montmorillonite clay (Product No: 682659; Brand: Aldrich, Product name: Nanoclay, hydrophilic bentonite; Formula: H2Al2O6Si; Molecular weight: 180.1g/mol; Appearance (Color): Conforms to Requirements Light Tan to Brown; Appearance (Form): Powder; Loss on drying: ≤18.0%; Density: 600-1100kg/m³; Bulk density: Avg. particle size: ≤25micron) supplied by Sigma-Aldrich Chemicals Pvt. Limited, Bangalore, India and the second is fumed nano silica as a nanofiller used AEROSIL®380, hydrophilic fumed silica with a specific surface area of 380 m²/g and an average

Fig. 2. TGA thermograms of epoxy with (a) FS series and (b) MC series.
Fig. 3. Effect of clay/silica content on (a) tensile strength and (b) tensile modulus.

Fig. 4. SEM tensile fracture surface of (a) pure epoxy (b) E + 1% FS, (c) E + 3% FS and (d) E + 10% FS.

2.2. Nanocomposite fabrication

Initially epoxy resin is heated at 100 °C for 1 h. Clay and silica were kept in the oven for about half an hour at 50°C to keep the moisture at bay. The clay is then gently added in to the resin bath. Mixing of clay and epoxy is carried out by mechanical shear mixer. The mixer rotates at 1000 rpm and mixing is carried out for 2 h. After uniform mixing of clay and resin, TETA hardener is added in to the resin/clay solution. The resin to hardener ratio is maintained 100:1 and then casted in the mold. The mold is kept at 80°C for 4 h until complete polymerization occurs. Hand-lay up technique was used to impregnate the composite structures. Same procedure is employed for fumed nanosilica composites developments as same clay variations. The nanocomposite specimens synthesized by this method are tested for various characterizations.

2.3. Characterization

Curing characteristics of epoxy and epoxy filled FS and MC series is studied using DSC. The sample (consisting of resin, clay and hardener prior to curing) of 5 mg is taken in an alumina crucible and heat is applied at the rate of 10° C min. Modal analysis is performed to calculate natural frequency and damping factor ‘n’. Damping factor is calculated using impulse hammer technique (IHT) and logarithmic decrement method (LDM). Specimens of 250 mm x 25 mm x 3 mm is prepared, in which one end of beam is clamped and other end is attached to accelerometer to obtain vibration modes. Natural frequency is determined by impulse loading at free end of the beam using impulse excitation (Rion PH 7117, modally tuned hammer). The signal received from accelerometer is displayed in Fast Fourier Transform (FFT), in which natural frequency is noted down for various modes. Damping factor ‘n’ using IHT is determined using half power bandwidth method. The expression for damping factor by half power width technique is given by

$$\zeta = \frac{\Delta \omega}{2 \omega_n}$$

where $\Delta \omega$ is bandwidth at half-power points of resonant peak for the nth mode and $\omega_n$ is resonant frequency. The
Fig. 5. SEM tensile fracture surface of (a) E + 1% MC, (b) E + 3% MC and (c) E + 10% MC.

Fig. 6. SEM of (epoxy + 10% FS) showing the existence of voids.

where $\xi$, the damping factor, $n+1$, the number of cycles, $\delta$ is the logarithmic decrement, $x_1$ and $x_{n+1}$ are the two displacement values at the time intervals $t_1$ and $t_2$, respectively. A Jeol JSM 840A Japan scanning electron microscope (SEM) was used to study the morphology of fractured surfaces of silica/clay nanocomposite samples at different magnifications. The fractured surfaces of tensile test specimens carried out using SEM. Samples were gold-coated initially subjecting it to SEM analysis. The scanning electron micrographs of different cross-sections of the nanocomposite samples of pure epoxy, filled with fumed silica and clay nanocomposites, are studied. Tensile strength was studied using an universal testing machine (UTM) supplied by Instron Corporation; 3369, series-9 automated testing machine was used with a crosshead speed of 5mm/min. Testing samples were prepared in dumb-bell shapes and these dimensions are 100 x 20 x 3 mm$^3$ based on the ASTM D 638 standards. In each case, five samples were tested and the average value tabulated. The thermal characteristics of the epoxy/clay/silica nanocomposites were measured using both differential scanning calorimetry (DSC-2010 TA Instrument) and thermogravimetric analyses (TGA) at a rate of 10°C/min under nitrogen flow. The FTIR spectra of the powders of the untreated and alkali treated fabric samples were run on an ABB-Bomem FLATA-2000 model spectrophotometer using KBr pellets. The concentration of the fabric powder was maintained at 1% in KBr.

3. Results and Discussions

3.1. Curing characteristics

DSC scans of epoxy filled with fumed silica (FS) and modified clay (MC) as a function of FS and MC are shown in Fig. 1. Pure epoxy shows a single exothermic peak at 325 °C. At 1 wt% FS addition, the peak position has decreased to lower temperature as shown in Fig. 1a. At higher silica content (>3 wt% FS), two distinct peaks are noticed, both at same temperatures. The higher temperature peak is noticed at curing peak of pure epoxy resin. The lower temperature peak corresponds to curing of epoxy resin in intergallery region, and higher temperature peak corresponds to curing at extra gallery (matrix) region. As silica content increases above 3 wt% the lower temperature peak decreases and for 10 wt% FS it is at 320 °C. Single exothermic peak is noted for epoxy with FS content up to 1 wt% FS and results in uniform polymerization. This uniform polymerization at intergallery region and extra DSC scans of epoxy, epoxy filled with FS and epoxy filled with MC clay are shown in Fig. 1. Pure epoxy shows a single exothermic peak at 325 °C. At 1 wt% FS addition, the peak position has decreased to lower temperature as shown in Fig. 1a. At higher clay content (>3 wt% FS), two distinct peaks are noticed, one at lower temperature and another at higher temperature. The higher
temperature peak is noticed at curing peak of pure epoxy resin. The lower temperature peak corresponds to curing of epoxy resin in intergallery region, and higher temperature peak corresponds to curing at extragallery (matrix) region. As silica content increases above 3 wt% the lower temperature peak decreases and for 10 wt% FS it is at 320 °C. Single exothermic peak is noted for epoxy with FS content up to 1 wt% FS and results in uniform polymerization. This uniform polymerization at intergallery region and extragallery (matrix) region could have favored the exfoliated structure. At higher FS content (>3 wt%), the uneven polymerization occurs in the nanocomposites and this leads to two same curing exothermic peaks. The intergallery polymerization occurs at lower temperature and cures quicker than that of the extragallery polymerization. Due to this quick intergallery polymerization, the clay nanolayers cannot separate further. This uneven polymerization leads to the formation of intercalated nanocomposite structure for higher FS content (above 3 wt%). The DSC curing scans of epoxy and epoxy filled with MC are shown in Fig. 1b. It is seen that the addition of MC does not shift the exothermic peak, which is noted for FS contents. The result suggests that addition of MC fillers in epoxy resin does not affect the curing of epoxy. The addition of FS and MC fillers in the epoxy decreases the intensity of exothermic peak. The clay addition continuously decreases the exothermic peak and this is due to decrease in concentration of epoxy resin on clay addition. It is stated that the presence of organo ions in FS increases the polymerization of epoxy by catalytic effect [21] and has to increase the curing temperature. However, the existence of nanolayers affects the polymerization of epoxy. Results show that the effect of nanolayers restricting the polymerization is more than the polymerization of organo ions with epoxy polymer, and hence decreases the curing temperature of epoxy resin.

3.2. Thermogravimetry analysis [TGA]

Figure 2 shows TGA of epoxy, epoxy filled with FS and MC particles. The TGA curve shows the decomposition of epoxy, epoxy with FS and epoxy with MC. Figure 2a shows decomposition of FS at 355°C and is due to dissociation of alkyl ammonium ions. Pure epoxy decomposition starts at 350°C. On addition of FS in epoxy polymer, the decomposition starts at higher temperature. In epoxy with 10% FS, decomposition starts at 370 °C. The hard nanolayers act as barrier for volatile degradation of polymer matrix. This effect causes such enhanced thermal properties for nanocomposites. The enhanced thermal stability of nanocomposites is noted even at higher temperature (>400 °C). At this high temperature (>400°C), the existence of inorganic phases (SiO₂, Al₂O₃, MgO, etc) [25, 26] dominates the nanocomposites and cause such enhanced thermal stability. Figure 2b shows decomposition of MC, epoxy and epoxy filled MC particles. The MC shows negligible decomposition up to 600°C. The addition of MC particles does not improve the decomposition of epoxy polymer. There is a marginal shift in decomposition temperature when MC is added in to the epoxy polymer. The improved thermal stability is noticed for epoxy filled UC series at higher temperatures (>400°C). The reinforcement of FS increases the decomposition of epoxy. The MC addition does not show any improvement in decomposition of epoxy polymer.

Fig. 8 SEM impact fracture surface of (a) pure epoxy (b) E + 1% FS and (c) E + 3% FS.
3.3. Tensile properties

Figure 3 shows the effect of clay addition on tensile strength (Fig. 3a) and tensile modulus (Fig. 3b). The tensile strength of pure epoxy is 61.1 MPa. It is seen that tensile strength increases up to 2 wt% of FS addition in epoxy polymer matrix. As the FS content increases further (>2%), strength decreases. Epoxy with 10% of FS, strength is 61.5 MPa but above the strength of pure epoxy polymer matrix. The formation of exfoliated nanocomposite structures up to 2% FS increases the strength of nanocomposites. The formation of intercalated structure, agglomeration, presence of voids, etc., above 2% FS have decreased the strength of the nanocomposites. The addition of MC decreases the tensile strength of epoxy material. It is seen from Fig. 3 that addition of clay decreases the strain at break. The low strain value is due to the formation of voids, agglomeration, etc. The effect of FS and MC addition on tensile modulus is seen in Fig. 3b. Tensile modulus of pure epoxy is 3GPa. On addition of FS, tensile modulus increases. It is observed that modulus of nanocomposites increases continuously with increasing FS content. An improvement in modulus of ~1.3 times is observed for the addition of 10% OC. The orientation of clay platelets and polymer chains with respect to loading direction can also contribute to reinforcement effects. The decreasing rate of modulus at higher clay content (>2% FS) is due to presence of unexfoliated aggregates in epoxy polymer matrix. In epoxy/MC composites, there is not much improvement in modulus due to the absence of clay distribution at molecular level, and hence does not contribute to molecular strength. The tensile fracture surfaces of epoxy and epoxy filled FS are shown in Fig. 4. Fig. 5 shows the fracture surface of epoxy filled FS and MC particles. If it seen from Fig. 4 that fracture surface of pure epoxy polymer is smooth due to brittle failure. However, on addition of FS particles, crack surface becomes rough (i.e. ductile nature). The roughness increases as FS content increases in the matrix. The fracture roughness indicates that the resistance of propagation of crack is large and the crack has not propagated as easily as seen in pure epoxy. The fracture surface roughness indicates that crack propagation is large and increased the tortuous path of propagating crack [24]. This effect results in higher stress to failure and caused improved strength of nanocomposites. Though the fracture roughness is predominant at 10% FS, the existence of unexfoliated aggregates, voids, etc. could have decreased the strength of nanocomposites. Figure 6 shows SEM picture of epoxy with 10% FS consisting of voids. The fracture surface of epoxy with MC series is seen in Fig. 5. Fracture surface of epoxy with 1% MC is rougher than pure epoxy. At 3% MC, the presence of voids is noted. This indicates that particles have peeled off from material as crack propagates, and create void at the positions where MC particles were there. This also indicates that bonding between matrix and MC particle is poor. For higher clay content (10% MC), though the fracture surface is rough, the existence of voids is clearly visible and has decreased the strength of the material. The poor bonding strength, smooth fracture surface, voids, etc. could decrease the tensile strength of the MC filled epoxy composites. In epoxy FS series, though the strength increases up to 2% FS, for higher clay content it decreases due to unexfoliated aggregates, voids, etc. However, the strength of nanocomposites for all clay content is higher than that of pure epoxy. It requires further investigation of the synthetic procedure to understand the methods of improving tensile strength for higher FS contents.

3.4. Impact properties

Impact results of FS and MC filled in epoxy polymer is shown in Fig. 7. It is seen that impact strength increases for values up to 5% FS and further addition of FS decreases the
impact strength. The decrease in impact strength at higher filler content is due to the existence of agglomeration, unexfoliated aggregates, voids, etc. The addition of MC in epoxy decreases the impact strength of pure epoxy polymer beyond 4% MC. The impact results show that fumed silica nanocomposites provide better impact properties than MC filled epoxy composites. Figure 8 shows the impact fracture surface of nanocomposites. The impact fracture surface provides the reason for impact properties in nanocomposites. Figure 8a shows that fracture surface of pure epoxy is smooth and indicates brittle failure. The addition of 1% FS, fracture surface shows a rough morphology. The roughness of the fracture surface increases as clay content increases to 5% FS. The existence of rough surface shows that crack propagation is difficult and could have increased the torturous path and led to high strength to failure. This has caused high impact strength of silica nanocomposites up to the addition of 5 wt% FS, and on higher addition. Impact results suggest that some additional energy absorbing mechanism is taking place when nano-particles are reinforced in matrix. Crack pinning, cavitation mechanisms, crack surface roughness, etc. [25] are the possible reasons for high impact strength for FS filled epoxy polymer.

3.5. Vibration characteristics

Table 1 shows the effect of FS addition on natural frequency of epoxy polymer. Pure epoxy shows natural frequency of 17.56, 118.53, 120.53 and 298.34Hz for 1st four modes of vibrations, respectively. On addition of FS, natural frequencies of all modes are higher than the epoxy polymer. For higher clay content (>3 wt %), natural frequency decreases, however higher than that of pure epoxy polymer. The large increase in stiffness due to the reinforcement of FS in the matrix causes such increased natural frequency. For higher clay content, the presence of unexfoliated aggregates of FS reduces the stiffness, which decreases the natural frequency. Table 2 shows the effect of MC addition on natural frequency of epoxy polymer. The MC addition does not improve the natural frequency of epoxy as is seen in the case of FS filler additions. A negligible effect in stiffness on reinforcement of MC in epoxy polymer matrix causes such low natural frequencies. Figure 9 shows the effect of FS and MC addition on damping characteristics of pure epoxy polymer. Damping factors measured by LDM and IHT methods for 1st and 4th mode of natural frequencies are presented. Damping factors measured for 1st mode of natural frequencies of epoxy with FS and MC series are shown in Fig. 9a. It is observed that both the FS and MC addition increases the damping factors of pure epoxy. The FS filled epoxy increases the damping factor than that of MC filled epoxy polymer. Damping factor measured by IHT shows higher values than that of measured by LDM. The free load during impact causes increased damping in IHT.

Table 1. Frequency dependence of epoxy and epoxy filled FS series.

<table>
<thead>
<tr>
<th>(E + wt.% FS)</th>
<th>Natural Frequency at mode 1, Hz</th>
<th>Natural Frequency at mode 2, Hz</th>
<th>Natural Frequency at mode 3, Hz</th>
<th>Natural Frequency at mode 4, Hz</th>
</tr>
</thead>
<tbody>
<tr>
<td>E+0</td>
<td>17.56</td>
<td>118.53</td>
<td>120.53</td>
<td>298.34</td>
</tr>
<tr>
<td>E+1</td>
<td>18.63</td>
<td>136.05</td>
<td>206.34</td>
<td>306.41</td>
</tr>
<tr>
<td>E+2</td>
<td>20.85</td>
<td>137.59</td>
<td>217.42</td>
<td>329.05</td>
</tr>
<tr>
<td>E+3</td>
<td>22.75</td>
<td>145.53</td>
<td>234.19</td>
<td>333.46</td>
</tr>
<tr>
<td>E+5</td>
<td>18.24</td>
<td>126.72</td>
<td>229.00</td>
<td>316.29</td>
</tr>
<tr>
<td>E+6</td>
<td>18.12</td>
<td>124.08</td>
<td>225.63</td>
<td>315.72</td>
</tr>
<tr>
<td>E+10</td>
<td>18.02</td>
<td>123.63</td>
<td>222.05</td>
<td>307.41</td>
</tr>
</tbody>
</table>

![FTIR spectra analysis](image.png)

**Fig.10.** FTIR spectra analysis of (a) E + 3% FS and (b) E + 5% FS.
Table 2. Frequency dependence of epoxy and epoxy filled MC series.

<table>
<thead>
<tr>
<th>(E + wt.% MC)</th>
<th>Natural Frequency at mode 1, Hz</th>
<th>Natural Frequency at mode 2, Hz</th>
<th>Natural Frequency at mode 3, Hz</th>
<th>Natural Frequency at mode 4, Hz</th>
</tr>
</thead>
<tbody>
<tr>
<td>E+0</td>
<td>17.66</td>
<td>117.10</td>
<td>194.00</td>
<td>297.00</td>
</tr>
<tr>
<td>E+1</td>
<td>17.23</td>
<td>122.07</td>
<td>183.75</td>
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<td>E+2</td>
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<td>20.40</td>
<td>135.08</td>
<td>204.63</td>
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<tr>
<td>E+5</td>
<td>16.20</td>
<td>111.04</td>
<td>214.06</td>
<td>294.26</td>
</tr>
<tr>
<td>E+6</td>
<td>15.24</td>
<td>110.50</td>
<td>201.75</td>
<td>291.43</td>
</tr>
<tr>
<td>E+10</td>
<td>15.23</td>
<td>110.45</td>
<td>201.43</td>
<td>291.23</td>
</tr>
</tbody>
</table>

From Fig. 10) the IR spectra of epoxy filled with FS for 3 wt. % and 5 wt. % respectively and the characteristic absorption band of the FS is at 1040 cm\(^{-1}\) and the characteristic absorption bands of C–H stretching are shown at 2850 and 2919 cm\(^{-1}\). From (Fig.11) the IR spectra of epoxy filled with MC for 3 wt. % and 5 wt. % respectively and the characteristic absorption band of the FS is at 1500-1600 cm\(^{-1}\) and the characteristic absorption bands of C–H stretching are shown at 2950 and 3000 cm\(^{-1}\).

4. Conclusions

Room temperature cured epoxy filled with fumed silica and modified clay nanoparticles are synthesized separately by adding TETA curing agent. DSC results shows that higher FS content (3% and above) leads to uneven curing and results in intercalated nanocomposite structure. The MC addition in epoxy matrix does not affect the peak exothermic curing temperature of epoxy resin. Tensile property of nanocomposites shows enhanced tensile modulus than that of pure epoxy resin and epoxy filled MC series. The addition of FS increases the tensile strength of epoxy polymer and MC addition decreases the tensile strength of epoxy polymer. Improved impact strength is noted for epoxy filled with FS series than those of MC filled epoxy polymer. Natural frequency of nanocomposites is higher than that of pure epoxy polymer. Damping factor is increased for epoxy/FS series than that of epoxy/MC series.

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