Dual strengthening mechanisms induced by carbon nanotubes in roll bonded aluminum composites

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The dual role of carbon nanotubes (CNTs) in strengthening roll bonded aluminum composites has been elucidated in this study. An increase in the elastic modulus by 59% has been observed at 2 vol.% CNT addition in aluminum, whereas tensile strength increases by 250% with 9.5 vol.% CNT addition. CNTs play a dual role in the strengthening mechanism in Al–CNT composite foil, which can be correlated to the degree of dispersion of CNTs in the matrix. Better CNT dispersion leads to improvement of elastic properties. In contrast, CNT clusters in the aluminum matrix impede dislocation motion, causing strain hardening and thus improvement in the tensile strength. Dislocation density of the composites has been computed as a function of CNT content to show the effect on strain hardening of the metal matrix–CNT composite.

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1. Introduction

Carbon nanotubes (CNTs) have found an increasing interest as reinforcement for aluminum matrix due to their excellent mechanical properties. Elastic modulus of CNT has been reported to vary between 200 and 1250 GPa [1,2], depending on different morphologies. Since the first study by Kuzumaki et al. [3], Al–CNT composite has been successfully processed through different techniques. Most of such investigations have dealt with the powder metallurgy route, often assisted with severe plastic deformation process [4–9]. Researchers have also explored thermal spray methods [10–12], magnetic sputtering [13], liquid infiltration [14], and high pressure torsion [15] for preparing Al–CNT composites. Dispersion of nanotubes is the most critical step towards enhancement of mechanical properties of the composite. Improved degree of CNT dispersion in Al matrix has been achieved through molecular level mixing [9,16] and growing CNT on Al powder particles through chemical vapor deposition [5] prior to consolidation. This improvisation has resulted in 350% increase in the strength [16], proving the role of better dispersion in enhancement of mechanical properties of the composite. Most of the above-listed processing technique involves high temperature for consolidation, which might result in inducing some damage to CNTs. Hence a room temperature technique, like roll bonding, could be a viable option for processing soft, ductile matrix, e.g., Al or Cu based CNT reinforced composites.

Roll bonding is already an established process for mechanical strengthening of aluminum [17,18]. This technique has been successfully used for fabrication of metal (Cu) [19] and ceramic (SiC) [20] reinforced aluminum matrix composites. Choi et al. [8] have shown the positive effect of severe deformation on good dispersion and alignment of CNTs in Al matrix. Li et al. [21] has published the only report on processing of Cu matrix–CNT composite through roll-bonding route. In that study, SWCNTs were acetone sprayed on 10 μm thick Cu foils and then 19 such foils, with intermediate layers of CNTs, were roll bonded to form the composite, with 13% increase in strength. But, till date, processing of Al–CNT composite through roll-bonding route has not been reported. Hence, in the present study, Al–CNT composites have been synthesized by roll-bonding technique.

Another aspect of interest is the role of CNTs in the strengthening mechanism of metal matrix. Apart from enhancement in elastic modulus, it has also been concluded that CNTs obstruct dislocation movement, thus creating high dislocation density in the metal matrix composites in case of soft matrix like Al [8], Mg [22–25], Cu [26] and lead free solder (SnAgCu) alloy [27]. Dislocation pile ups are created due to mismatch in the elastic modulus of CNT and the soft metal matrix [22–24,27]. This, in turn, increases the strain hardening of the matrix metal during plastic deformation. But, till date, no report is available on the “quantification” of dislocation density and strain hardening of the composite induced by CNTs.
In the current study, investigation has been carried out to clearly understand the function of CNTs in strengthening mechanism of roll bonded aluminum composite. The role of CNT content in altering the dislocation density and its effect on strain hardening and tensile strength of the Al-matrix composite is quantified. The suitability of the roll-bonding process has been estimated in terms of improvement in the elastic modulus and tensile strength of the composite.

2. Experimental

2.1. Preparation of the composite

CNTs of 40–70 nm diameter and 1–3 μm length with more than 95% purity were procured from Nanostructured & Amorphous Materials, Inc., Houston, USA. Commercial aluminum foil of 40 μm thickness and >99% purity was used as the matrix material. CNTs were ultrasonicated in acetone for 30 min and sprayed on Al-foil with a hand atomizer and dried. Four layers of Al foils with intermediate three layers of sprayed CNTs were stacked together and rolled in 2-high cold rolling machine (International Rolling Machine, RI-02860). A schematic of the process is shown in Fig. 1. Four passes of rolling were followed by annealing at 523 K for 1 h and further five passes of rolling to a final thickness of ~50 μm. All the rolling passes were given along same direction. The final composite goes through ~70% thickness reduction. The content of CNT in the composites was determined by weighing the as-received aluminum foil before and after spraying CNT on it. Subsequently weight percentage was converted to volume percentage using the density of Al and CNT as 2.7 and 2.1 gm/cm³ respectively. The roll-bonded composites thus obtained with CNT content of 0, 2, 7.5 and 9.5 vol.% will be referred as Al, Al–2CNT, Al–7.5CNT and Al–9.5CNT respectively.

2.2. Microstructural characterization

Composite foil cross sections, perpendicular to the rolling direction, were mounted in the epoxy resin and metallographically polished to observe the microstructure in Buehler-Versamet-3 Metallograph optical microscope. Transmission electron images of CNTs reinforced composite foils were recorded using Philips PW 6061 TEM system (model CM 200, Eindhoven, Netherlands). JEOL, JSM-6330F Field Emission Scanning Electron Microscope was used in secondary electron mode to observe the distribution of CNTs in the delaminated surfaces of the composite foils.

2.3. Evaluation of mechanical properties

Tensile samples were cut from the rolled composite foils with 50 mm length and 4 mm width. Thickness achieved was ~50 μm and the gauge length was 15 mm. These tests were carried using an EnduraTEC, ELF3200 series tensile machine with a maximum load of 245 N and maximum crosshead movement of 12 mm. The tests were carried out at a crosshead speed of 0.003 mm/min. Uniaxial load was applied during tensile testing along the rolling direction of the samples. The tensile tests were carried out as per ASTM E 345-93.

Hysitron Triboindenter, Minneapolis, MN, USA with 100 nm Berkovich pyramidal tip was used in the quasistatic nanoindentation mode to measure the elastic modulus of the composite foils on mounted cross sections perpendicular to the rolling direction. Tip-area calibration was performed using a standard fused quartz substrate of known modulus (69.6 GPa). The load function used for indentation was same for all the samples, with a peak load of 900 N, loading/unloading time of 10 s with a constant rate and 10 s hold time at the peak load. Elastic modulus (E) had been calculated from the slope of the unloading part of indentation load–displacement curves using Oliver–Pharr method [28].

2.4. X-ray diffraction study and line profile analysis

X-ray diffraction (XRD) study on the composite foils has been carried out using Cu Kα (λ = 1.542 Å) radiation in a Siemens D-500 X-ray diffractometer operating at 40 kV and 40 mA. A scan rate of 0.1°/min and a step size of 0.01° have been used to obtain XRD profiles. A fully annealed aluminum sample has been used to correct the instrumental broadening. (1 1 1), (2 0 0), (2 2 0) and (3 1 1) peaks of aluminum have been used for line profile analysis in this study. CMPR software [29] has been used for fitting the XRD peaks with pseudo-Voigt function. Breadth software by Balzar [30] has been used.
used to analyze the XRD profiles to get the crystallite size and microstrain present in the composite foils.

3. Results

3.1. Microstructure

Fig. 2a shows the optical micrograph of the cross-section of Al–9.5CNT composite foil. The effectiveness of roll-bonding in consolidation can be inferred as interfaces between four (4) layers of Al is not observed even for the highest CNT content (9.5 vol.%). The embedded CNTs in the same composite foil are clearly visible in the TEM image (Fig. 2b). These images are the preliminary proof of successful fabrication of Al–CNT composite through roll-bonding route.

Dispersion of CNTs as a function of the CNT content in the composite foil has been investigated by manually delaminating the roll-bonded layers and observing the debonded surface in SEM. A sharp blade was used to cut open the strongly bonded layers through cross-section to observe the roll bonded surface. Fig. 3a and b presents the SEM image of delaminated surface of Al–2CNT and Al–9.5CNT respectively, showing the distribution of CNTs in Al matrix. The delaminated surfaces revealed uniformly dispersed CNTs in the Al matrix in case of Al–2CNT composite. Al foil with higher CNT content of 9.5 vol.% shows the presence of big CNT clusters. Hence, the overall area coverage of the matrix by 9.5 vol.% CNTs (Fig. 3b) looks similar to area coverage by 2 vol.% CNTs (Fig. 3a). It can be concluded that at higher concentration, CNTs tend to agglomerate due to their high surface tension and roll-bonding method does not prove to be effective enough to disperse them. A high magnification image of Al–2CNT (Fig. 3c) reveals the protruded end of deeply embedded CNTs in the Al matrix, indicating the presence of strong interfacial bonding, which result into effective reinforcement. Some cracks and breakage in CNTs are also visible which could be due to high compressive forces to those CNTs which are not so deeply embedded in the matrix.

3.2. Mechanical properties

Mechanical properties of roll-bonded composites have been evaluated in terms of elastic modulus and tensile strength to have an overall understanding about the elastic and plastic part of the deformation and corresponding strengthening mechanisms.

3.2.1. Elastic modulus of the roll bonded Al–CNT composites

Elastic modulus for the roll-bonded composites has been evaluated through nanoindentation technique. More than 50 indents were made on the mounted cross-section of each of the compos-
The composite foil. Effect of CNT dispersion on elastic strengthening of CNTs in the matrix on enhancement of elastic properties of composite foils with CNT content is significant. This observation, when correlated with the standard deviation in the measured values, shown as error bar in Fig. 4b, the change in elastic modulus of 75 GPa whereas rolled Al–2CNT foil shows 59% increase in elastic modulus of 75 GPa whereas rolled Al–2CNT has the highest slope and Al–7.5CNT and Al–9.5CNT composite foils show indentation depth (∼100 nm) of indent for Al sample and minimum (150 nm) for Al–2CNT sample, which indicates that the deformation is easiest in the first one and most difficult in the latter. Al–7.5CNT and Al–9.5CNT composite foils show indentation depth in between reflecting their medium difficulty level in deformation. Indentation depth is the measurement of elastic and plastic deformation together. Hence, the difficulty in deformation is actually the cumulative result of enhancement in elastic and plastic strengthening of the composite foils. It is clear from the depth of indentation that CNT addition strengthens the roll-bonded Al-foils.

The slope of the unloading curve is the measure of elastic modulus in nanoindentation measurements. From Fig. 4a, it can be clearly noticed that the Al foils has the lowest unloading slope, whereas Al–2CNT has the highest slope and Al–7.5CNT and Al–9.5CNT remains in between. Elastic modulus values measured by nanoindentation are shown in Fig. 4b. Al–2CNT composite has the highest elastic modulus of 75 GPa whereas rolled Al–2CNT foil shows 59% increase in E with a homogeneous dispersion of CNTs in the matrix, whereas ∼25% enhancement of E has been observed for higher CNT content with 56 and 60 GPa for Al–7.5CNT and Al–9.5CNT composite foils respectively. Even considering the standard deviation in the measured values, shown as error bar in Fig. 4b, the change in elastic modulus with CNT content is significant. This observation, when correlated with the SEM investigation, shows a prominent role of dispersion of CNTs in the matrix on enhancement of elastic properties of the composite foil. Effect of CNT dispersion on elastic strengthening mechanism of these Al–CNT composites has been discussed in details later in Section 4.1. Fig. 4b also shows the computed E values using rules of mixture (ROM) all composites. There is a good match between the computed and experimentally obtained E values for Al and Al–2CNT. However, there is a large difference between the computed and experimental E values for Al–7.5CNT and Al–9.5CNT composites. The reasons for this difference will be discussed in terms of dispersion and strengthening mechanism, in Section 4.1.

3.2. Tensile strength of the rolled Al–CNT composite

Tensile strength of all the composite foils has been evaluated through tensile test on flat strips with a gauge length of 15 mm. Tensile strength of the composites has been plotted as a function of CNT content in Fig. 5. Tensile strength increases with increasing CNT content in the composite foil. Al–9.5CNT shows a tensile strength of 97 MPa, which is 250% increase from the tensile strength value of 28 MPa for the foil without CNT.

This observation, in conjunction with the trend of enhancement in elastic modulus for these composites, suggests that strain hardening increases with increase in the CNT content. Because, in spite of decrease in elastic modulus with CNT content beyond 2 vol.% the continual increase in tensile strength could only be possible due to increase in strain hardening. The role of CNTs in strain hardening mechanism will be discussed in details in Section 4.2. Small error bars in Fig. 5 (based on five test results) shows the homogeneous and consistent behavior of the composite foils at macro scale. Elastic modulus and ductility for the composite foils could not be calculated from the tensile tests because the displacement resolution for the tensile testing machine was not enough to record the small amount of strain associated with these experiments. Hence, no comment could be made on ductility of these composites as a function of CNT content.

4. Discussion

CNTs play a dual role in strengthening the Al–CNT roll-bonded composite by influencing the elastic and plastic part of deformation. The dispersion of CNTs in the matrix plays a very active role in this process. The following sections explain the CNT induced strengthening mechanism in the Al–CNT composite foil in details.

4.1. Role of CNT in elastic strengthening

Elastic modulus is an intrinsic property for a single phase material. Hence it depends on the structure of the material and can be enhanced uniformly when there is a homogeneous structural change. In the case of composites, when the different phases are uniformly distributed in the matrix material, rule of mixture
(weighted average) can provide a fair approximation of the overall elastic modulus of the composite material.

In Al–2CNT composite, CNTs are separated from each other and uniformly distributed in the Al matrix (Fig. 3a). Fig. 3c shows good reinforcement and strong interfacial bonding which helps in load transfer during deformation. Hence, it can be inferred that Al–2CNT roll bonded foil forms a composite which is capable to behave like a homogeneous material. Hence, in case of Al–2CNT composite, the load is uniformly shared by CNTs and matrix. Thus, it can withstand higher amount of stress before it starts deforming plastically. This results into higher elastic modulus for the Al–2CNT composite. This inference is further strengthened by the calculated and measured values of elastic constants, which are almost comparable for 2 vol.% composition (Fig. 4b).

In case of Al–7.5CNT and Al–9.5CNT composite foils, the CNTs are present as micron size clusters in the Al matrix forming CNT-rich and CNT-depleted regions and weak interfacial bonding, which hinders the effective load transfer. Elastic modulus for Al and CNTs being very different, these composites cannot behave like a structurally homogeneous material during deformation. Hence, with lower amount of stress application, plastic deformation starts at CNT depleted regions, as the $E$ value for Al [31] is much lower than CNTs [1,2]. As a result, the effective elastic modulus is much lower for higher CNT contents in the composite. Thus, the calculation of elastic modulus in these compositions by rule of mixture (ROM) results in huge overestimation as revealed in Fig. 4b. The dependence of enhancement of elastic modulus on degree of dispersion of CNTs has been explained by a schematic in Fig. 6.

4.2. Role of CNT in strain hardening

Strain hardening is associated with the plastic part of deformation and results into increase in the dislocation density in the material. Hence, strain hardening starts dominating the deformation mechanism and strength of the material after the onset of plastic deformation.

Presence of CNTs in soft metal matrix has been reported to generate dislocation pile ups in their neighborhood [22–24,27]. Choi et al. [8] have identified high dislocation density regions at the vicinity of CNTs in Al-matrix composite. The observation of the present study is also in full agreement with these reports. Fig. 7 shows the TEM image of the roll bonded Al and Al–9.5CNT composite foils. Al foil shows the presence of sub-boundaries with low dislocation density, whereas Al–9.5CNT composite foil contains high density dislocation loops. All the processing parameters being same, CNT content is the only difference between these two samples. Hence it can be inferred that presence of CNTs are responsible for such high dislocation density in Al–9.5CNT composite foil.

The effect of CNT on dislocation density and strain hardening has been estimated through different indirect methods. Goh et al. [23,24] concluded that increase in CNT content causes increase in the dislocation density and matrix strain. However, there was no quantification of dislocation density to validate the conclusion. The same group, in another study on deformation mechanism of Mg–CNT composite, has proven that presence of CNTs generates non-basal and cross slips—which implies difficulty in dislocation motion and pile-up [25]. This is also an indirect indication of

![Fig. 6. Schematic representation on effect of CNT dispersion on deformation behavior prior to strain hardening.](image)

![Fig. 7. TEM image of (a) Al roll bonded foil showing subboundaries with low dislocation density and (b) Al–9.5CNT roll bonded foil showing dislocation loop formation and high dislocation density.](image)
increase in the dislocation density in the presence of CNTs. Recently, Choi et al. [8] have calculated the yield strength of Al–CNT composite by calculating grain size through Scherrer formula on XRD peak broadening. But, Scherrer formula poses error in the calculated grain size in case of heavy deformation as it does not take into account the micro-strain term. Driven by the present scenario, an attempt has been made in the present study to find out the dislocation density as a function of CNT content in aluminum matrix composite and thus explaining the role of CNTs in the strain hardening mechanism.

4.2.1. Computation of dislocation density as a function of CNT content

Dislocation density for the composite foils has been calculated from broadening of the XRD peaks using diffraction line broadening analysis proposed by Balzar [32]. The broadening in the XRD peak for deformed structures can come from both the reduction in the crystallite size and the micro-strain generated in the lattice—both of which, ultimately, is a result of increased dislocation density.

To obtain the broadening effect accurately, the peak needs to be fitted by a function. It is well established that pseudo-Voigt function fits the tail portion peak nicely whereas Gaussian function takes better care of the profile shape near its maxima. Hence, pseudo-Voigt technique is the best fitting function for the XRD peaks. By fitting the diffraction peak profile with pseudo-Voigt function, the Cauchy and Gaussian contributions in the integral breadth ($\beta_C$ and $\beta_G$, respectively) could be determined separately [33,34]. CMPR software [29] has been used in the present study for fitting the XRD peaks with pseudo-Voigt function to calculate $\beta_C$ and $\beta_G$. For correction of the line profile from instrumental broadening effect, fully annealed Al has been taken as standard sample. The corrected line profile parameters have been calculated using the following equations [34].

\[
\beta_C^h = \beta_C^g - \beta_G^g \quad (1)
\]

\[\beta_C^g = \sqrt{\beta_C^2 - \beta_G^2} \quad (2)\]

where superscript $h$, $g$ to $\beta$ denotes integral breadths of the sample under study, standard sample and instrumentally corrected peaks, respectively.

Each of these Cauchy and Gaussian integral breadths can be further divided into their crystallite size and micro-strain term using the following relations [33].

\[
\beta_C^g = \beta_{CG} + \beta_{DC} \quad (3)
\]

\[\left(\beta_C^g\right)^2 = \left(\beta_{CG}\right)^2 + \left(\beta_{DC}\right)^2 \quad (4)\]

where $S$ and $S_0$ denotes $1/d$ value of the corresponding peak and first peak ($d$ = lattice spacing), respectively and subscript $D$ and $S$ to $\beta_C$ and $\beta_G$ denotes Cauchy and Gaussian integral breadth components influenced by micro-strain distortion and crystallite size effects, respectively.

Peak data from at least two peaks are required to solve the above equations to know the unknowns. For more than two peaks, a linear least square fit is used. These values are used directly to obtain the surface- and volume-weighted crystallite sizes ($D_{sur}$ and $D_{vol}$) and root-mean-square strain ($\varepsilon$) in the material [35].

\[D_{sur} = \frac{1}{2\beta_{SC}} \quad (5)\]

\[D_{vol} = \frac{\exp(k^2)}{\beta_{SC}} \text{erf}(k) \quad (6)\]

Table 1

<table>
<thead>
<tr>
<th>Vol.% CNT</th>
<th>Crystallite size (Å)</th>
<th>Strain</th>
</tr>
</thead>
<tbody>
<tr>
<td>0</td>
<td>460</td>
<td>0.0000109</td>
</tr>
<tr>
<td>2</td>
<td>405</td>
<td>0.0000142</td>
</tr>
<tr>
<td>7.5</td>
<td>361</td>
<td>0.0000072</td>
</tr>
<tr>
<td>9.5</td>
<td>273</td>
<td>0.00209</td>
</tr>
</tbody>
</table>

\[\varepsilon = \frac{1}{S_0} \left(\frac{\beta_{SC}}{2\pi} + \frac{\beta_{DC}}{\pi^2L}\right) \quad (7)\]

\[k = \frac{\beta_{SC}}{\pi^0.5\beta_{GC}} \quad (8)\]

where $k$ is the characteristic integral-breath ratio of a Voigt function and $L$ is the column length (distance between two cells in a real space) orthogonal to diffracting planes. In the present study, breadth programme by Balzar [30] has been used to calculate the crystallite size and strain in the material. Four peaks, namely (1 1 1), (2 0 0), (2 2 0), (3 1 1) of Al have been used for all the composites in this analysis. The calculated values of crystallite size and strain are presented in Table 1, which shows that with increase in the CNT content, the crystallite size decreases and strain increases, resulting in an increase in the dislocation density.

Volume weighted crystallite sizes have been used in this study for calculation of dislocation density as this gives better idea of the material behavior. Dislocation density can be calculated as a function of the contribution of micro-strain and crystallite size towards it using following relationships [36,37].

\[\rho_D = \frac{3}{(D_{vol})^2} \quad (9)\]

\[\rho_e = K\frac{(S)^2}{S_0^2} \quad (10)\]

\[\rho = (\rho_D\rho_e)^{1/2} \quad (11)\]

where $\rho$ is the dislocation density, $\rho_e$ = contribution of micro-strain to dislocation density, $\rho_D$ = contribution of crystallite size to dislocation density, $K$ = $6\pi$ is the factor related to Gaussian strain distribution, $b$ is the Berger vector and for FCC $b = a\sqrt{2}$, $a$ being the lattice parameter of the material.

Dislocation density, thus calculated is shown to follow an increasing trend as a function of CNT content in the composite as evidenced from Fig. 8. For Al–2CNT, dislocation density is not much different than Al, whereas a huge increase of the same is found in Al–7.5CNT and Al–9.5CNT samples. This trend in dislocation density can be adequately explained as an effect of dispersion of CNT
in the Al matrix. In case of Al–7.5CNT and Al–9.5CNT composite foils, the CNTs form micron size clusters in the Al matrix, which behave like precipitates during the plastic deformation. Due to the difference in strength between the CNT clusters and the Al matrix, a high stress field is generated around the clusters which hinder the flow of dislocations through matrix during plastic deformation. Due to CNT clustering, the matrix is much soft and depleted in CNT content. This results in high stress gradient at the boundary of CNT clusters/matrix. Hence, dislocation pileups and loops are formed at those highly stressed regions (Fig. 7b), increasing the dislocation density (Fig. 8) for these composites. In addition to hindering dislocation motion, CNT agglomerates with high stress field can also act as Frank Read source and generate dislocation loops as observed in Fig. 7b. Thus, Al–CNT composites with higher CNT content show high strain hardening. But in case of Al–2CNT composite, CNTs are uniformly distributed over the matrix (Fig. 3a). Dispersed CNTs also create smaller stress fields around them. However, the stress fields due to dispersed CNTs are homogeneously distributed throughout the matrix. Thus, it can withstand higher amount of stress before it starts deforming plastically. At the onset of plastic deformation, there is minimal chance of formation of dislocation pileups due to absence of stress concentration points. Hence, Al–2CNT composite shows low dislocation density and lower amount of strain hardening. In case of Al roll bonded foil, due to absence of CNTs, dislocations are not hindered and flows easily forming no pile-ups which is clearly reflected in calculated dislocation density (Fig. 8) and TEM image (Fig. 7a).

### 4.2.2. Tensile strength as a function of dislocation density

Tensile strength of the composite foils has been calculated as a function of dislocation density using Taylor's relationship [38] and properties of pure Al.

\[
\sigma_{\text{Taylor}} = \sigma_0 + \alpha M^2 \frac{G b}{2 \rho^{1/2}} \quad (12)
\]

where \(\sigma_{\text{Taylor}}\) is the flow stress, \(\sigma_0\) the friction stress (25 MPa), \(\alpha\) the constant (0.33), \(G\) the shear modulus of Al (26 GPa), \(b\) the length of the Burgers vector of dislocations (\(b = 0.2865\) nm), \(M^2\) is the Taylor factor (3 for untextured polycrystalline materials) [39]. Taylor's relationship is generally used for calculating the flow stress in single phase material strengthened by dislocations. However, it has been used in this study for composite material to emphasize that at higher CNT content, reinforcement effect is poor due to clustering. Thus, strengthening in high CNT containing composite is largely due to increased dislocation density and strain hardening. The reason for doing so is to separate out the effect of reinforcement in the enhancement of elastic modulus to its contribution towards strain hardening, i.e., dislocation generation. In this way, it is possible to find out the effect of CNT distribution on elastic property and strain hardening behavior separately.

**Fig. 9** shows a comparison of calculated and measured strength of composite foils as a function of their CNT content. As the calculation uses the properties of pure Al, the stress thus calculated is only the reflection of the dislocation density in the material and does not take in to account the enhancement of elastic property, if any, by CNT addition. Thus, the calculated values should give a fair idea of stress in Al foil. The closeness of both calculated and measured strength values in Al sample reflects the accuracy in assumption of the calculation parameters. For Al–7.5CNT and Al–9.5CNT composite foils also, calculated and measured values are fairly comparable. This observation indicates that the strength in these materials is enhanced due to increase in the dislocation density. Hence, strain hardening can be inferred to be the dominating deformation mechanism for Al–CNT composites with higher CNT content. In Al, Al–7.5CNT and Al–9.5CNT composite foils, calculated values are to some extent higher than the measured values, which must have resulted from the properties of pure Al assumed for calculation. In case of Al–2CNT, the trend is totally reversed, i.e., calculated value of strength is much lower than the measured value. This means, the strengthening achieved in this composite is not restricted to increase in the dislocation density. This observation firmly supports the previous observation on enhancement of elastic modulus for Al–2CNT composite. Strength of Al–2CNT composite is a combined effect of elastic modulus enhancement and strain hardening with the first one dominating. Also, there is a marginal increase in the dislocation density in Al–2CNT (Fig. 8) composite resulting in little enhancement in strain hardening. Thus, the mismatch between the calculated and measured value of strength for Al–2CNT composite foil is created due to the large enhancement in the elastic modulus for the same. These observations directly support the dispersion of CNTs in the matrix and their effect on flow of dislocation during deformation.

### 5. Conclusions

This study has explored the dual role of CNT in the strengthening mechanism in roll bonded Al–CNT composite. Roll bonded Al–CNT composite shows very good interfacial bonding and strong reinforcement of CNTs in the matrix. Deformation by roll bonding resulted in homogeneous dispersion of the CNTs in the matrix at 2 vol.%, whereas it forms agglomerates at 7.5 and 9.5 vol.%. Elastic modulus of the composite increases to a maximum of 59% in Al–2CNT composite whereas and tensile strength increases up to 250% in Al–9.5CNT composite.

CNTs play a dual role in strengthening mechanism depending on their degree of dispersion in the aluminum matrix. Uniformly dispersed CNTs enhance the elastic modulus of the composite by behaving like a homogeneous structure and resisting the plastic deformation at lower stress. Agglomerated CNT clusters for high CNT concentration behave as the precipitates and do not enhance the elastic modulus of the composite. CNT clusters inhibit the dislocation flow and results in an increased dislocation density with the increasing CNT content. As a result the composite gets strengthened by strain hardening.

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