Pure Al matrix composites produced by vacuum hot pressing: tensile properties and strengthening mechanisms

Fei Tang\textsuperscript{a}, Iver E. Anderson\textsuperscript{a,\ast}, Thomas Gnaupel-Herold\textsuperscript{b}, Henry Prask\textsuperscript{b}

\textsuperscript{a} Ames Laboratory, Iowa State University, Ames, IA 50011, USA
\textsuperscript{b} NIST Neutron Research Center, 100 Bureau Drive, Gaithersburg, MD 20899-3460, USA

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Abstract

Al matrix composites reinforced by spherical intermetallic particles, were consolidated from gas atomized elemental Al and Al–Cu–Fe alloy fine powders (<10 \(\mu\)m) by a vacuum hot pressing technique. The composites were made from two types of powders including commercial inert gas atomized powder (99.7\%) and high purity powder (99.99\%) produced by a gas atomization reaction synthesis technique. The microstructures and tensile properties of the composites with three different volume fractions of the reinforcement particles (15, 20 and 30 vol \%) were characterized. Microstructural analysis of the samples demonstrated that the quasicrystalline phase in the Al–Cu–Fe particles transformed to a crystalline phase, which has similar elastic modulus, CTE, and hardness properties, but a reduced density. All the composites appear to be fully dense, with strong interparticle bonding, and exhibit elastic modulus values approaching upper bound predictions by rule of mixtures. Tensile test results and neutron diffraction measurements allowed an assessment of the relative influence of direct and secondary composite strengthening mechanisms on the yield strength of this model composite system. The results suggest that for elemental Al matrix composite samples without precipitation strengthening and severe strain hardening during consolidation, the direct and relevant secondary strengthening mechanisms can be combined to predict accurately the yield strength increase of the composites.

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1. Introduction

It is attractive to use particulate reinforced aluminum (PRA) matrix composites in structural applications because of their excellent stiffness-to-weight and strength-to-weight ratios [1,2]. Such PRA materials also exhibit generally good wear resistance, thermal conductivity, and low thermal expansion, all of which makes them good multifunctional light-weight materials [3,4]. These superior properties suggest many possible uses in weight-sensitive components for aerospace or land transportation. A wide variety of fabrication techniques have been explored for metal matrix composites. These include powder metallurgy, molten metal methods, semi-solid casting, pressure infiltration, and spray deposition [1,4–6].

The powder metallurgy processing technique is attractive for several reasons [6–8]. This approach offers microstructural control of the phases that is absent from the various routes that involve a liquid phase. Powder metallurgy processing employs lower temperatures and, therefore, reduced diffusion rates with better control of interface reaction kinetics. Because of their basis as a powder, PRA composites often have been deformation processed after powder consolidation to develop the best properties. In this manner, the composites behave like high strength aluminum alloys made by the powder metallurgy technique: i.e., the prior particle oxide skins must be broken up by metal working before the true properties of the matrix metal and, hence, the composite can be achieved. The most common primary breakdown process has been extrusion. Other metal working processes such as rolling, forging, shear spinning and swaging have also been demonstrated. Also, the typical ceramic reinforcements (e.g., SiC) for PRA composites give rise to dulling of many common machine tools, decreasing the machinability of these composites. Thus, it would be quite beneficial if
a powder metallurgy process for high strength PRA composites with good ductility could be developed that avoided extensive mechanical deformation and permitted net shape die forming without machining.

Strengthening mechanisms of composites have been studied extensively for many years. The continuum shear lag model was first developed to predict the direct strengthening of continuous fiber reinforced composites originally by Cox [9]. The predominant direct strengthening factor is the volume fraction of reinforcement. Powder processed material tends to give somewhat higher strengths than melt processed composites, probably because of additional strengthening from residual oxide dispersoids from prior particle surfaces, and from the somewhat finer grain size. Unfortunately, for the low aspect ratio whisker or particulate reinforcement particles typically used in current metal matrix composites, the shear lag model underestimates the strength [10,11]. Nar done and Prewo suggested that better agreement could be obtained if the shear lag model was modified to allow for whisker or fiber end loading effects [12]. The theoretical prediction by means of this modified direct strengthening model was closer to the experimental results when the reinforcement aspect ratio is small. The difficulty with the continuum approach of Cox [9] is that it ignores the secondary strengthening influence of reinforcement particles on the micromechanics of composite deformation, such as the very high matrix work hardening at low strains. The improved shear lag model [12] still ignores detailed modifications in composite microstructures, such as dislocation density increases and residual stresses from processing effects [13].

A model to predict the yield strength of a particle-reinforced metal matrix composites by considering the dislocation density due to mismatch between thermal expansion coefficients (ΔCTE) of particle and matrix was developed by Arsenault and Shi [14,15]. It was proposed that the high matrix dislocation density caused by ΔCTE should account more closely for the observed secondary strengthening. While this enhanced dislocation density strengthening model is in agreement with the general trends of the strengthening results, it is not sufficient to account for the complex combined strengthening effects in many composite materials [16,17]. For example, Orowan strengthening is not a major factor with the 5 μm and larger reinforcement particles usually used, but particles of this size can result in quench hardening and enhanced work hardening because of elastic misfit back stress hardening [14,18]. Reinforcement particle shape, in terms of aspect ratio, also can influence composite strength, but for the typical SiC particulate aspect ratio range of up to 2:1, it is not expected to be a major factor.

In order to study predominantly the direct strengthening contributions to the total strength of Al alloy based composites, Krawieski and Chawla et al. [19,20] elected to use a series of complex thermomechanical treatments to balance the complex secondary strengthening contributions in both the unreinforced Al alloy and in the composite material. These treatments were needed to ensure that the starting density and distribution of the dislocations and precipitates were similar in the unreinforced Al alloy and in the composites in order to compare the direct and secondary strengthening more unambiguously. In this paper, elemental Al powder was chosen as the matrix material to avoid the complications of matrix heat-treatment designs to isolate the strengthening mechanisms of composite materials and to permit the use of a high temperature consolidation temperature, without promoting partial melting.

In recent research [16], results on elemental Al matrix composites, reinforced by 30 vol.% of spherical Al–Cu–Fe alloy particles and consolidated by quasi-isostatic forging, were introduced. Because of the fine (diameter <10 μm) matrix and reinforcement particle sizes that match closely and a homogenous spatial distribution, the yield strength (YS) of this model composite material was improved over the matrix properties by 220% for the commercial purity composite sample. Remarkably, for an equivalent composite material produced from higher purity metal powders, i.e., with much thinner oxide surfaces, the yield strength of the composite was improved over the matrix properties by 328%. Because these strengthening results were so far above previous observations, typically <100% [16], the time seems right to revisit the predictive capability of the previous strengthening models for particulate reinforced metal matrix composites. The elastic modulus values [16] of these composites, in both versions, were slightly below the calculated upper bound value from the rule of mixtures (ROM) for 30 vol.% loading, using ultrasonically measured values of the specific Al matrix phases and the (assumed) quasicrystalline reinforcement phase. Such high elastic modulus measurements are also very unusual and add further motivation to re-examine composite strengthening for such materials.

These results suggested also that the selection of potential reinforcement phases for PRA materials should be broadened from refractory ceramics to include, e.g., intermetallic compounds that also can form a strong bond with the Al matrix by high temperature solid-state sintering. Moreover, the results clearly indicated that PRA composites made from powders with a thinner oxide surface can achieve significant improvements in tensile properties over the same composites made from commercial powders with residual impurities and thicker oxide surfaces [21].

In light of the interesting mechanical property measurements of the previous work [16], a broader study of this type of metal matrix composites was undertaken for this report. In this paper, we will report the tensile properties of similar Al/Al–Cu–Fe composites made by vacuum hot pressing (VHP) with a wide range of reinforcement loadings. Compared to quasi-isostatic forging, VHP has a further decreased amount of interparticle shear, strain hardening rate and almost no grain anisotropy compared to the forging method [7]. This vacuum hot pressing approach can be viewed as one step closer to the goal of powder consolidation process simplification for direct net shape forming of PRA composite parts. New results also have demonstrated that
the Al–Cu–Fe, quasicrystalline reinforcement phase in the composite actually transforms during consolidation by both quasi-isostatic forging and vacuum hot pressing to a crystalline intermetallic phase of similar properties. This more complete set of tensile property results will be compared to the predictions of the direct and the (relevant) secondary strengthening models for the YS increase of the composite samples, as suggested above.

2. Experimental procedure

2.1. Materials

Powders used for the two kinds of composite samples are shown in Table 1. For the baseline experiments, commercial inert gas atomized (CIGA) Al and Al<sub>63</sub>Cu<sub>25</sub>Fe<sub>12</sub> (quasicrystal) powders were obtained. The CIGA Al powder (99.7% purity) had been evaluated thoroughly in earlier work [21–23] to characterize its surface oxide properties. A patented gas atomization reaction synthesis (GARS) technique was also used to produce 99.99% pure Al and Al–Cu–Fe quasicrystal powders [24] in our laboratory. The X-ray diffraction results of both CIGA and GARS Al–Cu–Fe powders [25] revealed that there are two phases in the as-atomized powder, the major Al<sub>63</sub>Cu<sub>25</sub>Fe<sub>12</sub> quasicrystal (icosahedral) phase and some β-Al(Cu, Fe) cubic phase.

The Al (matrix) and Al–Cu–Fe (reinforcement) powders were air classified to <10 μm. The Al–Cu–Fe powders from both sources were screened subsequently through a 20 μm screen to eliminate the residual large particles, typical of an air classification yield. The size distributions of Al and Al–Cu–Fe powders were reported elsewhere [26].

2.2. Consolidation processing

First, powders were blended homogeneously. Then, blended powders were compacted by cold isostatic pressing (CIP) with a pressure of 280 MPa, intending to produce a green density of about 90%. The green samples from CIP were machined to match the hot press die dimensions and vacuum (10<sup>−5</sup> Torr) hot pressed at 550 °C for 6 h, using a pressure of 175 MPa for 5 h (after the first hour at 550 °C).

Table 2

<table>
<thead>
<tr>
<th>Temperature (°C)</th>
<th>Soak at temperature (h)</th>
<th>Pressure (MPa)</th>
<th>Dwell at pressure (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>550</td>
<td>6</td>
<td>175 or 245&lt;sup&gt;+&lt;/sup&gt;</td>
<td>5</td>
</tr>
</tbody>
</table>

<sup>+</sup> Only for the second GFF-20 sample.

Because the first GFF-20 sample, initially VHP at 175 MPa, was not fully dense, a second GFF-20 composite sample was consolidated at 245 MPa by the VHP process to get full density and was used for the subsequent analysis. Table 2 summarizes the vacuum hot pressing parameters.

2.3. Other tests

X-ray diffraction characterization of composite samples was performed using a Philips PW1830 generator with Cu Kα radiation. The Archimedes technique was used to measure the density of each composite sample. The elastic modulus of each composite sample was measured by an ultrasonic method [27]. Elastic modulus measurement of the resulting tetragonal (α) reinforcement phase, Al<sub>63</sub>Cu<sub>25</sub>Fe<sub>12</sub>, and each Al matrix phase, both CIGA and GARS, were made by the same ultrasonic method [27]. Tensile tests of consolidated composites were performed on an Instron model 1125 tensile test machine under 1.27 mm per min monotonic loading. Tensile sample dimensions were 6.35 mm in diameter and 25.4 mm in uniform gauge length, according to ASTM recommendations [28].

Microstructures of the composites and fractography of the tensile samples were examined on an AMRAY 1845 field emission scanning electron microscope. The coefficient of thermal expansion (CTE) of the Al<sub>63</sub>Cu<sub>25</sub>Fe (α) phase was measured by a Perkin-Elmer TMA7 thermal–mechanical analyzer. Neutron diffraction (in situ) measurement procedures for determining load bearing stresses during tensile tests were reported in detail, elsewhere [29].

3. Results and discussion

3.1. General microstructures

As shown in the low magnification micrographs of Figs. 1 and 2, the reinforcement particles are spherical in shape and distributed quite uniformly in all samples. Some local clustering still can be found [30] in all the samples, especially in 30 vol.% loading samples. The density measurements of the composite samples are given in Table 3, which shows that all of the samples are essentially fully dense.

3.2. X-ray diffraction of composites

X-ray diffraction measurements of the consolidated composites revealed elemental Al matrix phase peaks, but the quasicrystal and β phase peaks, of the as-atomized
Al–Cu–Fe powder [25] were not observed in the diffraction patterns. Instead, there were diffraction peaks of an \( \text{\textmu} \) phase (Al-Cu-Fe), which is a crystalline phase with the density of 4.18 g/cm\(^3\) [10]. As often reported in previous research, the Al–Cu–Fe quasicrystal phase is stable when annealing at about 700\(^\circ\)C [31–33]. However, Koster et al. reported that Al–Cu–Fe quasicrystal phase, with a density of 4.7 g/cm\(^3\) [10], may react with Al phase to form \( \text{\textmu} \) phase during heat-treatment [34]. Because the VHP process lasts 5 h at 550\(^\circ\)C, it is reasonable to expect that Al atoms may diffuse into the Al–Cu–Fe quasicrystal particles and may react to form the \( \text{\textmu} \) phase.

### 3.3. Elastic modulus and static mechanical properties

The measured elastic modulus (\( E \)) values of the \( \text{\textmu} \) and quasicrystal phases were 168 and 182 GPa for \( \text{\textmu} \) phase and the Al–Cu–Fe quasicrystal phases, respectively, as shown in Table 4. The microhardness values (at 500 g load) of these two phases were also very similar, as shown in Table 4. The quasicrystal and \( \text{\textmu} \) phase materials used for the modulus.
The elastic modulus measurements were fully dense samples made by hot iso-
static pressing of gas atomized powders of the stoichiometric al-
lloys at 800 and 700°C, respectively. The CTE of the 012 phase, given in Table 4, is 15.45 × 10⁻⁶ K⁻¹, which is slightly larger than the CTE of the quasicrystal phase, 12.6 × 10⁻⁶ K⁻¹ [35]. The similarity of the properties between these two phases suggests that, although there may be no quasicrystal phase remaining in the reinforcement particles, the resulting 012 phase can still act as an effective reinforce-
ment for promoting the modulus and strength of the com-
posites.

The elastic modulus measurements for the composite ma-
terials ranged from 80 to 100 GPa, as 012 phase reinforce-
ment loading was increased from 15 to 30 vol.%. As given in Fig. 3, most of the modulus values of the composites are close to or at the upper bound model values. This observa-
tion demonstrates the effectiveness of the 012 phase reinforce-
ment in this composite material and is consistent with good bond-
ing between Al and reinforcement particle phases. The modulus of the 012 phase, the fully transformed reinforcement particulate, was used for all of the model calculations of the composite modulus.

### 3.4. Tensile properties

The tensile property values including elastic modulus, ul-
timate tensile strength (UTS) and yield strength (YS) of all the AFF and GFF composite samples can be seen in Figs. 3 and 4, Table 5. The YS and UTS of all GFF composite sam-
ple are much above the corresponding values of the corre-
sponding elemental Al matrix sample, whose YS is 53 MPa and UTS is 95 MPa [26]. The same is true for all of the AFF composite samples. The YS and UTS of the corresponding

<table>
<thead>
<tr>
<th>Samples</th>
<th>AFF-15</th>
<th>GFF-15</th>
<th>AFF-20</th>
<th>GFF-20</th>
<th>AFF-30</th>
<th>GFF-30</th>
</tr>
</thead>
<tbody>
<tr>
<td>Density (g/cm³)</td>
<td>3.0</td>
<td>3.0</td>
<td>3.09</td>
<td>3.09</td>
<td>3.28</td>
<td>3.28</td>
</tr>
<tr>
<td>Theoretical density (g/cm³)</td>
<td>3.0</td>
<td>3.1</td>
<td>3.3</td>
<td></td>
<td></td>
<td></td>
</tr>
</tbody>
</table>

Table 4  
Properties of Al₆₃Cu₂₅Fe₁₂ quasicrystal phase and Al₇Cu₂Fe (012) phase

<table>
<thead>
<tr>
<th></th>
<th>Elastic modulus (GPa)</th>
<th>Microhardness (Hₐ)</th>
<th>CTE (× 10⁻⁶ K⁻¹)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al₆₃Cu₂₅Fe₁₂</td>
<td>182</td>
<td>922 ± 25</td>
<td>12.6 [35]</td>
</tr>
<tr>
<td>Al₇Cu₂Fe</td>
<td>168</td>
<td>935 ± 80 [34]</td>
<td>15.45</td>
</tr>
</tbody>
</table>

Fig. 3 gives a clear comparison of YS and UTS between AFF and GFF samples. Fig. 4(a) shows that for both AFF and GFF samples, YS has a quite linear relationship with the reinforcement volume fraction. The YS of AFF samples with different volume fractions of reinforcement also are consistently higher than GFF samples. This is reasonable because the commercial purity AFF samples contain more oxide or dissolved impurities that would promote higher strength. From Fig. 4(b), it can be seen that the UTS of AFF is obviously higher than GFF at 15 vol.% reinforcement loading, but at the 20 and 30 vol.% loading AFF samples show lower or about the same UTS values as GFF samples. Thus, it appears that after tensile samples pass the yielding point, there will be more and more micromechanical damage accumulated inside the material along with the tensile strain increase. When the damage accumulates to some critical level, the tensile strain reaches the UTS point and the tensile sample fails. If the damage accumulation rate is faster, UTS will be reached earlier and its value will be lower. From this preliminary analysis, we can see that although GFF samples have a lower YS level than AFF samples, they may have lower or about the same UTS levels as AFF samples.

<table>
<thead>
<tr>
<th></th>
<th>UTS (MPa)</th>
<th>YS (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>GFF-VHP</td>
<td>154</td>
<td>109</td>
</tr>
<tr>
<td>AFF-VHP</td>
<td>176</td>
<td>135</td>
</tr>
</tbody>
</table>

Table 5  
Tensile properties of GFF and AFF composite samples
3.5. Fracture mode

Fig. 5 shows the fracture surfaces of AFF-30 and GFF-30 composites tensile samples, which are typical for all the composites samples. The fracture surfaces show a mixture of fracture modes. In the Al matrix area, there are small dimples and tearing ridges, which suggest a very ductile fracture in the matrix area. There are many reinforcement particles cracking on the fracture surface, but there are also some small spherical particles apparently on the surface, especially for AFF type composites. This indicates some interface debonding between matrix and finer reinforcement particles. However, no particle pull-out appears on the GFF-30 sample fracture surfaces, suggesting that the GARS powders can be more easily sintered than CIGA powders because of a much thinner oxide layer on the GARS powder surfaces, compared to the CIGA powder. Much more detailed experiments and analysis on the sintering behavior of these two kinds of powders can be found in many previous reports [16,21,22,26].

4. Testing of composite strengthening models

During the past two decades a large number of investigations have been carried out to reveal the strengthening mechanisms of metal matrix composites, where both continuum and micromechanical models have been developed. As a result of these investigations the major mechanisms that may contribute to the direct and secondary strengthening of a composite have been deduced:

(a) Direct strengthening results from load transfer from the matrix to the reinforcement via shear stresses at the
interface between the components (shear lag theory) [12].

Secondary strengthening from increased density of dislocations are generated because of the differential thermal contraction between matrix and reinforcement [14,38] during cooling from an elevated processing temperature.

Geometrically necessary dislocation (secondary) strengthening is common in precipitation strengthened materials [39] and occurs during initial plastic deformation in a micromechanical manner.

All the above strengthening mechanisms, from a theoretical point of view, seem suitable for particulate reinforced Al matrix composites. For this re-examination of composite strengthening, the simplified microstructures of elemental Al matrix composites consolidated by vacuum hot pressing seem to be a preferred material. The dwell time at high temperature of vacuum hot pressing is much longer and the sample cooling rate after full consolidation is much slower than high strain-rate consolidation methods, e.g., the quasi-isostatic forging process [16]. Thus, the effects of direct strengthening should be emphasized, since residual dislocation density in the sample matrix and secondary dislocation strengthening should be significantly reduced. Also, the Al matrix used in the composites was produced from either commercial purity (99.7%) or high purity (99.99%) elemental powder, which minimizes the possibility of a secondary precipitation strengthening effect. Most dislocations can also be tested. Therefore, this evaluation will include testing of these individual strengthening contributions and will also test a combination of direct strengthening, and the remaining secondary strengthening mechanisms, as mentioned above.

4.1. Load transfer mechanism

The continuum shear lag model was developed originally to predict the strength of continuous fiber reinforced composites by Cox [9]. The model centers on the transfer of tensile stress from matrix to a fiber by means of interfacial shear stresses. For the aspect ratio of SiC whiskers (10:1) and SiC flake fragments (2:1) typically used in particulate reinforced metal matrix composites, the continuum shear lag model underestimates the strength [10,11]. Nardone and Prewo proposed a modified shear lag theory for this direct strengthening and suggested that better agreement would be obtained if the equation was modified to allow for whisker or fiber end loading effects, giving the following equation for strength prediction [12].

\[
\sigma_{ef} = \sigma_{my} \left( \frac{1}{2} V_R (s + 2) + V_m \right)
\]

where \(s\) is the aspect ratio of the reinforcement, and \(\sigma_{ef}\) and \(\sigma_{my}\) represent the yield strength of composite and matrix material, respectively, and \(V_R\) and \(V_m\) represent the volume fraction of reinforcement and matrix, respectively. They also showed that the Orowan strengthening mechanism was insufficient to account for the increase in yield strength of the particulate reinforced Al alloy matrix composites [12].

One difficulty with testing and verification of this continuum approach [12] to load sharing prediction is that it ignores variations in the influence of reinforcement particles on the matrix microstructure, such as increased dislocation density and secondary strengthening caused by CTE difference, modulus and strain incompatibility between matrix and reinforcement phases, and composite processing procedures.

By carefully matching the matrix precipitate microstructures, Chawla [40] showed that there is a good correlation between the experimental data of SiC-reinforced T8-tempered Al alloy (Al2080) matrix composites and T6-tempered matrix samples of Al2080 using the above modified shear lag model. On the other hand, studies reported by Arsenault and collaborators [10] showed that the shear lag model only gave a very small predicted amount of YS increase, compared with experimental results. If we test the modified shear lag model of Eq. (1) on the data of this study, for example, for the composites with a 30 vol.% of reinforcement reported in this paper, the YS increase prediction would be only 15% above the YS of the elemental Al matrix. Note that this calculation assumes an ideal aspect ratio for the spherical reinforcement of 1.0, i.e., assumes a lack of reinforcement particulate agglomeration. If we use the specific YS data given for each type of matrix in this report, a numerical YS increase is predicted of only 8 and 13 MPa for the GFF-30 and AFF-30 samples, respectively, instead of 138 and 111 MPa, as given in Table 5.

Alternatively, the load transfer capability or partitioning between phases in composite materials can be directly measured in situ by X-ray or neutron diffraction methods [41,42]. From these measurements, the specific load partitioning between each phase in a composite can be found. Therefore, it can be clearly known specifically how much load is transferred to reinforcement phases and this data can be used to calculate actual load transfer strengthening, instead of using the shear lag model to estimate the values.

For making an accurate assessment of the load transfer strengthening contribution, the actual load bearing stresses in Al matrix and reinforcement particles were measured in situ while applying external tensile stress by a neutron diffraction method at the NIST Neutron Research Center. The detailed neutron diffraction measurement procedure is reported.
in another paper [29]. The applied tensile stress level during the load partitioning measurement of all the composites samples was well within the elastic region of the material. The neutron diffraction measurement results of matrix and reinforcement phase bearing stresses, $\sigma_M^C$ and $\sigma_R^C$, respectively, in all composite samples are shown in Fig. 6. The load bearing ratios, $\sigma_R^C/\sigma_M^C$, of reinforcement and matrix (Al) phases are equivalent to the slope values listed in each plot in Fig. 6 and were obtained by fitting linearly the data trends. These load bearing ratios [33] measured in the elastic range were used in the subsequent load transfer strengthening analysis, although the operative load bearing ratios may have some minor decrease due to the hardening of the Al matrix during the plastic yielding portion of tensile deformation. Note that
the load transfer measurement for the 30 vol.% loaded composites are for quasi-isotropically forged samples, assuming that they are equal to that of the vacuum hot pressed samples at the same loading. In other words, the vacuum hot pressed samples were not available for these measurements and we can assume that the processing method should not affect the load transfer effect as long as the matrix and the reinforcement particles are bonded well and each sample is fully dense.

Now with these measurement results of load partitioning and load bearing ratios, we can calculate directly how much direct load transfer strengthening these composite materials demonstrate. According to Ref. [43,44], the average phase stresses in an Al matrix and in the reinforcement particles can be measured by neutron diffraction and have the following relationship with the external applied stress:

\[ \sigma^A = V_k \sigma^R + (1 - V_k)\sigma^M \]

(2)

where \( \sigma^A \) is the external applied stress on the PRA composites, \( V_k \) is the volume fraction of reinforcement, and \( \sigma^R \) and \( \sigma^M \) are the average bearing stresses of the reinforcement particles and the matrix, respectively.

Therefore the yield stress increase in the PRA composites that results from the load transfer effect (\( \Delta \sigma^{LT} \)) can be calculated as follows:

\[ \Delta \sigma^{LT} = \sigma^A - \sigma^M \]

(3)

where \( \sigma^M \) is the yield stress of the unreinforced matrix material. Therefore

\[ \Delta \sigma^{LT} = V_k \sigma^R + (1 - V_k)\sigma^M - \sigma^M \]

(4)

where \( \Delta \sigma^{LT} \) is the YS increase from the load transfer mechanism. Assuming that the Al matrix bearing stress \( \sigma^M \) is approximately equal to the yield stress of the pure Al matrix material \( \sigma^M \) when the composite is yielding, the load transfer strengthening can be calculated by following Eq. (5).

\[ \Delta \sigma^{LT} = \left( \frac{\sigma^R}{\sigma^M} - 1 \right) V_k \sigma^M \]

(5)

where the results are summarized in Table 6. As shown in Table 6, the direct load transfer strengthening of the composites samples only contributed a small part of the observed (experimental) strengthening, which is also consistent with the results of Al/SiC composites that are reported by Arsenault et al. [10]. In addition, the measured levels of direct strengthening are significantly greater than the predictions of Eq. (1), i.e., the modified shear lag model, that were cited for the AFF-30 and GFF-30 composite, as an example.

### 4.2 Thermal expansion mismatch model

In addition to the direct strengthening from the load transfer effect, an important secondary strengthening contribution to be considered should be the plastic deformation effect caused by \( \Delta \mathrm{CTE} \) between the Al matrix and the reinforcement particles. Arsenault and Shi [14,15] showed that an increased dislocation density of the matrix does result from a difference in the coefficient of thermal expansion (CTE) of an Al matrix and SiC reinforcement particles and that the increased matrix dislocation density does promote a large portion of the observed increase in Al/SiC composite yield strength. In other words, because the coefficient of thermal expansion of a typical aluminum alloy is about ten times that of SiC, there can be relatively high residual stress around each SiC particle at ambient temperature after high temperature processing of such composites [14]. The equation proposed for prediction of yield strength increase by \( \Delta \mathrm{CTE} \) is, as follows [15],

\[ \Delta \sigma^{\mathrm{CTE}} = \alpha \mu b \frac{V_k}{1 - V_k} \left( \frac{B}{b} \right)^{1/2} \left( \frac{1}{T} \right)^{2/3} \]

(6)

where \( \alpha \) is a constant that is equal to 1.25, \( \mu \) is the shear modulus of the pure Al matrix (26.4 GPa), \( b \) is the Burgers vector (2.86 \times 10^{-10} m), \( V_k \) is the volume fraction of reinforcement, \( B \) is 12 for spherical particle reinforcement, \( \varepsilon \) is the misfit strain that is a function of processing and test temperatures due to the \( \Delta \mathrm{CTE} \), and \( T \) the average diameter of the reinforcement (about 5.5 \( \mu \)m) [26].

For analysis of the PRA composites of this study, the CTE of each reinforcement particle in the composite, \( \omega \), phase, is 15.45 \( \times \) 10^{-6} K^{-1} and the CTE of pure Al is 23.6 \( \times \) 10^{-6} K^{-1}. The processing temperature is 550°C and, assuming room temperature is 20°C, the temperature difference is 530°C, i.e., 530 K. Then, misfit strain \( \varepsilon \) can be calculated by Eq. (7) and is equal to 6.54 \times 10^{-6},

\[ \varepsilon = \sqrt{\Delta \mathrm{CTE} \times \Delta T} \]

(7)

A summary of the results calculated with Eq (6) are found in Table 6. Inspection of the results in Table 6 showed that \( \Delta \mathrm{CTE} \) strengthening represents a relatively large portion of the total observed (experimental) strength increase. This observation is consistent with the results from Arsenault [10,14,15], but does not account for any differences in the interparticle bonding strength caused by powder surface oxide thickness differences, for example [21]. In addition, the analysis based on Eq. (6) predicts a state of residual tensile stress in the Al matrix phase, using the assumption of a fully relaxed stress state at the start of cooling from the consolidation temperature. Later neutron diffraction results [29] have
indicated a residual compressive stress state in the matrix, in general, and have stimulated further study of this issue.

4.3. Geometrically necessary dislocation strengthening model

When an external tensile load is applied to a composite material and it arrives at the matrix yielding point, there will be geometrically necessary dislocations (GND) generated\(^{[39]}\) in the Al/reinforcement interface area because of the yield strength and elastic modulus differences between Al and reinforcement particles, as shown in Fig. 7. Otherwise, extensive interface debonding between Al and particles would take place in the composites, which is contrary to the microstructural evidence (Fig. 5) of the fracture surfaces\(^{[16]}\). At the 0.2% strain of the composite material samples, AFF and GFF, the Al matrix already had some plastic deformation, as can be seen in the stress–strain curves in Fig. 8. In order to compensate for the strain incompatibility between matrix and particles, extensive GND dislocations must be generated in the matrix\(^{[39]}\).

A GND density increase around each interface region can also cause a secondary (indirect) strengthening effect for the composite material\(^{[39]}\). This GND effect can be also seen in some precipitation strengthening materials\(^{[39]}\), where it acts to increase the population and spatial density of dislocation barriers far beyond the initial precipitate distribution. The yield strength increase caused by GND strengthening (\(\Delta \sigma_{GND}\)) can be calculated by the following equation\(^{[39]}\).

\[
\Delta \sigma_{GND} = \alpha\mu b V_R \varepsilon_y \sqrt{8D}
\]  

where \(\alpha\) is the constant that is equal to 1.25, \(\mu\) the shear modulus of the pure Al matrix (26.4 GPa), \(b\) the Burgers vector \((2.86 \times 10^{-10} \text{ m})\), \(V_R\) the volume fraction of reinforcement, \(D\) the diameter of the prismatic dislocation loop around the reinforcement particles, and \(\varepsilon_y\) the yielding strain. By using an approximate \(D\) value of 5.5 \(\mu\text{m}\), which is the average diameter of reinforcement particles\(^{[26]}\) and \(\varepsilon_y\) value of 0.2%, the yield strength increase from the GND effect for each sample was calculated. Again, the calculated results are found in Table 6. The contributions from the GND strengthening mechanism increases with increased reinforcement particle volume fraction. The previous study of Sekine\(^{[45]}\) also showed that the GND mechanism is a significant contribution to composite strengthening, which is consistent with our analysis results. Again, the analysis of Eq. (8) does not take into account any differences in the type of powders\(^{[21]}\) used to process the particulate reinforced Al matrix composites.

4.4. Combined effect of strengthening mechanisms

In the calculation results from each of the three well-accepted direct and indirect strengthening contributions shown in Table 6, one can observe an under-prediction of the experimental results for the YS increases in our composite samples, if applied individually. Consistent with the micromechanical aspects of current composite strengthening theory, the yield strength increase may be better predicted from a combination of both an elastic effect, i.e., direct load transfer, and plastic-type effects, i.e., \(\Delta\text{CTE}\) and GND effects. The plastic-type secondary effects can be added to account for the highly local yielding phenomenon that occurs between reinforcement particles and the matrix phase, prior to reaching the 0.2% yield strength, a macroscopic yielding point. Thus, a combined effect may be proposed as a simple summation, which can be calculated for the two kinds of Al/QXL composites by Eq. (9).

\[
\Delta \sigma_y = \Delta \sigma_{LT} + \Delta \sigma_{\text{CTE}} + \Delta \sigma_{GND}
\]  

Table 6 shows the calculation results for each composite sample, using Eq. (9). Comparing the experimental YS increase values with predicted values from the combined calculation, very good agreement with the relationship given in Eq. (9) was found for all the composite samples. The YS increase values in Table 6 were calculated based on the 0.2% YS values of elemental CIGA Al and GARS.
Al matrix materials [26], which are 85 and 53 MPa, respectively. Fig. 9 shows a graphical view of the agreement between the experimental and predicted values for AFF- and GFF-VHP samples, as a function of reinforcement content.

For the AFF samples, the predicted values agree with the experimental values so well that the predicted trend almost overlaps with the experimental one. For GFF samples, the general trend of the prediction and the actual YS increase values also agree quite well, except that the predicted values are all lower than the experimental values. All of the GFF composite materials show higher experimental YS increase values than the combined analysis of Eq. (9) predicts. This observation suggests that the use of high purity powders may allow operation of some additional, but unknown, strengthening effect that remains to be explored.

5. Conclusions

1. Microstructural analysis of elemental Al matrix composites reinforced by Al–Cu–Fe alloy particles demonstrated that the quasicrystalline phase in the as-solidified Al–Cu–Fe particles transformed during high temperature consolidation processing to a crystalline α phase, which has similar elastic modulus, coefficient of thermal expansion, and hardness properties, but a reduced density. Tensile test results also were collected from an extensive set of composite samples with reinforcement loadings from 15 to 30% (by volume).

2. The Al/Al–Cu–Fe (30 vol.%) composites produced by vacuum hot pressing have lower tensile properties than some equivalent composites produced earlier by a quasi-isostatic forging process. However, all of the vacuum hot pressed composites appear to be fully dense, with strong interparticle bonding, and exhibit elevated elastic modulus values, which agree well with upper bound predictions using the α phase elastic modulus.

3. Neutron diffraction measurements of the direct strengthening effects (load bearing ratio) allowed an assessment of the relative influence of direct strengthening and two secondary composite strengthening mechanisms on the yield strength of this model composite system.

4. The results suggest that for elemental Al matrix composite samples without precipitation strengthening and severe strain hardening during consolidation, the direct and relevant secondary strengthening mechanisms can be combined in a simple summation to predict accurately the yield strength increase of the composites.

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