Microstructure, plastic deformation and strengthening mechanisms of an Al–Mg–Si alloy with a bimodal grain structure

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Al6063 alloys with bimodal grain size distributions comprised of ultrafine-grained (UFG) and coarse-grained (CG) regions were produced via mechanical milling followed by hot extrusion. High-energy planetary ball milling for 22.5 h with a rotational speed of 350 rpm was employed for the synthesis of nanocrystalline Al6063 powders. The as-milled Al6063 powders were mixed with 15, 30, and 45 vol.% of the unmilled powders and then the powder mixtures were consolidated via extrusion at 450 °C with an extrusion ratio of 9:1. The microstructure of the bimodal extrudates was investigated using optical microscope, transmission electron microscope (TEM) and field emission scanning electron microscope equipped with an electron backscattered diffraction (EBSD) detector. The deformation behavior was investigated by means of uniaxial tensile tests. The bimodal Al6063 exhibited balanced mechanical properties, including high yield stress and ultimate tensile strength resulting from the UFG regions together with reasonable ductility attained from the CG areas. The fracture surfaces demonstrated a ductile fracture mode, in which the dimple size was correlated with the grain structure. The strengthening mechanisms are discussed based on the dislocation models and the functions of the CGs in the deformation behavior and ductility enhancement of bimodal Al6063 are explored.

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1. Introduction

Grain refinement is a significant issue in load bearing components due to the establishment of opportunities to design lighter and stronger structures [1–4]. Recent advancements in high-energy mechanical milling (HEMM) and further consolidation through thermomechanical processes (e.g. hot pressing, hot extrusion, and hot isostatic pressing) help to produce very fine grains in metallic microstructures, termed as ultrafine-grained (UFG, average grain size <1 μm) materials. These materials demonstrate a noticeable increase in strength compared to their coarse-grained (CG, average grain size >10 μm) counterparts [5–9]. Considering the Hall–Petch relationship for conventional CG polycrystals, the yield stress increases with a decrease in grain size [10,11]. However, the strength of UFG materials depends not only on the average grain size, but also on the grain size distribution [12,13]. On the other hand, although high strength materials are desirable, the grain refinement severely increases resistances to dislocation generation and motion and subsequently decreases work hardening capacity, plastic deformation, and ductility of UFG materials [12].

It has been observed that UFG materials exhibit a limited uniform elongation (usually less than 5% in tension) which makes them inappropriate for applications to load bearing components or for secondary forming processes [14]. The limited ductility is attributed to artifacts from processing [15], plastic instability in tension [16], and crack nucleation [17]. Various methods, including heat treatment [18,19] and accurately controlled manufacturing processes, have been proposed for considerably diminishing defects, such as voids and inclusions [20], and enhancing ductility without a significant degradation of strength in UFG materials. Because the heat treatment process may coarsen the grains in the UFG materials, the controlled manufacturing is costly and hard to implement on a large scale. Another promising approach is tailoring the grain structure by introducing soft inclusions of the same phase in the strong UFG matrix [21]. For example, bimodal microstructures can be produced through mixing hard mechanically-milled powders with soft unmilled powders followed by consolidation. The UFG regions retain their high strengths while high extent of deformations can occur in the soft CG regions. Tailoring the microstructure through powder metallurgy (PM) routes has been successfully used for fabrications of several bimodal metallic materials, such as Al [22], Al–Mg [16,17,23–26], Cu–Al [27], and Ni [28]. Witkin et al. [24] have
demonstrated 286% increase in the elongation and only a 13.6% decrease in the yield stress in the UFG Al–7.5% Mg alloy containing 30% of coarse grains. Han et al. [16] have demonstrated a significant increase in strength and ductility of a bimodal Al5083 sample along the extrusion direction compared to that of transverse direction.

Al6063 is a heat treatable aluminum alloy which is widely used as medium strength structural material possesses interesting properties like good weldability, corrosion resistance, and immunity to stress corrosion cracking [29]. Recently, we have prepared Al6063 with a high hardness and strength, but a low ductility [30]. In this paper, we describe the fabrication of UFG Al6063 alloy bulks with bimodal grain structures engineered for high strengths with enhanced ductilities. To the best of authors’ knowledge, tailoring the microstructure of Al–Mg–Si (6xxx series) alloys through the PM route has not been explored. The microstructural features and tensile properties of the bimodal grain structured Al6063 with various amounts of CG regions are investigated. Moreover, the deformation behavior in terms of strengthening and ductility enhancement mechanisms are systematically discussed.

2. Materials and methods

Gas-atomized Al6063 alloy powders with a chemical composition of Al–0.92 Mg–0.4 Si–0.33 Fe–0.01 Cu (in wt.%) was used as a starting material. 2 wt.% Stearic acid (CH₃(CH₂)₇COOH, Merck, Germany) was added to the Al6063 powders as a process control agent (PCA) and mixed for 1 h with a rotational speed of 100 rpm. HEMM of the Al6063 powders was performed in a planetary ball mill for 22.5 h under Ar atmosphere with a rotational speed of 350 rpm. High-chromium steel balls with diameters of 11 and 19 mm were used as milling media. The charge (ball-to-powder weight) ratio was fixed at 10:1.

The milled Al powders were mixed with 15, 30, and 45 vol.% of unmilled powders. The powder mixtures were degassed at 400 °C for 1 h in a tube furnace under Ar atmosphere. The powders were filled in 30 mm diameter Al cans and then cold compacted to attain a relative density of approximately 0.75. After preheating at 450 °C for 0.7 h under Ar atmosphere, the compacted powders were extruded with an extrusion ratio of 9:1 in order to produce cylindrical billets with a diameter of 10 mm and a length of 100 mm. The plain milled and unmilled powders were consolidated with the same procedure for comparison.

Morphology and size of the powders were characterized using a field emission scanning electron microscope (FESEM, Tescan, Czech Republic). Microstructural observations of the billets were performed using an optical microscope (OM, Olympus, Japan). The samples were prepared through a common metallographic procedure and then were etched in Keller’s solution (2.5% HNO₃, 1.5% HCl, 1% HF, and 95% H₂O). The grain structures of the samples were examined using a FESEM (Hitachi, SU-6600, Japan) equipped with a high resolution electron backscattered diffraction (EBSD) detector. The specimens were mechanically ground, pre-polished using 0.9, 0.3, and 0.1 μm diamond pastes, and finally polished using 0.04 μm colloidal silica suspensions. The orientation mapping was made on a rectangular grid with a step size of 20 nm. The EBSD data processing was performed using the OIM analysis software (version 5.2, EDAX, NJ). Clean-up with minimum confidence index of 0.1, minimum grain size of 20 nm, and grain tolerance angle of 2° was applied in order to eliminate the incorrectly indexed points. The precipitates in the microstructure were characterized by transmission electron microscope (TEM, JEM-2100F, JEOL, Japan) operated at 200 kV. TEM samples were prepared by dimpling (Model 656, Gatan, Inc., CA) and ion milling (Model 691, Gatan, Inc., CA) of thin foils until perforation.

Uniaxial tensile tests were performed in order to evaluate the mechanical behavior of the consolidated powders. Flat tensile specimens with a gauge length of 8.5 mm and a cross section area of 0.5 × 2 mm² were prepared through electro discharge machining (EDM). The tensile axis was parallel to the extrusion direction. Tensile tests were performed using a Hounsfield universal machine (H10KS, Tinius Olsen, Inc., PA) at a constant crosshead speed of 0.2 mm min⁻¹ according to the ASTM E8 standard. The tensile tests were repeated three times for assuring the accuracy of tensile properties and reproducibility of the testing. Elongation of the tensile specimens was determined by fitting ends of the fractured specimen together carefully and measuring the distance between the gauge marks. The fracture surfaces of the tensile test samples were explored using SEM.

3. Results

Morphologies of gas-atomized and 22.5 h mechanically-milled Al6063 powders are presented in Fig. 1. Spherical and teardrop morphologies of the gas-atomized particles (Fig. 1a) were changed into a relatively flake shape (Fig. 1b). The average size of the powders was increased from 80 μm to 90 μm due to the flattening effect of mechanical milling. The milled powders comprised of welded and fragmented particles due to the impacts of high energy balls during HEMM.

Fig. 2 presents the optical micrographs of the Al6063 alloys extruded from mixtures of the milled and 15, 30, and 45 vol.% of the unmilled powders along and perpendicular to the extrusion direction. The coarse grains were uniformly distributed in the UFG matrix, revealing the formation of bimodal grain structures. The coarse and ultrafine grains are observed in the micrographs as bright and dark areas, respectively, owing to the more susceptibility of the UFGs to the chemical reagent as a result of higher fraction of grain boundaries. While the CGs were elongated along the extrusion axis and formed narrow bands within the UFG matrix, the UFG regions were slightly deformed. This can be attributed to the small crystallite size and high hardness of the milled powders that make their deformation during extrusion complicated. In the
15 vol.% CG specimen, the CG regions were uniformly distributed with appropriate intervals so that individual particles did not agglomerate or coalesce (Fig. 2a and b). The bimodal alloy containing 30 vol.% CGs has a similar microstructure, however, it contains more CG regions with shorter distances from each other (Fig. 2c and d). On the contrary, in the 45 vol.% CG material, some CGs coalesced and formed bands, which were somewhat thicker along the transverse direction and irregularly branched and elongated along the longitudinal direction (Fig. 2e and f).

Complementary information on the grain structures was obtained using the EBSD analyses. Fig. 3 illustrates the color-coded EBSD and image quality (IQ) maps of the bimodal grain structured Al6063 alloy with 30 vol.% CG regions. The presence of grains with a large number of dissimilar colors in the EBSD map (Fig. 3a) discloses significant differences in the orientation of neighbor grains. Moreover, a bimodal grain structure consisting of ultrafine grains and coarse grains was observed. For clarification, a CG region is highlighted in the IQ map, as shown in Fig. 3c. In spite of the UFGs, the CGs were more elongated along the extrusion direction. The grain size distribution diagram obtained from the EBSD analysis exhibits a bimodal distribution with two peaks appeared at grain diameters of approximately 1 \( \mu m \) and 2.2 \( \mu m \) (Fig. 4a). The former is related to the UFG regions in which the grain size varied in the range of 0.08–1.4 \( \mu m \) with an average grain size of 0.63 \( \mu m \). The latter is connected to the CG regions in which very large grains in the order of several micrometers can be detected. In the IQ map, low-angle and high-angle grain boundaries (denoted by LABs and HABs) were defined by misorientations between adjacent grains of 2–15° and >15°, respectively. It should be noted that the boundaries with misorientations below 2° were omitted due...
to the less resolution in the EBSD analysis. The LABs and HABs were
presented in the IQ map with white and blue lines, respectively.
The frequency distribution of misorientation angles with a super-
-imposed theoretical distribution for aggregates of randomly orien-
ted grains is presented in Fig. 4b. The misorientation angle
distribution exhibits the presence of approximately 31% LABs
which were mostly located within the coarse grains. Meanwhile, 
ultrafine grains were mostly surrounded by the HABs so that a
large fraction of HABs (~69%) were present in the microstructure.
Consequently, a two-peak misorientation distribution is found
with one peak in the high misorientation angle range around 48°
and the other peak in the small misorientation angle range. Such
a bimodal misorientation angle distribution is typical for heavily
deformed samples [12,31]. The average misorientation angle was
found to be ~29°. The significant difference between the frequency
distribution and random distribution exhibits that the bimodal Al
alloy contains a strong texture. In order to assess the texture com-
ponents of the samples, the orientation distribution functions
(ODFs) obtained from EBSD analysis were plotted at

\[ h \quad i \quad 111 \]/C176

angles [32]. The P component is the predominant texture com-
ponent in hot consolidated Al6063 alloys. The P texture component
has a fraction of 38% and a high intensity of 24 times random in the
CG Al6063 since it has a fraction of 17% with an intensity of 10
times random in the UFG sample (Fig. 5a and b). The fraction and
intensity of the P component in the bimodal alloy containing
30 vol.% CGs are between the plain UFG and CG specimens (26%
and 13 times random, respectively), as shown in Fig. 5c. The copper
texture component arises from hot powder extrusion with a lower intensity compared to the recrystallization P component.
The intensity of the copper component is 3.5, 4, and 4.3 times ran-
dom in the CG, bimodal and UFG Al6063, respectively. Thus, it
appeared that the intensity of copper texture was amplified due to
HEMM processing which is in agreement with our previous
investigation [33]. Moreover, it has a low fraction of 7%, 8.4%,
and 9% in the CG, bimodal and UFG Al6063, respectively.

Fig. 6 shows a TEM image of the UFG alloy. Fine precipitates
(40–200 nm) which are located both within the Al grains and on
the grain boundaries are visible. The characterization of second
phase particles via energy dispersive X-ray spectroscopy (EDS)
analysis and selected area diffraction (SAD) patterns indicates that
the particles are Mg2Si, AlFeSi and Al13Mg2FeSi2 precipitates.

The tensile stress–strain curves of the bimodal and plain (UGF
and CG) Al6063 alloys are presented in Fig. 7. The CG sample had
a low strength, but a high fracture elongation of ~17.8%. On the
other hand, the UFG material exhibited high yield and tensile strengths (250 MPa and 280 MPa, respectively) with a very limited ductility (~1.3%). The tensile ductilities in the bimodal samples were enhanced to tensile elongations in the range of ~2.8–11.3%.
The appearance of work hardening region after yielding in the
bimodal samples is also noticed. However, the bimodal Al6063
alloys showed lower strengths compared to the UFG sample. Thus,
the bimodal Al6063 alloys demonstrated an attractive balance of
strength and ductility compared to their plain counterparts. Variations in strength and ductility of the Al6063 vs. the percentage of coarse grains are plotted in Fig. 8a. Although for the specimens with a low content of CGs (<30 vol.%), the strength decrease and the ductility increase were almost linear with an increase in CGs, the addition of 45 vol.% CGs caused a nonlinear behavior. In order to find the best combination of strength and elongation, tensile strength vs. elongation results, which is usually referred it as the “banana curve”, were plotted in Fig. 8b. The highest level of
strength × ductility is found in the 45 vol.% CG alloy.

From the macroscopic point of view, the CG sample showed a
significant necking during tensile loading and failed noticebly in
ductile manner with slant surfaces aligned ~45°. Little to no
necking occurred in the UFG alloy and the specimen failed abruptly
with a jagged fracture path perpendicular to the loading direction.
The fracture surfaces of the bimodal alloys showed a mixed frac-
ture mode with large shear lips at the edge and a flat region at
the center of samples. Fracture surfaces of the tensile specimens

Fig. 3. Electron backscatter diffraction maps of Al6063 alloy with a bimodal grain structure: (a) color-coded; (b) image quality; and (c) elucidation of CG region within the UFG matrix.
at a high magnification are represented in Fig. 9. Well-developed dimples over the entire fracture surface of the CG Al6063 were observed in Fig. 9a. On the fracture surface of the UFG sample (Fig. 9b), many dimples whose sizes are in the order of magnitude of one or more UFG grain diameters were detected. It is clear that the sizes of the dimples in the UFG material (~200–800 nm) are much smaller than those of CG alloy (~1–5 μm). The fracture surface of the bimodal Al6063 alloys (Fig. 9c–e) exhibited fine dimples related to the UFG regions and coarse dimples regarding the CG regions. Additionally, localized plastic deformation (necking) within the CG regions and delaminations at the CG/UFG interfaces were observed on the fracture surfaces of the bimodal alloys, revealing an extensive plastic deformation prior to fracture. Cavitations were also examined along the edge of the bimodal samples, where the major deformation concentrated near the fracture region. It is clear from the fracture surfaces that the height of the failed CG regions is larger than the UFG matrix level, indicating that the final failure occurred in the CG regions.

4. Discussion

4.1. Strengthening mechanism

The bimodal Al6063 alloys containing <30 vol.% CGs showed high strengths which are comparable with the strength of the UFG alloy. This high yield stress can be described in terms of the strengthening mechanisms. Because it is difficult to aggregate all of the strengthening mechanisms in a single constitutive equation, a simple linear superposition of strengthening mechanisms from grain boundaries ($\sigma_{GB}$), dislocations ($\sigma_d$), and second phase particles ($\sigma_p$) is considered for prediction of yield stress, as follows:

$$\sigma_y = \sigma_{GB} + \sigma_d + \sigma_p. \quad (1)$$

The contribution of the grain boundaries (HABs) can be stated by Hall–Petch equation [34,35]:

$$\sigma_{HP} = \sigma_t + \frac{k_p}{\sqrt{d}}, \quad \sigma = M \tau_c \sqrt{\frac{G b M}{d}}, \quad \sigma = \frac{G b M}{d^{1/2}}, \quad \sigma_{GB} = \sqrt{\frac{G b M}{d}}.$$  

where $\sigma_t$ is the lattice friction, $k_p$ the strengthening coefficient constant, $d$ the effective grain diameter, $M$ the Taylor factor, $\tau_c$ critical shear stress of a free single crystal, $\tau_{GRSS}$ critical resolved shear stress, $r$ the distance from the nearest dislocation piled-up to the dislocation source in the adjacent grain. The $\sigma_t, k_p, M$ parameters are considered 50 MPa, 70 MPa μm0.5 and 3.06 respectively [36,37]. While both the UFGs and CGs are concurrently present in the microstructure, the rule of mixtures was employed in order to calculate the effective grain size of the bimodal alloys, as expressed by:

$$d_e = f_{CG} \times d_{CG} + f_{UFG} \times d_{UFG}. \quad (3)$$

where $f_{CG}$ and $f_{UFG}$ are the volume fractions of the CG and UFG regions, respectively. $d_{CG}$ and $d_{UFG}$ are the average grain sizes of the CGs and UFGs, respectively. The average grain size of the equiaxed ultrafine grains ($d_{UFG}$) was directly measured from the EBSD maps. While the CGs are significantly elongated along the extrusion direction, their shape has been considered as an oval with a large diameter along the extrusion axis. Hence, the average grain size of the CGs was calculated by assuming that the area of oval grains is equivalent to a circle with a diameter of $d_{CG}$.

It is well known that the dislocation strengthening is proportional to the square root of the dislocation density. By assuming that the dislocations are located on the LABs, the dislocation strengthening can be estimated by [38]:

$$\sigma_d = MG \alpha \sqrt{1.5b\theta_s S_{\alpha,a}}. \quad (4)$$

where $G$ is the shear modulus (25.8 GPa [36,37]), $\alpha$ a constant (0.1 [36,37]), $b$ the Burgers vector (0.286 nm [36,37]), $\theta_s$ the effective misorientation of the LABs, and $S_{\alpha,a}$ the effective LAB length per unit area. According to the EBSD analysis, the average misorientation of the LABs and the average LAB length per unit area values are higher in the UFG regions compared with the CG areas. Thus, the rule of mixtures was utilized for obtaining $S_{\alpha,a}$ and $\theta_s$ as follows:

$$S_{\alpha,a} = f_{CG} \times S_{\alpha,CG} + f_{UFG} \times S_{\alpha,UFG}, \quad (5)$$

$$\theta_s = f_{CG} \times \theta_{CG} + f_{UFG} \times \theta_{UFG}, \quad (6)$$

where $S_{\alpha,CG}$ and $S_{\alpha,UFG}$ are the average low angle boundary lengths per unit area of the CG and UFG regions, respectively, and $\theta_{CG}$ and $\theta_{UFG}$ are the average misorientations of the LABs of the CG and UFG regions, respectively.

By assuming that most of the solute atoms are depleted from the matrix into the second phase particles, the effect of solute atoms on the strengthening is ignored. An estimation of Orowan strengthening from the second phase particles can be evaluated with the equation [39]:

$$\sigma_p = 0.156G\lambda \ln \left( \frac{0.816b}{\lambda} \right) \frac{1}{\sqrt{1 - \nu}}, \quad (7)$$
where \( \nu \) is the Poisson’s ratio (0.33 [36,37]) and \( f_{p,a} \) and \( \lambda_{a} \) are the effective volume fraction and diameter of second phase particles, respectively. \( f_{p,a} \) and \( \lambda_{a} \) values were also calculated by the rule of mixtures. All the parameters required in aforementioned equations are listed in Table 1. The contribution of each mechanism to the strengthening of the bimodal and plain Al6063 alloys are presented.
Although both the lattice friction and second phase particles have rather analogous shares on the yield stress of various alloys, the contribution of grain boundaries and dislocations in strengthening is declined with an increase in the volume fraction of CGs. The calculated yield stress values for the UFG and bimodal samples are in a good accordance with the experimental yield stress; the predicted yield stresses are about 2–16% higher than the experimental ones. Nevertheless, this model is unable to predict the yield stress of the CG alloy that may be attributed to the low density of dislocations in its microstructure. Moreover, it should be noted that texture evolutions during processing could affect mechanical strength of the hot extruded materials due to the changing of the geometry of different slip systems. In a material with a strong texture, the active slip systems may be unfavorably oriented with respect to the applied stress and thus less-easily activated slip systems may be forced to operate to maintain the sample integrity, resulting in an increase of the yield stress. It is known that occurrence of recrystallization during deformation reduces the final texture strength [40]. Here, a relatively intense recrystallization texture component (P component) together with a weak deformation texture component (copper component) was formed during hot powder extrusion, especially in the CG Al6063 (Fig. 5). The effect of the texture on strengthening can be considered by Taylor factor. The relationship between critical resolved shear stress and Taylor factor can be expressed as [40]:

\[
M = \frac{\sigma_y}{\tau_{CRSS}} = \frac{1}{\cos \phi \cos \lambda}.
\]

where \( \phi \) and \( \lambda \) are the angles between the direction of the applied force and the normal of the slip plane and slip direction, respectively. The mean Taylor factors for the CG, UFG and bimodal containing 30 vol.% CG samples are obtained using EBSD analysis as 2.85, 3.18 and 3.09, respectively. According to the Eq. (8), it is clear that the CG material with a smaller Taylor factor tends to flow easier; however, the UFG alloy with a higher \( M \) value shows higher strength. Interestingly, Taylor factor obtained for the bimodal alloy with 30 vol.% CG is in a reasonable agreement with the value calculated from the rule of mixtures (3.08). As the difference between the Taylor factor values with the theoretical \( M \) value (3.06) is reduced, the error in the calculation of the yield stress with the proposed model decreases.

Fig. 8. (a) Changes of yield stress, ultimate tensile strength and elongation of specimens as a function of the CGs content and (b) banana curve of Al6063 alloy.

Fig. 9. SEM images of the fracture surfaces of Al6063 alloys after tensile testing: (a) CG; (b) UFG; (c) UFG + 15 vol.% CG; (d) UFG + 30 vol.% CG; (e) UFG + 45 vol.% CG.
4.2. Deformation and fracture mechanism

It is well known that for a material with grain size of $>1 \mu m$, the dislocation mechanism of deformation dominates. In contrast, grain boundary processes e.g. grain boundary sliding and grain boundary rotation/migration prevail for a grain size $<20 \text{ nm}$. In the intermediate grain size range (20 nm–1 $\mu m$), the interaction between the dislocation glide and grain boundary shear is quiet efficient [41]. Meyers et al. [42] have shown that the core and mantle model can describe the deformation behavior in the materials with ultrafine grain structure (100 nm–1 $\mu m$). According to this model, the core, or grain interior, is subjected to a more homogeneous state of stress since in the mantle, or grain boundary region, dislocations are generated and work hardened grain boundary layers are formed. As the grain size is reduced, the mantle/core volume fraction ratio and thus yield stress increases. The results of this study indicate that for UFG regions with grain sizes in the range of 0.08–1.4 $\mu m$ and CG regions with grains in the order of several micrometers, Al6063 exhibits deformation that is more characteristic of dislocation control processes. Supporting evidence for this is the tensile stress–strain curves (Fig. 7) that reveals strain hardening region in which multiplication of dislocations occur. However, dislocations glide more easily in the CG regions than in the UFG regions due to the larger grain size and fewer obstacles to slip in CG areas.

Table 1
Required parameters for estimation of the yield stress of Al6063 alloy.

<table>
<thead>
<tr>
<th>Sample</th>
<th>$d_a (\mu m)$</th>
<th>$d_i (\text{rad})$</th>
<th>$S_{av} (\mu m^{-1})$</th>
<th>$\lambda_a (\text{nm})$</th>
<th>$f_{av}$</th>
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<tbody>
<tr>
<td>UFG</td>
<td>0.63</td>
<td>0.113</td>
<td>2.3</td>
<td>62.46</td>
<td>0.019</td>
</tr>
<tr>
<td>UFG + 15 vol.% CG</td>
<td>15.53</td>
<td>0.109</td>
<td>2.18</td>
<td>66.14</td>
<td>0.020</td>
</tr>
<tr>
<td>UFG + 30 vol.% CG</td>
<td>30.4</td>
<td>0.105</td>
<td>2.06</td>
<td>69.81</td>
<td>0.021</td>
</tr>
<tr>
<td>UFG + 45 vol.% CG</td>
<td>45.34</td>
<td>0.101</td>
<td>1.94</td>
<td>73.48</td>
<td>0.023</td>
</tr>
<tr>
<td>CG</td>
<td>100</td>
<td>0.087</td>
<td>1.5</td>
<td>86.96</td>
<td>0.027</td>
</tr>
</tbody>
</table>

The enhanced tensile ductility of the bimodal Al6063 alloys in comparison with the UFG counterpart is connected to the existence of the CGs that tends to deform extensively and contribute to the global uniform elongation. In fact, the CGs experienced significantly more plastic deformation than the UFGs did, and therefore necking occurred in the CG regions (Fig. 9d), representing the local ductile failure in these regions. The UFG matrix should also displace toward the deforming CGs in order to maintain the cohesion of the UFG/C3 boundary, and thus delaminations at UFG/CG interfaces perpendicular to the fracture plane were created. Delaminations are representatives of severe plastic deformation near the fracture surface that improve the ductility of bimodal materials [26]. Moreover, cavitations were formed on the fracture surface of bimodal materials due to the unaccommodated strain between the UFGs and CGs [16] (Fig. 9e).

Detection of necking, delamination, and cavitation on the fracture surface introduces a particular fracture mechanism for the bimodal Al6063 which is schematically shown in Fig. 10a. Under the tensile stress, firstly, the CG regions plastically deform while the UFG matrix elastically carries the applied stress. At a higher stress level, the UFG regions undergo a slight plastic deformation due to the limited dislocation activities in the UFGs. Because the UFG matrix bears most of the tensile stress, a little stress is transferred to the CG regions. Consequently, a slight decrease in the yield stress and increase in the ductility of the bimodal materials compared to the UFG alloy took place. Under increased tensile stress, microvoids are initiated in the UFG matrix and large plastic strains are localized in the CG regions. By further loading, whereas the CG regions straining, the UFG regions are stressed beyond their ultimate tensile strength, resulting in the formation of cracks in the UFG matrix and particularly at UFG/CG interfaces. By crack propagation in the UFG regions and approaching to the CG regions, the CGs blunts the crack tip (Fig. 10b). With acceleration of crack propagation at large applied stresses, crack expands through both the UFG and CG regions and finally, the overall failure occurs by

Table 2
Contribution of various strengthening mechanisms on the yield stress of Al6063.

<table>
<thead>
<tr>
<th>Sample</th>
<th>$\sigma_i$ (MPa)</th>
<th>$\sigma_{HP}$ (MPa)</th>
<th>$\sigma_D$ (MPa)</th>
<th>$\sigma_P$ (MPa)</th>
<th>Calculated $\sigma_y$ (MPa)</th>
<th>Experimental $\sigma_y$ (MPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>UFG</td>
<td>50</td>
<td>88.1</td>
<td>83.4</td>
<td>65.4</td>
<td>286.9</td>
<td>250</td>
</tr>
<tr>
<td>UFG + 15 vol.% CG</td>
<td>50</td>
<td>17.8</td>
<td>12.7</td>
<td>10.4</td>
<td>203.5</td>
<td>183</td>
</tr>
<tr>
<td>UFG + 30 vol.% CG</td>
<td>50</td>
<td>12.7</td>
<td>76.1</td>
<td>64.7</td>
<td>197.2</td>
<td>170</td>
</tr>
<tr>
<td>UFG + 45 vol.% CG</td>
<td>50</td>
<td>7</td>
<td>72.5</td>
<td>64.3</td>
<td>178.9</td>
<td>86</td>
</tr>
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</table>

Fig. 10. Schematic illustration of a bimodal Al6063 under tensile loading: (a) necking, cavitation and delamination; (b) crack initiation and crack tip blunting.
necking in CGs. The mechanism of failure in this study is in a reasonably good agreement with the mechanism proposed by Lavenalia and his colleagues [16,39].

5. Conclusions

Bimodal Al6063 alloys were fabricated via hot consolidation of a mixture of milled nanocrystalline and unmilled microcrystalline powders. The main findings can be summarized as:

1. Bimodal grain structures with equiaxed ultrafine grains (grain size range of 0.08–1.4 μm) and elongated coarse grains (grain size range of 1.8–100 μm) were formed. The frequency of misorientation angles showed a bimodal distribution with a large fraction of high-angle boundaries.

2. A superior tensile elongation (2.8–11.3%) was achieved in the bimodal Al6063 alloys without a remarkable degradation of strength. The enhanced tensile ductility was attributed to the occurrence of crack blunting and bridging as well as delamination during the plastic deformation.

3. The yield stress of the bimodal Al6063 alloy was analyzed through a linear superposition of strengthening from the lattice friction, high-angle boundaries, low-angle boundaries, and second phase particles. The predicted yield stress values were in a reasonably good agreement with the experimental data.

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