Crystal plasticity in fusion zone of a hybrid laser welded Al alloys joint: From nanoscale to macroscale

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HIGHLIGHTS

• Nano/microplasticity in fusion zone from our experiments can be used to predict the macro-plasticity of the joint.
• The nano/microplasticity in fusion zone depends significantly on sample's size.
• Nano/microplasticity is affected by orientation whose influence deceases in larger samples.
• Numerical models are constructed and used to assess the size and orientation effect.

GRAPHICAL ABSTRACT

Abstract

In this paper, we propose a novel approach to predict the plasticity of hybrid laser welded Al alloys joints based on the microplasticity obtained from the micropillar compression test. The micropillar test was performed on the single-crystal pillar with three orientations and various diameters (400 nm to 6.8 μm). It was found that independent of orientation, the yield strength of the pillar increased with the decrease of diameter below a critical length (3.3 μm). A numerical model was successfully built and used to explain the size effect on the pillar’s strength. Crystalline orientation did affect the yield strength, the orientation having higher Schmid’s factor showing lower yield strength, but the effect was reduced with the enlarged diameter. The macroscale yield strength achieved from crystal plasticity finite element simulation showed was found to have a good agreement with that from the experiment. The results here shed new insights both on the application of the micropillar study of alloys, and on prediction of strength in welded Al alloys joint.

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Multiscale study
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1. Introduction

Hybrid laser welding is increasingly used in joining structures of aluminium alloys, since this welding method has advantages of deep welding penetration, stable welding process, and high welding speed. Because of the quick thermal cycles, the microstructure in the welded joint is often heterogenous. More specifically, fusion zone (FZ) is filled with coarse microstructures (precipitates and grain size) and lack of alloying elements (Mg and Zn) [1–3], while heat affected zone (HAZ) is made of coarsened or dissolved precipitates. The direct outcome of such diverse microstructures is the heterogenous spatial distribution of mechanical properties across the welded joint. Despite the fact that the weakest part of the joint is dependent on several factors (such as...
filling materials and base metal), the results from existing literatures [1–10] showed that most of the joints are broken in FZ under tensile load. This indicates that the weakest part of the hybrid laser welded Al alloys joint is in FZ. As suggested by Puydt et al. [11], the weakest part of the welded joint would withstand almost all the strains under external loading while other parts remain elastic. Thus, to some extent, the mechanical behaviour of welded joint rest with that of FZ [11], especially at the early-stage plasticity. Thus, knowledge of the plasticity in FZ is crucial to accurately predict the mechanical properties of the hybrid laser welded joint.

To examine and predict the plasticity in FZ, the micro-tensile tests and integrated modelling have been developed [12]. The micro-tensile test sample, with dimensions actually in mm scale, was taken from FZ along the thickness direction and then subjected to tensile tests [11]. Afterwards, integrated modelling was used to predict the mechanical properties of the welded joint based on the macro stress-strain curves. The integrated model was made of several sub-models [12]: a thermal model, a precipitation model, a yield strength and strain hardening model, and a finite element model of a transverse tensile test coupled with a damage model. However, there are several limitations for integrated modelling to apply to the hybrid laser welded Al alloys joint: (a) It is difficult to accurately model the hybrid laser welding due to the complex interaction of laser-arc and laser-materials etc. [13], causing the difficulty to precisely predict the microstructure formed during welding. As a result, the accuracy of this model is not reliable for the hybrid laser welding process. (b) Micro-tensile testing method is based on the macro stress-strain curves as the input, but the information of how the microstructure reacting to the external loading at the grain/micro scale is unknown. (c) The industrial polycrystalline materials usually have crystalline textures, thus show anisotropic mechanical properties, but such model does not tackle this. Therefore, a new method is needed to accurately and effectively predict the plasticity of FZ in a hybrid laser welded Al alloys joint. Unlike the conventional finite elements [14–16], the crystal plasticity finite element (CPFE) simulation can precisely describe the anisotropic mechanical behaviour of single-crystalline and polycrystalline samples [17]. For instance, Zhang et al. [18] used CPFE to successfully predict the anisotropic yield behaviour of the polycrystalline materials. Pinna et al. [19] employed CPFE model accurately predicting the crystal texture and grain structure in FCC metal under elasto-plastic deformation. These cases prove that CPFE model can give accurate prediction of the plastic deformation of the metal.

The accuracy of the CPFE model is based on the mechanical-deformation data of a single crystal (SC), which determines the material-property parameters used in CPFE simulation. Currently, most of the mechanical experiments for the welded joint have been done using polycrystalline samples, causing difficulty to determine the mechanical properties of SC as the stress in the grains differs from each other [20,21]. One direct way to get the mechanical deformation of SC is to deform a macro SC sample using compressive or tensile test [22,23]. This is mainly applicable for pure metals that can be grown into a macro SC. For alloys, the SC sample may be obtained by Bridgman technique [20]. Nevertheless, the grain size in FZ is only a few tens of micrometres [24] because of the quick solidification process. Such small grains in FZ make it impossible to get mechanical properties of SC utilizing macroscopic testing methods.

Combination of focused ion beam (FIB) milling and nanindentation [25] provides a way to explore the plastic deformation of a SC sample at submicron/micron scales that is made from bulk polycrystal. In the last decades, extensive efforts have been given to understand the mechanical response of materials in submicron/micron region (see Ref. [26]). The mechanical properties of SC, such as the yield stress, elastic modulus, could be obtained via such advanced testing method. It should be noted that the flow strength of SC depends on the sample’s size at micro/submicron scales, known as “smaller is stronger”. Several theories have been proposed to understand this size effect. For nanoscale, the dislocation starvation [27] is applicable, while other models (such as source exhaustion [28], source truncation [29], and weakest link theory [30]) have been used to explain the size effect in the micron region. Although there is a clear size effect in the pure SC of metal, the role of the internal microstructure (such as alloying, precipitation) in the plastic deformation at micro scale is still not well understood. Some researchers reported that the size-dependent strength is still effective for alloys (e.g., FeCrMnNi high entropy alloy [31], austenitic-ferritic stainless steel [32], and aluminium alloys [33,34]), although the size effect is much weaker compared to that of the pure metallic SC. There are also literatures [35,36] arguing that no size effect exists on the strength in the alloying metallic systems. These researches imply that the strength at micro/submicron scale is determined not only by the sample’s size but also by the internal microstructure.

The FZ contains aluminium alloys with solute and precipitation strengthening. So, questions are: how would SC with different orientations in FZ perform under uniaxial compressive loading? Does size effect still work? How could we get the mechanical properties (i.e., yield stress) related to the bulk polycrystal sample from the small-scale experiment? How could we accurately predict the mechanical response of the welded joint using CPFE model based on the information obtained at submicron/micro scales? To answer these questions, we conducted micro-compression on pillars with the diameter from 0.4 μm to 6.8 μm with three orientations, [100], [111] and [−301], in FZ. The deformed microstructure is examined through scanning electron microscope (SEM) and transmission electron microscope (TEM). Theoretical modelling and molecular dynamic simulation are used for the explanation of the size and orientation effect in FZ. Finally, CPFE simulation based on the results of micro-compression test is employed to accurately predict the mechanical properties of the hybrid laser welded Al alloy joint. To our knowledge, this is the first time that the microplasticity in FZ has been studied. And this research sheds new insights both on the application of the micropillar study on alloys, and prediction of strength in welded Al alloys joint.

2. Experimental and simulation methods

2.1. Experimental methods

AA6061 aluminium alloys were joined using hybrid laser-MIG welding system, more details about the welding process can be found in Ref. [24]. A rectangular block was cut from the centre of FZ and mounted by resin. The sample was then metallurgically grinded and polished with the last step that was finished by 0.02 μm colloidal silica to remove the deformed layer on the sample’s surface. The microstructure of the sample was observed using Zeiss Ultraplus Field Emission Scanning Electron Microscope (FESEM). The chemical composition of the matrix was measured using energy dispersive spectroscopy (EDS). The results of the SEM observation can be found in Fig. S1 in the Supplementary materials. TEM experiment was performed at JEOL 2100F microscope at the acceleration voltage of 200 kV to observe the microstructure of the bulk material and the deformed pillars. The TEM results for the bulk material can be found in Fig. S2 in the Supplementary materials. EBSD test was performed utilizing the same parameters as Ref. [24], and the results about the texture, dislocation density, and the grain size are shown in Fig. S3 in the Supplementary materials.

Cylindrical pillars were made through FIB using FEI Helios 600 Nanolab. All pillars were machined at 30 kV with currents ranging from 9 nA to 0.21 nA for coarse milling and 45 pA for fine polishing to minimize the Ga+ damage on the surface of the pillar. The pillars’ aspect ratios of height-to-diameter were kept between 3:1 and 4:1. To study the orientation effect on the strength, orientations of [100], [111] and [−301] were selected. Since the phases could be etched way during FIB milling, all pillars were made in the phase-free area, as shown in Fig. S1a. Thus, only the solid solution strengthening was considered in the pillar.
Compression tests were conducted by a nanoindenter (TriboIndenter TI900, Hysitron) with 9 \(\mu\text{m}\)-diameter flat-end tip. All tests were carried out under displacement-control mode with the strain rate being \(1 \times 10^{-3} \text{ s}^{-1}\). The pillar was located by in-situ imaging method, then the tip was disengaged for \(~40\) min to reduce the thermal drift below 0.05 nm/s. Stress value at the first strain burst was considered.

Fig. 1. (a) Nanopillar for MD simulation, (b) model for single crystal in CPFE simulation, (c) model for polycrystal in CPFE simulation, the colour is shown according to the Euler angle \(\phi_1\).
(For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)

Fig. 2. Representative images for the pillars after compression, (a) 800 nm-diameter pillar with orientation [101], (b) 6.8 \(\mu\text{m}\)-diameter pillar with orientation [101], (c) 800 nm-diameter pillar with orientation [111], (d) 6.8 \(\mu\text{m}\)-diameter pillar with orientation [111].
as the yield stress of the pillar. At least three samples were tested under each condition. SEM images of the pillars were taken before and after the test to obtain the information of dimensions and the deformed morphology.

2.2. Simulation methods

To understand the influence of orientation on the deformation and dislocation behaviour at atomistic level, molecular dynamic (MD) simulation was performed using Large scale Atomic/Molecular Massively parallel Simulator (LAMMPS) [37,38] with the potential developed by Mendelev et al. [39]. The model for the compression in MD simulation is shown in Fig. 1a. The nanopillar had a height of 33 nm and a diameter of 11 nm, with a ratio of height-diameter of 3:1 that is close to the experimental condition. Mg atoms were randomly distributed in the pillar with the concentration of 1.27% (atomistic weight) that is the same as the results from EDS tests (Fig. S1b). Since the pillar was made in the solid-solution area, the interaction between the dislocation and phases was not studied in this research.

Before compression, the dislocation-free nanopillar was firstly relaxed by energy minimization (conjugate gradient method), then equilibrated by MD in the isobaric-isothermal (NPT) ensemble at a pressure of 0 bar and a temperature of 300 K for 100 ps. A constant strain rate of $1 \times 10^9$ s$^{-1}$ was applied along x axis at a temperature of 300 K under NPT conditions with periodic boundary conditions for all three directions. The timestep for the simulation was 1 fs. The results were visualized with the software of Ovito [40]. The common neighbour analysis (CNA) [41] techniques were used to identify the defect structure and its evolution during the simulations. The dislocation extraction algorithm (DXA) [42] was used to analysis the dislocation behaviour during the compression.

The CPFE simulation was performed using phenomenological crystal plasticity theory. The theory for phenomenological constitutive models is introduced here. The deformation gradient $F$ is given as:

$$ F = F_e F_p $$

where $F_e$ represents the crystal lattice rotation and elastic stretching and $F_p$ denotes the plastic part of the deformation gradient. Given that the slip is the only deformation process, the plastic deformation ($\dot{F}$) can be expressed by:

$$ \dot{F} = L_p F_p $$

$L_p$ is given as:

$$ L_p = \sum_{\alpha=1}^n \gamma^\alpha m^\alpha \otimes n^\alpha $$

where $\gamma^\alpha$ is the slip rate on the slip system $\alpha$, $m^\alpha$ and $n^\alpha$ are unit vectors of the slip direction and normal to the slip plane of the slip system $\alpha$.

![Fig. 3. Representative stress-strain curves of (a) [101]-orientated pillars and (b) [111]-orientated pillars; (c) yield strength vs. diameter for the tested pillars with different orientations.](image-url)
respectively; n is the number of slip systems, which is 12 for fcc metal. The slip rate $\dot{\gamma}^\alpha$ can be described as [18]:

$$\dot{\gamma}^\alpha = \frac{\tau^\alpha}{G} \left| \frac{\gamma_0}{g_c^\alpha} \right| \text{sgn} (\tau^\alpha)$$

(4)

where $\dot{\gamma}^\alpha$ is the slip rate of slip system $\alpha$, $\tau^\alpha$ is the applied resolved shear stress on the slip system $\alpha$, $g_c^\alpha$ is the slip resistance of slip system $\alpha$, $\gamma_0$ and $m$ are material parameters. The effect of any set of slip system (index $\beta$) on the hardening behaviour of the fix slip system ($\alpha$) rate can be expressed by:

$$g_c^\alpha = \sum_{\beta=1}^{n} h_{\alpha\beta} |\dot{\gamma}^\beta|$$

(5)

where the hardening matrix $h_{\alpha\beta}$ can be given as:

$$h_{\alpha\beta} = h_0 \left[ q + (1-q)\delta_{\alpha\beta} \right] \left| 1 - \frac{g_{\infty}}{g_\infty} \right| \text{sgn} \left( 1 - \frac{g_{\infty}}{g_\infty} \right)$$

(6)

where $h_0$, $a$, $g_{\infty}$ are slip hardening parameters that are determined from experiment, $q_{\alpha\beta}$ is the lateral hardening parameter that is given as:

$$q_{\alpha\beta} = \begin{cases} 1: \alpha, \beta \text{ coplanar} \\ 1.4: \text{otherwise} \\ \end{cases}$$

(7)

The CPFE theory has been implemented in DAMASK software [43] as a user subroutine of ABAQUS. The micropillar for SC used in CPFE is shown in Fig. 1b, the diameter and height of the pillar is 6.8 μm and 20.4 μm, respectively. The pillar was connected to the substrate with the same orientation. The top surface of the pillar was compressed by an indenter that is a flat rigid surface. The friction coefficient between the indenter and the top surface of the pillar was set to 0.14 to prevent the slip between the indenter and the pillar’s top [44]. The compression rate was $1 \times 10^{-3}$ s$^{-1}$ that matched the experimental condition. Based on the EBSD results (such as the grain size and Euler angle), the polycrystalline model was built using the software package Neper [45]. As shown in Fig. 1c, the built model with dimensions of $1.75 \times 0.4375 \times 0.07$ mm$^3$ is made of 157 grains with the grain size of 71 μm.

3. Results

3.1. Micropillar compression

As shown in the SEM images (Fig. 2), double slips could be observed for the [101]-orientated pillars with size of diameters from 400 nm to 6.8 μm. It can also be seen from Fig. 2a–b that the slip in orientation [101] was not affected by the sample diameter in this study, which is different from the previous reports [46,47]. The difference may be due to the microstructure factor (e.g., the alloying composition, orientation and dislocation density). Opposite to the [101]-orientated sample, the morphology of the compressed [111]-orientated sample is influenced by the sample’s diameter. For the 400 nm-diameter sample (Fig. 2c), only localized wavy slip line could be seen on the surface. But for larger samples (Fig. 2d), the single slip band could be seen on the surface. Fig. 3a–b shows representative stress-strain curves from the uniaxial compression of the pillars with two orientations in FZ. The discrete stochastic bursts can be seen after reaching the yielding point in the stress-strain curves, which is much like that of the SC of pure FCC.
metal. As illustrated in Fig. 3c, the strength is higher as the diameter of the pillar (≤ 3.3 μm) decreases for both orientations, existing “smaller is stronger” trend. The results also illustrate that the orientation can influence the strength of FZ at micro/submicron scales. For samples with size of the diameter bigger than 3.3 μm, the size effect on the yield strength is not effective for both orientations, indicating that the critical diameter for the size effect in our study is 3.3 μm. Another interesting thing is that the orientation effect on the yield strength is reduced along with an increase in the sample’s diameter (see Fig. 3c).

3.2. Dislocation behaviour

Fig. 4 shows SEM and TEM images for the deformed [101]-orientated 800 nm-diameter sample. As illustrated in the SEM micrograph in Fig. 4a, the TEM sample was taken along the cross section of the pillar. Fig. 4b shows that most of the deformation happens on the top of the pillar and strained parts are parallel to each other along the slip direction. From the enlarged view (Fig. 4d), most of the dislocations distribute in the centre with the place near the sample’s surface being dislocation-free. But the inverse fast Fourier transform image (Fig. 4f) shows that there are lattice distortions, meaning that the dislocation was there before deformation and escaped from the surface during deformation. The dislocation in Fig. 4g could be seen on the slip plane and close to the surface. This may be due to two reasons: (a) the cross slip locked the dislocation (Fig. 4c) and (b) dislocation annihilating during the compression. As reported previously [48,49], the FIB milling could damage the surface of the sample, which leads to amorphous layer or dislocation strengthening on the surface. As shown in Fig. 4c, the damaged layer is only ~9.6 nm with fine current (48 pA) during the FIB milling. Such small damaged layer should not influence on the mechanical properties of the pillar.

Fig. 5 is SEM and dark-field TEM images for the [111]-orientated 800 nm-diameter sample. The bright areas in Fig. 5b are the main deformed parts of the pillar. And the enlarged view of the area (Fig. 5c–d), shows the dislocation lines in the deformed sample. Comparing the TEM results for both orientations, it seems that the [111]-orientated sample has higher dislocation density than the [101]-orientated one. We have used the method described in Ref. [50] to measure the dislocation density in the deformed pillar. The calculated dislocation density for [101]- and [111]-orientated 800 nm-diameter sample is $3.2 \times 10^{14}$ and $5.32 \times 10^{14} \text{ m}^{-2}$, respectively. The lower dislocation density in the [101]-orientated pillar is possibly due to more slip events during the compression. More details will be discussed in Section 3.3.

3.3. Molecular dynamic simulation

The stress-strain curves for the orientation [111] and [101] pillars are shown in Fig. 6a. Similar to the experimental results, the [111]-orientated nanopillar shows higher yield strength than that of the [101] one. However, after yielding, the flow stress of the [101] nanopillar fluctuates in a wide range (0.12 GPa to 2 GPa), while the flow stress of the [111] nanopillar varies much milder. The fluctuation of the flow stress after yielding is due to the dislocation behaviour [51]. The dislocation density for both orientations during compression is shown in Fig. 6b. In the plastic strain, the dislocation density in [111]-orientated pillar is higher than that of the [101]-orientated one, which is consistent with the experimental results (Fig. 4 and Fig. 5).

The detailed microstructure of the pillar under compression is studied from snapshots. Fig. 6c presents the snapshots for the [101]-orientated sample. From the snapshots (Fig. 6c), most dislocations move along the stacking fault and then annihilate from the surface. With the forming of cross slip (Fig. 6c), some dislocations are locked
in the crossed area while no dislocations exist on the slip plane, indicating that dislocations escape from the surface during the slip. For the [111]-orientated sample (Fig. 6d), the stacking faults are distributed on the {111} planes bounded with perfect and partial dislocations. As shown in Fig. 6d, dislocations entangle with each other, causing them difficult to escape from the surface. As a result, the dislocation density is getting higher. Further, since there are too many slip directions and planes in the [111] sample (Fig. 6d), the annihilation of the dislocation is multi-directional, resulting in wavy surface.

3.4. Simulation based on crystal plasticity theory

Material parameters used in the CPFE model are determined by fitting the stress-strain curve from simulation to the one from micropillar experiments. As shown in Fig. 3c, the strength of the pillar becomes larger when the sample’s diameter decreases. To accurately predict the strength of the welded joint at the macroscale, the choose of the sample with a right diameter is crucial. As shown in Fig. 3c, the size effect on the strength of the pillar disappears when the sample’s diameter is larger than 3.3 μm. For safety, we use the experimental data of the [101]-orientated sample with diameter of ~6.8 μm to get the material parameters for CPFE simulation. The optimized material parameters are listed in Table 1.

The stress-strain curves from the CPFE simulation and experiment for three crystalline orientations, [101], [111], and [−301], are shown in Fig. 7. For three orientations, the elastic strain period from simulation and experiment does not match perfectly. This discrepancy is due to the misalignment between the pillar’s top and the flat-end tip [52,53]. However, the yield strength from simulation and experiment is very close, as demonstrated in Table 2. It is worth to note that the yield strength from the simulation is the strength at the strain of 0.2%, while the yield strength from the experiment is the strength from the first strain burst. The stress distribution of the Von Mises stress is shown in Fig. 8. For comparation, the SEM images of the deformed pillar are also shown. It can be seen from Fig. 8 that the slip can be accurately predicted by the simulation. The [101]-orientated sample shows double slips, while the other two orientations show single slip.

After obtaining and calibrating material parameters used in CPFE from SC, we then apply the same parameters to predict the tensile test behaviour of the polycrystalline bulk material. The predicted stress-strain curves and the experimental ones are shown in Fig. 7d. The well agreement between them is achieved, especially for the yield strength. The Von Mises stress distribution at 5% strain is shown in Fig. 8d. The stress is not evenly distributed in the sample with some grains showing higher stress.
4. Discussion

4.1. Orientation effect

As shown in Fig. 3c, the yield stress of the micro/submicron pillar in FZ is orientation-dependent, with the [111] orientation having highest yielding stress. During the FIB milling, the pillar was intended to be made in solid-solution area, see Fig. S1a, to not include phases. Fig. 4b shows that the pillar does not contain any phases. Thus, it is reasonable to consider the pillar made from FZ as a kind of FCC solid-solution material, of which the slip system(s) should be the one(s) having the highest Schmid factor among the \( \{111\} \langle 110 \rangle \) slip system. From EBSD test, the Schmid's factor of orientation \([101]\), \([111]\) and \([−301]\) is 0.44, 0.32, and 0.5, respectively. Thus, the [111] pillar shows highest stress when subjected to uniaxial loading.

Our finding about the orientation effect on the strength at micro/submicron scale is consistent with the research of Hagen et al. [54]. They reported that the strength of \( \alpha \)-Fe with orientation \([011]\) is higher than that of \([010]\). But it is different from Frick's report [55]. They reported that the SC Ni in orientation \([111]\) with less slip systems has lower critical resolved shear stress (CRSS) than that of in orientation of \([269]\) with more slip systems. They attributed this interesting phenomenon to that there may have more available slip planes for the \([111]\) Ni at micro/submicron scales, which is contrary with Schmid's law. It is also reported that the orientation has no effect on the CRSS at micro/submicron scales, like Nb [56]. These studies demonstrate that the Schmid's law could be broken down at the nanoscales [57]. At micro/submicron scales, the yielding strength is determined by the activated dislocation source. In other words, the slip system with lower Schmid factors could be involved in the deformation.

Another interesting finding is that the orientation effect on the yield stress decreases with the increase of pillar's diameter. As shown in Fig. 3c, the yield stress due to the orientation effect could be up to \(-115\) MPa (0.4 \( \mu \)m-diameter sample) or down to \(-25\) MPa (samples with diameter larger than 3.3 \( \mu \)m). The change may be attributed to the size effect. As shown in Fig. 9a–b, the contribution from dislocation-source strengthening decreases with the increase of sample's diameter. The dislocation-source strengthening is only about 5 MPa when diameter is larger than 3.3 \( \mu \)m but could be as high as 110 MPa with diameter being 0.4 \( \mu \)m.
4.2. Flow strength

Although there are phases and solid-solution areas in FZ, the pillar is made in the solid-solution area without phases. Thus, the phases strengthening is not considered in this flow strength. The flow shear stress of the sample in this study can be expressed as:

$$\tau_{\text{total}} = \tau_{\text{fs}} + \tau_{\text{ss}} + \tau_{\text{d}} + \tau_{\text{source}}$$ (8)

where $\tau_{\text{total}}$ is the total shear stress, $\tau_{\text{fs}}$ is the friction stress that is taken as 0 [57], $\tau_{\text{ss}}$ is the solid solution strengthening, $\tau_{\text{d}}$ is the strengthening from dislocation-dislocation interactions, $\tau_{\text{source}}$ is the dislocation source-controlled strength that is related to the size effect. The solid solution strengthening can be given as [58]:

$$\tau_{\text{ss}} = \sum_{j} K_j C_j^2$$ (9)

where $C_j$ and $K_j$ is the concentration and the scaling factor of the $j$th alloying element, respectively. According to the EDS results, only Mg can be considered as the solute element for the joint. Taking $K = 29 \text{ MPa wt}^{-2/3}$ [58], $C_{\text{Mg}} = 1.21\%$, the resulting solid solution strengthening ($\tau_{\text{ss}}$) is 33 MPa.

The strengthening contribution from dislocation interaction ($\tau_{\text{d}}$) can be quantitatively described as [59]:

$$\tau_{\text{d}} = \alpha |b| \mu \sqrt{\rho_{\text{total}}}$$ (10)

where $\alpha$ is a constant, $b$ is the magnitude of Burgers vector, $\mu$ is the shear modulus, and $\rho_{\text{total}}$ is the total dislocation density. Since the mobile dislocation in the small-scale sample prefers to escape from the sample's surface, the increment of dislocation density during compression is taken to be 0 in this study. Taking $\alpha = 0.5$ [29], $b = 0.3 \text{ nm}$ [57], $\mu = 26 \text{ GPa}$ [57], $\rho_{\text{total}} = 9.19 \times 10^{12} \text{ m}^{-2}$, the obtained increment from the dislocation strength is 11 MPa.

As suggested by Parthasarathy et al. [29], the doubled-end sources in the micron pillar would turn to single-ended source due to the interaction with the free surface. The stress to activate the single-ended source dislocation ($\tau_{\text{source}}$) can be given as:

$$\tau_{\text{source}} = \frac{K \mu b}{X}$$ (11)

where $X$ is equal to 1 [29], $\mu$ is the shear modulus, $b$ is the magnitude of the Burgers vector, $X$ is the effective source length, which can be obtained by:

$$X = \int_{0}^{R} \left(1 - \frac{n(R-L)(a-L)}{nka}\right)^{n-1} \times \left\{\frac{n[(R-L) + (a-L)]}{nka}\right\} n\lambda d\lambda$$ (12)

where $R$ is the radius of the sample, $a$ is given as $a = \frac{\theta}{\cos \beta}$ ($\beta$ is the angle between the primary slip plane and the applied stress axis), for our case $\beta = 35.3^\circ$ and $90^\circ$ for [101] and [111] orientated sample respectively; $n$ is pining number determined by the sample dimension and initial dislocation density, as obtained by:

$$n = \text{Integer} \left[\frac{nD^2 \rho_{\text{total}}}{8}\right]$$ (13)

where $\rho_{\text{total}}$ is the total dislocation density, $D$ is the diameter of the pillar.

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**Table 2**

Yield strength obtained from the simulation and the experiment, the unit is MPa.

<table>
<thead>
<tr>
<th></th>
<th>[101]</th>
<th>[111]</th>
<th>[−301]</th>
<th>Bulk polycrystal</th>
</tr>
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<tbody>
<tr>
<td>From simulation</td>
<td>132</td>
<td>182</td>
<td>128</td>
<td>173</td>
</tr>
<tr>
<td>From experiment</td>
<td>135</td>
<td>170</td>
<td>125</td>
<td>170</td>
</tr>
</tbody>
</table>

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**Fig. 8.** Von Mises stress of orientation (a) [101], (b) [111], (c) [−301], (d) polycrystal; SEM images for the deformed pillar of orientation (e) [101], (f) [111], and (g) [−301].
Using Eqs. (8)–(13), the total flow stress can be expressed as:

$$\tau_{\text{total}} = 44 + \frac{\mu b}{\pi R_{0}^{1-n}} \int_{0}^{R} \left(1 - \frac{\pi(R - \lambda) R_{2}^{n-1}}{\pi a_{0}^{n-1}} \right) n d_{\lambda}$$  \hspace{1cm} (14)

The calculated $\tau_{\text{total}}$ is shown in Fig. 9a–b together with the experimental results. The experimental results are about 20–30 MPa higher than the calculated ones, especially for the 0.8 μm-diameter and 2 μm-diameter samples. This is consistent with the previous report on the small-scale research of alloys [46]. The discrepancy between the calculation and the experiment may be due to the underestimation of the bulk strength:

1. The underestimation of friction stress, because the solute atom Mg can significantly cause the lattice distortion due to misfit volume and elastic mismatch between Al and Mg atoms. (2) Some other solute elements may not be detected using EDS, such as Fe (existing both in the base metal and filling material). It is reported [60] that element Fe has an important influence on the solute strengthening, even if its concentration is less than 3 ppm. But such low concentration is beyond the accuracy of the EDS used in our study.

4.3. Strength from nanopillar to bulk

One of the key interests in our study is to extract macroscopic mechanical properties from the small-scale test, such as the yield stress, which can be used to the design and assessment of the Al alloys structure joined by hybrid laser welding. As can be seen from Fig. 3c, all the samples from both tested orientations exhibit the trend of "smaller is stronger". As suggested by Dou and Derby [61], the relationship between the sample’s diameter and strength can be express by:

$$\frac{\tau}{\mu} = A \left(\frac{D}{b}\right)^{m}$$  \hspace{1cm} (15)

where $\tau$ is the critical shear stress, $\mu$ is the shear modulus, $D$ is the sample’s diameter, and $|b|$ is the magnitude of the Burgers vector, and $m$ is the size exponent.

Here we apply the power law, Eq. (15), to our experimental results. As shown in Fig. 9c, the value of $A$ for [101] and [111] orientated sample is 0.15 and 0.14, respectively. And the size exponent $m$ for [101] and [111] orientated sample is $-0.46$ and $-0.45$, respectively. The value of $m$ for both tested orientations is very close, implying that the size effect for the present study is not affected by the orientation. The value of $m$ in our study is smaller than that of pure SC Al, which is $-1$ [57] and $-0.63$ [62]. The smaller value is due to that the power law has only considered the external contribution from sample’s size, not considered the contribution from the internal changes [63] (i.e., solute strengthening). Moreover, the bulk yield strength obtained from this power law is too small.

Fig. 9. Calculated critical shear stress according to Eq. (14) for (a) orientation [101] and (b) orientation [111]. The red points are the calculated CRSS, while the black points are the experimental ones. The contribution to the strength from $\tau_{\text{source}}, \tau_{\text{ss}},$ and $\tau_{\text{d}}$ is illustrated in green, magenta and blue dash line, respectively. (c) The normalized strength vs. normalized diameter for the tested pillars in this study, (d) yield strength vs. diameter for orientation [101] and [111]. (For interpretation of the references to colour in this figure legend, the reader is referred to the web version of this article.)
Here we use a simple power law to evaluate the bulk strength [64]:

\[ \sigma_{\text{size}} = \sigma_{\text{bulk}} + A \cdot D^{-\mu} \]  

(16)

where \( \sigma_{\text{size}} \) is the yield strength of the micro sample, \( \sigma_{\text{bulk}} \) is the yield strength for the macro sample, \( A \) is a constant. The fitted curves are shown in Fig. 9d, both curves are presented with a high confidence interval, 99% for both orientations. The obtained bulk yield strength for [101] and [111] is 147 ± 3 MPa and 177 ± 8 MPa, respectively. These values are slightly higher than that obtained from the experiment (see the yield strength of 6.8 \( \mu \)-diameter pillar in Fig. 3c), which is 133 ± 3 MPa and 164 ± 7 MPa for [101] and [111], respectively. The higher value extracted from the micron test is also reported for the CP-Ti [65] and pure Zr [94]. The discrepancy can possibly be minimized by compressing the pillar with larger range diameter. But there is still a fair good agreement between the theoretical and experimental results. Moreover, the size exponent \( \mu \) obtained from this equation is -0.9, which is higher than the one calculated from the Dou- Derby power law (Eq. (15)), but close to that from the pure Al [57].

The concern/dispute for the micron test is the FIB machining, as Ga ion beam could induce damage on the surface of the sample in the form of defects [66], such as dislocation loops and amorphous layer. Some researchers [48,49,67,68] argued that the flow strength could be enhanced by such damaged layer due to the dislocation pile up in these damaged areas. But this effect is significant only when the damaged depth reaches 100 nm or more [48]. Using fine FIB milling current could reduce this damaged layer (only a few tens nanometres), and the mechanical response will not be influenced, as demonstrated in Ref. [69]. For our study, as shown in Fig. 4c, the Ga\(^+\) damaged layer is only 9.6 nm, which is considerably smaller than the 6.8 \( \mu \)-diameter pillar used to evaluate the bulk's yield strength. Therefore, it is reasonable to assume that the influence of the Ga\(^+\) damage can be neglected for the flow stress of the pillar in this study.

5. Conclusion

In this paper, we have performed micro-compression test on the pillars of fusion zone in a hybrid laser welded AA6061-T6 joint, with the size of diameters ranging from 400 nm to 6.7 \( \mu \)m. Based on the data obtained, following summaries are made:

1. Crystalline orientation has an influence on the microscale yield strength of the single crystal in fusion zone, which can be explained with Schmid’s law; the orientation-dependent strength decreases with the increase of pillar’s diameter, which could be explained by the dislocation source strengthening.

2. Trend of “smaller is stronger” is observed for both orientations with the exponent \( m \) being –0.45 when the pillar’s diameter is below 3.3 \( \mu \), the exponent \( m \) would be –0.9 if considering the bulk’s strength; the exponent is not influenced by the crystal orientation.

3. The observed critical size of pillar’s diameter is 3.3 \( \mu \), the size-dependent strength disappears beyond this value, and the yield strength could be used as the bulk’s strength if pillar’s diameter is larger than this value.

4. The bulk yield strength could be obtained by fitting the law of \( \sigma_{\text{size}} = \sigma_{\text{bulk}} + A \cdot D^{-\mu} \), with \( n \) being –0.9.

5. Base on the microplasticity of pillars with the size of diameter of 6.8 \( \mu \), CPF simulation could accurately predict the onset plasticity of the hybrid laser welded AA6061-T6 joint, with the predicted yield strength (173 MPa) very close to that of the experimental one (170 MPa).

Author contributions

Shaohua did all experiments and wrote the paper. Haiyang did the numerical modelling. Bobin, Li, and Shuang joined the discussion. Qinghua designed the whole work and joined the writing and analysis.

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Appendix A. Supplementary data

Supplementary data to this article can be found online at https://doi.org/10.1016/j.matdes.2018.09.031.

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