Ultrafine-Grained Microstructure and Improved Mechanical Behaviors of Friction Stir Welded Cu and Cu–30Zn Joints

Yan-Fei Wang¹ · Jian An¹ · Kun Yin¹ · Ming-Sai Wang¹ · Yu-Sheng Li² · Chong-Xiang Huang¹

Received: 24 September 2017 / Revised: 15 January 2018 / Published online: 9 April 2018
© The Chinese Society for Metals and Springer-Verlag GmbH Germany, part of Springer Nature 2018

Abstract

Ultra-strong joints of pure Cu and Cu–30Zn alloy were obtained by friction stir welding under flowing water. The effects of heat inputting condition and material characteristics on the morphologies, microstructures and mechanical properties of welding joints were studied. Defect-free stirring zones of pure Cu and Cu–30Zn were characterized by onion-ringed structure and plastic flowing bands, respectively. Both low stacking fault energy and fast cooling condition contributed to the formation of small recrystallized grains less than 1 μm in stirring zones. The welding joints in both materials exhibited enhanced mechanical performances due to ultrafine-grained microstructure in stirring zones and disappearance of soft heat-affected-zone. The technique of digital image correlation was used to study the tensile deformation behaviors of welding joints and verify the improved tensile properties.

Keywords Friction stir welding · Ultrafine grain · Mechanical properties · Stacking fault energy · Digital image correlation (DIC)

1 Introduction

Due to the high thermal/electrical conductivity, good formability, excellent resistance to corrosion and acceptable combination of strength and ductility, pure copper and copper–zinc alloys are used widely in industries [1, 2]. Whether used as structural component or functional material, there is an inevitable demand for the welding of these materials. However, the high thermal diffusivity implies high heat input to compensate rapid heat dissipation during welding. This is a stumbling block for conventional fusion welding technologies to get high strength joints, because extremely high heat input would cause

serious distortion, solidification cracking, high oxidation rate and residue stress in welding joints [3].

Fortunately, friction stir welding (FSW), a novel solid-state joining technology invented in 1991 [4], can weld various metals and alloys at a temperature far below their melting point to overcome these problems in fusion welding process [5, 6]. During FSW, a rotating tool was inserted into the workpiece and moved forward along the welding line, making the processed zone, i.e., the stirring zone (SZ), experience extensively high rate thermoplastic deformation. To date, the FSW technique has been successfully applied in Cu [2, 7, 8], Al [9–11], Mg [12–14], Ni [15], Fe and their alloys [5], etc.

The amount of heat input influences the quality and strength of FSW joints significantly. Generally, high heat input is necessary to overcome the rapid heat dissipation and reduce welding defects. On the other hand, long time and high temperature annealing process induced by high heat input will lead to rapid recrystallization and grain coarsening in the SZ, which deteriorates the strength of joints. Although numerous works on optimizing the microstructure and mechanical properties of metals and alloys have been tried, it is still a challenge to process high strength and defect-free joints in Cu and its alloys [3, 7, 8]. For instance, Barenji [7] got hard SZ with refined grains in

Available online at <http://link.springer.com/journal/40195>

✉ Yu-Sheng Li
liyusheng@njjust.edu.cn

✉ Chong-Xiang Huang
chxhuang@scu.edu.cn

¹ School of Aeronautics and Astronautics, Sichuan University, Chengdu 610065, China

² Nano Structural Materials Center, School of Materials Science and Engineering, Nanjing University of Science and Technology, Nanjing 210094, China

pure Cu by adjusting the transverse and rotational speed of tool. However, a soft heat-affected-zone (HAZ) near the thermo-mechanically affected zone (TMAZ) was also produced. Sun and Fujii [16] revealed that high applied load can refine microstructure effectively and obtained an average grain size of 3.8 μm in pure Cu by increasing the applied load to 1500 kg. Farrokhi et al. [3] obtained a small grain size of 7 μm in pure Cu and pointed out that the microstructure achieved under water cooling condition was finer than that in air. Successfully, Xue et al. [17] achieved an ultrafine-grained (UFG) Cu with the finest grain size of 400 nm by adopting flowing water. Heidarzadeh et al. [18] developed a relationship among the FSW parameters, grain size and hardness of joints by response surface methodology. They revealed that stacking fault energy (SFE) could significantly affect the refining mechanism during thermal-plastic stirring process. Compared with Cu, much finer grains ($\sim 1.6 \mu\text{m}$) were produced in Cu–Zn alloys with low SFEs [18]. Sun et al. [8] investigated the microstructure and mechanical properties of FSW Cu–30Zn in a wide range of processing parameters, and the resulted grain size in the SZ was ranged from 6.4 to 9.6 μm . In fact, the UFG microstructure has been obtained in Cu–Al alloys with low SFEs by FSW recently [19].

According to above studies, It is found that the material characteristics and thermoplastic history that decide the microstructure and mechanical properties of FSW joints. Low peak temperature and fast cooling rate can produce finer grains in the SZ [3, 20–22]. However, insufficient thermal cycle and plastic flow will result in the formation of porosity, tunnel cracking and other inner or surface defects [8, 23]. The current work is aimed to achieve ultra-strong welding joints with ultrafine-grained SZs in pure Cu and Cu–Zn alloy by FSW process which has proper amount of heat input, and to clarify the effects of thermoplastic history and material characteristics on the microstructures and mechanical behaviors of joints. In addition, the deformation and fracture behaviors of joints are characterized using the technique of digital image correlation (DIC).

2 Experimental

Commercially pure Cu and Cu–30Zn plates with dimension of 300 mm $\times$ 90 mm $\times$ 2 mm annealed at 600 $^{\circ}\text{C}$ for 2 h were used as base metals (BM). The FSW tool fabricated by H13 steel has a shoulder with diameter of 10 mm and a threaded cylindrical pin with diameter of 3 mm and length of 1.3 mm. In order to increase cooling rate and suppress post-annealing effect after stirring, FSW was conducted under cold flowing water. The traverse speeds of pin were 100 mm/min for pure Cu and 200 mm/min for

Cu–30Zn, respectively. A constant rotational speed of 1200 rpm was applied.

The microstructures of joints were characterized using optical microscopy, scanning electron microscopy (SEM) equipped with electron backscatter diffraction (EBSD) and transmission electron microscopy (TEM). The samples for SEM observation were cut in the direction perpendicular to the welding direction, polished and then etched with a solution of 8 g FeCl_3 and 50 ml H_2O . The samples for EBSD were electrochemically polished in a solution consisting of phosphoric acid, ethanol and deionized water with a volume ratio of 1:1:2. The samples for TEM were prepared by mechanically polishing to a thickness of $\sim 50 \mu\text{m}$ and then electronically polished at -10°C by twin-jet.

The Vickers microhardness measurements were carried out along a line beneath the top friction surface (500 μm far away) on the cross section of joints, using a load of 10 g for 15 s. The arrangement of indentations in TMAZ was denser than that elsewhere, in order to measure the hardness of transition area accurately. The minimum distance between two neighboring indentations was 100 μm , which is five times of the indentation diameter. Dog-bone-shaped tension specimens with a gauge length of 32 mm and width of 8 mm were machined perpendicular to the welding direction (transverse samples) by spark cutting, and thereafter polished away the redundant thickness layer mechanically from the bottom side of SZ to a final thickness of 1 mm. To evaluate the mechanical heterogeneity of joints during tensile deformation, the strain evolution of welding joints was recorded by digital image correlation method. In addition, miniature tensile specimens with the gauge dimension of 16 mm $\times$ 3 mm $\times$ 0.8 mm were cut from the SZ and used to evaluate the mechanical properties of the SZ. All tensile tests were conducted at a nominal strain rate of $5 \times 10^{-4} \text{ s}^{-1}$.

3 Results and Discussion

3.1 Macroscopic Morphologies

Figure 1a, c shows the optical morphologies of FSW joints. No groove-like surface defects, distortion or superfluous flash described by Sun et al. [8] are found in the current pure Cu and Cu–30Zn. The high speed stirring tool induced perfect veins on the friction surface, as shown in Fig. 1b, d. Such good surface primarily indicates that the heat input is controlled well.

The cross-sectional morphologies of stirring joints are shown in Fig. 2a–f. No inner defects, such as pores and cracks, are detected in the present samples. Besides, several important features can be summarized as follows. First,

Fig. 1 Optical images of FSW welds for pure Cu **a, b** and Cu-30Zn **c, d** for overall appearances **a, c** and stirring veins in SZs **b, d**. The coordinate system inserted in **c** denotes the principal directions of welding geometry, welding direction (WD), transverse direction (TD) and normal direction (ND)

Fig. 2 Cross-sectional SEM morphologies of FSW joints for pure Cu **a–c** and Cu-30Zn **d–f** in overall cross sections **a, d**, transition areas between the SZ and the BM **b, e**, and enlarged images for SZs **c, f**. AS and RS denote the advancing side and the retreating side of stirring geometry

both pure Cu and Cu-30Zn joints exhibit basin-shaped SZ, as shown in Fig. 2a, d, indicating that the shape of the SZs depended on material, tool geometry and processing parameters, such as rotation speed and cooling condition, etc. [16, 23]. More intensive stirring by the tool probe would cause a wider thermoplastic flowing area and then contribute to the formation of downward convex outline of the SZ, as appeared in Cu-30Zn in Fig. 2d, while for the pure Cu samples welded with relative high heat input (low traverse speed of pin), the high softening degree in deep SZ limited the scale of stirring affected zone, as shown in

Fig. 2a. It is found from the experiments conducted with different applied loads that the high pressure friction between tool shoulder and workpiece may play an important role in shaping the upward convex outline of the SZs [16].

Second, three sub-zones of the welding joint named as SZ, TMAZ and BM can be distinguished clearly by their different metallurgical features. The sharp contrast between the SZ and the BM shown in Fig. 2a, d suggests great difference between their microstructures, which will be discussed in the following content. Figure 2b, e presents

the transition area between the SZ and the BM, i.e., the TMAZ. It consists of a mixed microstructure of shear-deformed coarse grains (CG) and recrystallized fine grains. Generally, the TMAZ in the retreating side (RS) is wider than that in the advancing side (AS), as shown in Fig. 2a, d. Similar to the previous experiments conducted with additional cooling medium or optimized processing parameters [20, 24], the soft HAZs were not formed in the current samples, which could be attributed to the high thermal dissipation rate under flowing water. However, if the welding parameters are not controlled well, the HAZ can be easily produced in pure Cu [7, 16].

Third, obvious onion-ringed structure is observed within the SZ of pure Cu. As the enlarged image shown in Fig. 2c, the spacing between rings is wide inside and becomes narrow toward the outside. Different from previous investigations [9, 25, 26], the onion rings in the current Cu are fragmentized, which would be affected by the tool geometry and intense friction between workpiece and tool shoulder. So far, the formation mechanism of onion rings in FSW joint is not completely understood. Based on the clay model, Krishnan [9] explained that, due to the geometric effect, the cylindrical sheet of material was extruded by tool during each rotation and the ring spacing was equal to the forward motion of tool in one rotation, which is attributed to the circular flowing of material induced by threaded pin [27]. However, both our current work and previous study [26] show that the onion rings can be produced by nonthreaded tool pin. Xie et al. [25] found that onion rings disappeared when the input heat was increased. Different from the onion rings in Cu, the SZ morphology of Cu–30Zn is characterized by banded structure, as shown in Fig. 2f. These plastic flowing bands are almost parallel to the outline of SZ, and the band spacing changes slightly regardless of inside and outside, which implies a different thermoplastic flowing in Cu–30Zn from that in Cu.

3.2 Microstructure Characterization

Figure 3a–d shows the typical EBSD images for the BMs and SZs. The BMs are composed of CG with abundant annealing twins, which are the typical characteristics for annealed FCC metals with medium/low SFE. However, the microstructure in the SZs is dominated by equiaxed ultrafine grains (UFGs), with average grain sizes of 0.98 and 0.54 μm for Cu and Cu–30Zn, respectively, as statistically analyzed in Fig. 3e. There is a weak [001] texture in Cu and [101] texture in Cu–30Zn, as verified by the inverse pole figures in Fig. 3b, d. Such equiaxed UFGs and weak texture indicate a well-developed dynamic recrystallization (DRX) history in the current samples [16, 28].

The distribution of misorientation angles in the SZs is shown in Fig. 3f. Within the measured angle ($> 2^\circ$), the

fraction of high angle grain boundaries (HAGB, $> 15^\circ$) is accounted to $\sim 56\%$ in Cu. The highest peak of Cu locates at $\sim 2^\circ$, implying lots of substructures among the fine grains. As verified by TEM observation in Fig. 4a, many sub-grains and cell structures can be clearly identified in addition to the equiaxed UFGs. Similar results were also reported previously that high fraction of LAGBs or dislocation cells appeared in Cu processed by low-heat-input FSW [21, 28]. For the Cu–30Zn sample, the fraction of HAGBs ($\sim 86\%$) and $\Sigma 3$ twin boundaries ($\sim 18\%$) is far higher than that of pure Cu, though its average grain size in the SZ is much smaller. TEM examination of Cu–30Zn in the SZ is shown in Fig. 4b. The microstructure is mainly characterized by equiaxed UFGs with sharp, clear and straight grain boundaries. Besides, some annealing twins can be observed.

The refined grain size is one of the desired microstructures, because small grains can significantly improve the strength of the SZ. There are several mechanisms for grain refinement during friction stir processing. First, high speed stirring is a typical process of severe plastic deformation (SPD), which can effectively refine parent CG by dislocation subdivision and/or twin fragmentation [14, 29]. Second, thermoplastic process induces DRX [30, 31]. It has been reported that both continuous and discontinuous DRX occurred in Cu and Cu–30Zn during FSW, and the former plays a dominant role in those metals with medium/low SFEs [28, 32]. Third, dynamic recovery was also considered as the grain refinement mechanism during FSW of Al alloys [10]. In this study, the flowing water speeds up the cooling rate of joints and suppresses the post-annealing cycles of material behind rotating tool. Therefore, it is not surprise that UFG microstructure was formed in the SZs of both Cu and Cu–30Zn samples if the appropriate processing parameters are used. Besides, it is believed that the occurring mechanisms and their relative contribution to final microstructure depend heavily on material properties, such as SFE [10, 28, 32].

Empirically, the dynamically recrystallized grain size (d) can be related to the Zener–Hollomon parameter (Z) through the following expression [12, 13]:

$$\ln d = A - B \ln Z, \quad (1)$$

where A and B are positive constant. The parameter Z can be used to evaluate deformation conditions, such as temperature (T) and strain rate ($\dot{\epsilon}$), according to Ref. [12]:

$$Z = \dot{\epsilon} \exp(Q/(RT)), \quad (2)$$

where Q is the activation energy and R is the gas constant. It is clear from above equations that a higher Z value contributes to smaller grain size after recrystallization, and Z can be increased by increasing $\dot{\epsilon}$ and decreasing T . Therefore, the small grain size in the SZs can be reasonably

Fig. 3 EBSD micrographs showing microstructures of FSW joints: **a** BM of Cu; **b** SZ of Cu; **c** BM of Cu-30Zn; **d** SZ of Cu-30Zn; **e** grain size (d) distributions in SZs; **f** distributions of grain boundary misorientations in SZs (d^* represents the average grain size. The insets in **b**, **d** show the inverse pole figures of the SZs)

attributed to the high $\dot{\epsilon}$ induced by high rotational speed and the low processing temperature decreased by flowing water. In addition, the higher transverse speed leads to a lower processing temperature as well, which is one of the reasons for the smaller grains achieved in Cu-30Zn sample than that in Cu.

It should be noted that the processing parameters used in Cu-30Zn have also been applied in pure Cu in this study. Although the average grain size was further refined to

$\sim 0.71 \mu\text{m}$, many micro-holes were produced along the boundary between the SZ and the TMAZ in the AS. This phenomenon suggests that the intrinsic material characteristic is another dominant factor to shape the plastic flowing and final microstructure in the SZ. As revealed by Morishige et al. [14], increasing the Mg content in Al-Mg alloys led to a decrease in grain size in the SZ because of a reduced SFE. According to the model proposed by Mohamed [33], the effect of SFE (γ) on the steady state

Fig. 4 TEM images in SZs of **a** pure Cu, **b** Cu–30Zn

Fig. 5 Hardness distributions in welding joints (the indentations were carried out along a line 500 μm far away beneath the friction surface on the cross section)

minimum grain size ($d_{\min}$) induced by SPD can be expressed as:

$$d_{\min}/b = C(\gamma/(Gb))^q, \quad (3)$$

where b and G are Burgers vector and shear modulus, respectively; C is a constant; q is a exponential coefficient determined by experiment, i.e., the slope of the experimental data of a double log plot of $d_{\min}/b$ versus $\gamma/(Gb)$. It is evident

that smaller grain size can be obtained in materials with lower SFEs under a similar FSW condition.

3.3 Mechanical Properties

The hardness distribution obtained on the cross section of joints is presented in Fig. 5. Every point was averaged from 3 indentations. The hardness in the SZs distributes uniformly, and the values (~ 108.4 HV for Cu and ~ 155.2 HV for Cu–30Zn) are far higher than their BMs (~ 71.3 HV for Cu and ~ 96.7 HV for Cu–30Zn). However, the hardness in the TMAZs oscillates seriously because of the mixed microstructure of refined grains and severely deformed parent grains. The soft HAZ is not detected from the profile of hardness distribution (Fig. 5), which is consistent with the above microstructural examinations (Fig. 2). It is worth noting that the width of the TMAZ can be measured according to the hardness distribution. For example, the widths of TMAZs in the AS and RS of Cu–30Zn joints are measured as ~ 420 and ~ 1060 μm , respectively.

The typical engineering stress–strain curves of the samples cut from welding joints are shown in Fig. 6a, b. It is measured that the yield strength of the UFG SZs (275 MPa for Cu and 408 MPa for Cu–30Zn) is several times higher than that of CG BMs. Importantly, the current

Fig. 6 **a** Tensile engineering stress–strain curves of FSW pure Cu, **b** engineering stress–strain curves of FSW Cu–30Zn, **c** strain hardening rate (σ_T/ϵ) and true stress (σ_T) versus true strain for ultrafine-grained SZs

Fig. 7 Macroscopic appearances and DIC-characterized strain fields of transverse samples for pure Cu **a1–a3** and Cu-30Zn **b1–b3**: **a1, b1** tensile specimens before and after test. The white frames mark the effective DIC measured areas; **a2, b2** strain maps in the stage of nominally uniform plastic deformation; **a3, b3** strain maps in the necking stage. ϵ_y denotes the strain along tensile direction Y, while ϵ_x denotes the strain along transverse direction X. The numbers on the top of each strain subgraph indicate the applied strain; **c1, c2** evolution of local strain (locations of A, B and C marked in **a2, b2**) with the tensile strain

UFG SZs exhibit prominent strain hardening ability and considerable uniform ductility (10.5 and 20.1% for pure Cu and Cu–30Zn, respectively), which are much higher than those of most UFG materials produced by SPD methods [29, 34]. There are several approaches to improve the strain hardening capability of UFG materials, such as decreasing initial dislocation density [35], increasing the fraction of HAGBs [36], introducing nanotwins and/or nanoprecipitates into ultrafine grains [37, 38], etc. In the case of current FSW, the material in the SZ experienced complex thermoplastic process, including severely plastic deformation, high temperature cycling and rapid cooling by flowing water. Accordingly, grain refinement, extensive DRX and controlled post-annealing coupled during FSW, resulting in UFG microstructure with both high fraction of HAGBs and low density of dislocations, as confirmed in Figs. 3 and 4. Therefore, it is not surprise that strong strain hardening can occur in the current SZ samples (Fig. 6c).

It is noted that the strain hardening rate of the SZ of Cu–30Zn is much higher than that of Cu, which could be attributed to the higher volume fraction of HAGBs, the solute-induced multiplication of dislocations and drag effect [39, 40]. As similar as the results revealed in Cu and Cu–Al alloys [17, 19], the strength-ductility synergy of the UFG materials produced by friction stir processing is usually superior to that synthesized by conventional SPD methods. These results may also indicate that friction stir processing technology possesses the potential application of producing bulk UFG or performing surface modification on bulk materials to improve the mechanical properties.

As to the joints, they exhibit almost the same yield and ultimate tensile strength as that of CG BMs, but a reduced ductility, as confirmed by the stress–strain curves of transverse samples in Fig. 6a, b. The fracture appearances are shown in Fig. 7a1, b1 for Cu and Cu–30Zn, respectively. Clearly, the fracture occurred in the BMs far away from the welding point. In order to explore the plastic deformation of joints in detail, the technique of DIC is used and the strain images taken from the uniform elongation stage and necking stage are shown in Fig. 7a2–3, b2–3. It can be seen that the BMs undertake much higher plastic strain in both loading direction (ϵ_y) and transverse direction (ϵ_x) than the SZs. These results show that the failure process of current joints is dominated by their BM, which confirms the high strength of SZ and no appearance of soft HAZ.

The evolutions of local true strain in the BMs and the SZs (A, B and C marked in Fig. 7a2, b2) are plotted against the entire tensile true strain in Fig. 7c1, c2. It is seen that the strain of the BMs increases quickly with increasing applied strain, while the SZs stand always in the elastic range during the whole tension process, suggesting inhomogeneous deformation for joints. As shown in Fig. 7a2,

b2, a steep strain gradient is set up in a narrow area of TMAZ in order to coordinate the strain incompatibility between the BM and the SZ. Importantly, the accumulation of a large number of geometrically necessary dislocations is needed to accommodate this strain gradient, which can result in a significant strengthening effect for this area [41, 42]. Moreover, the initial straight transition line between the BM and the SZ, i.e., the TMAZ observed under optical microscope (Fig. 2), is bent gradually with increasing the applied tensile strain, which is possibly due to triaxial stress status and complex strain gradient distribution across the transitional zone [41].

4 Conclusions

In this work, ultrafine-grained microstructures and improved mechanical behaviors were observed in friction-stir-welded Cu and Cu–30Zn alloy. Based on the experimental results, the following conclusions can be drawn:

1. Defect-free joints with ultrafine-grained stirring zones can be produced by friction stir welding under flowing water.
2. During friction stir welding, the controlled heat input and intrinsic material characteristics determined the microstructure of joint. Onion-ringed structure appeared on the cross section of stirring zone in pure Cu, while plastic flowing bands were formed in the stirring zone of Cu–30Zn. Low stacking fault energy and low heat input were conducive to reduced dynamically recrystallized grains in the stirring zone.
3. A very good combination of strength and ductility was obtained in the stirring zones of both pure Cu and Cu–30Zn, possibly due to their recrystallized ultrafine grains with both high fraction of HAGBs and low density of dislocations. The technique of DIC was used to characterize the plastic deformation process of FSW joints during uniaxial tension, confirming improved mechanical performances of welding joints.

Acknowledgements This work was supported financially by the National Natural Science Foundation of China (Nos. 11672195 and 51301092) and Sichuan Youth Science and Technology Foundation (No. 2016JQ0047).

References

- [1] X.X. Xu, F.L. Nie, J.X. Zhang, W. Zheng, Y.F. Zheng, C. Hu, G. Yang, *Mater. Lett.* **64**, 524 (2010)
- [2] R.M. Leal, I. Galvão, A. Loureiro, D.M. Rodrigues, *Int. J. Adv. Manuf. Technol.* **80**, 1655 (2015)
- [3] H. Farrokhi, A. Heidarzadeh, T. Saeid, *Sci. Technol. Weld. Join.* **18**, 697 (2013)

- [4] W.M. Thomas, E.D. Nicholas, J.C. Needham, M.G. Murch, P. Templesmith, C.J. Dawes. International Patent Application, No. PCT/GB92/02203 (1991)
- [5] R.S. Mishra, Z.Y. Ma, *Mater. Sci. Eng. R* **50**, 1 (2005)
- [6] S.B. Aziz, M.W. Dewan, D.J. Huggett, M.A. Wahab, A.M. Okeil, T.W. Liao, *Acta Metall. Sin. (Engl. Lett.)* **29**, 869 (2016)
- [7] R.V. Barenji, *Trans. Indian Inst. Met.* **69**, 1077 (2016)
- [8] Y.F. Sun, N. Xu, H. Fujii, *Mater. Sci. Eng. A* **589**, 228 (2014)
- [9] K.N. Krishnan, *Mater. Sci. Eng. A* **327**, 246 (2002)
- [10] N. Kumar, R.S. Mishra, C.S. Huskamp, K.K. Sankaran, *Mater. Sci. Eng. A* **528**, 5883 (2011)
- [11] D. Wang, B.L. Xiao, D.R. Ni, Z.Y. Ma, *Acta Metall. Sin. (Engl. Lett.)* **27**, 816 (2014)
- [12] L. Commin, M. Dumont, J.E. Masse, L. Barrallier, *Acta Mater.* **57**, 326 (2009)
- [13] C.I. Chang, C.J. Lee, J.C. Huang, *Scr. Mater.* **51**, 509 (2004)
- [14] T. Morishige, T. Hirata, T. Uesugi, Y. Takigawa, M. Tsujikawa, K. Higashi, *Scr. Mater.* **64**, 355 (2011)
- [15] P. Xue, B.L. Xiao, Z.Y. Ma, *Mater. Des.* **56**, 848 (2014)
- [16] Y.F. Sun, H. Fujii, *Mater. Sci. Eng. A* **527**, 6879 (2010)
- [17] P. Xue, B.L. Xiao, Z.Y. Ma, *Sci. Eng. A* **532**, 106 (2012)
- [18] A. Heidarzadeh, T. Saeid, *Rare Met.* (2016). <https://doi.org/10.1007/s12598-016-0704-9>
- [19] P. Xue, B.L. Xiao, Z.Y. Ma, *Scr. Mater.* **68**, 751 (2013)
- [20] A. Azizi, R.V. Barenji, A.V. Barenji, M. Hashemipour, *Int. J. Adv. Manuf. Technol.* **86**, 1985 (2016)
- [21] P. Xue, B.L. Xiao, Q. Zhang, Z.Y. Ma, *Scr. Mater.* **64**, 1051 (2011)
- [22] N. Xu, R. Ueji, Y. Morisada, H. Fujii, *Mater. Des.* **56**, 20 (2014)
- [23] S. Emami, T. Saeid, *Acta Metall. Sin. (Engl. Lett.)* **28**, 766 (2015)
- [24] N. Xu, R. Ueji, H. Fujii, *Mater. Sci. Eng. A* **610**, 132 (2014)
- [25] G.M. Xie, Z.Y. Ma, L. Geng, *Scr. Mater.* **57**, 73 (2007)
- [26] H. Pashazadeh, J. Teimournezhad, A. Masoumi, *Int. J. Adv. Manuf. Technol.* **88**, 1961 (2017)
- [27] A.P. Reynolds, *Sci. Technol. Weld. Join.* **5**, 120 (2000)
- [28] S. Mironov, K. Inagaki, Y.S. Sato, H. Kokawa, *Philos. Mag.* **95**, 367 (2015)
- [29] Z.J. Zhang, Q.Q. Duan, X.H. An, S.D. Wu, G. Yang, Z.F. Zhang, *Mater. Sci. Eng. A* **528**, 4259 (2011)
- [30] T. Sakaia, A. Belyakov, R. Kaibyshev, H. Miura, J.J. Jonas, *Prog. Mater. Sci.* **60**, 130 (2014)
- [31] K.V. Jata, S.L. Semiatin, *Scr. Mater.* **43**, 743 (2000)
- [32] A. Heidarzadeh, T. Saeid, V. Klemm, *Mater. Charact.* **119**, 84 (2016)
- [33] F.A. Mohamed, *Acta Mater.* **51**, 4107 (2003)
- [34] M.A. Meyers, A. Mishra, D.J. Benson, *Prog. Mater. Sci.* **51**, 427 (2006)
- [35] C.X. Huang, W.P. Hu, Q.Y. Wang, C. Wang, G. Yang, Y.T. Zhu, *Mater. Res. Lett.* **3**, 88 (2015)
- [36] P. Xue, B.L. Xiao, Z.Y. Ma, *J. Mater. Sci. Technol.* **29**, 1111 (2013)
- [37] K. Lu, L. Lu, S. Suresh, *Science* **324**, 349 (2009)
- [38] B. Dang, X. Zhang, Y.Z. Chen, C.X. Chen, H.T. Wang, F. Liu, *Sci. Rep.* **6**, 30874 (2016)
- [39] Y.Z. Chen, X.Y. Ma, X.H. Shi, T. Suo, C. Borchers, K.H. Zhang, F. Liu, R. Kirchheim, *Scr. Mater.* **98**, 48 (2015)
- [40] Y.Z. Chen, H.P. Barth, M. Deutges, C. Borchers, F. Liu, R. Kirchheim, *Scr. Mater.* **68**, 743 (2013)
- [41] X.L. Wu, P. Jiang, L. Chen, F. Yuan, Y.T. Zhu, *Proc. Natl. Acad. Sci.* **111**, 7197 (2014)
- [42] H. Gao, Y. Huang, W.D. Nix, J.W. Hutchinson, *J. Mech. Phys. Solid* **47**, 1239 (1999)