Influences of welding speed on microstructure and mechanical properties of friction stir welded Al–Mg alloy with high Mg content

Yingli Li1, Hongge Yan1,2, Jihua Chen1,2, Weijun Xia1,2, Bin Su1,2, Tian Ding1 and Xinyu Li1
1 School of Materials Science and Engineering, Hunan University, Changsha 410082, People’s Republic of China
2 Hunan Provincial Key Laboratory of Spray Deposition Technology & Application, Hunan University, Changsha 410082, People’s Republic of China
E-mail: yanhg68@163.com

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Abstract
In this work, the influences of welding speed on microstructure and mechanical properties in friction stir welding (FSW) of the hot-rolled Al–9.2Mg–0.8Mn–0.2Zr–0.15Ti alloy plates has been investigated. Microstructures and mechanical properties of the joints are characterized by electron backscatter diffraction (EBSD), transmission electron microscopy (TEM), scanning electron microscopy (SEM) equipped with electron dispersive spectroscopy (EDS), hardness and tensile testing. The results show that all the joints are free of volume defects at the welding speed of 25–100 mm min−1 and the fixed rotation rate of 1000 rpm. And all the nugget zones (NZs) are characteristics of fine grains with the high angle grain boundaries (HAGBs) fraction higher than 90% at the welding speed of 25–100 mm min−1. The mean grain size in NZs and tensile properties of joints exhibit a parabolic relation with the welding speed. Furthermore, the optimal welding parameters are the welding speed of 50 mm min−1 and the rotation rate of 1000 rpm. The as-prepared joint at 50 mm min−1, featured with the smallest grain size of 3.02 μm and a uniform distribution of the fine second phase particles in NZ, exhibits the highest elongation to rupture about 45% higher than the base metal (22.2 ± 1.6%) and the highest ultimate tensile strength efficiency of 87.4%. It can be attribute to the synergetic effect of the fine-grain structure with the high HAGBs fraction and the small second phase particles with a uniform distribution.

1. Introduction

High Mg alloyed Al–Mg alloys are an essential structural material in the field of transportation and aerospace application due to high strength and light weight structure [1, 2]. Welding and joining are indispensable in the manufacture of structural components. Being an innovative solid-state joining technique, friction stir welding (FSW) can obtain the high-quality joints lack of hot cracking and porosity defects [3]. In recent years, FSW has achieved a notable success in joining and processing of the structural alloys, particularly Al alloys [4]. Welding variables such as the welding speed, the rotation rate and the tool geometry have significant impacts on microstructure evolution and mechanical properties, which are dominated by the thermal cycle and the material flow behavior during the FSW processing. The welding speed is the more important parameter affecting the mechanical properties of the joints among the above parameters [5].

Many studies focus on the microstructure evolution and mechanical properties of the FSWed Al–Mg alloys with the low and medium Mg content [6–9]. According to the study of Chhangani et al [6], the AA5024 aluminum alloy FSW joint possesses finer grains and smaller intermetallic precipitates in NZ with the increasing welding speed, which are associated with the decrease of heat input. El Rayes et al [7] have pointed out that the increased welding speed can refine grains and enhance the fragmentation and the dispersivity of second phase particles in NZ, but has a slight impact on the tensile properties of the AA5754-H111 FSWed joints. Peel et al [8] have reported that the heat input rather than the plastic deformation dominates the mechanical properties of the FSWed AA5083 alloy joints when welded at various welding speeds. Farahani et al [9] have...
investigated the correlation between the microstructure evolution and mechanical properties of the FSWed 5086–H34 alloy joints at various welding speeds and have found that the higher welding speed is beneficial to refine grains and improve both hardness and mechanical properties of the joints. Some researchers have also conducted modeling of FSW process to account well for the effect of process parameters on the microstructure and mechanical properties of FSWed Al–Mg alloy joints \cite{10,11}. Shash et al \cite{10} have studied the effect of welding speed on the peak temperature of AA5083–O FSWed joints by developing finite element modeling. And they have found that the increased welding speed can cause lower peak temperature. Grujicic et al \cite{11} have reported that a proper modeling can be used to explain qualitatively well the influence of welding speed on the heat/material flow and strength of FSWed AA5083 alloy.

So far, few studies have been conducted on the Al–Mg alloys with the Mg content exceeding 7 wt%. Thus it is worth exploring the feasibility of FSW in joining the high Mg alloyed Al–Mg alloys and to elucidate the relationships between microstructure and mechanical properties of the FSWed joints under the various FSW parameters. In this paper, the high Mg alloyed Al–Mg alloy sheets are welded by FSW at various welding speeds to evaluate the weldability and to investigate the influences of the welding speed on microstructure and mechanical properties of the joints. These preliminary results are expected to promote the development and application of FSW in the high Mg alloyed Al–Mg alloys.

### 2. Experimental procedure

The hot-rolled Al-9.2Mg-0.8Mn-0.2Zr-0.15Ti alloy sheets with the thickness of 2 mm were used for FSW. The experimental alloys were homogenized at 400 °C for 24 h followed by water quenched, then were hot rolled to a 86.7% thickness reduction. The prepared hot-rolled sheets were butt-welded by FSW perpendicular to the rolling direction (RD). Four welding speeds and the rotation rate of 1000 rpm were adopted on the basis of the previous exploration experiments. A tool with a 2.5° tilt angle, a shoulder 10 mm in diameter and a conical threaded pin (3.5 mm in root diameter, 2.5 mm in tip diameter and 1.8 mm in length) was used. The inserted depth of the shoulder to the sheet was controlled at about 0.15 mm. The detailed FSW parameters were summarized in table 1.

All the specimens for macrostructural and microstructural analysis were sectioned on the cross-section perpendicular to the welding direction (WD). The specimens for macrostructure examination were etched with Keller’s reagent for 60–90 s and were observed by stereoscopy (OLYMPUS DSX510). The distribution of the second phase particles in the joints were observed by scanning electron microscope (SEM, FEI Quanta 600) equipped with electron dispersive spectroscopy (EDS). The grain microstructural features of the joints were analyzed by electron backscatter diffraction (EBSD) on SEM (FEI Quanta 600) equipped with the TSL OIM software at the operating potential of 30 KV and by transmission electron microscopy (TEM, Titan G2 60–300). The samples for EBSD and TEM examination were prepared by electro-polishing and twin-jet electro-polishing in a mixture of 70% CH3OH + 30% HNO3 at 15 V and −30 °C, respectively. The grain size was determined by a linear intercept method.

The hardness profile of the joint was measured along the mid-thickness of the cross-section with the interval of 0.5 mm by a hardness tester (HVS-1000A) under a load of 200 g for 15 s. Tensile specimens with the gauge length of 15 mm and the width of 4 mm were cut from the welded sheets perpendicular to the WD and the weld zone was located in the center of each tensile specimen. Tensile testing was performed on the tensile machine (Instron 3369) with the strain rate of $1.1 \times 10^{-3}$ s$^{-1}$ at room temperature in triplicate. The tensile fracture surface was observed by SEM (FEI Quanta 600).

### 3. Results and discussions

#### 3.1. Macroposical features of the FSWed joints

The transverse cross-sectional macrographs of the FSWed joints are shown in figure 1. No volume defects are detected in the joints prepared at various welding speeds, indicating that FSW processing is feasible for the

| Samples | Welding speed (mm min$^{-1}$) | Rotation rate (rpm) |
|---------|-----------------------------|---------------------|
| 1       | 25                          | 1000                |
| 2       | 50                          | 1000                |
| 3       | 75                          | 1000                |
| 4       | 100                         | 1000                |
preparation of the sound high Mg alloyed Al–Mg alloy joints. The boundary between the nugget zone (NZ) and the thermo-mechanically affected zone (TMAZ) at the advancing side (AS) is evident. However, it is relatively indistinct at the retreating side (RS). Such observation is typical in FSW [12, 13]. With increasing the welding speed, the width of NZ decreases because of the reduced heat input. This result agrees with previously reported results for FSWed Al–Mg alloys [10, 11, 14]. As indicated by the arrows in figure 1(d), the kissing bond is featured with a dark wavy line in the NZ at the welding speed of 100 mm min$^{-1}$. It is reported that the kissing bond originates from the oxide layer on the initial butt surface and experiences the broken-up, extrusion and stirring during FSW [15, 16]. The weaker stirring associated with the low heat input causes an insufficient scattering of the broken pieces and the kissing bond is formed.

3.2. Microstructure of the FSWed joint

The joint prepared at 75 mm min$^{-1}$ and 1000 rpm is selected for further study. The BM and EBSD maps of the joints are shown in figure 2. The regions located at ‘A’, ‘B’, ‘C’, ‘D’ in figure 1(c) correspond to the base metal (BM), the heat-affected zone (HAZ), the TMAZ and the NZ, respectively. The misorientation angle distribution characteristics in HAZ, TMAZ and NZ are shown in figure 3. The HAGBs (with misorientation angles ($\theta$) $>$15$^\circ$) and the low angle grain boundaries (LAGBs, $2 < \theta \leq 15^\circ$) are marked by the bold black lines and the thin red lines, respectively.

As shown in figure 2(a), the BM exhibits a fibrous structure composed of the elongated grains along the RD. As shown in figure 2(b), the HAZ consists of a mixed microstructure containing the coarse elongated grains and the equiaxed grains (with a mean grain size of 6.53 $\mu$m), which indicates that static recrystallization occurs in HAZ. The hot-rolled alloy with high stored energy is prone to static recrystallization during thermal exposure and HAZ experiences thermal cycles without plastic deformation during FSW. As shown in figure 3(a), the HAGBs fraction is 66.3% and the average grain boundary misorientation angle ($\theta_{AV}$) is 28.4$^\circ$ in HAZ.

As shown in figure 2(c), the TMAZ at AS is characteristics of the recrystallized grains with the average size of 5.13 $\mu$m and a small number of the elongated grains in an upward flow pattern. It should be noted that dynamic recrystallization (DRX) occurs in TMAZ, which can be ascribed to the following two points. Firstly, the high content of Mg solute atoms in the Al–Mg alloy reduces the dynamic recovery (DRV) rate and promotes DRX by retarding dislocation motion [17–19]. Secondly, the combination of the plastic deformation and high temperature in TMAZ induce DRX during FSW processing [20]. As compared with HAZ, the HAGBs fraction and the $\theta_{AV}$ in TMAZ increase to 86.7% and 36.3$^\circ$, respectively.

As shown in figures 2(d) and 3(c), the NZ presents fine-equiaxed grains because of DRX induced by both high temperature and maximum strain [21] and the mean grain size is 3.25 $\mu$m. The HAGBs fraction and the $\theta_{AV}$ further increase to 91.9% and 38.3$^\circ$, respectively. Besides, the subgrains in NZ and TMAZ are surrounded by LAGBs. And the gradual increase in the HAGBs fraction from HAZ to TMAZ and NZ manifests that continuous dynamic recrystallization (CDRX) rules the grain structure formation in NZ with the gradual transformation from LAGBs to HAGBs from HAZ to TMAZ and NZ [22–24].

3.3. Changes of DRX grain size and misorientation angle with the welding speed in NZs

The EBSD maps, the DRX grain size distribution and the misorientation angle distribution in NZs at various welding speeds are shown in figure 4. All the NZs present the fine-equiaxed grains at the welding speed of 25–100 mm min$^{-1}$ because of the occurrence of DRX. As shown in figures 4(b), (e), (h) and (k), the mean grain size in NZs shows a parabolic relation with the welding speed. The grain size in the joint at 50 mm min$^{-1}$ is the smallest, which is due to the combined effects of the appropriate heat input and the intense material deformation during FSW. The higher welding speed causes a drop in both the peak temperature of heat input and the deformation degree, which are the determining factors of grain size [25, 26]. On the one hand, the decreased peak temperature reduces the driving force for the grain growth [24]. One the other hand, the reduced

Figure 1. Cross-sectional macrographs of the joints prepared with different welding speeds at 1000 rpm.
deformation degree leads to the increase of the grain size according to the recrystallization general principles [24]. Thus, the change of the grain size in NZs with the welding speed during FSW depends on which factor dominates. The refined grain in NZ from 25 to 50 mm min$^{-1}$ has relation to the reduction of heat input per the unit weld length, the shortened interval available for grain growth and the faster cooling cycle. However, the coarsening of grains in NZ from 50 to 100 mm min$^{-1}$ is due to that the deformation degree is the dominating factor at higher welding speeds. The result is similar to the published study [26]. The higher welding speed causes the weaker stirring, which reduces both dislocation density and the DRX nucleation sites and attains coarser grains.

Figure 2. OM image of BM (a) and EBSD maps in HAZ (b), TMAZ (c) and NZ (d) of the joint (75 mm min$^{-1}$, 1000 rpm).

Figure 3. Misorientation angle distribution in HAZ (a), TMAZ (b) and NZ (c) of the joint (75 mm min$^{-1}$, 1000 rpm).
Moreover, as shown in figures 4(c), (f), (i) and (l), each NZ possesses the HAGBs fraction higher than 90%. Besides the fine-equiaxed grains, the high fraction of HAGBs is formed during FSW. Almost complete DRX occurs in NZs at each welding speed. It is worth noting that the high Mg alloyed Al–Mg alloy used in the present study has a smaller grain size in NZ relative to the Al–Mg alloys with the low or medium Mg content under the similar conditions [7, 9]. On the one hand, the increased Mg solubility in Al achieves the stronger solid solution effect and causes the reduction of dislocation mobility and DRV rate. On the other hand, the high solute Mg content lowers the stacking fault energy of Al, impeding the dislocation slip and DRV. The combined roles lead to the acceleration of DRX and the formation of finer grains [17–19].

Figure 4. EBSD maps, DRX grain size distribution and misorientation angle distribution of the NZs at different welding speeds (Da indicates the average DRX grain size): (a)–(c) 25 mm min⁻¹; (d)–(f) 50 mm min⁻¹; (g)–(i) 75 mm min⁻¹; (j)–(l) 100 mm min⁻¹.
3.4. TEM microstructure of the joint

The joint prepared at 50 mm min\(^{-1}\) and 1000 rpm shows the finest grain size in NZ and thus it is selected for TEM study with BM for comparison. As shown in figure 5(a), the deformed grains in BM contain a great deal of dislocation tangles, indicating that high dislocation densities are involved. As shown in figure 5(b), the subgrain boundaries are formed by migration and rearrangement of dislocations in HAZ since the welding thermal cycle enhances the dislocations mobility. As shown in figure 5(c), the fine grains with clear boundaries and free of dislocations are detected, further confirming that static recrystallization in HAZ is driven by the high stored deformed energy and the welding thermal cycle. As shown in figure 5(d), dislocation tangles structures with high dislocation density and subgrains are formed in TMAZ. As shown in figures 5(e) and (f), NZ possesses the fine-equiaxed recrystallized grains with clear boundaries. Besides, dislocations in NZ are rearranged to gradually form the subgrain boundaries and are absorbed into the subgrain boundaries. FSW is a hot deformation process because of simultaneous existence of deformation and heat during FSW processing, resulting in formation of new dislocations inside the DRX grains\([27, 28]\). The microstructural characteristics in NZ manifest that dislocations are constantly induced by the further intense plastic deformation and subgrains are formed by DRV. These subgrains grow and develop into the fine-equiaxed DRX grains by absorbing dislocations into the subgrain boundaries. The processes are characteristic of CRDX\([29, 30]\).

3.5. Variation of second phase particles with the welding speed

The STEM image and elemental mapping of the particles in NZ prepared at 50 mm min\(^{-1}\) and 1000 rpm are shown in figure 6. These particles are rich in Mn and Fe, indicating they are the intermetallic phase composed of Mn and Fe. The morphologies and the sizes of the second phase particles in BM and joints at various welding speeds are shown figure 7 and the coarse plate-shape particles in BM distribute along the RD. These particles in BM and NZ at 50 mm min\(^{-1}\) (as marked by the circle in figures 7(a) and (c)) are examined by EDS and the results are shown in figure 8. In the light of the elemental mapping (figure 6) and EDS analysis (figure 8), these particles mainly consist of Al, Mg, Mn and Fe elements. Some researchers have pointed out that the second phase particles rich in Al, Mn and Fe are the Al\(_6\)(Mn, Fe) phase in the Al–Mg alloys\([31–33]\).

For one thing, the particles experience the broken-up and the redistribution owing to the mechanical stirring. For another, the temperature in NZ rises to 0.85–0.9 \(T_m\) of the alloy melting point due to the FSW heat, which makes the pre-existing particles grow up but these particles are not dissolved since the solvus temperature of this phase (about 635 \(^{\circ}\)C) exceeds the peak temperature that can be reached during FSW\([34, 35]\).

As shown in figure 7, the particles in NZs become fragmented and redistributed at every welding speed during FSW and refined particles in NZ at 50 mm min\(^{-1}\) exhibit a homogeneous distribution. The particles in
NZ at 25 mm min$^{-1}$ become coarsening due to the higher deformation temperature and lower cooling rate. However, a combination of the decreased deformation degree and the reduced material flow ability at higher welding speed of 75 and 100 mm min$^{-1}$ weakens the fragmentation and dispersion effects of the particles, causing the clustering and coalescence of the particles. As shown in figures 7(f)–(h), the particles in TMAZ at AS and RS are also broken up and are distributed in an upward flow streamline pattern. The particle in HAZ shows almost no difference with that in BM.

### 3.6. Hardness profile of the FSWed joints at various welding speeds

The hardness profiles of the joints measured along the cross-section centerline at different welding speeds (marked with the dotted line in figure 1(c)) are shown in figure 9. As compared with BM, the decreased hardness value in the weld region indicates the occurrence of the softening phenomenon since the reduction or the elimination of the preexisting work hardening effect occurs during FSW [36]. The softening zones are observed in the NZ, the TMAZ, and the HAZ, and the hardness minimum locates the HAZ adjacent to the TMAZ at the AS and RS. The hardness of the joint increases at first and then decreases with the increase of the welding speed and reaches the maximum at 50 mm min$^{-1}$. The hardness variation is primarily related to the grain size, the size and the distribution of second phase particles and the dislocation density for Al–Mg alloys [21, 37]. The increased degree of material softening and the coarsening of grains and second phase particles result in the joint at 25 mm min$^{-1}$ with the minimum hardness and the widest softened region. With the welding speed increasing to 50 mm min$^{-1}$, grain refinement and the homogeneous distribution of the smallest second phase particles are the reasons for the maximum hardness of the joint. As the welding speed further increases to 75 and 100 mm min$^{-1}$, the joint shows a slight decrease in the hardness due to the coarser grains and the clustered particles. In addition, the hardness variation of the joint with the welding speed corresponds to the microstructure features.

### 3.7. Fracture location of the FSWed joints at various welding speeds

The fracture locations of the joints at various welding speeds are shown in figure 10. It is noted that, except for the joint at 100 mm min$^{-1}$, the tensile specimens of the other joints fracture at the HAZ adjacent to the TMAZ at the AS or RS, which coincides with the lowest hardness location. However, the joint at 100 mm min$^{-1}$ fractures in NZ during the tensile testing. The loose structure and the coarse second phase particles caused by the insufficient flow at a higher welding speed result in stress concentration and crack initiation of the joint [38, 39].
Moreover, the kissing bond observed in NZ (figure 1(d)) is possibly the stress concentration center and significantly degrades the tensile properties of the joint.

3.8. Tensile properties of the FSWed joints at various welding speeds

The average tensile properties of the joints at different welding speeds and BM are summarized in table 2. The joints exhibit lower ultimate tensile strength (UTS) and yield strength (YS) than BM, which can be attributable to
the loss of the working hardening effect caused by recrystallization and annealing softening during FSW [36].

The YS values are relatively constant (around 267 MPa) for all the joints. The typical engineering stress-strain curves for the joints at different welding speeds and the UTS and elongation to rupture (Er) of the joints as a function of the welding speed are shown in figure 11. With the welding speed increasing, the UTS and Er of the joints increase at first and then decrease and reach the maximum value in the joint at 50 mm min $^{-1}$.

The joint prepared at 50 mm min $^{-1}$ and 1000 rpm exhibits the highest UTS of 477 ± 4 MPa (with the joint efficiency of 87.4%), which can be mainly ascribe to the refined grains with high HAGBs fraction and the homogeneous distribution of the small second phase particles. The fine-grain structure implies that a great number of grain boundaries act as the main obstacles to dislocations motion and enhance the localized plastic deformation resistance [40]. The high HAGBs fraction suggests the large misorientation among the adjacent grains, which creates more barriers against the dislocation movement and enhances resistance against grain boundary disruption by dislocations, causing grain boundary strengthening [41, 42]. Moreover, the fragmented second phase particles also provide barriers that impede the dislocation movement, which improves the strength of the joint [33]. Nevertheless, the reason for the poor UTS of the joints at the higher welding speeds can be explained by the loose bonding structure and the clustered second phase particles resulted from the reduced flowability and the lower deformation degree [38].

It should be noted that some joints have lower Er values than BM but Er of the joint at 50 mm min $^{-1}$ is higher than BM (22.2 ± 1.6%, with the joint efficiency of 145.1%). The reasons are as the following. Firstly, the fine grains with HAGBs exerts a significant influence on the Er value. According to Huo, Terlinde and Luetjering

### Table 2. Tensile properties of the BM and joints prepared with different welding speeds.

| Welding speed (mm min $^{-1}$) | UTS (MPa) | YS (MPa) | Joint efficiency in UTS (%) | Er (%) | Joint efficiency in Er (%) |
|-------------------------------|-----------|----------|-----------------------------|--------|--------------------------|
| BM                            | 546 ± 5   | 357 ± 2  | —                           | 15.3 ± 1.2 | —                       |
| 25                             | 455 ± 3   | 267 ± 2  | 83.3%                       | 13.7 ± 0.8 | 89.5                    |
| 50                             | 477 ± 4   | 267 ± 3  | 87.4%                       | 22.2 ± 1.6 | 145.1                   |
| 75                             | 436 ± 8   | 269 ± 5  | 80.0%                       | 11.3 ± 0.9 | 73.9                    |
| 100                            | 404 ± 10  | 264 ± 3  | 74.0%                       | 8.8 ± 0.6  | 57.5                    |

Figure 9. Hardness distribution characteristics of the joints prepared with different welding speeds at 1000 rpm.

Figure 10. Fracture locations of the joints prepared with different welding speeds at 1000 rpm.
the fine-grain structures are greatly beneficial for the enhancement of compatibility among grains and deformation in coordination and thus the elongation is improved. More uniform slips occur and some dislocations are existent in the fine grains, which improve the work hardening capacity during the tensile deformation. Secondly, the fine second phase particles with uniform distribution exert a certain impact on the increased elongation. Imam et al have concluded that the uniformly distributed small particles can promote the additional dislocation accumulation and thus enhance the work hardening capacity [45]. Thirdly, the fracture strain occurs not only in the NZ, the TMAZ but also in the HAZ. Because the HAZ comprises of many fine equiaxed grains, which can accommodate the newly introduced dislocations during the tensile deformation. Consequently, the Er of the joint is improved under these combined effects.

As shown in figure 12, the fracture morphology of the tensile specimen at the welding speed of 50 mm min⁻¹ features by the abundant tearing ridges and deeper dimples, which is in good consistency with the higher elongation. The shallower dimples for the joints at 75 and 100 mm min⁻¹ imply their lower elongations.
4. Conclusions

FSW of Al-9.2Mg-0.8Mn-0.22Zr-0.15Ti alloy is conducted at various welding speeds and a fixed rotation rate and the influences of welding speed on the microstructure and mechanical properties of the joints are studied.

(1) The hot-rolled high Mg alloyed Al–Mg alloy plates with the thickness of 2 mm are successfully welded by FSW with the welding speed from 25 to 100 mm min\(^{-1}\) and the joints free of volume defects are achieved.

(2) All the NZs have the fine-equiaxed grains with the high HAGBs fraction higher than 90% at the welding speed of 25–100 mm min\(^{-1}\). The mean grain size presents a parabolic relation with the welding speed and the joint at 50 mm min\(^{-1}\) exhibits the smallest grain size (3.02 μm).

(3) The second phase particles in NZ are fragmented and redistributed at various welding speeds. The refined particles in NZ reveal a uniform distribution at 50 mm min\(^{-1}\) and become clustering and coalescence at 75 and 100 mm min\(^{-1}\).

(4) The UTS and Er of the joints increase at first and then decrease with the increasing welding speed. The joint at 50 mm min\(^{-1}\) reveals the optimal tensile properties, with UTS of 477 ± 4 MPa (with the joint efficiency of 87.4%, and Er of 22.2 ± 1.6% (about 145.1% of BM). It can be attributable to the synergetic effects of the fine-equiaxed DRX grains with the high fraction of HAGBs and the presence of the small second phase particles with uniform distribution.

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ORCID iDs

Hongge Yan 🏷️ https://orcid.org/0000-0001-8659-5145

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