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Abstract

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A new finite element model for multi-cycle accumulative roll-bonding process and experiment verification

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Abstract: The modelling of multi-cycle accumulative roll-bonding (ARB) is challenging due to the repetitive cutting, stacking and roll-bonding. In this study, a finite element model was proposed and by which the real ARB process was successfully simulated for 5 cycles for the first time. Mapping solution, a remeshing analysis technique, was adopted to transfer the deformation solution from the deformed mesh to a new mesh between cycles, which not only enabled the simulation of discontinuous processes, but also alleviated the mesh distortion. Moreover, with this technique, tensile tests of ARB processed materials were also simulated. The predictions have been validated by the corresponding experimental observations. The deformation behaviours, in terms of texture, hardness, tensile strength and fracture, and plasticity instability were studied. The evolution of shear strain explained the heterogeneity and transition of textures. The distribution of hardness was in the same pattern as equivalent plastic strain through the thickness at first and then it tended to be uniform due to microstructural saturation. The investigation of tensile fracture and plastic instability showed that their occurrence was related to the difference in strength between the two components of composites.

Keywords: accumulative roll-bonding; finite element method; hardness; tensile test; plastic instability; texture

1. Introduction

As a relatively new severe plastic deformation (SPD) method, accumulative roll-bonding (ARB) has attracted extensive scientific research interests (e.g. [1, 2]) since its introduction in 1998 [3]. In the ARB process, two stacked sheets are firstly roll-bonded at a 50% reduction and then they are cut into two halves, which are subsequently stacked together in preparation for the next cycle [3]. The process can repeat for a potentially unlimited number of cycles, as the geometry of the sheet remains unchanged.

Due to the repeated cutting-stacking and rolling, the evolution and distribution of strains in ARB are complicated, where the highly strained surface is simply compressed after being transferred to the centre of the next cycle and then the shear deformation increases as it moves toward the surface in the following cycles. To measure the shear strain, the embedded-pin method has been adopted in experiments only in the first cycle of ARB [2, 4], i.e., conventional rolling (or roll-bonding), since measuring the curves of deformed pins after a large strain is not easy. In contrast, finite element method (FEM), as an alternative, can predict precise strains [5]. However, only a few studies concerning FEM modelling of ARB have been attempted, because the cutting-stacking makes multi-cycle ARB a discontinuous process [2]. The discontinuity is the first challenge and due to this, ARB was simulated to only one [2, 5-7] or two cycles [8]. The cumulative distortion of FEM meshes, which would cause convergence and accuracy problems, is the second challenge. For instance, a severely deformed FEM mesh resulted from the modelling method adopted in Ref. [9] and the ARB simulation was restricted to a low cycle number of three. To the authors' best knowledge, no FEM model that has been used to simulate the real ARB up to a high number of cycles (≥ 5) has been proposed and thus, no in-depth analysis of ARB deformation behaviours based on FEM simulations has been performed to explain common experimental findings.

In this study, an FEM model following the real ARB process was developed and the simulation was successfully conducted up to 5 cycles. Prior to the simulation, corresponding experimental research was carried out too, which was used to verify the prediction. Texture evolution, hardness, tensile strength, and plasticity instability were investigated based on the numerical study.

2. Materials and experimental procedure

The materials used in this work were commercial aluminium alloy AA1050 (Al-0.05Cu-0.4Fe-0.05Mg-0.04Mn-0.25Si-0.03Ti-0.05Zn-0.05V- wt%) and AA6061 (Al-0.22Cu-0.42Fe-0.87Mg-0.05Mn-0.58Si-0.04Zn-0.21Cr- wt%) sheets (1mm in thickness). Before ARB, AA1050 was annealed at 450°C for 1 h, while AA6061 at 500°C for 2 h. A fully annealed homogeneous microstructure was achieved, with an average grain size of 96 μ m for AA1050 and 36 μ m for AA6061. According to the combination of the two materials, three categories, AA1050/AA1050, AA6061/AA6061 and AA1050/AA6061, were deformed. The ARB process followed the standard 4-step (wire-brushing, stacking, roll-bonding and cutting) procedure [10]. Two sheets of the starting materials were stacked after being cleaned with acetone and wire-brushed, and then they were joined together by metal wires at the four corners to avoid relative displacement. The joined sheets were rolled (50% reduction) under dry conditions with acetone-cleaned rolls (125mm in diameter) and then cut into two halves. The processing proceeded up to 5 cycles for all three categories. Embedded-pin tests [1, 2] were conducted to measure the shear strain in the first cycle of AA1050/AA1050 and the

flection of the curved embedded-pins was used to calibrate the surface friction in the simulation, exactly the same method as that in Ref.[2].

Hardness and tensile tests were carried out to test mechanical properties of all three composites. Vickers micro-hardness (load: 25g, dwell time: 12s) was measured after each cycle, and about 30 points were taken along the normal direction (ND) on the rolling direction (RD)-ND plane. Tensile samples (gauge length: 25mm, gauge width: 6mm) were machined along the RD, and the tests (initial strain rate of 0.001/s) were conducted at room temperature using an Instron 5566 testing machine. After tensile testing, the tensile fracture on the RD-ND plane was observed by optical microscopy (OM) after mechanical grinding. OM was also used to observe the bonded interfaces in the longitudinal cross-section of ARBed AA1050/AA6061, before which the surface was ground and polished (finish step: OPS), and finally etched with Barker's reagent. Electron backscattered diffraction (EBSD) scans were also made with a field emission scanning electron microscope (JEOL JSM-7001F) to characterize the texture of the 1-, 2- and 3-cycle of AA1050/AA1050. For preparing EBSD samples, the longitudinal sections from the processed sheets were mechanically ground, and then polished with diamond powder up to 1 μ m and finally electropolished (16°C, 25V) in A3 Struers electrolyte with a Struers LectroPol-5. EBSD scans (accelerating voltage: 15kV, working distance: 15mm) were extended as close to the surfaces as possible and HKL Channel-5 software was used to determine crystallographic orientations.

3. FEM modelling of ARB process

The experimentally obtained strain-stress curves of fully annealed AA1050 and AA6061 were fitted up to a large strain (Fig.1a) according to the model [11] expressed as

$$\sigma = \sigma_0 + \sigma_1 [1 - \exp(\frac{-\varepsilon^n}{\varepsilon_c})] \quad (1)$$

where σ_0 is the initial yield stress, ε_c is the characteristic strain, n is an exponent, σ_1 is a constant, and $\sigma_0 + \sigma_1$ is the saturation (upper limit on the value of) stress. The values of σ_0 , σ_1 , ε_c and n for both AA1050 and AA6061 are listed in Table 1. Microstructural refinement and saturation (no further refinement) are taken into account in this model [11], so the stress saturation at large strain can be reasonably expected.

Table 1. Parameters used in Eq.1 for AA1050 and AA6061.

	σ_0 (MPa)	σ_1 (MPa)	ε_c	n
AA1050	25	170	0.7	0.5
AA6061	58	350	1.25	0.5

The simulation, designed to match the experiment conditions, was performed by commercial FEM code Abaqus/Standard ver.6.9, and a two-dimensional FEM model under

the assumption of plain strain conditions was used. The sheet was meshed into 81680 elements and 82782 nodes after mesh calibrations. The rolls were considered as analytical rigid bodies and their rotation drove the sheet to deform through surface friction. The coefficient of friction between the sheet and rolls was tested by matching the deformed FEM mesh to the curved embedded-pins in the experiment [1, 2], and thus a coefficient of 0.25 was chosen. Plane strain four-node elements with one integration point (element id: CPE4R) were used, which can provide an efficient and fast numerical formulation. Enhanced hourglass control was applied to increase the resistance to the hourglassing problem and accuracy to the displacement solution.

Fig.1. (a) Experimentally measured and fitted strain-stress curves of fully annealed AA1050 and AA6061, (b) ARB FEM model.

The two challenges, i.e., discontinuous process and cumulative FEM mesh distortion, were well addressed in the proposed ARB FEM model, as illustrated in Fig.1b. Different from the experiments, only three steps were considered in this model, and they are 'Rolling', 'Cutting' and 'Stacking'. The 'Wire-brushing' step was omitted because its effect is limited to a superficial layer [12, 13]. The two stacked sheets were simplified to a single-layered one, i.e., using conventional rolling to replace roll-bonding, since the relative movement between sheets was prohibited in the experiment by joining them with metal wires. After the first step 'Rolling', 'Cutting' was realized by mapping solution, a remeshing analysis technique. The deformed mesh was completely replaced by a new mesh after mapping solution (Fig.1b), but the deformation dependent solution was transferred to the new mesh via interpolation. In the 'Cutting' step, the rolled single-layered sheet was mapped into two separated ones (Fig.1b), and then they were stacked by translating along the RD and ND rigidly in the 'Stacking' step. The two stacked sheets were mapped into one integrated sheet again before the next cycle, and then boundary conditions were applied. With the aid of mapping solution, tensile tests of ARB processed materials were also simulated for the first

time. The deformed sheets were mapped to reconstructed tensile samples and were then displaced along the RD by tension at the two ends.

4. Results

In the following, the thickness position is defined by t/t_0 , where t is the distance from the upper surface to a thickness position and t_0 is the whole thickness. Therefore, $t/t_0=0$ and 1 correspond to the upper and lower surfaces, respectively.

4.1. Evolution of stress and strain

Fig.2 shows the distribution of von Mises stress in the rolling bite of AA1050/AA1050. In the first cycle, the stress at surfaces is higher than at the centre because of surface friction, and the surface stress is relatively low at the neutral point. The distribution of stress in 2- and 3-ARB clearly reflects the cutting-stacking pattern, where the highly stressed surface in the first cycle was moved to the centre and quarter position in the second and third cycle, respectively, as marked by the arrows. Stress became uniform after 5-ARB due to its saturation.

Fig.2. Distribution of von Mises stress in the rolling bite of AA1050/AA1050.

The cumulative equivalent plastic strain (ε_{eq}) of AA1050/AA1050 is plotted in Fig.3a. The ε_{eq} is defined as

$$\varepsilon_{eq} = \frac{2\sqrt{3}}{3} \int_0^t \sqrt{\left(\frac{d\varepsilon_{xx}}{dt}\right)^2 + \frac{1}{4}\left(\frac{d\gamma_{xy}}{dt}\right)^2} dt \quad (2)$$

where $d\varepsilon_{xx}$ is the nominal strain along the X direction (RD in Fig.1b) and γ_{xy} is the shear strain on the X-Y (RD-ND) plane. The nominal strain ε_{xx} is 0.8, equal to that along the Y direction, ε_{yy} . After 1-ARB, ε_{eq} distributes symmetrically about the mid-thickness and increases from 0.8 at the centre to 3.2 at the surface, where 0.8 corresponds to the value of ε_{eq} theoretically calculated under plain strain compression after a 50% reduction, while the

large ε_{eq} at the surface was caused by the surface friction. At the beginning of the second cycle, two 1-ARB processed sheets were stacked and mapping solution was applied. By comparing the value of ε_{eq} after 1-ARB and before 2-ARB, a small discrepancy can be found at the surface and centre of 2-ARB, since the strain gradient at the surface of 1-ARB is very high. After the second cycle, the surface underwent the same deformation twice with ε_{eq} reaching 6.2 ($=3.0+3.2$), which, without the discrepancy, is supposed to be 6.4 ($=3.2+3.2$). The ε_{eq} at the centre is 3.8 ($=3.0+0.8$), because the highly strained surface was then deformed by compression. The minimum ε_{eq} locates at the quarter position with a value of 1.8 ($=0.8+1.0$). The large variation of ε_{eq} formed 5 and 17 peaks through the thickness after the third and fifth cycle, respectively. The strain at a certain thickness position can be calculated by adding the strain imposed in the current cycle to the pre-existing strain cumulated from the previous cycles. The shear strain γ_{XY} developed in a single cycle, not the cumulative value, is similar (Fig.3b). It is positive ('+') in the upper half, while it is negative ('-') in the other half, as indicated by the deformed FEM mesh in the inserted figure (Fig.3b). The maximum value, 0.48, locates at the subsurface, but not the exact surface, because the direction of shear stress reversed after passing the neutral point. By the strain adding method, the cumulative γ_{XY} after 2- and 3-ARB was obtained and is shown in Fig.3c and d, respectively. Before 2-ARB, the distribution of the cumulative γ_{XY} through the thickness is in a '+ - + -' pattern (Fig.3c). Shear strain in the '+ -' pattern, nearly the same as that in the first cycle (Fig.3b), was introduced one more time after the second cycle. In the region from $t/t_0=0$ to 0.25, the γ_{XY} developed in 1- and 2-ARB alone is unidirectional (marked by 'Uni' in Fig.3c), so the cumulative γ_{XY} increased. In contrast, the directions of γ_{XY} in 1- and 2-ARB are opposite in the second layer ($t/t_0=0.25$ to 0.5), and this means the γ_{XY} developed in the second cycle reversed ('Rev' in Fig.3c) the shear deformation evolved in the first cycle, so the cumulative value decreased. Reversed and unidirectional shear strain evolved in the third and fourth layers of 2-ARB, respectively. The reversal of cumulative shear strain also developed in 3-ARB (Fig.3d). Only surface experienced consecutively unidirectional shear strain. The shear strain reversal in ARB occurred due to the repeated rolling and cutting-stacking.

Fig.3. Distribution of (a) cumulative ϵ_{eq} , and (b) shear strain γ_{XY} evolved in a single cycle of AA1050/AA1050. Distribution of cumulative γ_{XY} before and after (c) 2-ARB, and (d) 3-ARB. Inserted figure in (b): shear strain and deformed FEM mesh after 2-ARB.

A similar evolution of strains was also observed in AA6061/AA6061. However, unlike the monotonic material composites, an asymmetrical distribution of strains (Fig.4) and the deformed FEM mesh (inserted figure in Fig.4b) developed in AA1050/AA6061, and sharp changes appeared at the bonded interfaces. The ϵ_{eq} at the upper surface, i.e., AA1050 layer, is about 2.7 times that of the lower surface (AA6061) after the first cycle (Fig.4a), and this ratio increased to 5.6 after 4-ARB. Moreover, the strains in the AA1050 and AA6061 layers of AA1050/AA6061 are higher and lower, respectively, than that at the same thickness position in AA1050/AA1050. This means, interacting with each other at the interfaces, the material flow in the softer and harder material was enhanced and reduced, respectively. This kind of asymmetrical distribution of shear strain has been observed by the embedded-pin method in the first cycle of ARB deformed Cu/TA [14].

Fig.4. Distribution of (a) cumulative ε_{eq} , and (b) shear strain evolved in a single cycle of AA1050/AA6061. Inserted figure in (b): shear strain and deformed FEM mesh after 2-ARB.

4.2. Effect of strain to texture evolution

The evolution of shear strain at a position in AA1050/AA1050 was traced and is shown in Fig.5. It is point A1, as marked in Fig.3b, at the upper surface of 1-ARB. A large shear strain developed at this point though its value significantly dropped after passing the neutral point. A1 either stayed at the upper surface (B2a in Fig.3c) or moved to the centre (B2b) in the second cycle. The large surface shear strain was applied one more time at B2a, while it remained unchanged at B2b (Fig.5). The mid-thickness point B2b moved to the upper and lower quarter position in the third cycle, i.e., C3a and C3b (Fig.3d), respectively, and a shear strain of medium magnitude developed at these two points (Fig.5).

Fig.5. Evolution of cumulative shear strain at a position.

The effects of distribution and change of shear strain to textures were investigated. Fig.6a-c show the distribution of ideal texture components of AA1050/AA1050 after 1-3 cycle. Here, $\{112\}\langle 111 \rangle$, $\{123\}\langle 634 \rangle$ and $\{110\}\langle 112 \rangle$ are defined as rolling texture [15], while shear texture includes $\{001\}\langle 110 \rangle$, $\{111\}\langle 112 \rangle$ and $\{111\}\langle 100 \rangle$. A tolerance of 15° was used to classify crystallographic orientations. The whole thickness was divided into 6 equal regions and the fractions of rolling and shear textures in them are shown in Fig.6d. Shear texture

dominated at the surface after the first cycle because of the large shear strain (point A1 in Fig.5). The shear texture was preserved when the surface remained surface (B2a in Fig.5), but transferred into rolling texture after moving to the centre (B2b in Fig.5). When the centre region moved to the quarter position, a combination of both shear and rolling textures developed due to its medium shear strain (C3a and C3b in Fig.5). It can be seen that the texture transition between shear texture and rolling texture is closely associated with the evolution of shear strain. The distribution of texture in all three cycles is similar (shear texture at surface and rolling texture at centre), corresponding to the shear strain imposed in a single cycle being similar (Fig.3b). This means texture distribution is mainly determined by the currently imposed shear strain [1, 2, 15], while the influence of previously developed textures is weak.

Fig.6. Distribution of texture after (a) 1-, (b) 2-, (c) 3-ARB. (d) Fractions of shear and rolling textures.

4.3. Prediction of mechanical properties

Hardness was calculated based on the flow stress that was identified from the cumulative ϵ_{eq} according to Eq.1, where hardness is 1/3 of the flow stress [16]. It can be seen from Fig.7a-c that the predicted through-thickness hardness matches well with the experimental measurement. After 1-ARB, the hardness is high at the surface, but relatively low at the centre in all three composites. In the following cycles, the hardness increased slightly at the surface of AA1050/AA1050 (Fig.7a and d) because the stress had nearly reached saturation at the surface of 1-ARB (Fig.2), while it increased obviously in AA6061/AA6061 (Fig.7b and d). As for AA1050/AA6061 in Fig.7c, sharp changes of hardness evolved at the bonded interfaces. Because of microstructural saturation, the hardness in the AA1050 layers of AA1050/AA6061 is close to that at the same thickness position in AA1050/AA1050, though

ε_{eq} of the former is much higher than the latter (Fig.3a and Fig.4a). However, the hardness in the AA6061 layers of AA1050/AA6061 is noticeably smaller than its counterpart in AA6061/AA6061 due to the relatively lower ε_{eq} of the former, which was especially pronounced at the lower surface of the first cycle (Fig.7b and c).

Fig.7. Distribution of hardness through the thickness of (a) AA1050/AA1050, (b) AA6061/AA6061, and (c) AA1050/AA6061. (d) Evolution of hardness with strain.

Tensile tests of ARB processed composites were simulated, and the obtained representative strain-stress curves of 3-ARB processed composites agree well with the experimental measurements in Fig.8a, which shows the accurate fit of strain hardening (Fig.1a). The strength of AA1050/AA6061 locates mid-way the two monotonic material composites, which is in accord with the rule of mixture [17] that the strength of AA1050/AA6061 ($\sigma_{1050/6061}$) can be fitted by $\sigma_{1050/6061} = V_{1050}\sigma_{1050} + V_{6061}\sigma_{6061}$, where V_{1050} and V_{6061} are the volume fractions of AA1050 and AA6061, and σ_{1050} and σ_{6061} are the tensile strengths, respectively. The tensile fracture of 3-ARB AA1050/AA6061 is in the shear direction (Fig.8c), the same as that observed in the experiment (Fig.8b). However, it is necking in 1-ARB AA1050/AA6061 (Fig.8d and e). The difference in fracture behaviours between 3- and 1-ARB will be discussed later. Incidentally, in Fig.8d and e, the relatively slow necking of AA1050 demonstrates its high ductility.

Fig.8. (a) Strain-stress curves obtained from tensile tests of 3-ARB processed composites. Tensile fracture of (b) 3-, and (d) 1-ARB AA1050/AA6061 observed by OM in the experiment. Simulated σ_{xx} in the tensile sample of (c) 3-, and (e) 1-ARB AA1050/AA6061.

4.4. Plastic instability in AA1050/AA6061

In the AA1050/AA6061, the bonded interfaces were straight after up to the third cycle (Fig.9a) and started to become wavy during the fourth cycle (Fig.9b and d) in both simulation and experiment, where the OM images of 3-5 ARB AA1050/AA6061 were obtained under polarized light (Fig.9a-c), and Fig.9e and f show the enlarged figures of separated AA6061 and AA1050 layers in the simulation, respectively. The thickness of AA6061 layers varies significantly along the RD with maximum and minimum values of $98\mu\text{m}$ and $29\mu\text{m}$ (Fig.9e), respectively. This means necking occurred in AA6061 layers. In contrast, the change of thickness in AA1050 layers is relatively low (Fig.9f) though the interfaces are not straight. Moreover, the interfaces at centre were wavier than those close to the surfaces (Fig.9b, c and f). The necking eventually evolved into fracture in 5-ARB (Fig.9c), which was not simulated due to the irregularity of the interfaces.

Fig.9. OM images of RD-ND plane of AA1050/AA6061 after (a) 3-, (b) 4-, and (c) 5-ARB in the experiment. Distribution of σ_{xx} of 4-ARB (d) in the rolling bite, (e) after rolling of AA6061, and (f) AA1050 layers in the simulation.

5. Discussion

In ARB, 5 cycles are generally required to reach microstructural saturation and other phenomena such as the plastic instability in this study only appear after a large strain. The proposed FEM model has the potential to simulate ARB for an unlimited number of cycles. Mapping solution is crucial to the success of the FEM model. It not only enables the simulation of discontinuous processes, but also eliminates the FEM mesh distortion developed in a single cycle, which potentially allows the simulation to propagate. The replacement of FEM mesh and boundary conditions slightly alters the constraint conditions and accordingly, it is important to determine when to perform mapping solution. A severe solution discontinuity suggests that an earlier application of mapping solution is desired, since the severe discontinuity is caused by using interpolation to large gradients of deformation dependent solution. This severe discontinuity could be reduced by improving mesh resolution, i.e., utilizing more elements to describe large gradients. In the current research, acceptable accuracy of mapping solution has been achieved (Fig.3a). A simplification adopted in this FEM model is by using rolling to replace roll-bonding. The accurate prediction of the occurrence of wavy interfaces in AA1050/AA6061 shows that the

main deformation mechanism in ARB has been captured due to two reasons. Firstly, the two sheets were joined by metals wires at four corners and no bending was observed during processing, which means no, or little (if any), relative slide occurred. Secondly, the sheet surfaces to be bonded were wire-brushed, which enhanced their roughness and dramatically increased friction between them. The high friction in turn reduced relative displacement. Another two factors that influence the accuracy of prediction are the friction coefficient and stress-strain curve that have been validated by the experimental observations.

5.1. Texture and microstructure evolution in ARB

Texture evolution is mainly determined by the starting texture, i.e., previously developed texture, and the newly imposed strain [18]. In ARB, texture that had reached stability became unstable after being moved to another thickness position (due to cutting-stacking), since the through-thickness shear strain varies. This is how texture transition occurs in ARB. Texture evolution is associated with strains, especially shear strain, in two aspects: magnitude and direction. The texture transition from shear texture to rolling texture in ARB was quick and complete (Fig.6), which means, as previously stated, the newly imposed strain played a dominant role over the starting texture. However, the transition made the texture intensity very low compared to that after conventional rolling [15, 19, 20]. The fraction of rolling texture (<50%) in the whole sample (Fig.6) is significantly smaller than that of 75% in conventional rolling [19]. This reduction resulted from the effect of previously developed texture in ARB, or more specifically, the transition from shear to rolling texture being not complete. As shown in Fig.5 and 6, the change from large shear strain at the surface to low shear strain at the centre is the reason for the textural transition [15, 21], so this transition is located only at the centre [2, 15, 21]. Another example to show the effect of shear strain change to texture is that the maximum intensity of shear texture was observed at the subsurface [1], but not the exact surface that experienced shear strain reversal after passing the neutral point. Moreover, a close inspection of all four reports [22-25] concerning ARB processed single crystals revealed another type of texture transition that has not been reported, as shown in Fig.10. The positions that texture reversal (rotating toward the initial orientation, but not transitioning from shear to rolling texture) occurred were marked by 'Rev' in Fig.10, and where these positions came from in the last cycle were also marked by brackets. For Cube $\{0\ 0\ 1\}<1\ 0\ 0>$ in Fig.10a, the thickness of the layer marked by brackets in the second cycle is supposed to be half in the third cycle due to another 50% reduction. However, the fraction of the initial orientation Cube increased slightly at the regions marked by 'Rev'. It is the same for S $\{2\ 1\ 3\}<3\ 6\ 4>$ in Fig.10b and clearer in Fig.10c and d. It is worth noting that the ARB in Refs. [22-24] was conducted under lubricated conditions, in which the variation of shear strain through the thickness is very low [1, 5] and a uniform distribution of texture was expected [1, 26]. The positions of texture

reversal in Fig.10 are not only limited to the centre. Therefore, the change from shear deformation to compression is not the reason for this kind of texture reversal. It is known that, compared to polycrystals, single crystals are highly anisotropic and sensitive to deformation [27]. Though the shear strain is low in lubricated ARB, the state of shear strain would be changed after the material moves to a new position, and the direction of crystal rotation would be altered accordingly [28]. It seems that the partial shear strain reversal (Fig.3c and d) is the reason for the partial texture reversal in these single crystals. The shear strain evolved in single crystals is definitely different from that in the current simulation (Fig.3), where isotropic strain hardening was assumed. However, partial shear strain reversal due to cutting-stacking is believed to occur in single crystals. Texture modelling by crystal plasticity models, especially crystal plasticity FEM, is needed to quantify the correlation between strain evolution and texture transition in ARB.

Fig.10. EBSD images of texture components in ARB processed single crystals with orientations of (a) Cube $\{1\ 0\ 0\} \langle 0\ 0\ 1 \rangle$ [22], (b) S $(2\ 1\ 3)[3\ 6\ 4]$ [24], (c) Cu $(1\ 1\ 2)[1\ 1\ 1]$ [23], (d) $(15\ 12\ 5)[9\ 10\ 3]$ [25].

Microstructure refinement and hardness are generally related to the magnitude of plastic strain [5] measured by ϵ_{eq} . In the first cycle, the value of ϵ_{eq} at surface (3.2) is about 4 times that at the centre (0.8) (Fig.3a), which corresponds the experimental observation that the fraction of high-angle grain boundaries at the surface of 1-ARB IF steel is about 4 times that at the centre [1]. Also, this microstructural parameter at the surface of 1-ARB under dry rolling conditions is about 5 times of that after lubrication [1], where lubricated ARB generated a homogeneous microstructure [1, 26]. This shows that the large shear strain at the surface due to dry rolling conditions enhanced microstructure refinement. After multiple cycles, the highly strained surfaces were transferred into the inner regions (Fig.3), and a heterogeneous distribution of strain and hardness was observed. This transfer is an important reason for the fast microstructural refinement in ARB. Besides the magnitude of

strain [5], deformation history also influences microstructural evolution [1, 9]. Compared to conventional ARB, the microstructure refinement was enhanced in cross-ARB [29, 30], in which the sheet was rotated about the ND by 90° between cycles. However, the effect of shear strain reversal (Fig.3c and d) and change from shear deformation to compression upon microstructure in ARB has not been well understood, but it is relatively clear in another SPD process, equal-channel angular pressing (ECAP). It has been demonstrated by Gholinia et al [31] that route A, producing a monotonic shear, is the most effective deformation model in ECAP, while route C, causing reversed shear deformation, is the least effective. In ECAP, the direction of shear deformation is almost completely reversed in route C. However, in ARB, the shear strain reversal is partial (Fig.3b and c) and also combined with a compression deformation, which makes it difficult to quantify the correlation between microstructure and strain. Though grain refinement by unidirectional strain is higher than that by reversal strain, reversal strain also refines microstructure, not coarsens it. This can be seen from the fact that the maximum hardness locates at the exact surface (Fig.7) though it experiences shear strain reversal. Hardness became nearly constant through the thickness at large strain, which means a further introduction of strain would not refine the microstructure anymore. The steady-state of microstructure is a dynamic balance of the generation and annihilation of dislocations and/or grain boundaries [32, 33].

5.2. Plastic instability in composites

Fig.11. Evolution of (a) σ_{XX} , and (b) ε_{eq} in the element D and E in the AA1050 and AA6061 layers (Fig.9), respectively.

Plastic instability in the harder material has been widely observed in ARB processed composites with dissimilar metals [34-37] that have different initial strengths and strain

hardening rates. Due to these differences, the stress imbalance at interfaces evolves (Fig.9d), where the imbalance along the RD is more important [38], since it is the main direction of material flow. Here, the evolution of σ_{xx} and ε_{eq} at a pair of elements, element D in AA1050 layer (Fig.9f) and E in AA6061 (Fig.9e) in the centre region, was traced and is plotted in Fig.11. The σ_{xx} of element D is negative in almost the whole rolling period, which means AA1050 experienced compression deformation along the RD. Though σ_{xx} of element E changed from positive in 'Region1' to negative in 'Region2' and then back to positive again in 'Region3', σ_{xx} in AA6061 was always smaller than in AA1050. Thus, relative to each other, AA1050 was compressed along the RD, while AA6061 experienced tension, because the faster material flow of AA1050 was hindered by the bonded AA6061 with the lower flow rate. The difference in flow rate is due to their difference in strength, where the reductions were ~49% and ~51% for AA6061 and AA1050 in the first cycle, respectively. The different material flow induced a sharp change in strain at the bonded interface at the centre (Fig.4) even for uniform deformation in the 1-3 cycles. The difference in strength increased with increasing strain and after reached a threshold, AA6061 first started to neck (Fig.9e), since it underwent tensile deformation. To accommodate the strain incompatibility, the softer material, AA1050, protruded and filled in the necked regions of AA6061 (Fig.9e). This is consistent with the prediction of the tensile test of 1-ARB AA1050/AA6061 that AA1050 moved toward AA6061 (marked by the white arrow in Fig.8e). The occurrence of plastic instability is also associated with the change of stress state, where the sudden change of strain at about 2.21s (Fig.11b) is accompanied by the stress σ_{xx} starting to decrease (Fig.11a). The decreased σ_{xx} means the compression deformation along the RD was reduced, which was beneficial to trigger the occurrence of tensile necking. In this study, 4 cycles were required for the difference in strength to reach the threshold. In contrast, a small strain is enough for greatly dissimilar metals, e.g., 1 cycle for Al/Ti/Mg [34] and Cu/Ti [35], while it is relatively large for metals of high similarities, e.g., 5 cycle for AA2219/AA5086 [37]. Besides, the hindered material flow in AA6061 decreased its work-hardening and accordingly, delayed the appearance of wavy interfaces. The plastic instability was also the reason for the shear fracture of 3-ARB AA1050/AA6061 (Fig.8c). After 3-ARB, the difference in strength was close to the threshold, and plastic instability occurred when an extra tensile strain was applied. In contrast, their strengths were similar before 3-ARB, so a uniform deformation at interfaces of 1-3 cycles (Fig.9) and necking of the 1-ARB AA1050/AA6061 (Fig.8e) were observed.

6. Conclusions

1. An FEM model, following the real ARB process, was developed and the simulations were conducted for 5 cycles. Mapping solution was used to transfer the deformation solution from the deformed mesh to the new mesh between cycles. The accurate predictions that

have been validated by the experimental observations suggest that the assumptions adopted in the FEM model are valid.

2. The strain developed in a single cycle is non-uniform through the thickness of monotonic material composites and it varies greatly after multiple cycles due to the repeated rolling and cutting-stacking. In AA1050/AA6061, the deformation in the AA1050 and AA6061 layers is, relative to each other, enhanced and reduced respectively, creating sharp changes of strains at the interfaces.

3. Shear texture developed at the surface and rolling texture at the centre due to the shear and compression deformation, respectively. The change from shear deformation to compression caused textural transition. It seems that the partial shear strain reversal is the reason for the partial texture reversal observed in ARB processed single crystals.

4. Hardness distributes in the same pattern as equivalent plastic strain at first, but it becomes uniform through the thickness after microstructural saturation. Compared to its counterpart in AA6061/AA6061, the hardness in the AA6061 layers of AA1050/AA6061 increased slowly due to its hindered deformation.

5. Tensile tests of ARB processed composites were simulated for the first time and successfully predicted the experimentally measured strain-stress curves and tensile fractures. The strength of AA1050/AA6061 was found to be between the two monotonic material composites.

6. Plastic instability occurred in AA1050/AA6061 during the fourth cycle due to the increasing difference in strength between its two components. AA6061 first started to neck, since it experienced tension along the RD, and AA1050 filled in the necked regions.

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