



Tailoring AZ91 laminates for combination of strength and ductility

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Abstract: Three types of laminates were designed by alternately stacking AZ91 extruded sheets in different states for extrusion to improve the mechanical properties. The tensile tests revealed that the combination of solid-solution-treated sheets with the aging-treated sheets achieved high tensile strength and ductility, i.e., ultimate tensile strength of ~386 MPa and elongation of ~19.8%, respectively. Electron backscatter diffraction (EBSD) and TEM results indicated that the aging-treated layers with more nano-sized precipitates and small grain size provided high strength and reasonable ductility, while the solid-solution-treated layers with low dislocation density facilitated strain hardening. Also, the strong interface bonding between the successive layers played an important role in the enhanced ductility.

Key words: AZ91 Mg alloy; solid solution treatment; aging; extrusion; mechanical properties

1 Introduction

Mg alloys, as lightweight metals, have a wide range of applications in aerospace, automobile, digital 3C and other fields owing to their low density, high specific strength, good recyclability and damping performance [1–3]. Among the presently available Mg alloys, AZ91 has received considerable attention due to its relatively high strength, good corrosion resistance and low cost [4,5]. However, the hexagonal close-packed (HCP) crystal structure of AZ91 makes its room-temperature plasticity unsatisfactory compared with the face-centered cubic (FCC) metals such as aluminum and copper. For instance, the tensile strength of AZ91 Mg alloy can reach more than 400 MPa, while the elongation is generally less than 5% [6]. Accordingly, the imbalance between strength and ductility creates a huge obstacle to its

application.

In recent years, some strategies on the superior combination of strength and ductility in Mg alloys were proposed, including laminated structure [7], bimodal structure [8], gradient structure [9], the introduction of nano-precipitation [10], etc. For instance, ZHA et al [8] achieved a bimodal structure in AZ91 via controlling deformation temperature during the hard-plate-rolling (HPR) process. The tensile test result indicated that the AZ91 processed at 350 °C had a superior strength–ductility balance. SUN et al [11] found that the combination of high-temperature ECAP and short-time aging improved the strength and ductility of AZ91. The main reason was that the combined processing induced high-density precipitates, which facilitated the back stress strengthening and hardening.

Among the above strategies, laminated structure is likely to be the most effective way to achieve large-scale industrial preparation since it

can be produced via simple processing technologies, such as rolling and extrusion. Till now, different laminated structural materials, including Cu/Al alloys [12], Al/Al alloys [13], Mg/Al alloys [14], etc, have been designed to synthesize the advantages of each component, achieving superior performance. However, it should be emphasized that these metallic laminated composites are commonly fabricated via stacking dissimilar metals (namely different compositions) in alternate layers. So, the question arises: whether the laminates stacked by the constituent layers with the same composition and different states can achieve the desired performance? To the best of our knowledge, it has not been studied before. Herein, we prepared three types of AZ91 laminates by means of extrusion, in which the aging-treated sheets were stacked alternatively with the other three states (as-extruded state, solid-solution-treated state and aging-treated state) respectively. The microstructure evolution and mechanical properties of three kinds of AZ91 laminates were investigated and compared in detail. And, the corresponding mechanisms of strength and toughness and fracture feature were also discussed.

2 Experimental

2.1 Materials and preparation process

In this work, commercially extruded AZ91 (Mg–9Al–1Zn, wt.%) sheets (3 mm in thickness) were used. The as-extruded (AE) samples were firstly subjected to solid-solution treatment (SST) at 683 K for 12 h and then quenched in water immediately. Subsequently, the SST samples were aged at 453 K for 10 h and recorded as AG. The AE, SST and AG samples were cut into the strips with the dimensions of 35 mm (ED) $\times$ 15 mm (TD) $\times$ 3 mm (ND) (where ED, TD and ND referred to the extrusion, transverse and normal directions, respectively). After surface treatment and degreasing, the AE, SST and AG strips were separately stacked with AG strips to form three different laminates, i.e., AE/AG, AG/AG and SST/AG alternating composite laminates. Once the temperature of the square extrusion container reached 523 K, the three stacked laminates were respectively placed in it and held for 5 min, and then subjected to extrusion. The schematic diagram of the extrusion is shown in Fig. 1, and a similar

scenario was described in our previous study [13]. The ram speed and extrusion ratio were ~ 6 mm/s and ~ 12.8 , respectively. During the extrusion processing, the true strain can be calculated using the equation: $\varepsilon = \ln \lambda$, where λ is the extrusion ratio [15]. A strain of ~ 2.5 was produced after extrusion. The final thickness of the extruded sheet was ~ 1.5 mm. The three designed laminates were marked as “AE+AG”, “AG+AG”, and “SST+AG”, respectively.

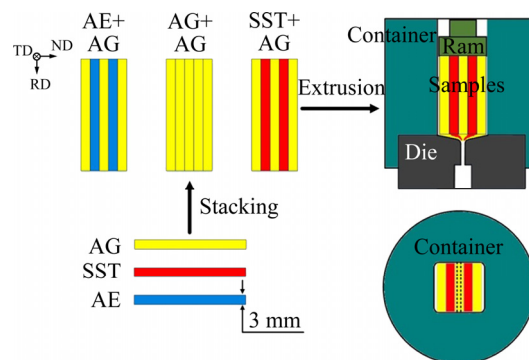


Fig. 1 Schematic diagram showing preparation of three types of laminates

2.2 Microstructure and mechanical properties characterization

The microstructure was characterized by the optical microscope (OM), field emission scanning electron microscope (SEM, JOEL JSM 7800F) and transmission electron microscope (TEM, ThermoFisher Scientific Talos F200S) operated at 200 kV. The AZtecHKL software was applied to EBSD measurement. The EBSD step size for deformed samples was $0.5 \mu\text{m}$, while that of SST and AG samples was $0.8 \mu\text{m}$. The TEM samples were cut from the longitudinal section of extruded sheets, mechanically polished to $\sim 50 \mu\text{m}$ in thickness, and then Ar ion beam thinning was performed with Gatan 695 ion polishing system at -70°C .

Vickers microhardness (HV) tests were conducted along the ND direction on the ED–ND plane using a load of 100 g and dwell time of 10 s. Uniaxial tensile tests along ED were conducted on a CMT6305-300KN testing system at room temperature. The strain rate was $1 \times 10^{-3} \text{ s}^{-1}$. All the samples were dog-shaped with a cross-section of $4 \text{ mm} \times 1.5 \text{ mm}$ and a gauge length of 24 mm. During tensile tests, a mechanical extensometer with a gauge length of 12.5 mm was used.

3 Results

3.1 Microstructure characteristics before combined extrusion

Prior to combined extrusion, the SEM micrograph and corresponding microhardness distribution of the samples in three different states (AE, SST and AG) are shown in Fig. 2. In Fig. 2(a), it was seen that the AE sample contained fine precipitates which were located in the grain interior and at grain boundaries, and these precipitates streamlined along the ED direction. The spectrum from Point 1 showed that the precipitates were β -Mg₁₇Al₁₂ (Fig. 2(d)). After SST, the number of β -phase particles decreased clearly, and the streamlined distribution characteristics were still retained, as shown in Fig. 2(b). For the AG sample, the high-density micro-sized precipitates were observed obviously, especially the dense lamellar precipitates which almost covered the whole grains in some local regions, marked by the white arrows in Fig. 2(c). The lamellar precipitates were related to the discontinuous precipitation at the subsequent

aging. A similar finding was reported in the previous literature [10]. Moreover, note that the size of precipitates grew up significantly. The microhardness values of the three samples as a function of distance are plotted in Fig. 2(e). One can see that despite the slight fluctuation, the hardness values of three samples were different, which further indicated that there was a microstructure discrepancy among the three samples.

To further reflect the microstructure discrepancy among the three samples, EBSD analysis was carried out. Figure 3 shows the inverse pole figures (IPF), (0001) pole figures and corresponding misorientation distribution of three samples. Obviously, from the IPF point of view (Figs. 3(a, d, g)), the grain size of the three samples was different. The measured average grain sizes of AE, SST and AG samples were ~ 6.6 , ~ 13.1 and ~ 11.9 μm , respectively. The obtained result showed that significant grain growth occurred after SST and aging. The corresponding (0001) pole figures (Figs. 3(b, e, h)) revealed that three samples exhibited a strong basal texture. Note that the SST

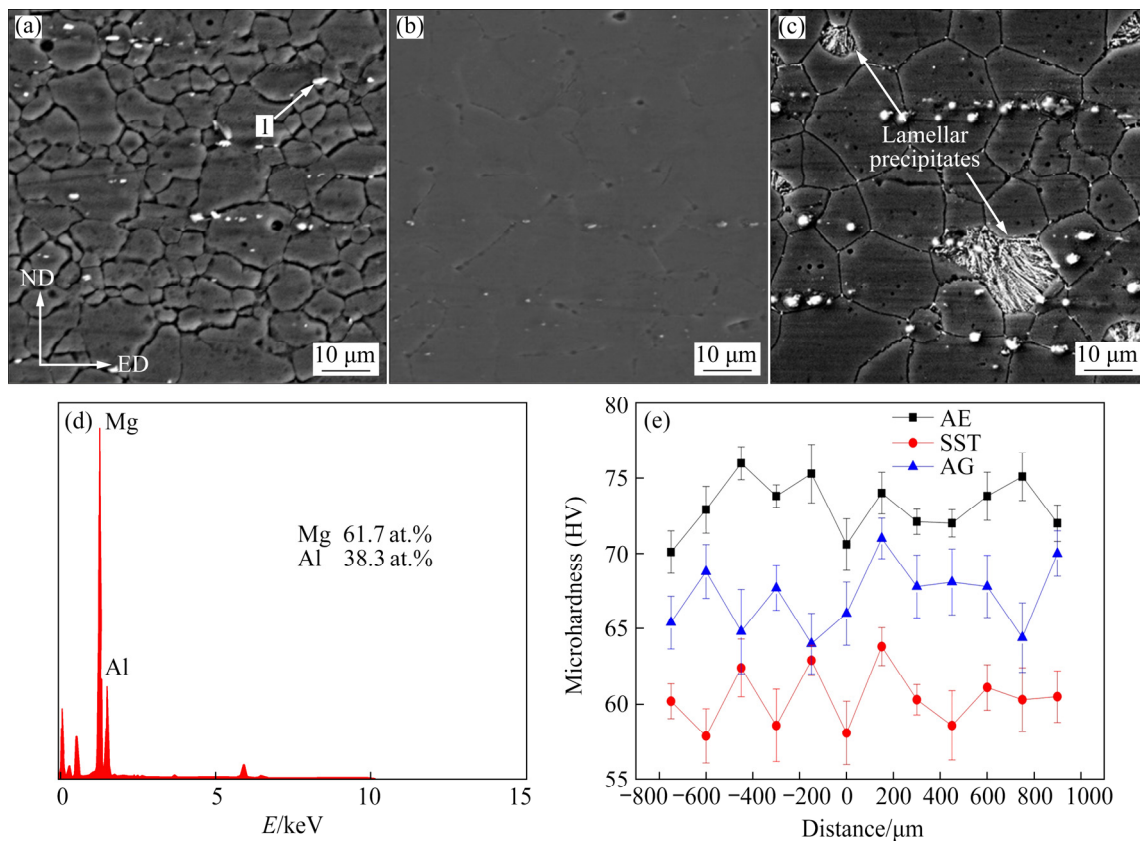


Fig. 2 SEM micrographs of AE (a), SST (b) and AG (c) samples, corresponding spectrum from Point 1 (d) and microhardness distribution of three samples (e)

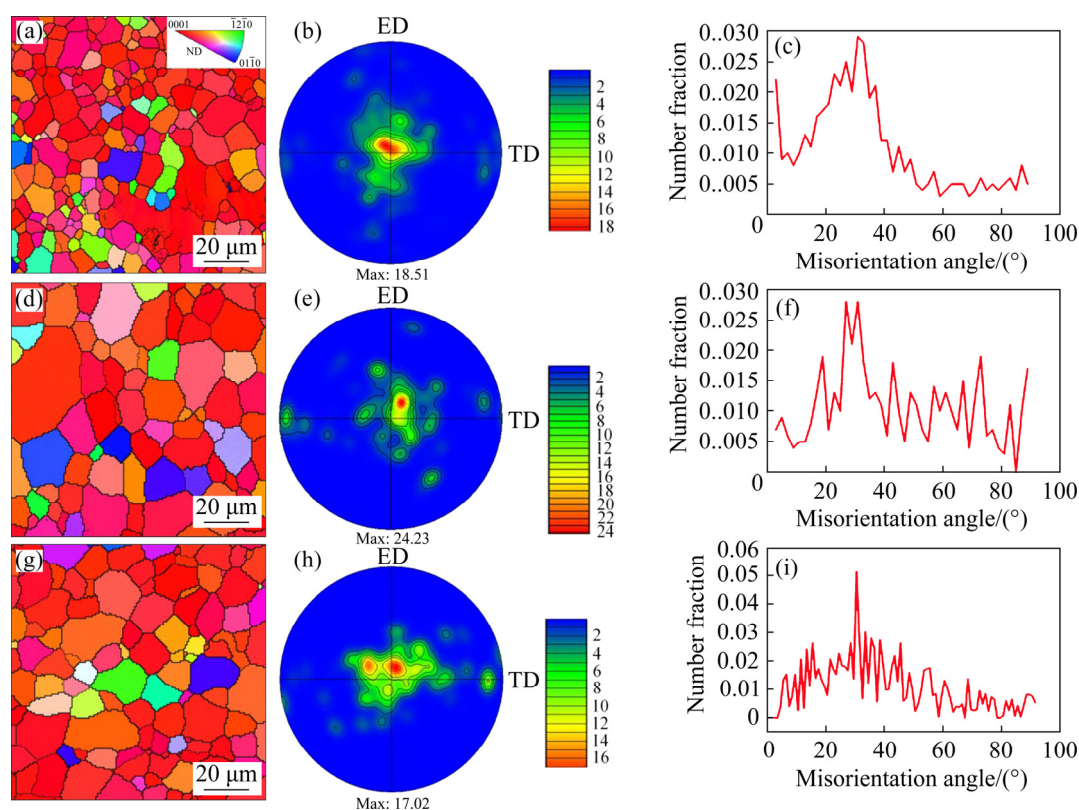


Fig. 3 EBSD inverse pole figures (IPF) maps (a, d, g), (0001) pole figures (b, e, h) and misorientation distribution (c, f, i) of AE (a, b, c), SST (d, e, f) and AG (g, h, i) samples

sample presented a maximum texture intensity. This was caused by the existence of large grains and a limited number of grains. The comparison of misorientation distribution (Figs. 3(c, f, i)) among the three samples showed that the SST and AG samples displayed a relatively uniform misorientation distribution mainly due to the preferential nucleation and growth during recrystallization [16]. Based on the above observation, it can be inferred that the differences in the microstructure of three samples were expected to make the designed laminates produce different mechanical properties responses.

3.2 Microstructural evolution of three designed laminates

After extrusion, three designed laminates were firstly characterized by OM and SEM. The OM and SEM images of three laminates are presented in Fig. 4. It was found from the OM pictures (Figs. 4(a, b, c)) that the interfaces between layers were straight after etching. The layer thickness of laminates was relatively uniform, with a single layer thickness of ~300 μm. The magnified SEM

images revealed that the pre-existing micron-sized β -phase particles were finer due to extrusion deformation, especially for AG layers where the pre-existing lamellar precipitates vanished. This was possibly because some lamellar precipitates were dissolved during the heat preservation process before extrusion deformation. Careful observation showed that the AG layers for the three laminates contained relatively more precipitates compared to AE and SST layers. The difference in the number of micron precipitates undoubtedly gave rise to the intra-granular heterogeneity, which was expected to have an effect on the mechanical responses [17].

Figure 5 displays the HAADF-STEM images of individual layers and corresponding EDS mapping. Apparently, during extrusion, many nano-sized β phase particles with the sizes of 10–50 nm occurred, in which the AG layer encompassed more nano-sized precipitates, and the number and size of nano precipitates were much larger than those of AE layer and SST layer. By a careful inspection, it was found that there were not only spherical nano precipitates but also block nano precipitates in the AG layer. The area fraction of

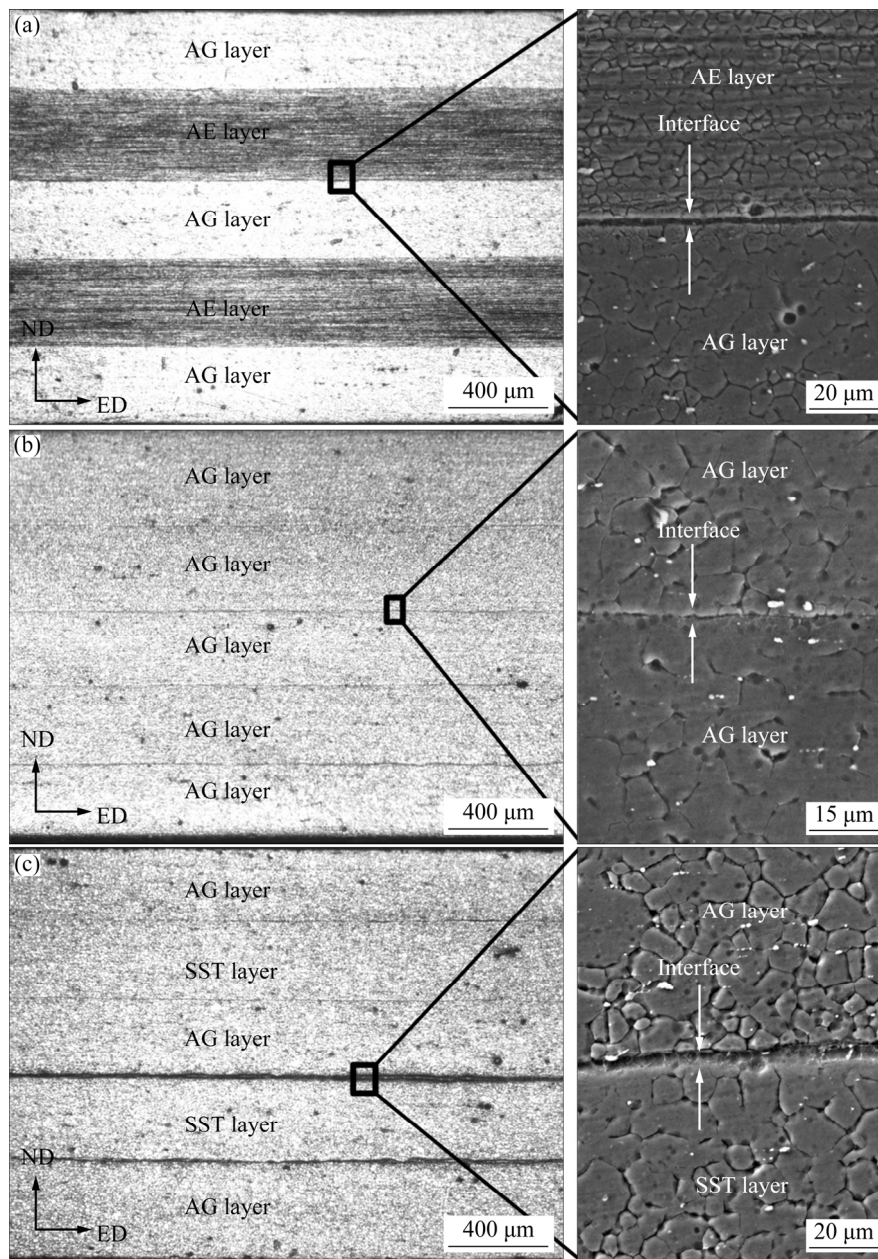


Fig. 4 Optical micrographs and corresponding SEM images of three designed laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

these nano-sized precipitates was counted from the STEM images, as shown in Fig. 6. It was found that the nano-precipitate in the AG layer was the most, accounting for 3.1%. On the contrary, 0.7% and 0.6% of nano-precipitates were responsible for the AE and SST layers, respectively, which had little effect on the strengthening of the alloy.

EBSD analysis is applied to investigating the microstructure of three designed laminates. The grain morphology and grain size distribution of different layers are shown in Fig. 7. One can see that the microstructure of all laminates was

composed of approximately equiaxed grains due to the occurrence of dynamic recrystallization (DRX). The grain size of individual layers was counted from EBSD data. The average grain size (AGS) of AG layers for the three laminates was ~ 5.8 , ~ 5.6 and ~ 6.1 μm , respectively, which had no significant difference due to the same processed condition. In contrast, the AGS of AE layer (~ 4.4 μm) and SST layer (~ 6.6 μm) showed a slight difference. The results suggested that the AGS of three laminates was refined after extrusion compared with the corresponding counterparts before deformation.

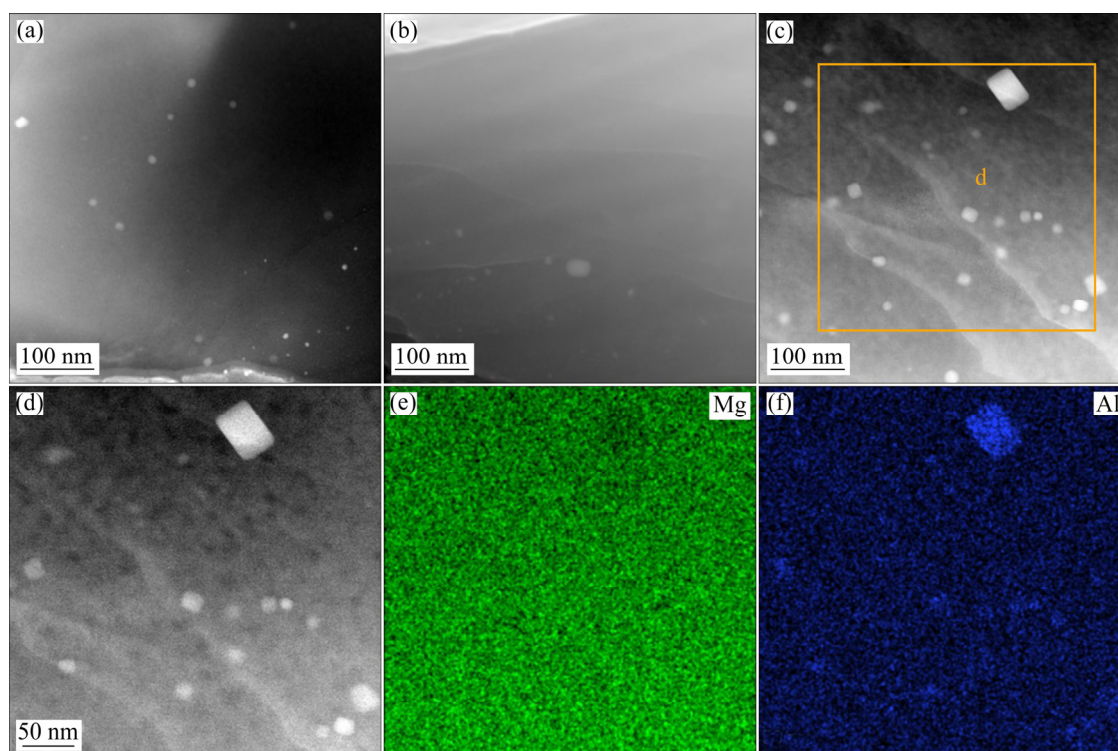


Fig. 5 High angle annular dark field-scanning transmission electron microscopy (HAADF-STEM) images of AE layer (a), SST layer (b) and AG layer and corresponding energy dispersive spectrometer (EDS) mapping (c–f)

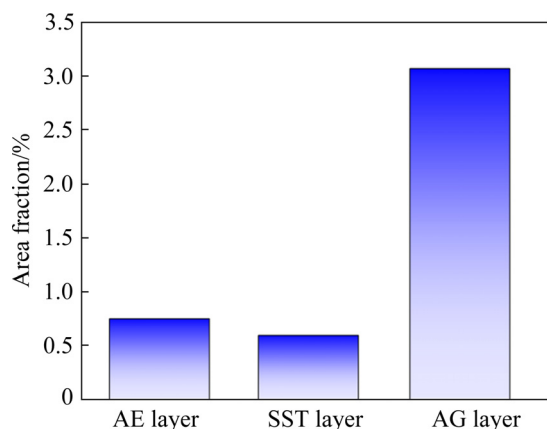


Fig. 6 Area fractions of nano precipitates in individual layers

The thin red lines in band contrast maps (Fig. 7) represent low angle grain boundaries (LAGBs, microstructure orientation angle 2° – 15°), while the thick black lines indicate high angle grain boundaries (HAGBs, microstructure orientation angle $>15^{\circ}$). The microstructure orientation angle distribution of individual layers in the three laminates is plotted in Fig. 8. Obviously, the microstructure orientation angle distribution in the range of 15° – 60° was prominent for all layers. The fraction of HAGB (f_{HAGB}) of SST and AG layers

remained at a relatively high level, indicating a higher extent of DRX. In contrast, the f_{HAGB} of the AE layer was slightly low, being responsible for 87.2%, which was mainly due to the highly accumulative deformation.

Figure 9 shows the typical (0001) pole figures of individual layers in the three laminates. It was clearly found that individual layers showed a relatively weakened texture where the c -axis of grains was slightly tilted toward the ED direction compared with the initial AE, SST and AG samples. Specifically, the AE layer of Sample AE+AG exhibited a stronger texture than the AG layer in which the maximum texture intensity was ~ 14.09 . For the SST+AG laminates, the texture type of SST and AG layers had no significant difference, but the SST layer seemed to have a higher texture intensity than AG layer, probably because of the relatively large grain size. Similarly, Sample AG+AG obtained a similar texture component, which was very common for extruded magnesium alloys.

In general, the Kernel average misorientation (KAM) maps can reflect the dislocations density in various grains [18]. Thus, the KAM maps and corresponding values are plotted in Fig. 10. As for the AE+AG laminate, it can be clearly seen that the

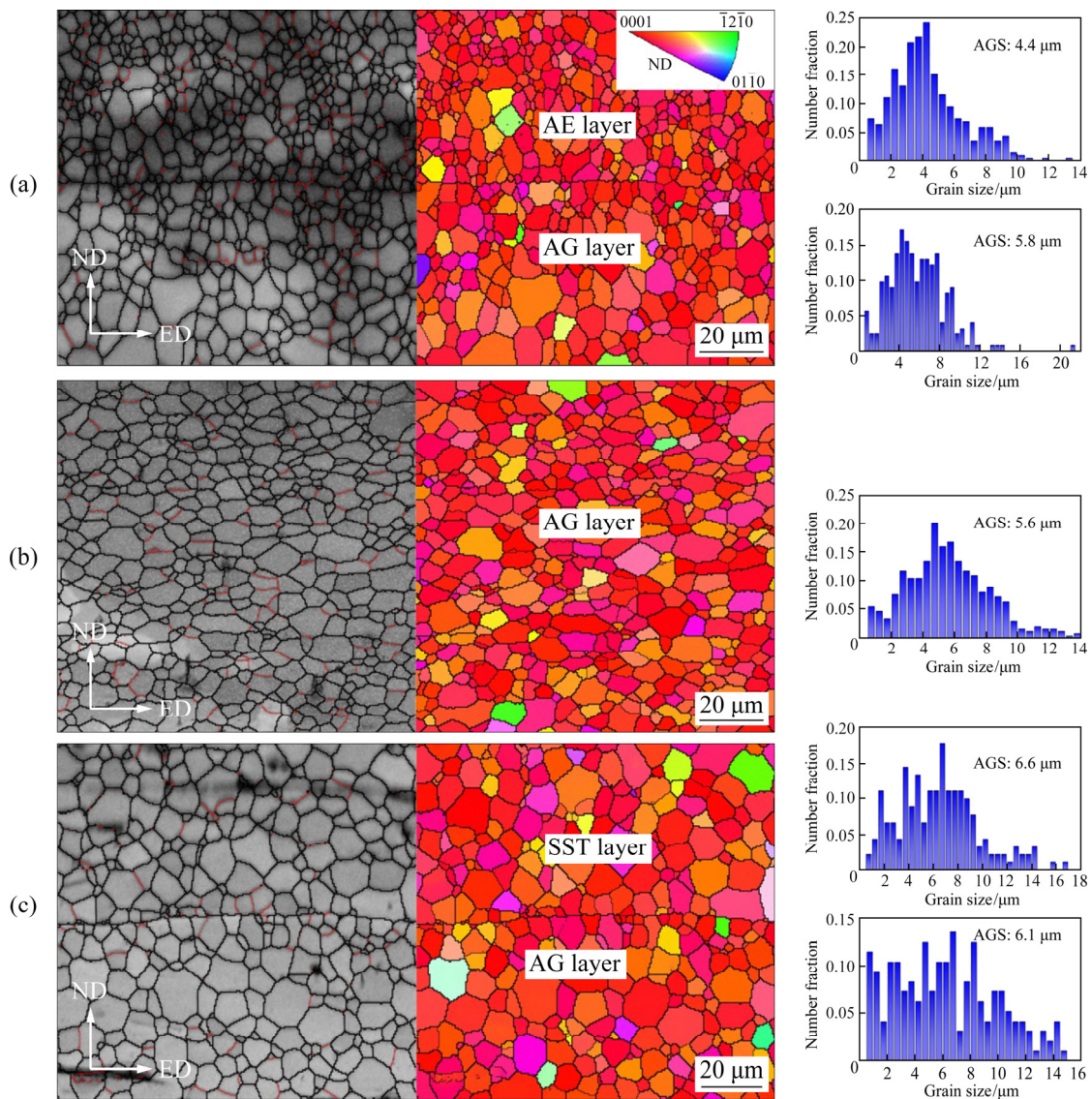


Fig. 7 Band contrast maps, IPF figures, and corresponding grain size distribution of three types of laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

KAM value was higher in the AE layer than in the AG layer, suggesting a high dislocation density. Meanwhile, the region near the interface in the AG layer contained a relatively higher dislocation density than the area far away from the interface (Fig. 10(a)), which indicated that there was high internal stress near the interface due to the mechanical incompatibility. In contrast, the average KAM values of AG+AG and SST+AG laminates showed no obvious difference and remained relatively low dislocations density in grains, implying the occurrence of well-developed DRX [19].

To quantitatively evaluate the dislocation density of individual layer, KAM method based on

local misorientation is applied to calculating the geometrically necessary dislocations (GNDs) density using the equation [20]: $\rho_{\text{GND}} = 2\theta/(ub)$, where θ indicates the average misorientation which is obtained from the defined nearest plus the second-nearest neighbor points (local misorientation above 2° is excluded), u represents the step size of EBSD and b is the magnitude of Burger vector (3.21×10^{-10} m for Mg [21]). Figure 11 displays the histogram distribution of global GNDs density of individual layers. It was seen that the distribution characteristics of GNDs in each layer were similar, in which the AE layer remained the highest GNDs density ($\sim 1.21 \times 10^{14} \text{ m}^{-2}$), while the other layers had a relatively reduced dislocation

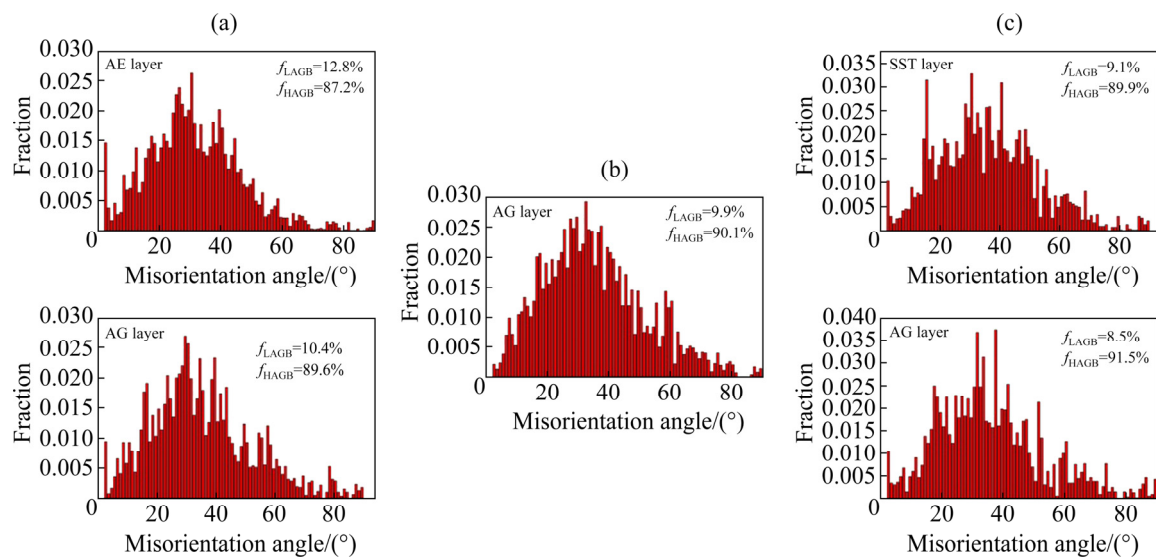


Fig. 8 Microstructure orientation angle distributions of individual layers in three designed laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

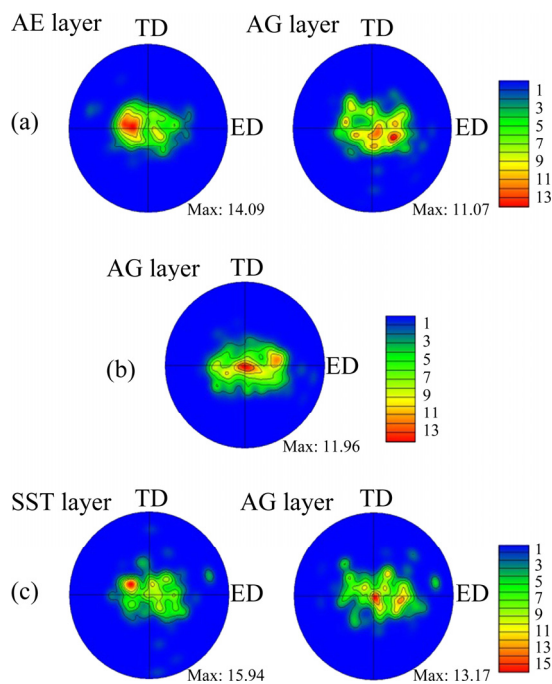


Fig. 9 (0001) pole figures showing individual layers of three kinds of laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

density which was an order of magnitude smaller than the value of AE layer. This may be due to the annihilation of some dislocations in the process of SST and aging treatment. This result was consistent with the KAM maps in Fig. 10. The crystal orientation affects the GNDs density, thereby giving rise to the discrepancy of GNDs distribution from grains to grains [22].

3.3 Mechanical properties

The typical stress–strain curves of the AE, SST and AG samples, as well as three designed laminates are plotted in Fig. 12. As shown in Fig. 12(a), it was noteworthy that the three engineered laminates presented a good combination of strength and ductility compared with the initial AE, SST and AG samples. The yield strength (YS), ultimate tensile strength (UTS) and elongation to failure (EL) of AZ91 samples in all states are listed in Table 1. One can see that the UTS and EL of three tailored laminates were obviously improved. Among the three laminates, the SST+AG laminate possessed the highest strength while simultaneously obtaining a maximum EL. The YS, UTS and EL of SST+AG laminate were 185 MPa, 386 MPa, and 19.8%, respectively. Figure 12(b) shows the strain hardening rate versus true strain curves, and the true stress–strain curves were also drawn to check the instability of the intersecting point based on the Considère criterion [23]. Clearly, the strain hardening rate of SST+AG laminate can be retained within a large strain range, thereby delaying plastic instability (necking during the tensile test). Quite the opposite, the AE sample showed a premature fracture.

Figure 13 summarizes the mechanical properties (UTS and EL) of AZ91 produced by different processing methods as well as some representative Mg–RE alloys. It was obviously seen that AZ91 processed by SPD or conventional

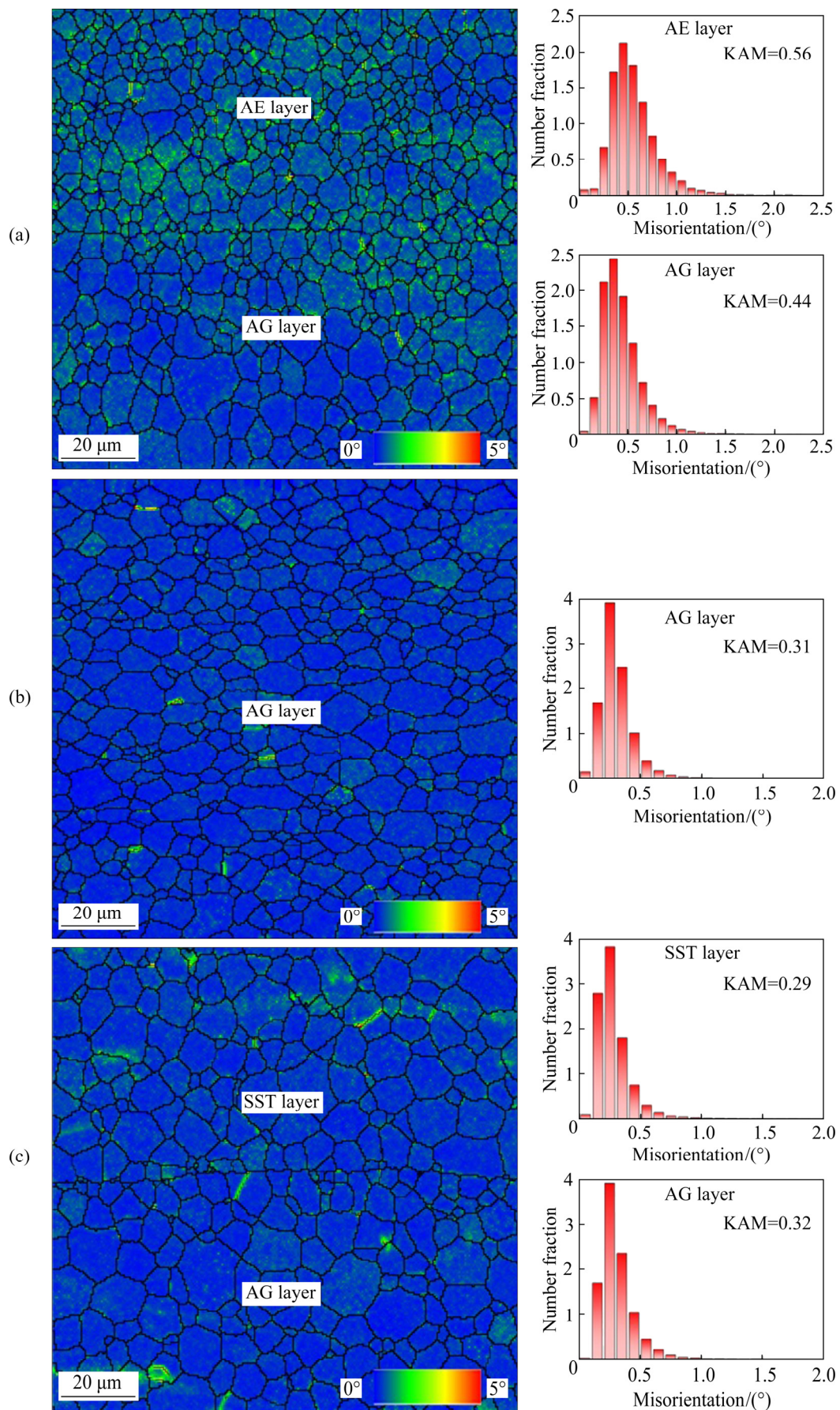


Fig. 10 Kernel average misorientation (KAM) maps of three kinds of laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

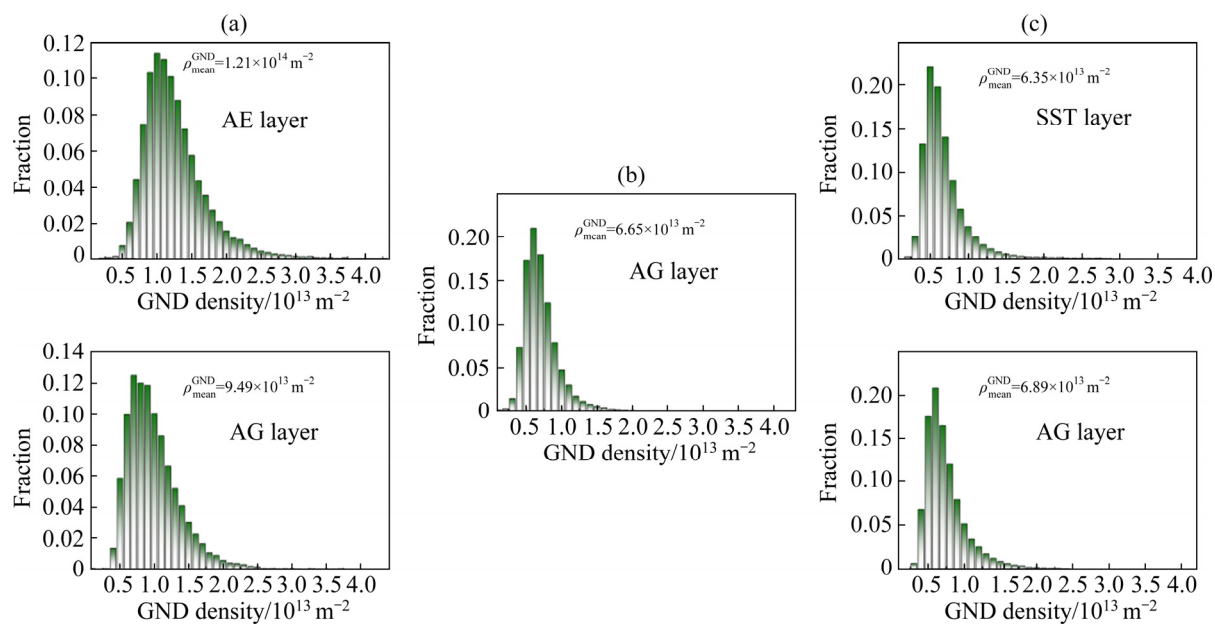


Fig. 11 Global GNDs distribution maps (acquired from EBSD orientation data) showing individual layers of three kinds of laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

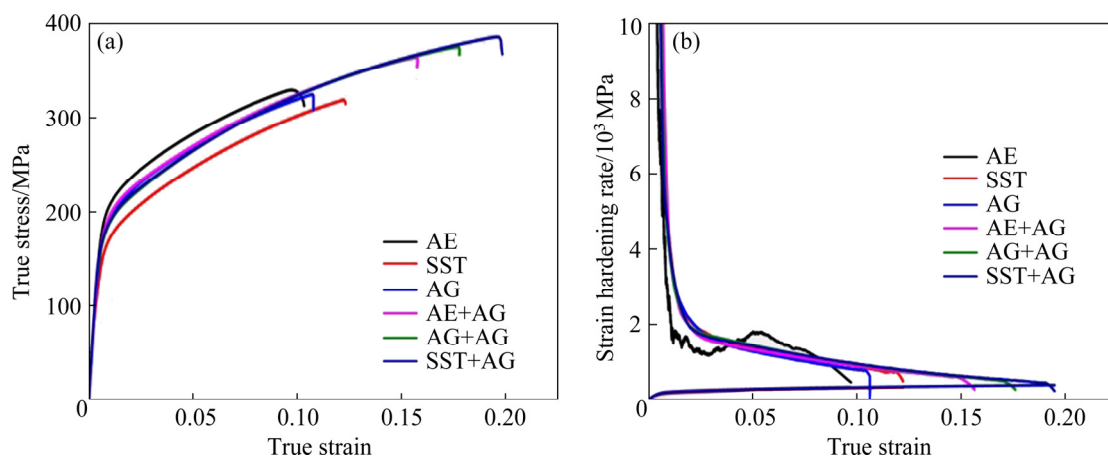


Fig. 12 Tensile true stress (σ)–strain (ε) curves of AZ91 Mg alloy samples (a) and strain hardening rate ($d\sigma/d\varepsilon$) curves versus true strain (b)

Table 1 Variations of YS, UTS, and EL of AZ91 Mg alloy samples

Sample	YS/MPa	UTS/MPa	EL/%
AE	200±3.6	329±7.8	10.3±0.5
SST	166±8.2	316±6.0	12.2±0.8
AG	188±5.0	324±4.5	10.7±1.4
AE+AG	189±5.5	364±7.5	15.7±1.2
AG+AG	183±5.2	375±7.3	17.7±1.0
SST+AG	185±6.5	386±8.2	19.8±1.3

deformation technologies frequently revealed a strength–ductility trade-off. On the contrary, the current strategy achieved an excellent strength–

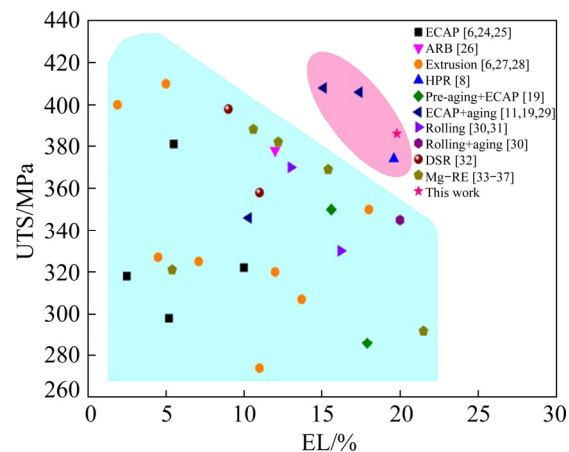


Fig. 13 Comparison of mechanical properties of AZ91 Mg alloy processed by current approach and other technologies

ductility combination, which was comparable or even superior to that reported in previous literatures [8,11,29]. Moreover, compared with those Mg–RE alloys which usually add costly rare earth elements to improve mechanical properties, the cost-efficient AZ91 prepared by the present method was quite competitive in terms of the improvement of strength and ductility.

3.4 Fracture morphology

Figure 14 provides the SEM fracture images of three laminates. As seen in Fig. 14(a), the bonding interfaces of AE+AG laminate were easily distinguished, as if delamination occurred between the interfaces. In contrast, most of the bonding interfaces of AG+AG and SST+AG laminates were invisible, as shown in Figs. 14(b, c), suggesting a relatively good interface bonding quality. It can be noted from the magnified fracture images that there

were many dimples in each layer of three laminates, implying a ductile fracture feature. A closer look revealed that the diameter of dimples was approximately the same as the size of the grains, indicating that the fracture mode was an inter-granular fracture. Also, it was noted that there were some micron-sized particles in the center of dimples, as shown by the white arrows.

To further reveal the fracture characteristics of three laminates, the longitudinal section of the fracture is selected for observation, as shown in Fig. 15. It was apparent that the fracture feature of three laminates was discrepant. For the AE+AG laminate (Fig. 15(a)), a delamination phenomenon of about 100 μm in length occurred at the interfaces, and the crack propagation direction at the interfaces presented an angle of $\sim 40.6^\circ$ (or 48.9°) to the ED direction. On the contrary, this delamination phenomenon was not observed in the AG+AG and

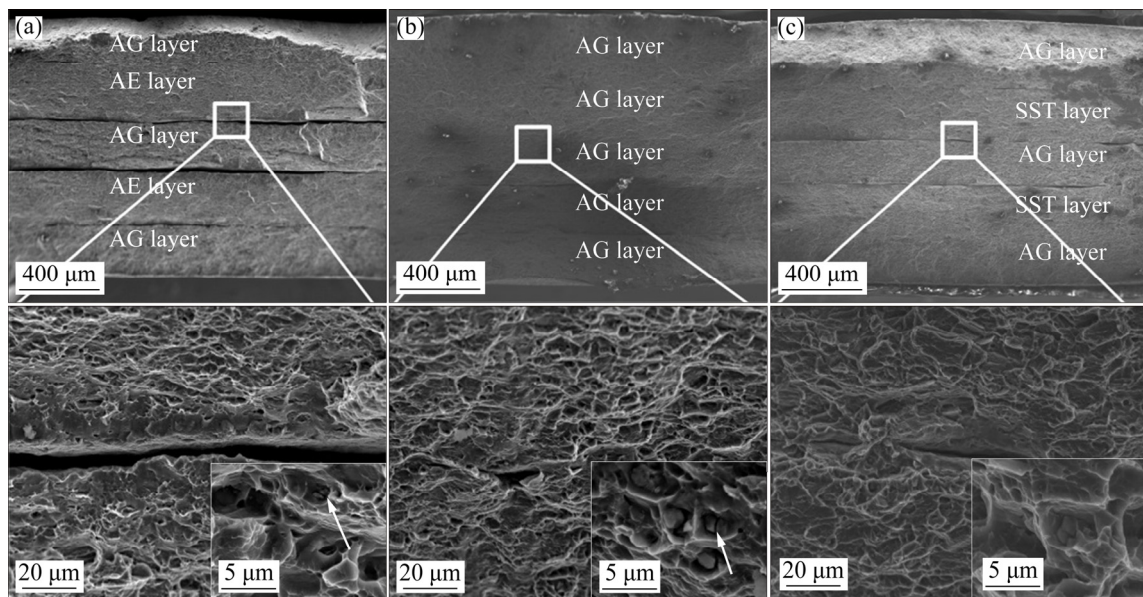


Fig. 14 Tensile fracture micrographs and corresponding magnified images of three kinds of laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

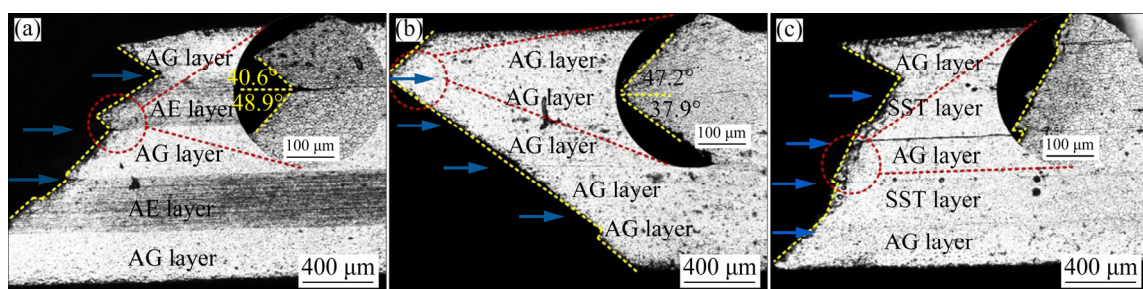


Fig. 15 OM images showing tensile fracture surface of three kinds of laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG (The insets show the morphology near the interface; The yellow dot lines refer to crack propagation, and the blue arrows indicate the bonding interfaces between layers)

SST+AG laminates. It was suggested that the interfacial bonding of AE+AG laminates was not as good as that of the last two laminates. Generally, good interfacial bonding indicated a decent bonding strength, which had a positive effect on mechanical properties.

4 Discussion

4.1 Mechanism of strengthening and toughening

The variations in mechanical properties are usually related to the microstructure. Thus, an important issue to be solved in the current work is what gives rise to such different mechanical responses among the three engineered laminates. In general, the factors affecting the mechanical properties of Mg alloys primarily contain grain size, texture, precipitation phases as well as dislocation density. Compared with the initial AE, SST, and AG samples, the AGS of three laminates after extrusion decreases significantly. Thus, grain refinement contributes to the improvement of the strength of three laminates. Moreover, as for the three laminates, although the total average grain size of AE+AG laminate is slightly lower than that of AG+AG and SST+AG laminates, such a subtle difference does not lead to a significant change in mechanical properties, which can be reflected from Fig. 12(a).

In terms of the texture, the individual layers of three laminates have a relatively weak texture component with a *c*-axis of grains inclining toward ED direction compared with the initial AE, SST and AG samples. Based on the EBSD data, the schmid factor (SF) distribution for basal slip and corresponding (0001) pole figures of individual layers are provided in Fig. 16. It was found that the basal plane texture mainly distributed in the middle red areas of (0001) pole figures. This indicated a low SF value. To specifically illustrate the effect of SF, the mean SF values of individual layers were calculated. The results revealed that the mean SF values of three laminates were about 0.25, implying that the influence of texture on the mechanical response of three laminates is almost the same.

In addition, SEM and STEM observations show that the number of micron- and nano-sized β phase particles in individual layers of three laminates presents a significant difference, in which the AE and SST layers contain a small number of

particles, while the AG layers retain relatively more precipitates. It is expected that the inhomogeneous precipitations distribution in the three laminates has an effect on the mechanical properties, as the microstructural heterogeneity may induce back stress strengthening and hardening caused by the accumulation of GNDs near the interface between layers, which can render additional strength and ductility [38,39]. On the other hand, a substantial amount of nano-sized precipitates inside the grains of AG layers interact with mobile dislocations, and impede the movement of dislocations within the grains, thus bringing about an improved strain hardening. KIM et al [40] used two methods, i.e., conventional hot extrusion + equal channel angular pressing (ECAP), and ECAP + low-temperature aging, to introduce nano-sized β phase particles to grains interior of ultrafine-grained AZ91 Mg alloy. The samples processed by the two methods exhibited very high strength and decent strain hardening behavior.

Also, the dislocation density is responsible for the difference in mechanical properties of three laminates. Combined with the KAM result in Fig. 10 and GNDs distribution in Fig. 11, it is apparent to see that the AE layer contains much more dislocation density than the SST and AG layers. It is reported that the high dislocation density serves as barriers, which make them difficult to move [19]. In other words, it gives rise to a low strain hardening capacity, and thus a low tensile ductility. In contrast, the AG and SST layers with the reduced dislocation density have an improved ability to accommodate dislocation during plastic deformation, thereby enhancing the strain hardening ability of laminates.

According to the above discussion, it can be concluded that because the initial state of samples is different, the microstructures of the constituent layers show differences when subjected to the same deformation process. Specifically, the AE layer has a small grain size and high dislocation density, while the SST layer has a relatively large grain size and low dislocation density, indicating that the ability to accommodate dislocations for the SST layer is higher than that of the AE layer during plastic deformation, contributing to the increased ductility. Moreover, the AG layers of three designed laminates have roughly similar microstructure characteristics, including the size, number and

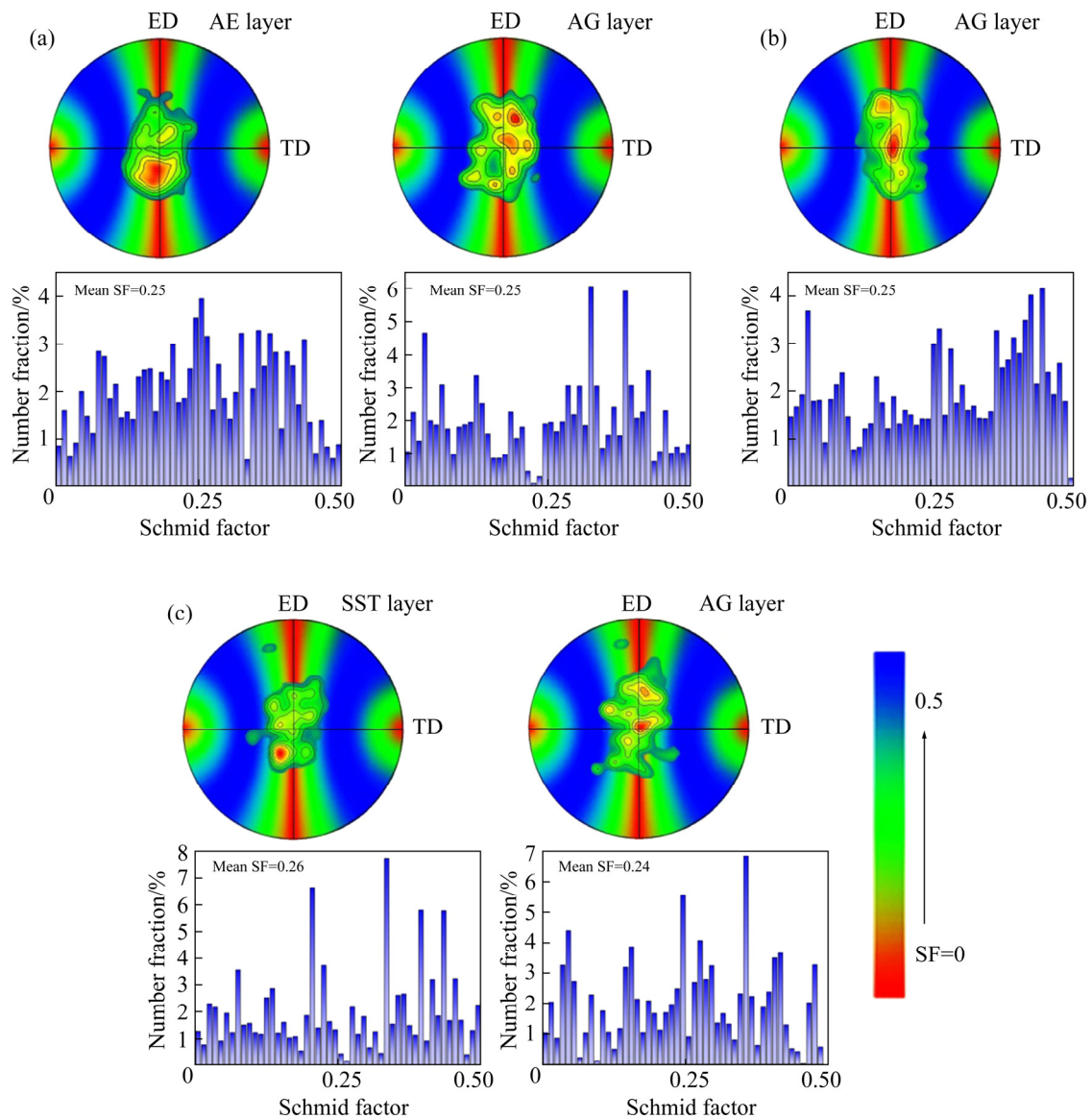


Fig. 16 Schmid factor (SF) distributions and mean SF values of individual layers in three laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG

morphology of second phase particles, the average grain size, the texture component as well as the dislocation density. This means that the contribution of this layers to the mechanical properties of three laminates is the same. Hence, the SST+AG laminate achieves a superior strength–ductility combination.

4.2 Fracture feature of laminates

According to Fig. 15, the tensile fracture surface of three laminates shows different behaviors, and thus a sketch of crack propagation in three laminates is drawn in Fig. 17 to illustrate the fracture characteristics. Here, the black lines in three laminates refer to crack propagation paths. For

the current three laminates, their deformation follows the equal strain mode based on the lamination theory during tensile deformation [41]. More specifically, in the elastic period, mutual deformation and restriction happen between the successive layers. Once entering the plastic deformation stage, cracks are formed in the stress concentration areas, such as interfaces, and impurity particles. It is clear to see from Fig. 17(a) that the delamination at the interfaces occurs after tensile deformation, which suggests that these interfaces are weak. Generally, interfaces between layers play a significant role in the mechanical properties of laminates [23]. However, when there exist impurities or oxide particles at the interfaces,

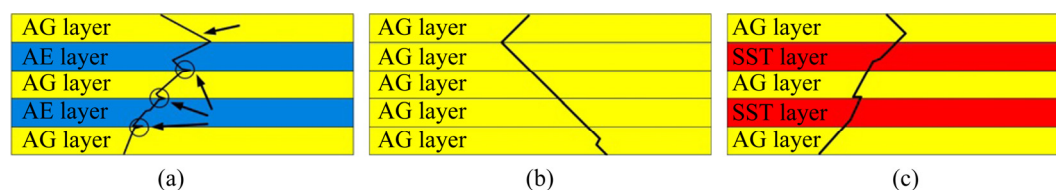


Fig. 17 Schematic illustrations of crack propagation in three laminates: (a) AE+AG; (b) AG+AG; (c) SST+AG (The black lines represent crack propagation)

interfaces may be easy to induce crack initiation, causing the nucleation and propagation of crack. In addition, there are some micron-sized precipitates in the AG layers besides the nano-sized precipitates. Once the stress concentration is formed around these large precipitates, micro-cracks may occur. Therefore, the crack sources are possibly formed at the interfaces or inside the interior of individual layers. For AG+AG laminate, despite no delamination, once the crack is formed, it passes through the constituent layers rapidly by shear. Unlike the above two laminates, the SST+AG laminate exhibits a relatively strong interface bonding. At the same time, the crack propagation path changes between the successive layers, which suggests that the SST/AG interfaces seem to play a crucial role in constraining the crack propagation. Therefore, the SST+AG laminate obtains a distinct strain hardening ability.

5 Conclusions

(1) After extrusion deformation, the average grain size of three laminates was significantly decreased. The grain refinement had a positive effect on the improvement of their strength.

(2) Although the texture characteristic of three laminates showed no obvious difference, the number of precipitated phases and dislocation density in the constituent layers presented significant differences.

(3) Among the three designed laminates, the SST+AG laminate attained a mutual improvement of strength and ductility, which was mainly related to a large number of nano-sized precipitates dispersed in the AG layers and low dislocation density in the SST layers. Additionally, the intra-granular microstructure heterogeneities (such as precipitated phases) may induce back stress strengthening and hardening.

(4) The fracture characteristics of three

laminates showed differences, in which the crack propagation in SST+AG laminate was relatively difficult, contributing to a relatively higher strain hardening capacity.

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设计 AZ91 层合板以实现强韧性结合

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摘 要: 通过交替堆叠不同状态下的 AZ91 挤压板设计三种类型的层合板, 以改善其力学性能。拉伸试验表明, 固溶处理后的板材与时效处理后的板材相结合, 可获得较高的抗拉强度和塑性, 即极限抗拉强度和伸长率分别约为 386MPa 和 19.8%。电子背散射衍射(EBSD)和透射电镜(TEM)结果表明, 时效处理层中析出的纳米颗粒较多, 晶粒尺寸较小, 具有较高的强度和合理的塑性, 而固溶处理层中位错密度较低, 有利于应变硬化。此外, 连续层间的强界面结合对延展性的提高起着重要作用。

关键词: AZ91 镁合金; 固溶处理; 时效; 挤压; 力学性能

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