Multi-directional Forging of Large-Scale Mg-9Gd-3Y-2Zn-0.5Zr Alloy Guided By 3D Processing Maps And Finite Element Analysis

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Abstract

The three-dimensional (3D) processing maps of cast Mg-9.0Gd-3.0Y-2.0Zn-0.5Zr alloy were established based on isothermal compression tests and dynamic material model (DMM). The stable and power efficient forming domains were determined by considering both the instability and power dissipation efficiency maps. Multi-directional forging (MDF) was then simulated by employing finite element (FE) analysis in the Deform-3D software, using the 3D power dissipation efficiency maps as input. The optimal forging parameters were thus obtained for a large-scale ingot with 430 mm in diameter and 440 mm in height, i.e. a forging temperature of 450 °C and forging speed of 10 mm/s. Finally, a Mg-9.0Gd-3.0Y-2.0Zn-0.5Zr cake-shaped forged part with 900 mm in diameter and 100 mm in height was produced. After T6-heat treatment, the edge and center of the forged part exhibit homogeneous microstructure and relatively consistent properties, with the tensile strength, yield strength and elongation being about 400 MPa, 320 MPa and 14.0% respectively. Using transmission electron microscopy, the main strengthening phases were revealed to be the dense nano-scale β' phases that are uniformly distributed inside the material.

1. Introduction

The rapid developments of aerospace and transportation industry bring accelerated products upgrade. Therefore, the demand for lightweight, high-performance, and large-scale load-bearing structural materials is becoming more and more urgent. Magnesium alloys have extensive application prospects in the fields of aerospace and transportation due to the advantages of low density, high specific strength and high specific modulus [1–3]. However, many magnesium alloys exhibit a poor ductility and narrow forming window due to the limited slip systems in the hexagonal close packed (hcp) crystal structure. Especially, large-scale parts are easy to crack during forging process, which limits their broad application.

The forging of magnesium alloys is traditionally carried out by single-direction and incremental multi-pass forging processes, but it would lead to inhomogeneous deformation, especially for the large-scale forged parts whose microstructures and properties vary with positions. On the other hand, the multi-directional forging (MDF) process can eliminate the casting defects, break the micron-scale secondary phases etc., thus overall homogenize the microstructures and produce consistent properties of the forged part.

Previous studies have shown that MDF process can improve the properties of Mg-Al-Zn system [4, 5], Mg-Zn-Zr system [6, 7] and Mg-RE-(Zn) system [8–10] alloys. In addition, some researchers have emphasized the efficacy of MDF process for grain refinement and texture control. For example, Li [4] et al. adopted MDF process to refine the grain size of AZ61 magnesium alloy with dimensions of 20 mm × 28.3 mm × 40 mm by increasing deformation passes and rate, after four MDF passes, the average grain size of 6.1 μm was achieved. Tong [6] et al. revealed that the MDF process can refine the grain size for ZK60 magnesium alloy with dimensions of 30 mm × 30 mm × 60 mm due to the occurrence of continuous dynamic recrystallization and weak the conventional basal texture which was replaced by a non-basal

texture. The work of Xia [8] et al. indicated that the grain size of Mg-Gd-Y-Nd-Zr alloy with dimensions of 80 mm × 80 mm × 120 mm was decreased from 200 μm to 5.1 μm by employing MDF process and the basal texture was weakened due to the synergistic effect of dynamic precipitation and changes in the direction of the MDF force. Dong [9] et al. performed a decreased-temperature MDF process for Mg-13Gd-4Y-2Zn-0.5Zr (wt%) alloy with dimensions of 130 mm × 130 mm × 160 mm and found that the average grain size of the forged part was refined remarkably to 4.0 μm and the basal texture transformed from strong basal texture to a random distribution gradually. However, the magnesium alloy forged parts in previous reports are usually small-scale laboratory samples, cannot reflect the beneficial effects of MDF process on homogeneity of microstructure and consistency of properties.

The present work aims to produce a large industrial-scale magnesium alloy forged part with 900 mm in diameter and 100 mm in height and investigate the effect of MDF process on microstructures and properties, which provides a guidance for manufacturing the large industrial-scale forged part and promotes the application of magnesium alloy in the fields of aerospace and transportation. The Mg-Gd-Y-Zn-Zr alloy that shows attractive mechanical properties [11–15] was used.

2. Experimental Procedures

2.1 Material preparation

The ingot of Mg-9.0Gd-3.0Y-2.0Zn-0.5Zr (wt.%, nominal composition) alloy with 450 mm in diameter and 3000 mm in length was made by semi-continuous casting. Based on the preliminary work, the large interspacing LPSO lamellae inside grains of the furnace-cooled Mg-Gd-Y-Zn-Zr alloy can effectively promote the dynamic recrystallization (DRX) inside the grains and make the alloy have good formability. Therefore, the ingot homogenized at 515°C for 40 hours and cooled in furnace to 400°C at an average cooling rate of ~1.5 °C/min followed by air quenching, the initial microstructures are shown in Fig. 1. It can be observed from Fig. 1 that the as-homogenized alloy is mainly composed of α-Mg matrix, block-shaped LPSO phases at grain boundaries and intragranular lamellar LPSO phases inside the grains which run through the grains and are parallel to each other. These block-shaped LPSO phases are difficult to remelt and rich in Gd/Y/Zn elements. In order to ensure the consistency of the microstructure, the specimens for compression tests with dimensions of 10 mm in diameter and 15 mm in height were machined from the homogenized material at the radius of 190 mm.

2.2 Hot compression, multidirectional forging and tensile tests

The hot compression experiments were carried out using a mechanical testing machine (INSTRON 5982) equipped with an environmental chamber. The samples were compressed at the temperatures of 350°C, 400°C, 450°C and 500°C, and strain rates of 0.001 s⁻¹, 0.01 s⁻¹, 0.1 s⁻¹ and 1 s⁻¹ up to a strain of 1.0. Prior to compression tests, the top and bottom of the sample were coated with graphite to reduce surface

friction, and all the samples were held at the target temperature for 5 minutes. After compression, the sample was immediately quenched in room temperature water to retain the deformed microstructure.

The MDF adopted three passes upsetting and drawing-out, and the deformation per pass was 30% - 40%. The schematic diagram of the MDF process is shown in Fig. 2. Finally, the forged part was trimmed to the target dimensions of 900 mm in diameter and 100 mm in height.

Tensile tests were conducted on the INSTRON 5982 machine at a speed of 1 mm/min at room temperature, using specimens of dog-bone shape with a gauge length of 25 mm and a diameter of 5 mm. Three specimens were tested for each region (edge and center) to ensure the reliability of the results.

2.3 Microstructure characterization

For microstructural characterization, specimens were sectioned along the compression direction (CD), polished and then etched in picric acid solution (4.2 g picric acid + 10 ml acetic acid + 70 ml alcohol + 10 ml distilled water). Micrographs were taken near the center of the cross-sectioned compressed specimens using optical microscopy (OM, LEICA MEF4M). Fracture morphologies were characterized by scanning electron microscopy (SEM, Tescan VEGA 3). Electron back-scattered diffraction (EBSD) analyses were carried out on Tescan Mira3 operated at 30kV and a scanning step of 0.5 μm . The data were analyzed using TSL OIM Analysis 7 software. JEM 2100 transmission electron microscopy was used for TEM characterization. The TEM samples were prepared by precision ion polishing system (PIPS).

3. Results And Discussion

3.1 True stress-strain curves

Figure 3 shows the true stress-strain curves of the samples under different deformation conditions. Obviously, these curves belong to a typical type of DRX. There is an obvious work hardening stage caused by the sharp augmentation of the dislocation density at the initial stage of deformation. When the stress increases to a certain peak with the increase of strain, the DRX softening effect including both dynamic recovery and dynamic recrystallization plays a dominant role until it reaches the steady-state, i.e. the work hardening and DRX softening reach a balance. Besides, it can be seen from Figure 3 that the deformation temperature and strain rate have a significant effect on the flow stress. The stress value of the alloy decreases with the deformation temperature increases or the strain rate decreases. High deformation temperature and low strain rate are favorable for DRX, resulting in considerable DRX softening effect.

3.2 3D processing maps

Based on the dynamic material model (DMM), the two-dimensional (2D) processing maps established by Prasad [16] have become one of the effective methods for analyzing the formability of materials, which were widely used in metal materials such as steel, aluminum alloy and titanium alloy. It provides a strong basis for formulating and optimizing forming process parameters. The 3D processing maps, i.e. the

strain is introduced on the basis of 2D processing maps. The 3D processing maps can reflect the influences of forming temperature, strain rate, and strain on the power dissipation efficiency and flow instability domain of the studied material. In recent years, 3D processing maps including the 3D power dissipation map and 3D flow instability map have been widely used in magnesium alloys [1, 2, 17–19].

3.3.1 3D power dissipation maps

Assuming that the flow stress conforms to the DMM at the given forming temperature and strain, the workpiece can be regarded as a power dissipator during plastic deformation. The flow stress is expressed as [1, 2, 16]:

$$\sigma = K\dot{\epsilon}^m \quad (1)$$

where K is a material constant and m is the strain rate sensitivity index which is given by:

$$m = \partial \ln \sigma / \partial \ln \dot{\epsilon}$$

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Based on the DMM, the total dissipation power P absorbed by the workpiece during thermal plastic deformation can be expressed by the two complementary parts: G content and J co-content, which are expressed as follows:

$$P = \sigma \cdot \dot{\epsilon} = G + J = \int_0^{\dot{\epsilon}} \sigma d\dot{\epsilon} + \int_0^{\sigma} \dot{\epsilon} d\sigma \quad (3)$$

Where G content represents the power dissipation during the thermal plastic deformation of the workpiece, most of which is converted into viscoplastic heat; J co-content represents the power dissipation caused by the microstructure evolution (e.g. dynamic recovery, dynamic recrystallization, internal cracks and so on) during the plastic deformation process.

At the given forming temperature and strain, the strain rate sensitivity index m can be used to express the proportion of power dissipation between G constant and J co-constant during the plastic deformation of the workpiece:

$$\frac{dJ}{dG} = \frac{\dot{\epsilon}}{\sigma} \frac{d\sigma}{d\dot{\epsilon}} = \frac{\epsilon \sigma}{\sigma \dot{\epsilon}} \frac{d \ln \sigma}{d \ln \dot{\epsilon}} = \frac{\partial \ln \sigma}{\partial \ln \dot{\epsilon}} = m \quad (4)$$

The power dissipation efficiency η can be used to quantify the ability of the workpiece to dissipate power through microstructure evolution, which is defined as:

$$\eta = \frac{J}{J_{\max}} = \frac{2m}{m+1} \quad (5)$$

Where $J_{\max}$ means that under ideal linear dissipation ($m = 1$), the power dissipation caused by the microstructure evolution can reach the maximum value. The power dissipation efficiency η represents the

ability of the power dissipation of the workpiece through the microstructure evolution. Different microscopic mechanisms correspond to different values of η , i.e. η value reflects the microscopic response of the workpiece under different deformation parameters.

It corresponds to a specific stress value at a given deformation rate, temperature and strain value based on the stress-strain curve. Therefore, the η value under different deformation conditions can be obtained by calculation. Fig. 4 shows the three-dimensional color grid map of the power dissipation efficiency value η , different colors represent the various efficiency of power dissipation. Generally, the larger the power dissipation efficiency is, the larger the proportion of energy input to the evolution of the microstructure, i.e. these areas are conducive to plastic deformation [2, 20]. It can be seen from Fig. 4 that as the strain increases, the power dissipation efficiency increases at high deformation temperature and low strain rate, decreases at low deformation temperature and high strain rate. Additionally, as the deformation temperature increases, the power dissipation efficiency increases. But with the strain rate increases, the power dissipation efficiency decreases. The 3D processing map of this alloy present a peak region of power dissipation efficiency with the average value of 40% where the temperatures are 430°C - 500 °C, the strain rates are 0.001 s⁻¹ - 0.06 s⁻¹.

3.3.2 3D flow instability maps

In the power dissipation maps, it is not that the larger the power dissipation efficiency, the better the formability of the workpiece due to the power dissipation efficiency may also be higher in the processing instability domain. Therefore, flow instability maps are also indispensable in the process of analyzing the formability. The Murty instability criterion is not an empirical formula and it is applicable to any type of flow curves [1, 2, 21, 22]. Therefore, the Murty instability criterion was used to construct the 3D flow instability maps of Mg-Gd-Y-Zn-Zr alloy, which is expressed as follows:

$$2m - \eta < 0 \quad (6)$$

The 3D flow instability maps of Mg-Gd-Y-Zn-Zr are shown in Fig. 5. Blue regions represent the safety domains while red regions represent the instability domains. Deformation temperature, strain rate, and strain have a great influence on the instability domain. The instability domain gradually increases with the decrease of deformation temperature, increase of strain and strain rate. It can be seen from Fig. 5 that the instability domains are domain Ⅱ of the deformation temperatures of 350 °C -400 °C, strain rates of 0.1 s⁻¹ - 0.8 s⁻¹ and strain values of 0.5 - 1, domain Ⅲ of the deformation temperatures of 350 °C - 450 °C, strain rates of 0.8 s⁻¹ - 1 s⁻¹ and strain values of 0.5 - 1.

3.3 Microstructure evolution during compression

Figure 6 shows the optical microstructures of the samples under different deformation conditions. It can be seen that an adiabatic shear band (indicated by the blue arrows) appears in the samples at deformation temperature of 350 °C, strain rates of 1 s⁻¹ and 0.1 s⁻¹. The crack was caused by severe local deformation and there were a lot of fine DRX grains near the crack (Figure 6 (b)). Microcracks appeared at the interface between the block-shaped LPSO phases and the matrix duo to the severe local

deformation when the samples were deformed at temperatures of 400 °C, 450 °C and strain rate of 1 s^{-1} (indicated by the red arrows in Figs. 6 (e) and (i)). The optical microstructures above the blue dotted line in Figure 6 reflect the characteristics of instability, which are consistent with the results of 3D flow instability maps shown in Fig. 5.

DRX behaviors have a great dependence on the deformation temperature and strain rate. With the increase of deformation temperature or decrease of strain rate, the number and size of DRX grains gradually increase, i.e. the greater the power dissipation efficiency is, the easier DRX will occur. In the process of compression deformation, the block-shaped LPSO phases were broken, sheared and bent while the intragranular lamellar LPSO phases were kinked and bent. Apparently, grain boundaries, block-shaped LPSO phases, kinked LPSO phases and Mg matrix between intragranular LPSO lamellae are all favorable nucleation sites for DRX. The fine DRX grains were distributed in chains at the grain boundaries of the deformed grains when the samples were deformed at the deformation temperature of 350 °C, the strain rates of 0.01 s^{-1} and 0.001 s^{-1} (Figs. 6(c) and (d)). The alloy underwent incomplete DRX and the size range of DRX grain was $1.9 \mu\text{m} \sim 4.8 \mu\text{m}$ when the samples were deformed at the deformation temperatures of 400 °C, 450 °C, and the strain rates of 0.001 s^{-1} - 1.0 s^{-1} . When the temperature increased to 500°C, complete DRX occurred in almost all alloys and the DRX grain size was increased significantly due to the increased temperature which contributed to DRX.

4. Fe Simulation Of Multi-directional Forging Process Of Cake-shaped Forged Part

The optimal deformation domain ($450 \text{ °C}/0.001 \text{ s}^{-1} - 0.06 \text{ s}^{-1}$) was obtained combined with the 3D processing maps and optical microstructures of the compression samples. The average power dissipation efficiency was about 40% and the alloy was prone to form the homogeneous and fine DRX grains when the optimal deformation parameters were adopted (Figs. 6 (j)-(l)). The MDF simulation of the cake-shaped forged part was performed in the light of the forging schematic diagram shown in Fig. 2. According to the favorable deformation strain rate, forging speeds of 5 mm/s, 10 mm/s and 15 mm/s were selected to further optimize the value. The forging temperature was 450 °C and the deformation degree per pass was about 40%.

Figure 7 shows the distribution of effective strain and power dissipation efficiency during each pass of forming process with different forging speeds. It can be seen from Figs. 7 (a) - (i) that effective strain in the center of the forged part is large while in the upper and bottom faces is small due to the influence of friction. As the pass of deformation increases, the cumulative effective strain value continues to increase. The distribution of effective strain is more homogeneous after three passes of upsetting and drawing-out, which means that the deformation is more uniform. However, the forging speed has little effect on the distribution of effective strain while has a greater effect on the distribution of power dissipation efficiency. Figs. 7 (j) - (r) indicate that the power dissipation efficiency at different forging speeds are all greater than zero, i.e. there is no instability occurred during the forging process. From the simulation

results, the power dissipation efficiency first increases and then decreases as the forging speed increases. The largest power dissipation efficiency with an average value of 45% was obtained when the forging speed was 10 mm/s.

Figure 8 shows the distribution of effective strain and power dissipation efficiency after final upsetting at different forging speeds. It can be seen from Figs. 8 (a) - (c) that the effective strain value was uniformly distributed when the forging speed was 10 mm/s, which meant that the workpiece was deformed uniformly. The power dissipation efficiency after the final upsetting is greater than zero (Figs. 8 (d) - (f)), it reaches a maximum of 48% at the forging speed of 10 mm/s. The large power dissipation efficiency indicates that a lot of energy input to the evolution of microstructure.

5. Multi-direction Forging Experiment, Microstructures And Mechanical Properties Of The Cake-shaped Forgings

The optimal MDF parameters were obtained based on the 3D processing maps and the Deform-3D MDF simulation, i.e. the forging temperature was 450 °C, forming speed was 10 mm/s, and deformation degree per pass was 40%. To verify the reasonability of the optimal MDF parameters, the as-homogenized ingot with 430 mm in diameter and 440 mm in length was hold at the forging temperature of 450 °C for 12 hours then the MDF was performed at a forging speed of 10 mm/s and single pass deformation of 40%. Finally, the cake-shaped forged part with 900 mm in diameter and 100 mm in height was successfully prepared (Fig. 9 (a)) and the size after machining was about 850 mm in diameter and 90 mm in height (Fig. 9 (b)). The wrought microstructures of the center and edge of the cake-shaped forged part are shown in Figs. 9 (c) and (d). We can observe that the block-shaped LPSO phases are broken and distributed intermittently along the TD direction. The wrought microstructures of the edge and center are relatively uniform. The DRX grains size in center is larger than that in edge due to the temperature of the center is higher than that of the edge in the forging process. The average DRX grain sizes of the edge and center of the forged part are 6.54 μm and 9.21 μm , respectively.

In order to coordinate the strength and ductility, the cake-shaped forged part was subjected to T6 (solution + aging) heat treatment. The tensile curves (the loading direction is along TD direction) and work hardening rates are shown in Fig. 10. After T6 heat treatment, the tensile strength, yield strength and elongation of the edge sample are 406 MPa, 327 MPa and 16.0% while those of the center sample are 391 MPa, 318 MPa and 13.3%, respectively. The work hardening rates are shown in Fig. 10 (b), which means that there is little difference in work hardening rate between edge and center samples.

Figure 11 shows the TEM images of the T6-specimen of the edge sample. From the TEM bright-field image (Figure 11(a)) and dark-field image (Figure 11(b)), it can be seen that there are high-density nano-scale strengthening phases homogeneously distributed inside the matrix. These strengthening phases are β' phases according to the selected-area electron diffraction (SAED) pattern (Figure 11(c)) taken along $[11\bar{2}0]$ direction [15, 25]. These β' phases play a great strengthening effect.

Figure 12 shows the EBSD analyses of the edge and center samples in the T6-state. It can be seen from the inverse pole figures that the microstructures of the T6-sample are homogeneous (Figs. 12(a) and (b)) and the average grain sizes of the edge and center samples are 19.2 μm and 30.2 μm , respectively. The maximum texture intensities of the edge and center samples are similar (about 4.4). The good ductility should be attributed to the homogeneous microstructures and low texture intensity. The strength and ductility difference of the edge and center samples may be related to the grain size difference. According to the Hall-Petch formula [23, 24]:

$$\sigma_y = \sigma_0 + k d^{-1/2} \quad (1)$$

where σ_y is the yield stress, σ_0 is the friction stress when dislocations glide on the slip plane, d is the average grain size, and k is the stress concentration factor. The k value of the Mg-Gd-Y-Zn-Zr alloy, k is generally 280-330 $\text{MPa} \cdot \mu\text{m}^{1/2}$ [23]. The yield strength of the edge sample is 10-12 MPa higher than that of the center sample by the calculation based on the Hall-Petch formula. According to the tensile curves, the yield strength of the edge sample is 9 MPa higher than that of the center sample. Therefore, the small difference in performance between the edge and center samples is mainly attributed to the difference in grain size.

Figure 13 shows the morphology of the tensile fracture of the edge and center specimens. There are some micron-size block-shaped LPSO phases distributed at the fracture (Figs. 13 (a) and (c)). The interface between the block-shaped LPSO phases and the matrix is prone to stress concentration, leading to crack nucleation. Figs. 13 (b) and (d) show the ductile fracture characteristics. The good ductility is attributed to the abundant dimples and tearing edges distributed on the macroscopic fracture of the alloy despite there are a few cleavage faces.

6. Conclusion

In this study, the 3D processing maps of cast Mg-9Gd-3Y-2Zn-0.5Zr alloy were established and the MDF simulation was performed by Deform-3D software. Finally, the large-scale cake-shaped forged part with 900 mm in diameter and 100 mm in height was produced. The microstructures and properties of the forged part were studied. The results can be concluded as follows:

(1) According to the 3D processing maps, the instability domains for cast Mg-9Gd-3Y-2Zn-0.5Zr alloy were domains of 350 $^{\circ}\text{C}$ - 400 $^{\circ}\text{C}$ / 0.1 s^{-1} - 0.8 s^{-1} and 350 $^{\circ}\text{C}$ - 450 $^{\circ}\text{C}$ / 0.8 s^{-1} - 1 s^{-1} . The stable and power efficient forming domains was in the temperature range from 430 $^{\circ}\text{C}$ to 500 $^{\circ}\text{C}$ and strain rate range from 0.001 s^{-1} to 0.06 s^{-1} .

(2) The results of FE simulation of the cake-shaped forged part indicate that the power dissipation efficiency first increases and then decreases as the deformation rate increases. The optimal forming speed is 10 mm/s.

(3) After T6-heat treatment, the edge and center of the large-scale forged part exhibit homogeneous microstructure and relatively consistent properties. The dense nano-scale β' phases play a great strengthening effect.

(4) The good ductility is attributed to the abundant dimples and tearing edges distributed on the macroscopic fracture of the alloy.

Declarations

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Code availability Not applicable

Ethics approval Not applicable

Consent to participate Not applicable

Consent for publication Not applicable

Author Contributions Jiyu Li and Shuai Dong and Jian Zeng and Jie Dong conceived and designed the experiments, Jiyu Li performed the experiments and wrote manuscript, Chaoyu Zhao performed the EBSD characterization, Jian Zeng and Li Jin and Fenghua Wang checked the data, Jian Zeng and Fulin Wang proofread manuscript.

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Figures

Figure 1

Microstructures of the as-homogenized alloy: (a) optical micrograph, (b) SEM image, (c) the mapping corresponding to SEM

Figure 2

Schematic diagram of multi-direction forging

Figure 3

True stress-strain curves of the as-homogenized alloys at different deformation conditions

Figure 4

The 3D power dissipation maps: (a) at strains of 0.27, 0.5, 0.77, 1.0; (b) at temperatures of 350 °C, 400 °C, 450 °C, 500 °C; (c) at strain rates of 1 s⁻¹, 0.1 s⁻¹, 0.01 s⁻¹, 0.001 s⁻¹

Figure 5

The 3D flow instability maps: (a) at strains of 0.27, 0.5, 0.77, 1.0; (b) at temperatures of 350 °C, 400 °C, 450 °C, 500 °C; (c) at strain rates of 1 s^{-1} , 0.1 s^{-1} , 0.01 s^{-1} , 0.001 s^{-1}

Figure 6

Microstructures of the samples at strain of 1.0, strain rates of (a, e, i, m) 1.0 s^{-1} , (b, f, j, n) 0.1 s^{-1} , (c, g, k, o) 0.01 s^{-1} , (d, h, l, p) 0.001 s^{-1} , and different temperatures (a-d) 350 °C, (e-h) 400 °C, (i-l) 450 °C, (m-p) 500 °C

Figure 7

Distributions of effective strain (a-i) and power dissipation efficiency (j-r) after upsetting deformation in each pass with different forging speeds

※ All the slicing of cloud maps are in the upsetting completion state.

Figure 8

Distributions of effective strain and power dissipation efficiency during final forming of the cake-shaped forged part

Figure 9

Macro-images of the cake-shaped forged part before (a) and after (b) machining, optical microstructures at the edge (c) and center (d) of the forged part

Figure 10

The stress-strain curves and work hardening rates of the edge and center samples

Figure 11

TEM images of the T6-specimen of the edge: (a) bright-field image, (b) dark-field image and (c) the corresponding selected-area electron diffraction (SAED) pattern taken along $[112]_{\alpha\text{-Mg}}$ direction

Figure 12

EBSD analyses of the edge (a, c, e) and center (b, d, f) samples: (a, b) IPF map, (c, d) average grain size and (e, f) pole figure

Figure 13

Fracture analyses of the edge (a, b) and center (c, d) samples: (a, c) backscattered electron (BSE) image, (b, d) secondary electron (SE) image