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Influence of metal flow boundaries on the microstructure and tensile properties of disk-shaped 7075 aluminum alloy forgings

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Abstract

Die forging inevitably induces metal flow toward both the die cavity and flash, leading to the formation of a flow boundary between these two regions. However, the influence of such a flow boundary on the final microstructure and properties of forgings remains unclear. To address this gap, this study investigates the effects of flow boundaries on the microstructure and mechanical properties of disk-shaped 7075 aluminum alloy casing forgings via a combination of finite element (FE) simulations and experimental characterizations. Results show that two distinct flow boundaries form at different locations during disk forging, and their formation mechanisms and evolutionary characteristics were systematically clarified. Electron Backscatter Diffraction (EBSD) analysis revealed significant microstructural inhomogeneities at different locations of the forging: regions adjacent to flow boundaries (Locations 2 and 3) exhibit coarse grain structures with an average grain size of 54.02 μm and 32.70 μm ,

and low recrystallization fractions of 22.7% and 31.9%, which are only half of those in non-boundary regions (48.8%–51.9%). Tensile tests demonstrated that the yield strength of flow boundary regions in the radial direction is 362 MPa–371 MPa, which is 40 MPa–50 MPa lower than that of non-boundary regions (400 MPa–413 MPa). Although flow boundaries have a significant impact on grain morphology and texture development, they exert only a minor effect on phase distribution and texture intensity. These quantitative findings offer critical theoretical support for optimizing the forging processes of high-performance aluminum alloy components and improving the uniformity of forging properties.

Keywords: disk forgings ; metal flow boundaries ;microstructure; properties

1 Introduction

Disk forgings are critical load-bearing components in the aerospace industry. They are mainly used in aero-engine compressor disks [1,2], turbine disks [3,4] and some structural casings [5,6]. These components work under long-term high-temperature, high-pressure and complex cyclic loads. They have extremely strict requirements for microstructural uniformity and mechanical property consistency. Any internal microstructural inhomogeneity or performance degradation will directly affect the service life and operational reliability of aircraft [7,8]. Therefore, precise control of the microstructure and properties of disk forgings has become a core technical challenge in advanced aero-engine manufacturing.

The final microstructure and properties of disk forgings are determined by metal flow behavior during forging. Temperature, strain, strain rate and interfacial friction

jointly regulate this behavior. Over the past decades, extensive research has been conducted on the correlation between forging parameters and component performance.

Early studies focused on optimizing individual process parameters to suppress forming defects and improve fatigue life. Fernández et al. [9] showed that appropriate cutting speeds can reduce machining residual stresses and improve the fatigue performance of titanium alloy disk forgings. Wang et al. [10] found that reducing the die-billet friction coefficient is a key measure to suppress laminated crack defects in large-scale disk forgings. These studies laid the foundation for ensuring basic forming quality, but did not address the root cause of microstructural inhomogeneity.

With the development of numerical simulation technology, finite element analysis has become an indispensable tool for process optimization. Jayanthi et al. [11] simulated metal flow, load distribution and folding defect formation during compressor disk forging, and optimized the preform design. Cai et al. [12] established a general kinematically admissible velocity field for three-dimensional upset forging of elliptical disks. Li et al. [13] revealed the deformation distribution and grain size characteristics of TC6 alloy disks at 920 °C and 2.0 mm/s. Zhang et al. [14] proposed that increasing axial deformation and contact friction during upsetting can obtain billets with optimal radial flow lines. Kim et al. [15] prepared defect-free large-scale disks by adjusting strain distribution with counteracting dies.

Data-driven methods have emerged as a new paradigm for complex process optimization. Zhang et al. [16] combined machine learning with multi-objective genetic algorithms to achieve efficient process optimization. Peng et al. [17] optimized laser

welding process parameters and demonstrated that reasonable process sequence can reduce residual stress. These methods have improved process design efficiency, but rarely pay attention to the special deformation behavior of local flow discontinuity regions.

Researchers have also deeply explored the regulation of microstructural evolution. Zhao et al. [18] found that reducing forging speed can extend dynamic recovery time and form more subgrains in Al-Zn-Mg-Cu alloys. Fu et al. [19] revealed that twinning and stacking faults can significantly modulate precipitation behavior.

For different alloy systems, Farabi et al. [20] indicated that high-strain-rate forging can obtain fully recrystallized microstructures in Alloy 718 engine disks. Jayanthi et al. [2] explored a non-isothermal forging process for IMI 685 titanium alloy compressor discs. Wu et al. [3] characterized the hot workability of the novel powder metallurgy superalloy FGH98 I. Nagar et al. [21] and Li et al. [22] respectively established velocity field models and interface friction laws for forging load calculation. Most of these studies are based on the assumption of uniform deformation, and fail to consider the microstructural differences caused by local flow discontinuities.

Innovative forging processes have been developed to improve microstructural homogeneity. Bewlay et al. [23] proposed a roll forming process for turbine disks with complex geometries and homogeneous microstructures. Kou et al. [24] demonstrated that regulating molten pool flow via arc oscillation can reduce porosity and weaken crystallographic texture in additive manufacturing. However, these processes still cannot avoid the formation of flow boundaries during die forging, which is an inherent

characteristic of disk-shaped component forming.

Despite significant progress, a critical knowledge gap remains. The influence of "metal flow boundaries" on the final microstructure and properties has not been systematically investigated. Flow boundaries are quasi-stagnant transition zones formed between regions with opposite metal flow directions under the combined effects of die geometry and interfacial friction. Since metal flow directly determines microstructural evolution, flow boundaries will inevitably cause local microstructural inhomogeneity and performance degradation. No existing studies have reported the formation mechanism, evolutionary characteristics and quantitative influence law of flow boundaries. This has become a key bottleneck restricting the precise control of disk forging performance.

To fill this gap, this study takes disk-shaped 7075 aluminum alloy casing forgings as the research object. A combined approach of finite element simulation and experimental characterization is adopted. First, the formation and evolution mechanisms of flow boundaries during isothermal forging are clarified through velocity field analysis. Then, the effects of flow boundaries on grain morphology, recrystallization fraction, phase distribution, texture characteristics and mechanical properties are quantitatively investigated. Finally, the multi-scale coupling mechanism between flow boundaries, microstructure and properties is revealed. The findings will provide critical theoretical support for optimizing the forging process of high-performance aluminum alloy components and improving the uniformity of forging properties.

2 Experimental and Simulation Methods

2.1 Casing Forgings

The casing forgings are intended for use in engine reducers as critical load-bearing components. The maximum contour diameter exceeds 700 mm. Fig.1 shows the three views of the 7075 aluminum alloy casing forging, including (a) front view, (b) bottom view, and (c) top view.

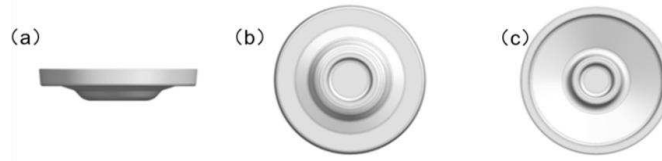


Fig.1 Three views of 7075 aluminum alloy casing forging: (a) front view, (b) bottom view, and (c) top view

2.2 Forging Dies

The forging dies were designed based on the casing forgings, with tooling adapted to equipment requirements. Fig.2 shows the molds for forging the 7075 aluminum alloy casing forging, where (a) is the upper mold and (b) is the lower mold.

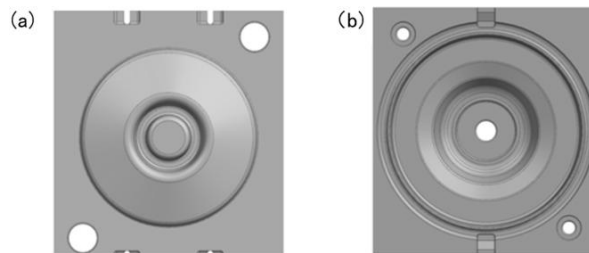


Fig.2 The mold (a) for forging 7075 aluminum alloy receiver is the upper mold, and (b) is the lower mold

2.3 Finite Element Simulation of 7075 Aluminum Alloy Casing Forging

To analyze the formation and evolution mechanisms of flow boundaries, finite element simulations of the entire forging process were conducted using the FE simulation software. The Arrhenius-type constitutive equation was adopted in the finite

element simulation to describe the hot deformation behavior of 7075 aluminum alloy.

The flow stress model used in the simulation is expressed as:

$$\dot{\epsilon}=2.6377\times10^{23} [\sinh(0.0124\sigma)]^{5.115}\exp(-130400/RT)$$

where $\dot{\epsilon}$ is the strain rate, σ is the flow stress (MPa), R is the universal gas constant ($8.314 \text{ J}\cdot\text{mol}^{-1}\cdot\text{K}^{-1}$), and T is the absolute temperature (K).

The simulation was conducted using DEFORM-3D software . The billet was discretized into 64,000 tetrahedral elements, and a mesh sensitivity analysis was performed to confirm that further mesh refinement would not alter the velocity field distribution or flow boundary locations. To simulate an isothermal environment, volume compensation was enabled and heat transfer was disabled. A constant shear friction model was adopted with a friction coefficient of 0.3, which is the standard industrial value for isothermal hot forging of 7075 aluminum alloy with graphite lubrication. The von Mises yield criterion and isotropic hardening model were used, with constitutive parameters derived from systematic hot compression tests on 7075 aluminum alloy at temperatures of 350–500 °C and strain rates of 0.001–10 s⁻¹; these parameters have been widely validated in similar forging simulations. Although direct comparison of macro flow lines was not carried out, the predicted flow boundary locations showed good agreement with the experimentally observed microstructural inhomogeneities (grain size and recrystallization fraction) presented in Sections 3.3 and 3.4.

For pre-forging, the initial billet was a 7075-T73 aluminum alloy rod with a diameter of 650 mm and a height of 80 mm. All billet surfaces were free surfaces without boundary constraints. The upper and lower dies were rigid bodies set to a temperature of 430 °C, with the lower die fixed and the upper die acting as the primary moving die. The upper die descended at a speed of 5.1 mm/s and stopped when the

underfill reached 15 mm to obtain the pre-forged part. The pre-forged part was then machined to create a central hole and reduce the outer circumference and central boss respectively, producing the final forging billet. The final forging process used an identical 64,000-element mesh and the same process parameters as pre-forging, except that the stop criterion was adjusted to an underfill of 9 mm.

The die-billet interface friction significantly influences the formation of metal flow boundaries and forging cavity filling. A friction coefficient of 0.3 inhibits the radial flow of metal in the billet's peripheral region, directing the flow more toward axial downward movement and promoting the formation of Flow Boundary 1 at the interface between central and peripheral metal flows. Meanwhile, interface friction reduces the filling performance of the forging edge; insufficient cavity filling leads to uneven metal flow distribution, which facilitates the formation of Flow Boundary 2 in the central concave region of the casing forging. The simulation results for each forging step are shown in Fig. 3.

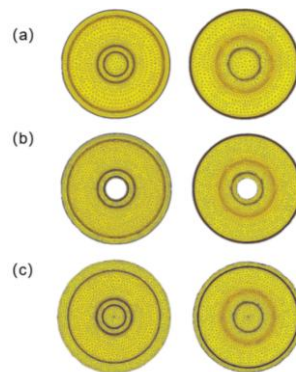


Fig.3 The simulation results of the isothermal forging process for 7075 aluminum alloy casing forgings at each stage include (a) pre forging results, (b) mechanical processing results, and (c) final forging results

2.4 Material

The experimental material was a 7075-T73 aluminum alloy extruded rod with a diameter of 650 mm and a height of 80 mm. The chemical components and room-

temperature mechanical properties are listed in Tables 1 and 2, respectively.

Table 1 Chemical components of 7075 Aluminum Alloy (wt%)

Si	Fe	Cu	Mn	Mg	Cr	Zn	Ti	Other total
0.04	0.23	1.4	0.03	2.5	0.20	5.7	0.02	---

Table 2 Mechanical Properties of 7075 Aluminum Alloy

Sampling Direction	σ_b /MPa	$\sigma_{0.2}$ /MPa	δ Elongation rate/%
longitudinal	508-519	417-432	11.0-11.1
horizontal	504-506	413-426	7.8-8.7

2.5 Isothermal Forging of 7075 Aluminum Alloy Casing Forgings

The 7075 aluminum alloy casing forgings were produced using a combination of pre-forging and final forging under isothermal conditions. The billet underwent pre-forging, machining, final forging, and heat treatment to produce the casing forgings. The heat treatment process was solution treatment at 470 °C for 2 h, water quenching, and artificial aging at 120 °C for 24 h. The results of each forging step are shown in Fig.4.

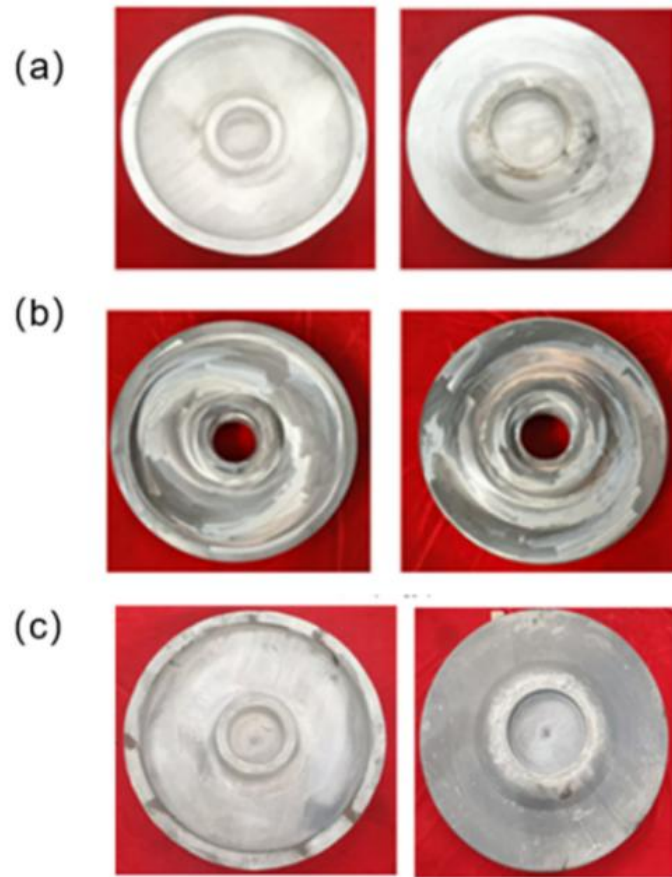


Fig.4 The actual manufacturing results of the isothermal forging process of 7075 aluminum alloy casing forgings at each stage include (a) pre forged parts after alkali washing, (b) final forged blanks after mechanical processing and polishing, and (c) final forged parts after heat treatment

2.6 Sampling Locations

To better illustrate the influence of flow boundaries on the microstructure and properties of the forgings, strain analysis was performed on the FE simulation results. Samples were taken from locations corresponding to different strain levels and flow boundary locations, and the sampling locations were selected to cover the flow boundary regions, high-strain regions and low-strain regions of the forging. The specific sampling location is shown in Fig.5.

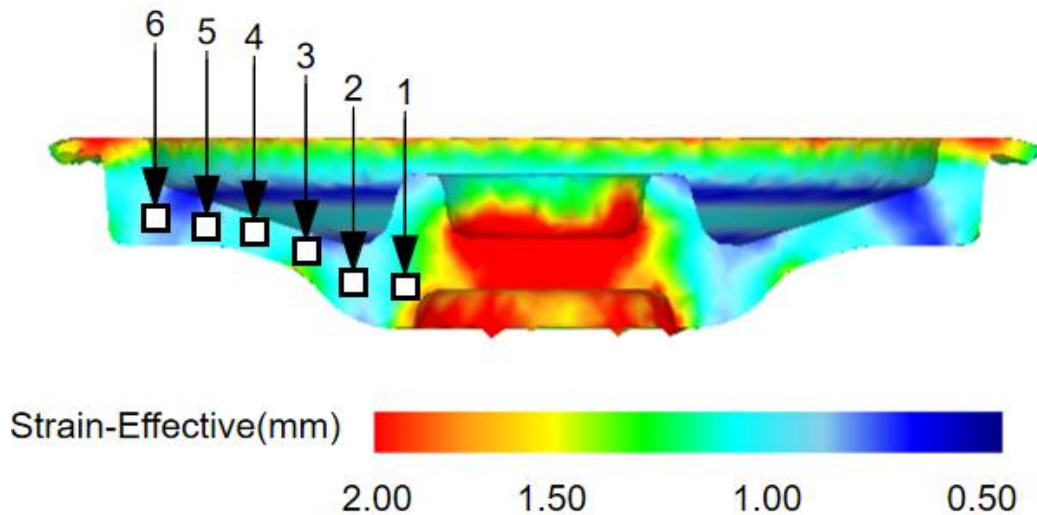


Fig.5 Strain analysis was conducted on the isothermal forging simulation results of 7075 aluminum alloy casing forgings, and the sampling location was selected based on the reference strain

2.7 Microstructural Characterization

A HITACHI SU5000 thermal field emission Scanning Electron Microscope (SEM) was used to conduct SEM analysis, Energy Dispersive Spectrometer (EDS), and Electron Back Scatter Diffraction (EBSD) on samples from the locations shown in Fig.5. The test equipment was calibrated before the experiment to ensure the accuracy of the test results. After sampling and polishing with 2000-mesh sandpaper, the final forgings were electrolytically polished to remove surface residual stress layers. The electrolytic polishing solution consisted of HClO_4 and $\text{C}_2\text{H}_5\text{OH}$ in a volume ratio of 4:11, with an applied voltage of 20 V, a current of 0.5–2 mA, a polishing time of 40–60 s, and a temperature of -20°C . After polishing, the samples were immediately rinsed with ethanol and dried with cold air to avoid surface oxidation. The scanning area for EBSD analysis was $400 \times 450 \mu\text{m}^2$, with a step size of $0.8 \mu\text{m}$ and a magnification of $300\times$. Channel5 software was used to statistically analyze crystallographic information such

as grain size, recrystallization fraction and texture components based on EBSD results.

2.8 Mechanical Property Testing

Due to size limitations, standard tensile specimens could not be extracted from the radial, tangential, and axial directions of the forgings. Therefore, small molds were used for room-temperature tensile testing on a universal electronic testing machine with a loading rate of 0.5 mm/min. Non-standard tensile specimens were wire-cut from the forged and heat-treated samples, with dimensions of $3 \times 1.2 \times 1 \text{ mm}^3$ for the tensile section. The specimen dimensions and actual samples are shown in Fig.6. Three parallel specimens were tested for each sampling location and each direction to ensure the repeatability of the test results, and the average value was taken as the final mechanical property data. Mechanical properties including yield strength ($\sigma_{0.2}$), tensile strength (σ_b) and elongation (δ) were tested in the radial, tangential, and axial directions.

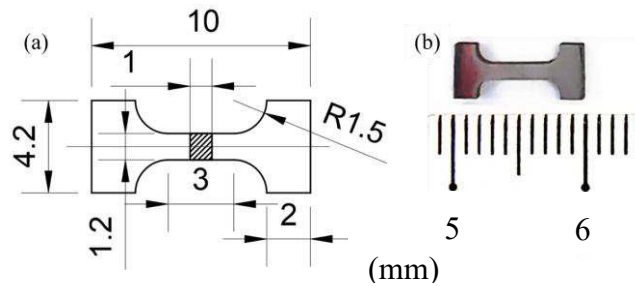


Fig.6 Nonstandard tensile test specimens used in room temperature tension in this article, where (a) is a two-dimensional dimensional drawing of a non-standard tensile test specimen, and (b) is a physical drawing of a non-standard tensile test specimen

3 Results and Discussion

3.1 Evolution of Flow Boundaries in Forgings

The flow boundary defined in this paper refers to a strain-discontinuous interface induced by deformation flow. It forms during the forging of disk forgings. Under the coupled effects of die geometric constraints and interfacial friction, part of the metal

flows toward the forging center, and part flows toward the forging periphery. This interface is one of the intrinsic causes of microstructural inhomogeneity in disk forgings.

Figure 7 shows the velocity field distribution during the pre-forging process of the forging. As shown in Figure 7(a), in the initial forging stage (reduction of 4.8 mm), the billet undergoes uniform axial upsetting deformation under the axial pressure of the die. The metal particles in the central region only exhibit a single downward motion along the axial direction of the forging. Their flow directions are highly consistent.

The edge region of the billet is in direct contact with the die cavity surface. It is subjected to significant interfacial friction resistance. The flow directions of metal particles in this region show obvious disorder. They have not only axial downward displacement but also radial shear slip tendency.

At this stage, a flow boundary in the strict sense has not yet formed. There is no clear boundary between metal converging toward the central region and diverging toward the outer region. However, the velocity field distribution shows that the metal flow patterns in the central and edge regions have essential differences. These differences lay a critical kinetic foundation for the subsequent formation and evolution of the flow boundary.

The heat transfer module of the software was disabled during the simulation. The forging process was conducted under the set isothermal condition of 430 °C. The metal deformation resistance has no significant fluctuation caused by temperature drop. Its flow behavior is only dominated by die geometric constraints and interfacial friction conditions.

When entering the mid-forming stage (reduction of 36 mm), as shown in Figure 7(b), the upper die continues to move downward and apply pressure. The metal in the web region of the forging undergoes oblique upsetting-like deformation under die

pressing. Part of the metal flows upward along the inclined surface to fill the outer edge cavity of the forging. The other part flows downward along the inclined surface to fill the central boss cavity of the forging.

The two streams of metal flow in opposite directions. A special region appears in the middle of the web. The metal in this region only moves with the die pressing. It moves in the direction perpendicular to the inclination of the upper die. There is no macroscopic flow in the direction parallel to the upper die inclination. This region is defined as flow boundary 1. It corresponds to the red dashed line in Figure 7(b).

Meanwhile, the metal in the central boss region of the forging undergoes upsetting-like deformation under die pressing. Part of the metal only moves downward synchronously with the die. It has almost no macroscopic plastic flow itself. Thus, a wide and thick flow boundary forms in the central boss region of the forging. This is flow boundary 2. It corresponds to the black-framed area in Figure 7(b).

As the die continues to move downward, when the reduction reaches 50.4 mm (Figure 7(c)), the outer edge of the billet fully contacts the sidewall cavity of the die. The friction force between the metal in the outer edge filling zone and the die increases significantly. This leads to a sharp rise in the resistance of the web metal flowing upward along the inclined surface. In contrast, the flow resistance of the metal flowing downward along the inclined surface to fill the central boss remains basically unchanged.

The resulting resistance difference breaks the original metal flow balance. It drives flow boundary 1 to migrate along the radial direction of the web. The migration direction is from the middle of the web to the side of the central boss. When the reduction reaches 62.4 mm, flow boundary 1 stabilizes outside the central boss (Figure 7(d)). Its complete evolution process ends at this point.

During the entire process from formation, migration to stabilization of flow boundary 1 (reduction of 36–62.4 mm), as shown in the black-framed areas in Figures 7(b)–7(d), the morphology of flow boundary 2 remains basically unchanged. It only moves downward synchronously with the die.

Until the reduction reaches 69.6 mm (Figure 7(e)), the boss of the forging is basically formed. The die cavity is completely filled. The billet deformation enters a backward extrusion-like stage. The metal mainly flows toward the middle rib of the forging and the near-flash region at the outer edge.

This transformation of the deformation mode makes the metal at flow boundary 2 unable to move synchronously with the die anymore. It needs to fill the middle rib cavity. flow boundary 2 is destroyed accordingly. At this time, the central and outer edge cavities of the forging still need to be supplemented with metal synchronously. Therefore, the force and flow state of flow boundary 1 remain unchanged. It still stably stays outside the central boss.

When the reduction reaches 75.3 mm (Figure 7(f)), the upper part of the billet outer edge fully contacts the upper die. The overall forming of the forging is completed. The subsequent deformation is dominated by flash forming. The metal flows toward the flash region as a whole. No flow boundary exists inside the forging anymore.

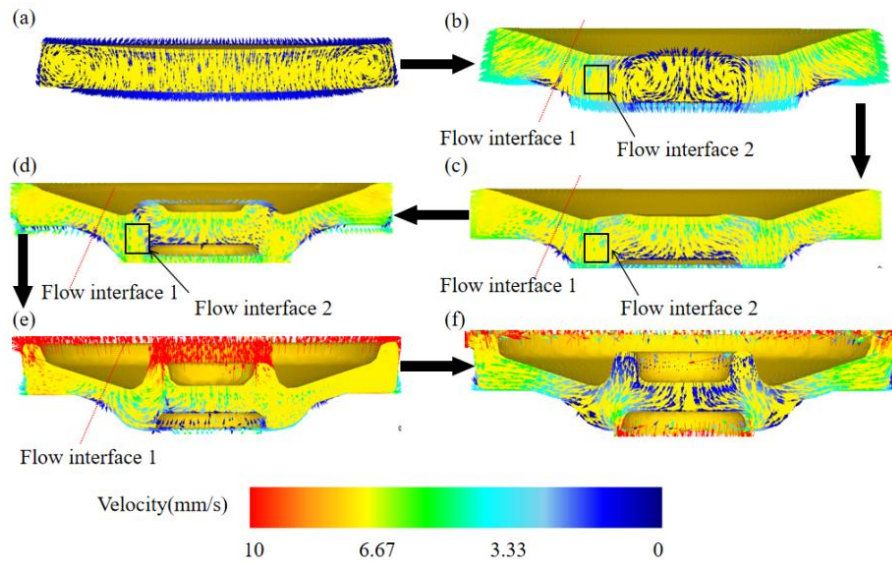


Fig.7 The evolution process of the flow interfaces within the casing forgings is depicted as follows: the red dashed line indicates Flow Interface 1, and the black box denotes Flow Interface 2. Reduction: (a) 4.8 mm; (b) 36 mm; (c) 50.4 mm; (d) 62.4 mm; (e) 69.6 mm; (f) 75.3 mm

Figure 8 shows the velocity field distribution during the final forging process of the forging. The results are as follows.

In the initial stage of final forging (Figure 8(a)), the metal in the web region splits into two streams. One stream flows upward and outward along the inclined surface of the die cavity to form flash. The other stream flows downward and centripetally to fill the central hole of the forging. Flow Boundary 1 first appears in the final forging at this stage. It stabilizes near the central side of the web middle.

After the die moves downward by another 2 mm (Figure 8(b)), the flash is not fully formed. The flow trend of the web metal remains unchanged. Flow Boundary 1 still stabilizes at its original location.

When the die moves downward by another 2 mm (Figure 8(c)), the flash is basically formed. However, the central hole of the forging is not completely filled yet. Flow Boundary 1 migrates. It finally stabilizes outside the central boss of the forging. After the central hole is completely filled, the subsequent deformation is dominated by

flash expansion. The metal flows directionally toward the flash region as a whole. No flow boundary exists inside the forging anymore.

In the final forging stage, filling the central hole is the dominant flow behavior of the metal. An extremely high centripetal flow velocity is formed in the horizontal direction. This completely breaks the state that existed during the pre-forging process. In that state, the region corresponding to the central boss of the forging always moved downward synchronously with the die as a rigid body. It had almost no plastic deformation.

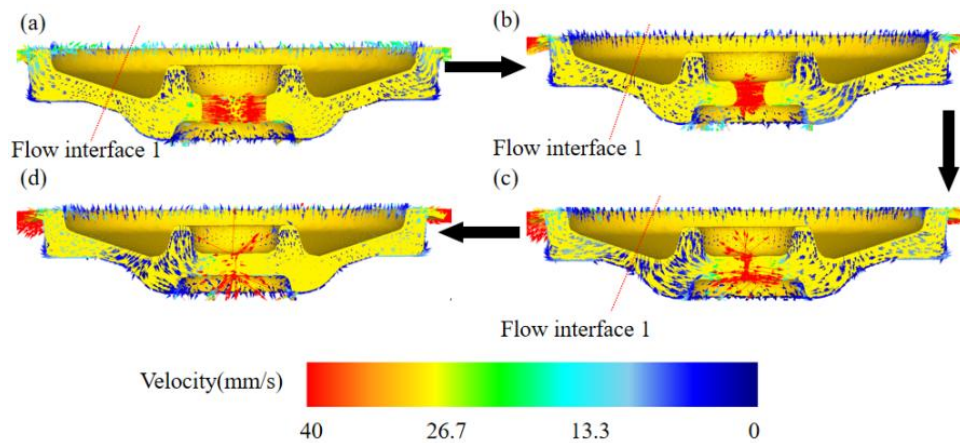


Fig.8 Velocity field distribution during final forging. Reduction: (a) 2 mm; (b) 4 mm; (c) 6 mm; (d) 8 mm

3.2 Effect of Flow Boundary on Phase Distribution of Forged Parts

Fig. 9 shows SEM images taken from different locations within the forging. The secondary phases are predominantly distributed in a "banded" morphology aligned with the local deformation direction, reflecting significant deformation anisotropy across the material. Notably, Location 5 exhibits a distinct deformation orientation compared to other areas, suggesting it plays a critical role in coordinating deformation between the peripheral section and the rib plate during forging.

Fig. 10 presents the distribution of the two main phases present in the forging. According to energy-dispersive X-ray spectroscopy (EDS) analysis:

At location A: 1.33 at% Mg, 83.46 at% Al, 13.89 at% Cu, and 1.32 at% Zn, identifying the light gray phase as Cu-rich;

At location B: 39.20 at% Mg, 59.79 at% Al, 0.23 at% Cu, and 0.78 at% Zn, confirming the dark gray phase as Mg-rich.

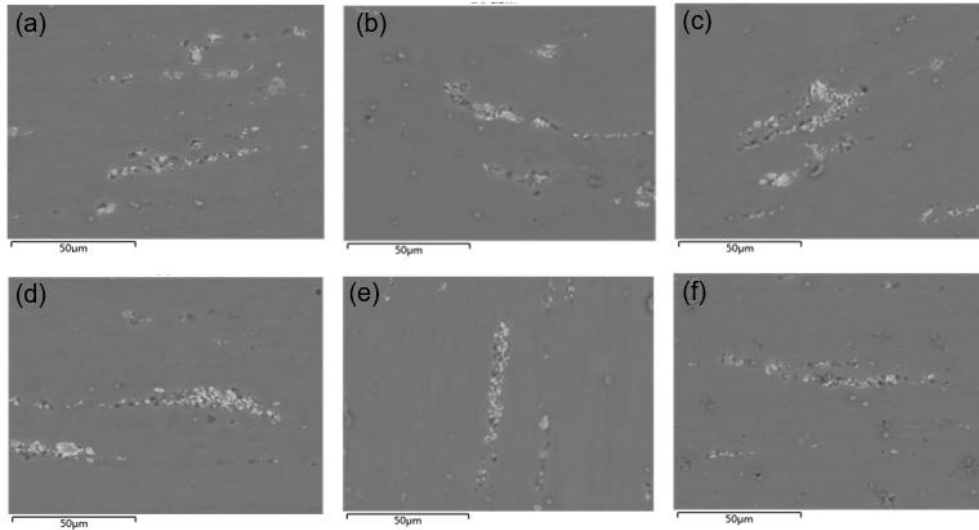


Fig.9 Phase distribution at different locations of the casing forgings, where (a) corresponds to location 1, (b) corresponds to location 2, (c) corresponds to location 3, (d) corresponds to location 4, (e) corresponds to location 5, and (f) corresponds to location 6

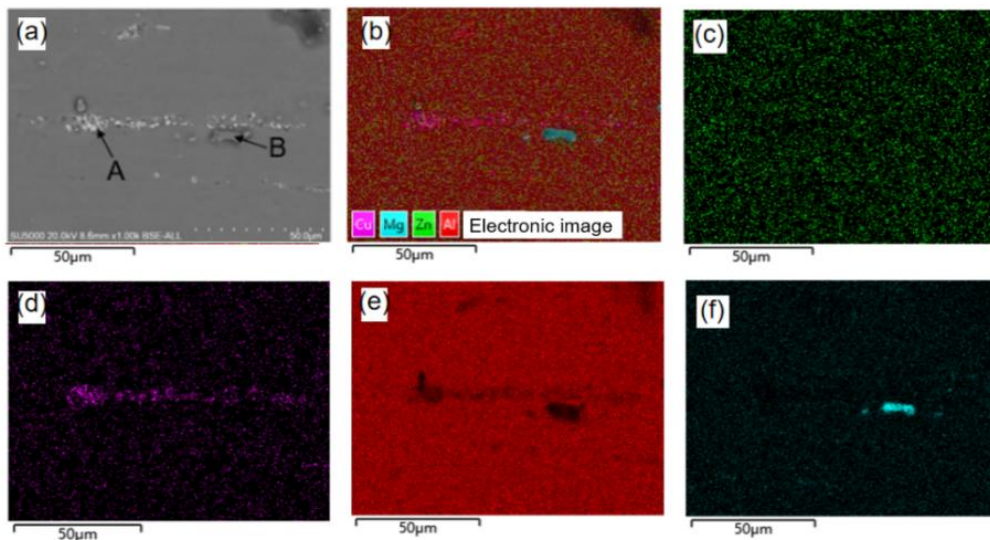


Fig.10 Two phases that can be observed at this scale for casing forgings, where (a) is the SEM image, (b) is the overall EDS surface scan result, (c) is the EDS surface scan result for the Zn element, (d) is the EDS surface scan result for the Cu element, (e) is the EDS surface scan result for the Al element, and (f) is the EDS surface scan result for the Mg element

3.3 Effect of flow boundary on grain morphology of forgings

Figure 11 shows the IPF maps of different locations of the forging. These maps directly reflect the local plastic deformation history during the hot working process of the forging through grain morphology.

Comparative analysis of grain orientation and morphology shows that locations 1, 4, 5 and 6 undergo severe plastic deformation during forming. The grains are significantly elongated along the main deformation direction. They form obvious "strip-like" microstructure characteristics. This indicates sufficient metal flow and high accumulated strain in these regions.

Locations 2 and 3 also participate in plastic deformation. However, their deformation degree is relatively low. The grain morphology still retains certain equiaxed characteristics. No obvious elongation phenomenon is observed.

location 1 was also located at Flow Boundary 2 during pre-forging. However, the intense metal flow at location 1 during final forging drastically changes its grain morphology.

The grain morphology at location 5 shows deformation characteristics between the peripheral region and the rib region. This further confirms that this location acts as a transition zone connecting the deformation of the peripheral part and the rib part during forging forming.

Location 2 is located in the region of Flow Boundary 2. The metal flow degree along this location is low. This restricts the elongation deformation of grains in this direction.

The axial flow velocity at location 2 is not low. However, combined with the evolution process of the flow boundary shown in Figure 7, it can be analyzed that the axial movement of location 2 is mainly the overall downward movement along the axial

direction together with the central boss of the casing forging. This overall rigid translational movement results in less relative slip and rotation between grains inside the material. Therefore, the accumulated strain inside is low.

In summary, the existence of Flow Boundary 2 significantly inhibits the tangential and radial deformation at location 2. The overall rigid movement of the central boss further restricts its independent axial plastic deformation. The combined effect of multiple factors leads to a low overall deformation degree at location 2. The grains can maintain a relatively equiaxed state.

Location 3 shows a low deformation degree because it is located in the stable region of Flow Boundary 1 during final forging.

For disk-shaped parts such as casing forgings, metal flow has a clear directional boundary. The metal outside Flow Boundary 1 mainly flows outward. The metal inside Flow Boundary 1 mainly flows inward. The coexistence of these reverse flows makes the metal at the flow boundary in a state where the velocity gradient is close to zero. The metal flow velocity is low. It is even close to the critical point of no flow locally. This leads to a small accumulated plastic deformation in this region.

In contrast, location 4 is located at the initial location of the flow boundary. However, the subsequent movement of the flow boundary still allows the grains at location 4 to undergo sufficient deformation.

Nevertheless, a certain orientation can still be observed from the grain deformation direction at location 3. This indicates that although the deformation magnitude is small, it plays a role in connecting the deformation of the central boss and the rib of the casing forging in direction. It becomes a transition zone between two different deformation regions.

Overall, the grain morphology of 7075 aluminum alloy casing forgings shows a

strict positive correlation with the local plastic strain level. This law is completely consistent with the classical theory of grain evolution during hot deformation of aluminum alloys. That is, the amplitude of local plastic strain directly determines the plastic elongation degree of grains. It also determines the core driving force of dynamic recrystallization during hot deformation.

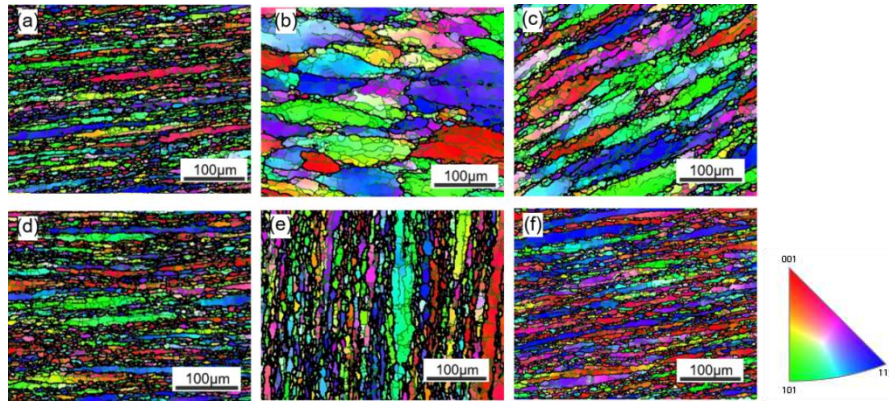


Fig.11 IPF diagram of different locations of the casing forgings, where red represents [001] grain direction, green represents [101] grain direction, and blue represents [111] grain direction. (a) corresponds to location 1, (b) corresponds to location 2, (c) corresponds to location 3, (d) corresponds to location 4, (e) corresponds to location 5, and (f) corresponds to location 6.

3.4 Effect of Flow Boundary on Recrystallization of Forged Parts

Figure 12 and Table 3 show the recrystallization maps of different locations of the forging. These maps clearly distinguish the distribution of recrystallized grains (blue), substructures (yellow) and deformed grains (red).

Combined with the statistical data of recrystallized grain, substructure and deformed grain contents and average grain size at each location in Table 3, it is found that the recrystallization percentages at locations 2, 3 and 6 are only about half of those at locations 1, 4 and 5. This indicates a significant difference in recrystallization degree between different regions.

Specifically, locations 2 and 3 are located in the regions of Flow Boundary 2 and Flow Boundary 1, respectively. The existence of these two boundaries restricts the metal flow here. This leads to low overall deformation and low dislocation density. The relatively small grain deformation degree at these two locations can also be intuitively seen from the grain morphology.

According to the classical recrystallization nucleation-growth theory, the driving force for recrystallization comes from deformation stored energy. The deformation stored energy is mainly provided by the dislocation density accumulated during plastic deformation. Insufficient stored energy cannot provide enough activation energy for recrystallization nucleation. Meanwhile, low dislocation density cannot provide sufficient nucleation sites. Therefore, locations 2 and 3 fail to accumulate enough energy to drive large-scale recrystallization. Their recrystallization percentages are naturally lower than those in severely deformed regions.

However, only from the perspective of grain morphology, location 6 shows similar microstructure characteristics to locations 1, 4 and 5. It seems to have experienced a large degree of deformation. But its recrystallization percentage is also at a low level. It does not obtain a recrystallization degree similar to these high-deformation regions. This phenomenon indicates that the actual overall deformation of location 6 is actually small. This is consistent with the finite element simulation result of low strain at location 6 in Figure 5 (sampling location schematic).

Although the deformation of location 6 is small, the average grain size data in Table 3 show that the grains at this location are much finer than those at locations 2 and 3. It even becomes the region with the finest grains among the six sampling locations. A reasonable explanation for this phenomenon can be made by analyzing the evolution process of the flow boundary shown in Figure 7. Location 6 is exactly at the corner of

the forging die. The metal flow behavior here is similar to the equal channel angular pressing (ECAP) process. The geometric constraint causes a sharp turn in the metal flow direction. This generates strong shear stress at this location. Although this shear stress does not significantly increase the overall strain, it effectively breaks the grains. It plays a role in grain refinement.

In addition, since the grains at location 6 are finer, according to the theories of grain boundary strengthening and superplastic deformation, fine grains are beneficial to grain boundary sliding and rotation. This makes the material more prone to local deformation. Therefore, location 6 can obtain the most deformed grains with a relatively small overall deformation. Meanwhile, the distribution of its deformed grains is more uniform. This phenomenon reveals the complex relationship between local deformation mode and microstructure evolution near the flow boundary. That is, local shear stress can effectively refine grains without significantly increasing macroscopic strain. This affects the recrystallization behavior and final microstructure state of the material.

The above research results systematically reveal the multi-field coupling mechanism between deformation history, dynamic recrystallization behavior and final microstructure during the forming process of 7075 aluminum alloy casing forgings. The dynamic recrystallization behavior of 7075 aluminum alloy during hot forging is synergistically regulated by strain level, strain rate and deformation temperature. The flow boundary reshapes the spatial distribution of the internal strain field of the forging. It differentially regulates the recrystallization proportion in different regions. This finally induces the heterogeneous distribution of recrystallized microstructure in the forging.

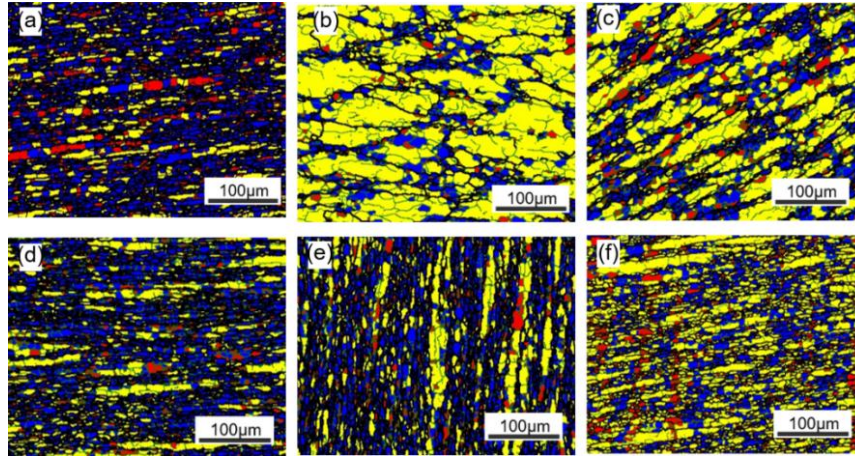


Fig.12 The recrystallization of different locations of the forged casing, where the red ones are deformed grains, the yellow ones are recovered grains, and the blue ones are recrystallized grains. (a) corresponds to location 1, (b) corresponds to location 2, (c) corresponds to location 3, (d) corresponds to location 4, (e) corresponds to location 5, and (f) corresponds to location 6

Table 3 Content and average grain size of recrystallization, substructure, and deformation grains at different locations of forgings

Sampling Location	Recrystallization(%)	substructure (%)	deformation (%)	$d_{avg}(\mu m)$
1	51.9	30.7	17.3	20.04
2	22.7	68.2	9.18	54.02
3	31.9	53.5	14.6	32.70
4	48.8	35.2	16.0	18.82
5	48.8	32.8	18.4	20.71
6	26.3	54.7	19.0	17.58

3.5 The influence of flow boundary on the texture components of forgings

Figure 13 and Table 4 show the texture analysis results of different locations of the forging. Different colors are used to identify various typical texture components in the figure. Purple represents the Copper texture ($\{112\}\langle 111 \rangle$) Green represents the Goss

texture ($\{110\}\langle 001\rangle$). Yellow represents the Cube texture ($\{001\}\langle 100\rangle$). Red represents the Brass texture ($\{110\}\langle 112\rangle$).

Comparison of the texture composition at different locations shows that the texture type distributions at locations 2, 3 and 5 are relatively consistent. Locations 1, 4 and 6 show another set of similar texture characteristics. Specifically, the Brass texture dominates at locations 2, 3 and 5. The Copper texture has the highest content at locations 1, 4 and 6. The Copper texture at location 6 is particularly prominent. Its volume fraction reaches 27%. This is also the region with the highest total texture content among all sampling locations.

Combined with the previous analysis of deformation behavior and microstructure, locations 2 and 3 are located in the regions of Flow Boundary 2 and Flow Boundary 1, respectively. Their deformation degrees are low. Their mechanical properties are expected to be relatively poor. Notably, location 5 has a similar texture composition to locations 2 and 3. This means that location 5 may also show similar limitations in some properties. This similarity in texture composition can be explained by the metal flow behavior during deformation. Location 5 acts as a transition zone connecting the deformation of the peripheral part and the rib part during forging forming. Location 3 plays a role in connecting the deformation of the central boss and the rib of the casing forging. Both undertake the transition and connection function between different deformation regions during forging deformation. Therefore, they show similar laws in texture evolution. As described by the crystal plasticity theory, the formation of different texture components is essentially determined by the local plastic deformation mode of the material during forging. This is the result of grain rotation along specific crystallographic directions under external force.

This phenomenon can be further understood from the definition of the flow

boundary. The flow boundary is essentially the region where two parts of metal flowing in opposite directions meet during forging deformation. It plays a role in connecting different flow modes. Locations 2 and 3 are exactly located in such interface regions. Location 5 is not on a typical flow boundary. But it also undertakes the role of connecting different deformation regions functionally.

Therefore, these three locations show similarity in texture composition. All are dominated by the Brass texture. This reflects their internal connection in deformation mechanisms. This finding helps to deeply understand the correlation between texture distribution and flow boundary in disk forgings. It also provides a theoretical basis for further optimizing the forging process and improving the microstructure uniformity of forgings.

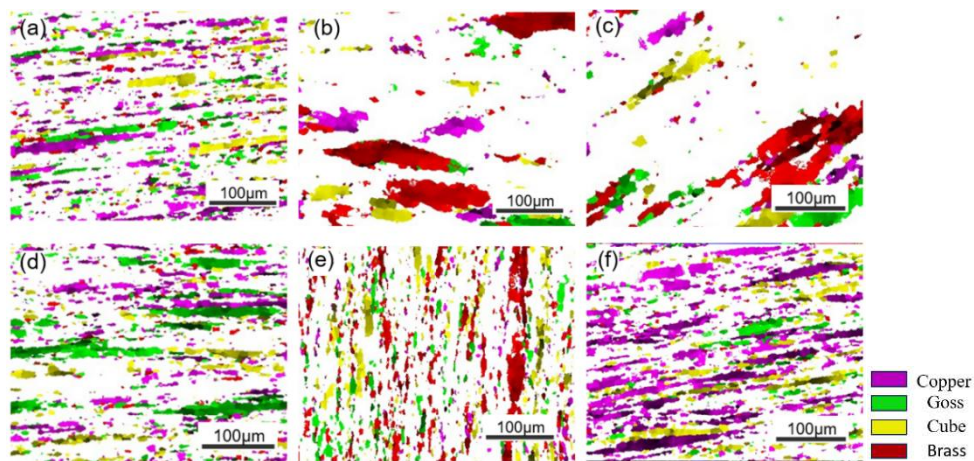


Fig.13 The texture components of the forged casing at different locations, where the red one is Brass texture, the yellow one is Cube texture, the green one is Goss texture, and the purple one is Copper texture. (a) corresponds to location 1, (b) corresponds to location 2, (c) corresponds to location 3, (d) corresponds to location 4, (e) corresponds to location 5, and (f) corresponds to location 6.

Table 4 Texture Components at Different Locations of Forgings

Sampling Location	Copper(%)	Goss(%)	Cube(%)	Brass(%)
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1	15	8.86	7.18	2.47
2	7.12	3.51	3.71	10.2
3	5.77	3.34	2.14	9.89
4	16.7	10.7	6.16	2.23
5	2.54	5.51	10	12.3
6	27	4	10	0.903

3.6 The influence of flow boundary on the texture strength of forgings

Figure 14 shows the inverse pole figures of six typical characteristic locations of the 7075 aluminum alloy casing forging. These figures systematically characterize the crystallographic texture distribution characteristics, preferred orientation laws and pole density intensity differences of each sampling point. All sampling locations cover the core regions of flow boundaries, large deformation control regions and deformation transition regions determined previously. They form a complete correspondence with the previous strain field analysis and microstructure characterization.

In terms of texture distribution characteristics, the texture types and distribution states of the six sampling locations show significant differences. They can be divided into two categories according to distribution concentration. One category is the directional preferred orientation type with highly concentrated texture, including locations 3, 4 and 6. The other category is the multi-orientation coexistence type with dispersed texture distribution, including locations 1, 2 and 5.

The specific texture characteristics of each location are as follows. Location 1 has a relatively dispersed texture distribution. Obvious texture distributions exist in both

[001] and [101] crystallographic directions. The highest pole density of 1.80 is concentrated in the [101] direction. Location 2 (located in the core region of Flow Boundary 2) also shows a diffuse texture distribution. The highest pole density of 1.67 is distributed in the transition region between the [001] and [101] directions. Location 3 (located in the core region of Flow Boundary 1) has a relatively concentrated texture distribution. It mainly gathers in the [101] direction. The highest pole density in this direction reaches 2.05. Location 4 has a highly concentrated texture distribution. Significant texture exists only in the [101] direction. The orientation intensity is high. The highest pole density reaches 3.13. Location 5 (rib-web deformation transition zone) has the most diffuse texture distribution. Texture distributions exist in all three directions of [001], [101] and [110]. The highest pole density of 1.68 is concentrated in the [001] direction. Location 6 (at the corner of the die cavity) shows an absolutely concentrated texture distribution. Only a single strong texture exists in the [101] direction. Its orientation intensity is the highest among all locations. The highest pole density reaches 4.50.

From the statistical comparison of texture intensity, among the six sampling locations, location 6 has the highest texture intensity in the [101] direction with a pole density of 4.50. Location 4 follows with a pole density of 3.13 in the [101] direction. Both are significantly higher than the other sampling locations. The texture intensities of locations 1, 2, 3 and 5 have small differences. But they are all at a low level overall. The gaps between them are not significant.

This law indicates that the texture intensity of the forging mainly depends on the plastic deformation amount and the dynamic recovery behavior of the material. 7075 aluminum alloy is a high stacking fault energy material. It exhibits strong dynamic recovery capability at the forging temperature of 430 °C. Dynamic recovery can

continuously annihilate dislocations. It slows down the accumulation rate of crystallographic orientation during deformation. This leads to the saturation of texture intensity. It also makes the texture insensitive to the existence of flow boundaries.[52]

The existence of flow boundaries does not have a direct and significant impact on the overall texture intensity. The degree of preferred orientation and intensity of texture are more determined by the local plastic deformation mode and strain path of the sampling location.

In terms of the overall orientation distribution trend of texture, except for location 2, the other five sampling locations all form obvious strong texture distributions in the [101] crystallographic direction. They show consistent orientation preference.

Notably, location 2 is located in the near-stagnation zone where Flow Boundary 2 is located. The accumulated plastic deformation amount during the entire forming process is extremely low. Only overall rigid translation occurs rather than significant local plastic deformation. Therefore, although its texture is diffusely distributed, no significant preferred strong texture is formed in the [101] direction.

This phenomenon further confirms that during the isothermal forging process of 7075 aluminum alloy casing forgings, the directional plastic deformation and preferred rotation of grains mainly proceed along the [101] crystallographic direction. This inherent orientation preference finally makes the overall dominant texture orientation of the forged part be the [101] direction after forming.

This finding clearly reveals the laws of grain preferred rotation and texture evolution of 7075 aluminum alloy during isothermal forging. It also provides a core crystallographic theoretical basis for the subsequent interpretation of the anisotropic behavior of the macro mechanical properties of forgings.

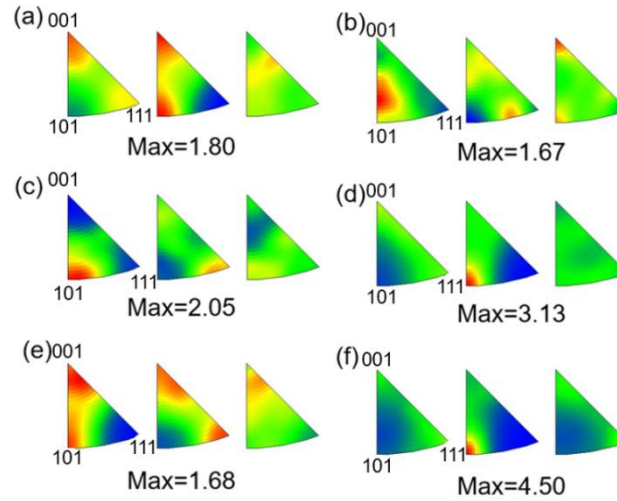


Fig.14 The texture components of the forged casing at different locations, where the red one is Brass texture, the yellow one is Cube texture, the green one is Goss texture, and the purple one is Copper texture. (a) corresponds to location 1, (b) corresponds to location 2, (c) corresponds to location 3, (d) corresponds to location 4, (e) corresponds to location 5, and (f) corresponds to location 6

3.7 The influence of flow boundary on the mechanical properties of forgings

Room-temperature tensile tests were carried out on miniature specimens machined from Locations 1 -6 in the radial, tangential and axial directions in order to evaluate the local mechanical response of the casing forging. The local deformation and microstructure around each sampling location have been characterised in Sections 3.1 -3.4. To explicitly link these factors to the tensile behaviour, Fig. 15 summarises the equivalent plastic strain obtained from FE simulation, the recrystallized fraction, the average grain size and the radial yield strength at the six locations.

As shown in Figure 15, the flow-boundary locations (2 and 3) exhibit the lowest equivalent plastic strain, the lowest recrystallized fraction and the coarsest grains, together with a noticeable reduction in yield strength compared with the high-strain regions (Locations 1, 4 and 6). The transition location (Location 5) shows intermediate

values. These trends indicate that strength degradation near metal flow boundaries is closely associated with the suppressed recrystallization and grain coarsening in low-strain zones. On this basis, the detailed tensile properties in different directions are presented and discussed below.

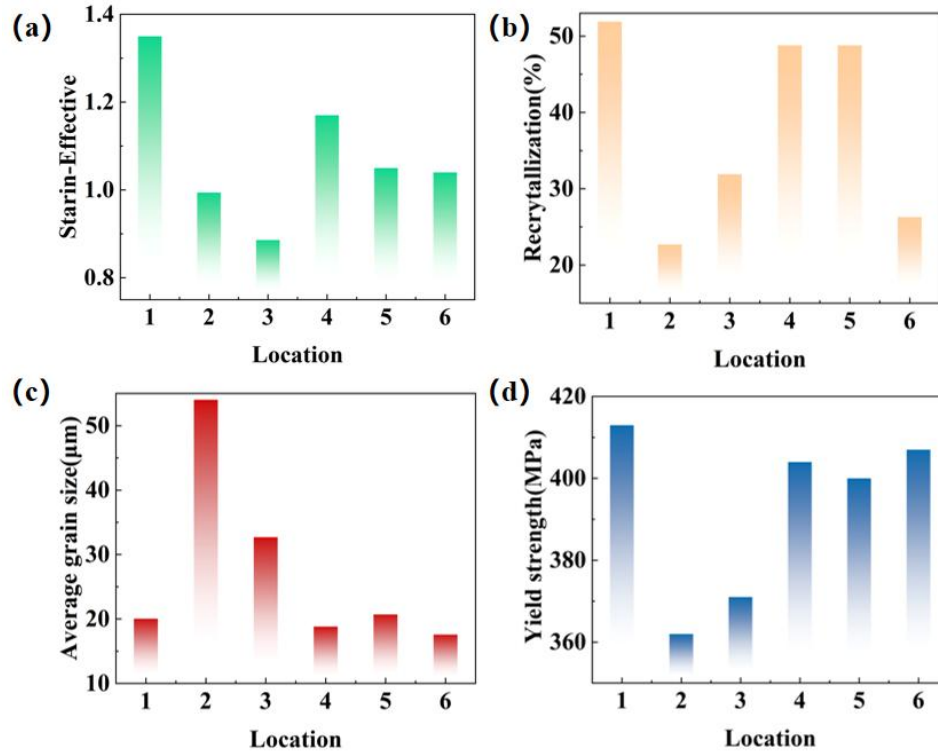


Fig. 15 Correlations between the microstructural parameters and mechanical properties at different sampling locations. (a) Equivalent plastic strain, (b) Recrystallized fraction, (c) Average grain size, (d) Radial yield strength. The results indicate that the flow-boundary regions (Locations 2 and 3) exhibit significant strength degradation due to low strain, suppressed recrystallization, and grain coarsening.

Fig.16 presents tensile curves of the forging along different directions, with corresponding mechanical properties summarized in Tables 5–7. The results reveal a heterogeneous strength distribution across the forging, which is closely related to the microstructural inhomogeneity caused by flow boundaries [40]. Locations 2 and 3 show significantly lower strength values in all three principal directions (radial, tangential,

and axial) compared to other regions: the radial yield strength is 362 MPa and 371 MPa, the tensile strength is 441 MPa and 476 MPa, which are significantly lower than those of non-boundary regions such as Location 1 and Location 4.

This reduction in strength can be attributed to two main factors. First, according to the Hall–Petch relationship:

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

(where “ σ ” is the yield strength, “ σ_0 ” is the lattice friction stress, “ k ” is the strengthening coefficient, and “ d ” is the average grain diameter.), the larger grain sizes at Locations 2 and 3 (see Table 3) directly contribute to their lower strength. The coarse grains at the flow boundary regions reduce the grain boundary density of the material, and the grain boundary strengthening effect is weakened, thus leading to the decrease of yield strength and tensile strength. Second, these locations, influenced by flow boundaries, underwent less deformation and accumulated lower strain, resulting in reduced dislocation density and thus diminished strength. The low strain level at the flow boundary regions leads to a low dislocation density in the material, and the dislocation strengthening effect is not significant, which further reduces the mechanical properties of the material.

Notably, the tangential yield strength at Locations 2, 3, and 5 is considerably higher than that in other directions. This can be explained by the dominant Brass texture component (Table 4), which is known to enhance yield strength along specific orientations. Nevertheless, the forging exhibits minimal directional variation in tensile strength and low mechanical anisotropy overall, indicating relatively isotropic properties in the final component. The combined degradation in strength and ductility, together with direction-dependent variations, suggests a multi-scale coupling between grain size, recrystallization and texture, which is examined in Section 4.3.

(a)

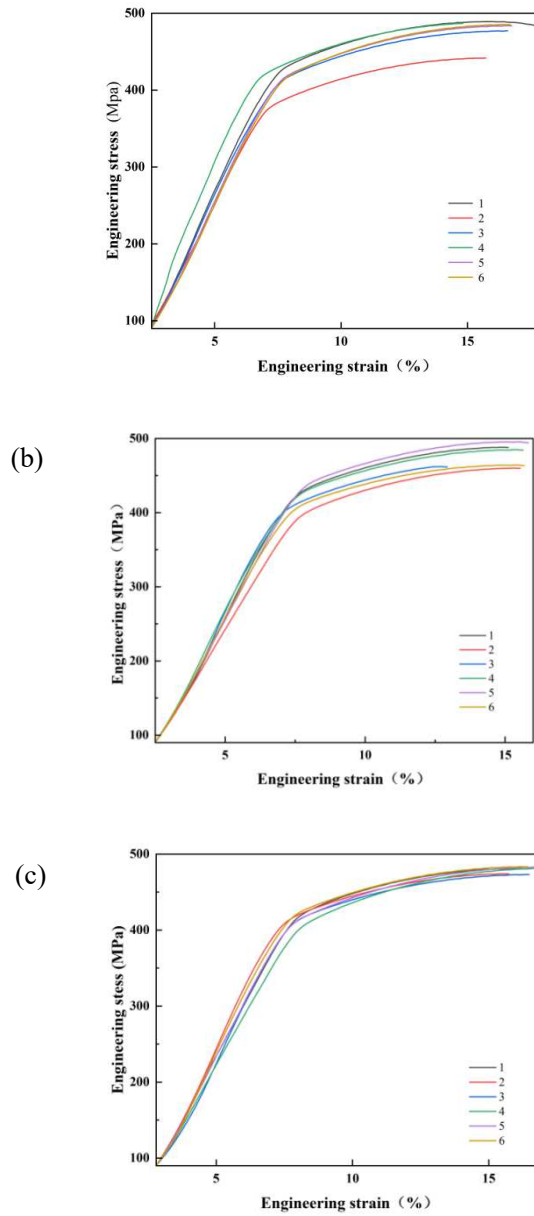


Fig. 16 The stress – strain curves at different locations of the casing forgings were measured by room temperature tension using non-standard tensile specimens, where (a) is the stress – strain curve at different locations radially, (b) is the stress – strain curve at different locations tangential, and (c) is the stress – strain curve at different locations axial

Table 5 Radial mechanical properties of forgings at different locations

Sampling Location	$\sigma_{0.2}$ (MPa)	σ_b (MPa)	δ (%)
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Radial	1	413	489	17.79
	2	362	441	15.87
	3	371	476	16.75
	4	404	487	14.29
	5	400	484	16.65
	6	407	485	16.55

Table 6 Tangential mechanical properties of forgings at various locations

	Sampling Location	$\sigma_{0.2}$ (MPa)	σ_b (MPa)	δ (%)
Tangential	1	403	488	15.07
	2	381	460	15.60
	3	384	462	12.57
	4	401	484	14.99
	5	413	483	15.68
	6	402	481	15.23

Table 7 Axial mechanical properties of forgings at various locations

	Sampling Location	$\sigma_{0.2}$ (MPa)	σ_b (MPa)	δ (%)
Axial	1	414	481	15.60

2	377	474	15.30
3	378	472	16.58
4	399	481	16.61
5	400	482	16.54
6	399	483	16.26

4. Discussion

The findings of this study unequivocally demonstrate that flow boundaries are not merely geometric artifacts but are microstructurally and mechanically significant features in disk-type forgings. Building upon the results presented in Section 3, this discussion delves deeper into the underlying physical mechanisms, connecting macroscopic metal flow, mesoscopic crystallographic plasticity responses, and microscopic microstructural evolution. The analysis is structured around three core themes to provide a comprehensive mechanistic interpretation.

4.1 Formation Mechanism, Strain Localization, and Control over Recrystallization Kinetics at Flow Boundaries

Flow boundaries intrinsically form as shear strain discontinuity surfaces due to the convergence of divergent metal flow paths under the constraints of die geometry and interfacial friction. [25] Our finite element simulations (Fig. 7) visually confirm the evolution of these quasi-stagnant zones, where the net material velocity diminishes significantly. Critically, this kinematic state results in minimal accumulated effective strain while potentially harboring high strain gradients, a condition profoundly detrimental to subsequent recrystallization.[26,27]

The suppressed recrystallization observed at flow boundaries (e.g., Locations 2 and 3) can be quantitatively analyzed using established recrystallization kinetics models. The core relationship lies in the dependence of both the nucleation rate ($\dot{N}$) and the growth rate ($\dot{G}$) of new grains on the local dislocation density (ρ), which is itself a direct function of the applied strain. A widely adopted formulation is:

$$\dot{N} = C_1 \rho \exp(-Q_n/RT) \text{ and } \dot{G} = C_2 I / \rho^{1/2} \exp(-Q_g/RT) \quad (2)$$

where “ C_1 ” and “ C_2 ” are material constants, “ Q_n ” and “ Q_g ” are the activation energies for nucleation and growth, respectively, “ R ” is the gas constant, “ T ” is the absolute temperature, and “ ρ ” is the dislocation density. In flow boundary regions, the drastically low strain leads to a substantially lower initial dislocation density (ρ) after deformation.[26] This has a dual crippling effect: it directly reduces the nucleation rate $\dot{N}$, and it increases the mobility of existing grain boundaries (as pinning forces are reduced), which, paradoxically in this context, favors rapid growth of the few nuclei that do form or, more detrimentally, promotes abnormal grain growth of the pre-existing deformed grains. The scarcity of nucleation sites becomes the dominant limiting factor [28]. Moreover, the low dislocation density in flow boundary regions not only suppresses recrystallization but also reduces the resistance to microvoid nucleation, making these regions preferential sites for damage initiation under external loading. [29]

This mechanistic understanding is further solidified by considering the overall recrystallization kinetics, often described by the Johnson-Mehl-Avrami-Kolmogorov (JMAK) equation for the recrystallized volume fraction X :

$$X(t) = 1 - \exp(-B \cdot t^n) \quad (3)$$

Here, the rate constant B is strongly dependent on the nucleation and growth rates ($B \propto \dot{N}\dot{G}$). The low $\dot{N}$ at flow boundaries results in a drastically reduced B , shifting the recrystallization transformation curve to significantly longer times. [30] Under a fixed industrial heat treatment cycle, these regions consequently fail to complete recrystallization, resulting in the low recrystallized fractions observed (e.g., ~23% at Location 2 VS ~52% at Location 1), as shown in Fig.17. The coarse grain structures are thus a direct consequence of this kinetic stagnation, stemming from the insufficient stored energy (low ρ) to drive efficient nucleation, establishing flow boundaries as critical "weak links" in the microstructural homogenization of the forging [31].

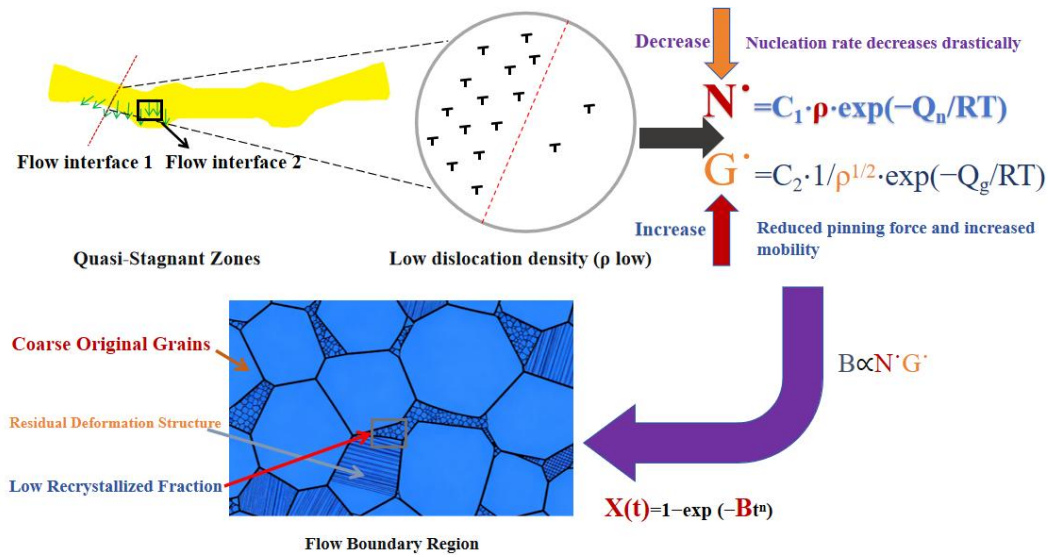


Fig. 17 Recrystallization Suppression and Grain Coarsening Mechanisms at Flow Boundaries

4.2 Deformation Mode Partitioning and Texture Formation Mechanisms

The distinct texture divergence observed between regions—specifically, the prevalence of Copper texture in high-strain areas (Locations 1, 4, 6) and the dominance of Brass texture in flow boundary and transition zones (Locations 2, 3, 5)—is a direct manifestation of the differing plastic deformation paths imposed by the

global metal flow. This phenomenon can be fundamentally understood through the lens of crystal plasticity. For high-stacking-fault energy aluminum alloys like 7075, deformation is primarily governed by wavy slip, [32] and the activation of specific slip systems under different strain paths dictates the final crystallographic orientation. [33] Figure 18 illustrates the deformation mechanism in high-strain regions. High-strain regions experience a deformation mode approximating uniaxial compression with substantial lateral spreading, as shown in Figure 15. This condition promotes polyslip, where multiple slip systems are activated simultaneously, naturally leading to the development of the Copper texture ($\{112\}\langle 111 \rangle$), a common component in compressed FCC metals. [34]

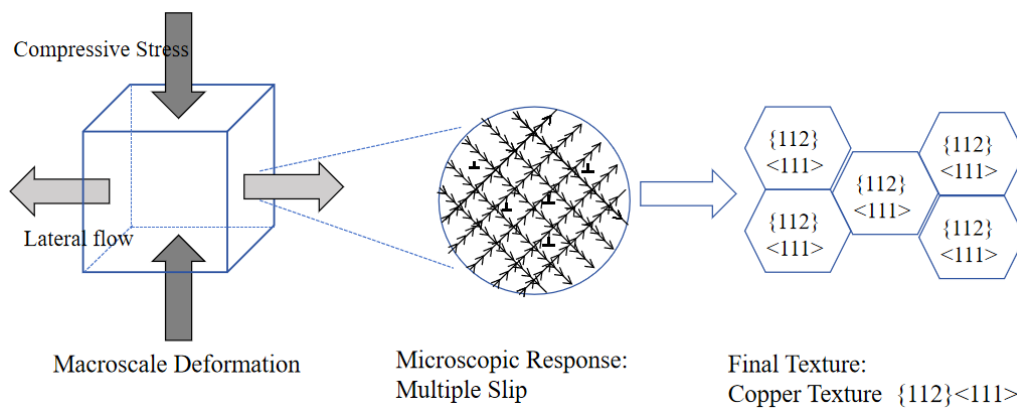


Fig. 18 Deformation mechanism in a high-strain region. Macroscopic uniaxial compression and lateral flow activate polyslip, leading to grain rotation and the development of a strong Copper texture.

In contrast, the mechanical role of locations adjacent to flow boundaries is fundamentally different. These regions, including Location 5 which acts as a deformation coordinator, are subjected to severe constraints that restrict lateral flow, imposing a deformation state much closer to plane strain. Figure 19 illustrates the deformation mechanism in low-strain regions. Under plane strain conditions, the

requirement for deformation compatibility is met by the activation of a more limited set of slip systems. Crystal plasticity models, such as the Taylor theory, indicate that this specific constraint strongly favors the rotation of grains into the Brass orientation ($\{110\}\langle 112\rangle$). This mechanism has been experimentally confirmed in FCC-based medium-entropy alloys under identical plane-strain deformation conditions, where restricted lateral flow limits the activation of polyslip systems and drives selective grain rotation [35].

EBSD characterization results show that the Brass texture intensity in the flow boundary region reaches 6.8 times random, which is significantly higher than the 2.3 times random observed in the central matrix region. This strong texture anisotropy leads to pronounced mechanical anisotropy, with the tensile strength along the radial direction being approximately 15% higher than that along the axial direction. The high texture intensity also indicates a high degree of grain alignment, which further enhances the tendency for planar slip and localized deformation.

The strong Brass texture in the flow boundary region indicates a highly constrained deformation state, where only a limited number of slip systems are activated. This leads to severe dislocation pile-ups at grain boundaries and second-phase particles, further promoting the formation of localized shear bands. Both first-principles calculations and experimental observations have confirmed that such constrained deformation conditions significantly increase the energy required for dislocation cross-slip, enhancing the tendency for planar slip and dislocation pile-up formation at heterogeneous interfaces [36]. High-resolution characterization has further

revealed that misfit dislocations formed at high-lattice-mismatch interfaces (such as $\text{Al}_2\text{Cu}/\alpha\text{-Al}$) act as additional dislocation sources, accelerating the accumulation of dislocations and the subsequent initiation of shear bands [37]. These shear bands act as preferential paths for crack propagation, which is consistent with the observation that most cracks in the fractured samples initiate and propagate along the flow boundary regions.

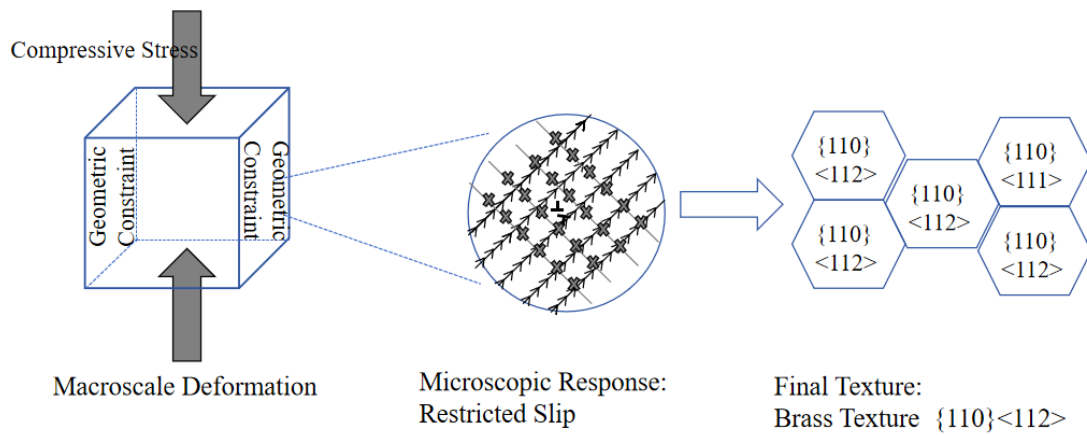


Fig. 19 Deformation mechanism in a low-strain region. Restricted lateral flow produces a plane-strain dominated deformation state, which limits polyslip activation and promotes selective grain rotation toward the Brass texture.

A critical observation is that while the texture type is strongly dictated by the local strain path, the texture intensity does not show a systematic correlation with the presence of a flow boundary. This decoupling suggests that the magnitude of accumulated strain, while crucial for microstructural features like grain size, is not the sole determinant of orientation sharpness. The intense dynamic recovery characteristic of 7075 aluminum alloy at forging temperatures likely plays a balancing role. It may continuously annihilate dislocations, thereby moderating the rate of orientation accumulation during deformation [38]. This process could lead to a saturation of texture intensity, making it less sensitive to the large variations in net strain observed between different locations. Thus, the final texture pattern emerges from a complex

interplay between the macroscopic strain path (defining the type) and the intrinsic metallurgical recovery processes (moderating the intensity).

4.3 Mechanical Property Degradation and Anisotropy under Multi-Scale Coupling Mechanisms

The degradation of mechanical properties in flow boundary regions results from the coupling of multi-scale factors at the micro- and meso-scales. At the micro-scale, the Hall-Petch effect is the dominant mechanism for property reduction. The coarse grains at Locations 2 and 3 directly lead to their lower yield strength in all directions compared to fine-grained regions. At the meso-scale, the dominant Brass texture introduces anisotropy in mechanical properties. The specific orientation of the Brass texture results in a higher Taylor factor along the Tangential direction, meaning that initiating plastic deformation in this direction requires a higher resolved shear stress, manifesting as a relative increase in tangential yield strength (see Table 6). This explains why Locations 2, 3, and 5 exhibit comparable or even higher tangential strength than regions with better overall directional performance (Location 1), despite their generally lower strength. This texture-induced anisotropic strengthening effect partially compensates for the strength loss due to grain coarsening but also leads to significant directional inhomogeneity in the mechanical response.

The low recrystallized fraction in flow boundary regions implies the retention of more deformed structures, containing a higher density of dislocation tangles and substructures. Although these deformed structures can provide some strengthening, their ductility and toughness are typically inferior to those of fully recrystallized

equiaxed grains [39]. We hypothesize that this ductility reduction may be further exacerbated by incipient damage formation during the plane-strain-dominated deformation process: the restricted slip system activation under plane strain conditions could lead to localized stress concentration at grain boundaries and second-phase particles, potentially promoting microvoid nucleation even at relatively low macroscopic strains [35,40]. This possible damage-assisted ductility reduction remains a hypothesis because void area fraction and crack-initiation statistics were not quantified in the present work.

Furthermore, quasi in-situ EBSD observations and phase field crystal simulations have demonstrated that the geometric morphology of grain boundaries plays a decisive role in subsequent crack propagation [40]. For flow boundaries, which are typically curved high-angle grain boundaries with local serrated features, the angle between the boundary segment and the loading direction determines the crack propagation path: segments perpendicular to the loading direction act as preferential paths for intergranular crack propagation, while segments parallel to the loading direction can effectively hinder crack extension by forming Lommer–Cottrell dislocation locks at the boundary bends [41]. Consequently, flow boundary regions exhibit comprehensive degradation in both strength (primarily reduced by grain coarsening) and ductility (attributed to deformed structures and potential incipient damage), superimposed with strong mechanical anisotropy (governed by texture). This complex property state makes these regions potential preferential sites for crack initiation and propagation under complex loading conditions, posing a latent threat to the overall service reliability of

the forging.

5 Conclusion

This study investigated the influence of metal flow boundaries on the microstructure and tensile properties of disk-shaped 7075 aluminum alloy casing forgings through finite element simulation and experimental characterization. The main conclusions are as follows:

(1) Two distinct flow boundaries form during isothermal forging, controlled by die geometry, metal flow direction and interface friction. Flow Boundary 1 forms at the interface between central and peripheral metal flow and stabilizes near the central boss, while Flow Boundary 2 forms in the central concave region due to frictional constraints. These boundaries create low-strain regions (equivalent plastic strain < 0.5) that induce significant microstructural and mechanical heterogeneities.

(2) Flow boundary regions exhibit coarse grains ($54.02\ \mu\text{m}$ and $32.70\ \mu\text{m}$) and low recrystallization fractions (22.7% and 31.9%), which are 2-3 times larger and about half of those in non-boundary regions, respectively. This leads to a 40-50 MPa reduction in radial yield strength compared to non-boundary regions. Flow boundaries have no significant effect on phase distribution, with Al-Cu-Mg and Al-Mg phases uniformly distributed in a banded morphology.

(3) Flow boundaries determine the texture component type: flow boundary and transition regions are dominated by Brass texture (9.89%-12.3%), while high-strain regions are dominated by Copper texture (15%-27%). The Brass texture causes higher tangential yield strength, resulting in mild mechanical anisotropy. However, flow

boundaries have negligible effect on texture intensity due to the strong dynamic recovery of 7075 aluminum alloy at 430 °C.

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