Microstructures and mechanical properties of dispersoid-strengthened 6082 aluminum alloy plate produced by vacuumed centrifugal casting

Xiaoming Qian¹, Bosi Zhang¹, Zhaodong Wang¹*, Yong Li¹,²*, Jiadong Li¹

¹ State Key Laboratory of Rolling and Automation, Northeastern University, Shenyang, 110819, China
² Guangxi Advanced Aluminum Processing Innovation Center Co. LTD, Nanning, 530007, China
*Correspondence author: zhwdwang@mail.neu.edu.cn; liyong@smm.neu.edu.cn

Abstract: A dispersoid-strengthened 6082 alloy was prepared using vacuum centrifugal casting. The casting pipe was split into two halves along the axial direction, then every single half of the pipe was flattened and was treated as a slab and was subsequently generated to the conventional rolling to produce plate. The microstructure and the mechanical properties of the inner, middle, and outer layers of the pipe was studied. The results shows that the intermetallic in the as-cast microstructure was gradient distributed, the number density of intermetallic decreased and the average size increased from outer layer to inner layer. After heat treatment, area fraction of the α-Al(Fe, Mn)Si dispersoid zone and the number density of dispersoid increased gradually from outer layer to inner layer. After rolling, the high temperature strength of the plates decreased and the elongation decreased from the outer layer to the inner layer.

1. Introduction

Aluminum and its alloys have the advantages of low density, high specific strength, good thermal conductivity, and good corrosion resistance. It has been widely used in aerospace, transportation, and industrial departments. Conventionally, the forming processes of wrought aluminum alloy mainly include casting, and deformation such as extrusion, rolling, stamping. Among these processing, rolling is one of the most economical and effective methods for producing plates.

Industrially, the aluminum plates are generally produced by direct chill (DC) casting with subsequent rolling. Since the DC casting of aluminum alloy was successfully introduced into the aluminum industry in the late 1930s, it has been one of the pillars of the aluminum industry[1,2]. Despite the DC casting technology is simple to prepare aluminum alloy plate, the issues such as shrinkage cavity, oxide inclusion and coarse core grain in large-sized flat ingot are still limiting the further improvement of performance[3].

Centrifugal casting is mainly used to produce discs and pipes, and can also be used to produce special-shaped parts such as impellers, seamless pipe blanks. Because of its simple casting process, high utilization rate of molten metal and fine microstructure of castings, it has been widely in recent years. Centrifugal casting process can produce thin-walled cast pipe parts, and the size of cast pipe is determined by the inner diameter of the mold. Research shows that centrifugal force can not only refine the grain, but also strengthen the feeding of aluminum alloy products by centrifugal casting[4,5]. The vacuum pouring technology, which was firstly used in 1980s, was carried out to produce castings.

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with very complex shapes and part with thin thickness that no shrinkage cavity and porosity could be generated.[6].

Theoretically, the mechanical properties could be improved if the DC casting ingot is replaced by centrifugal casting pipe blank with better as-cast microstructures. However, the actual problem is that rolling could be performed on the DC casting (usually rectangular), but not on the centrifugal casting pipe blank. Therefore, the problem hereby became how to switch the pipe to a plate or a slab. A novel idea is here by brought up by the authors of this study: to split a casting pipe into two halves along the axial direction, then flatten every single half of the pipe, and the flattened pipe could be treated as a slab and be subsequently generated to the conventional rolling to produce plate/sheet (Fig. 1). In this way, slabs produce from centrifugal casting were obtained and the microstructural advantages of centrifugal casting could be reserved.

Fig. 1 Schematic of the process switch the cast pipe to the slab used for rolling

A novel dispersoid-strengthened 6082 alloy was reported in a study carried out by Qian [7], α-Al(Fe, Mn)Si exhibited outstanding potential on improving the mechanical properties of aluminum alloys at elevated temperatures. In the 6082 alloy with 0.5Mn addition, α-Al(Fe, Mn)Si dispersoids started to precipitate at 350 °C and reached the highest number density at 400–450 °C. Correspondingly, the microhardness of the alloy increased substantially. When the heat treatment temperatures were over 500 °C, the dispersoids coarsened, and the number density sharply decreased with time. Accordingly, the microhardness dropped severely, indicating the loss of the dispersoid strengthening effect. Li [8] conducted a quantitative study on the strengthening effect of α-Al(Fe, Mn)Si in Al−Mn−Mg−Si alloys, and the results showed that after heat treatment at 375 °C for 24 h, α-Al(Fe, Mn)Si dispersoid strengthening was the predominant strengthening mechanism for Al−Mn−Mg−Si alloys. The quantitative dispersoid contribution to the yield strength reached as high as 49% among all the strengthening mechanisms (Al matrix, solid solution strengthening and dispersoid strengthening). In a recent study by the author of this study [7], the activation energies for hot deformation of 6082 alloys increased considerably from 191.2 kJ/mol for dispersoid-free alloys to 315.4 kJ/mol for α-Al(Fe, Mn)Si dispersoid-strengthened alloys owing to the dispersoid strengthening effect.

In this study, centrifugal casting was applied to prepare dispersoid-strengthened 6082 aluminium alloy. Microstructure of inner, middle, and outer layer of the casting pipe was analyzed, and the mechanical properties was tested at elevated temperature. The focus was placed on the different microstructure and mechanical properties at the inner, middle, and outer layer of the pipe. The primary of the innovation of this study is to switch a casting pipe to a slab that used for rolling, and the exploration of combining the advantages of vacuum centrifugal casting to the traditional rolling.

2. Experimental
Investigations were conducted using an Al-0.8Mg-1.0Si-0.75Fe-1.0Mn 6082 aluminium alloy. Vacuum centrifugal casting was performed on a self-developed equipment (Fig.2). Smelting was conduct in an induction heating furnace. The smelting and casting were performed under an air
pressure less than 10Pa. The temperatures of smelting and casting were set as 730 and 690℃, respectively, the rotation speed of the mold was set as 850r/min.

![Fig. 2 Schematic of the experimental process](image)

A vacuum centrifugal cast pipe blank with a length of 250 mm, an outer diameter of 200 mm and a thickness of 25 mm was prepared. The cast pipe blank is subjected to homogenization heat treatment at 425℃ for 6 hours. The homogenized pipe was cut along the wall direction and was divided into 3 layers, the thickness of each layer was 8mm, as is shown in Fig. 3. After then, every layer of the pipe was split into two halves along the axial direction, and every single half of the pipe was then flatten, the flattened pipe hereby was treated as a slab and be generated to the rolling and plates with a 1.5mm thickness were obtained.

![Fig. 3 Sampling schematic diagram](image)

DMI5000M optical microscope (OM), SSX-550 scanning electron microscopy (SEM) and FEI G2 F20 transmission electron microscopy (TEM) were used for microstructural observation. The tensile test at high temperature was carried out on CSS-44100 universal material tensile machine, the temperature of the tensile test was 300℃, the strain rate was 0.001s⁻¹.

3. Results

3.1 Microstructures

3.1.1 As-cast

Fig. 4 shows the as-cast microstructure of experimental alloy, in which (a), (b) and (c) of represents the microstructure of outer layer, middle layer and inner layer respectively. It can be seen that the as-cast structure was consisted of aluminum matrix and intermetallic compounds distributed along dendrite boundary. The intermetallic compounds present plate-like and irregular skeleton-like morphology, which were the commination of AlFeMnSi and primary Mg2Si[9-11]. Quantitative analysis was performed on the intermetallics and the results are shown in Fig. 5. The area fraction of
The intermetallic in cast pipe blank gradually decreases from 4.83% at the outer layer to 4.12% and 3.93% of the middle and inner layer. The equivalent length of the intermetallics gradually increases from 4.57μm of the outer layer to 6.58μm and 9.98μm of the middle and the inner layers.

![Fig. 4 Optical microstructure of cast pipe blank (a) outer layer (b) middle layer (c) inner layer](image)

**Fig. 4** Optical microstructure of cast pipe blank (a) outer layer (b) middle layer (c) inner layer

![Fig. 5 Number density and mean size of inter-metallic compounds in different layers](image)

**Fig. 5** Number density and mean size of inter-metallic compounds in different layers

### 3.1.2 After homogenization

Fig. 6 shows the microstructure of middle layer after homogenization at temperature of 425°C for 6h. Comparing with Fig. 4, it can be found that, the intermetallics remained nearly unchanged, but at the polished surface of the sample, large amount of black spots emerged, as indicated by the dot circle.

![Fig. 6 Microstructure after homogenization heat treatment](image)

**Fig. 6** Microstructure after homogenization heat treatment

The samples of after homogenization were etched with 0.5%HF for 40s to further reveal the detail of the microstructures and images are shown in Fig. 7, in which (a), (b) and (c) are representing the OM images of outer, middle, and inner layers, respectively, the (d), (e) and (f) stand for the SEM
photographs of outer layer, middle layer and inner layer of alloy respectively. It can be seen that a large number of fine particles emerged inside the dendrites, and these particles were confirmed as α-Al(Fe, Mn)Si dispersoids in the references[12,13]. Non-uniform distribution of the dispersoids were observed, the dispersoids were mainly distributed inside the dendrites, leaving the regions of near dendrites arms the precipitation free zone. The SEM images of Fig. 7d, e, f shows the closer observation images, the morphology of the dispersoids shows the plate-like or the short rod-like shapes in the two-dimensional image. Fig. 8(a) shows the TEM image of α-Al (Fe-Mn) Si sample, the size of the dispersoids around 20-50nm. Fig. 8b shows the quantitative results of precipitation area fraction and number density based on the analyzing of SEM images. The area fraction of dispersoid gradually increases from 60.7% of the outer layer to 70.1% of middle layer and 74.2% of inner layer. The quantity number density of dispersoid gradually increased from 5.4μm^{-2} in the outer layer to 5.9μm^{-2} in the middle and 7.3μm^{-2} in the inner layer respectively.

Fig. 7 Etched surface of homogenized microstructures of OM (a) outer layer, (b) middle layer samples, (c) inner layer; and the SEM image of (d) the outer layer, (e) the middle layer, and (f) the inner layer.

![Fig. 7](image)

Fig. 8 (a) TEM image of homogenized sample of middle layer, (b) quantitative data of area fraction of precipitation zone and α-Al(Fe, Mn)Si quantity number density after homogenization

3.2 Mechanical properties
The tensile test was performed at 300℃ with a strain rate of 0.001s^{-1} and the results are shown in Fig. 9. The yield strength gradually increased from 41.2MPa in the outer layer to 45.2MPa in the middle layer and 54.0MPa in inner layer; The ultimate strength gradually increased from 56.5MPa in the outer
layer to 60.8MPa in the middle layer and 65.6MPa in inner layer; The elongation decreased gradually from 54.5% in the outer layer to 49.0% and 48.0% in the middle and inner layers.

![Graph showing yield strength, ultimate strength, and elongation for different layers.](image)

**Fig. 9** Yield strength, ultimate strength, and the elongation of the different layer of the rolled plates

### 4. Discussion

#### 4.1 Microstructure

During centrifugal casting, the feeding of the melt was significantly improved under the action of centrifugal force. Casting defects such as shrinkage porosity and shrinkage cavity could be effectively avoided (Fig. 4). The solidification proceeded from the inner wall of mold to the inner surface of cast pipe step by step. The directional solidification resulted in the different features of the casting pipe in the different layers.

Centrifugal force and the cooling rates could be attributed to the differences between different layers. Along with the centrifugal casting, the centrifugal force was reduced gradually owing to the less liquid was left which was generating reduced centrifugal force. The pressure performed on the solid-liquid interface was accordingly decreased, which was believed to be one of the coarsening tendencies for the intermetallics from the outer to the middle and outer layers (Fig. 4). At the beginning of the casting when the melt with certain overheating encountered the mold, the heat dissipation direction of was from inside to outside along the radial direction the metal mold. With the decreasing temperature, the temperature of the outer layer of melt firstly dropped below the eutectic point, the intermetallics formed. Along with the on-going solidification, the cooling rate decreased gradually, which brought a consequence of the coarsened intermetallics. In addition, the Mg, Mn, Fe and Si were released from the solid-liquid interface continually due to the eutectic reaction, which resulted the increased solid solution levels from the inner to the middle and outer layers. During the subsequent homogenization stage, more dispersoids of α-Al(Fe, Mn)Si dispersoids was formed at the inner layer (Fig. 8) due to the overall more sufficient alloying elements storage in the supersaturated solid solution, which provided a stronger driving force for the dispersoids precipitation for the inner layer than that of the out layer.

#### 4.2 Mechanical properties

The traditional hardening mechanism of 6082 aluminum alloys is through the precipitation of fine nano-scale Mg2Si precursor precipitates to attain superior mechanical properties at room temperature [14]. However, for a service temperature exceeding 200 °C, the mechanical properties deteriorate rapidly owing to the coarsening and dissolution of Mg2Si precursors [15]. In this study, α-Al(Fe, Mn)Si dispersoids were introduced as the strengthening particles, the alloy was thermal stable at 300°C.
The increased yield strength and ultimate strength from the outer layer to the inner layer (Fig. 9) was closely related to the dispersoids number density and its precipitation area fraction. In Fig. 7 and 8, the dispersoids of outer layer was in highest number density and the highest area fraction. During the tensile test, the dislocations were retarded at the points where they encountered the dispersoids, and thus the dispersoids acted as stronger barriers to deformation. Furthermore, the piled-up dislocations during the deformation, were strongly pinned by the dispersoids, resulting an increased strength of the materials [7]. Therefore the overall higher strength of the inner layer than the middle and outer layer was attributed to the stronger pinning effect of the dispersoids on dislocation slips.

A better ductility was achieved in the outer layer than that of the middle and inner layer (Fig. 9). The elongation of tensile tests is usually closely related to the large heterogeneous particles in the deforming matrix. During plastic deformation, the brittle coarse particles were not able to adjust the deforming matrix, where stress concentrations formed and accumulated in the microneighbour of coarse particles. Along with the progressive deformation, once the deformation in the coarse particle region exceeded the endurance limit of the Al matrix, microcracks emerged [16]. The larger the size of the particle is, the easier it is for microcracks to emerge.

In Fig. 4 and 5, intermetallic particles are larger at the inner layer than the middle and outer layers, which were remained unchanged during and homogenization (Fig. 6). Thus, less severe stress concentration could be generated in the outer layer than those of middle and inner layers. During hot deformation, the large eutectics in the inner layer led to an increased risk of cracks and their growth; therefore, under the same strain conditions, fracture took place earlier in the inner layer than those in middle and outer layers.

5. Conclusion
A dispersoid-strengthened 6082 alloy was prepared using vacuum centrifugal casting. The casting pipe was split into two halves along the axial direction, then every single half of the pipe was flattened and was treated as a slab and was subsequently generated to the conventional rolling to produce plate, the microstructure and the mechanical properties of the inner, middle, and outer layers of the pipe was studied. The following conclusions were drawn:

1. The intermetallic in the as-cast microstructure was non-uniformly distributed, the number density of intermetallic decreased and the average size increased from outer layer to inner layer. The cooling rate and the centrifugal force during the casting was attributing to the nonuniform microstructures.

2. After heat treatment, area fraction of the α-Al(Fe, Mn)Si dispersoid zone and the number density of dispersoid increased gradually from outer layer to inner layer.

3. After rolling, the high temperature strength of the plates decreased and the elongation decreased from the outer layer to the inner layer, which was resulted from the corresponding increased area fraction of the α-Al(Fe, Mn)Si dispersoid zone and the number density of dispersoid.

Acknowledgements
The authors gratefully acknowledge the support of the National Key R&D Program of China (No. 2017YFB0306400), the National Natural Science Foundation of China (Grant No. 51790485), the Fundamental Research Funds for the Central Universities (No. N2007008), Nanning science and technology major special projects (No. 20191002), the Key research and development project of Shandong province (the first batch of competitive selection of major scientific and technological innovation projects, No. 2019JZZY010401), the Fundamental Research Funds for the Central Universities (No. N2007005).

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