Article

Effects of On-Line Vortex Cooling on the Microstructure and Mechanical Properties of Wire Arc Additively Manufactured Al-Mg Alloy

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Abstract: A novel on-line vortex cooling powered by low-cost compressed air was proposed to reduce common defects such as low forming precision, coarse grains, and pores caused by heat accumulation in the Wire Arc Additive Manufacturing (WAAM) of aluminum alloy. The impacts of interlayer cooling (IC), substrate cooling (SC), on-line cooling (OL), and natural cooling (NC) processes were compared on the morphology, microstructure, and mechanical properties of as-deposited walls, revealing that the OL process significantly lowers the interlayer temperature and improves forming precision. The high cooling rate produced by the OL process reduced the absorption of hydrogen in the molten pool, lowering porosity. Furthermore, the grains are refined due to the developed undercooling. However, the high cooling rate enhanced the segregation potential of Mg element and raised the content of the β phase. Conclusively, the maximum tensile strength, elongation, and microhardness of the as-deposited wall are achieved via the OL process, and the fine-grain strengthening mechanism plays an important role in improving mechanical properties. The OL process is cheaper and poses a significant effect; it is highly suitable for the additive manufacturing of complex components compared with other forced cooling processes.

Keywords: wire arc additive manufacturing (WAAM); on-line vortex cooling; microstructure and mechanical properties

1. Introduction

The newly introduced digital additive manufacturing technology works by depositing metal parts in layerwise [1]. Since it is less costly and saves time [2], many people have increasingly regarded additive manufacturing as a solution to meet personalized customization. In recent years, compared with Laser Depositing Manufacturing (LDM) [3] and Selective Laser Melting (SLM) [4], Wire Arc Additive Manufacturing (WAAM) has attracted far more attention in the field of additive manufacturing, for example, in the aerospace field [5,6]. This is attributed to its lower initial cost, higher forming efficiency, and higher material utilization [5,6].

With the rapid advancement of WAAM, a wide range of studies on WAAM aluminum alloy have been initiated, where a variety of aluminum alloy wires have been adopted as the feeding materials of WAAM, Al-Mg alloys, among others [7]. In particular, Al-Mg alloys are employed in aerospace because they resist corrosion and have efficient welding properties. However, the melting point of aluminum alloy is low (~ 660 °C), but its fluidity is satisfactory and it is easily overheated between

layers during the additive process. This results in poor forming precision, coarse structure, degradative property, and occurrence of pores.

A study by Derekar et al. [8] on the influence of different interpass temperatures through controlling the substrates preheated temperatures, 50 and 100 °C for wall-1 and wall-2, revealed that the deposits had increased number of pores at high interpass temperature. Ren et al. [9] studied interpass cooling (200, 160, 120 and 80 °C) of Al–6.3Mg alloy which adopted the cold metal transfer advanced process during WAAM process. The results demonstrate that reducing the interpass temperature effectively improved the mechanical properties and reduced the difference between the horizontal and vertical properties of the deposits. Besides, Geng et al. [10] studied the influence of different interpass temperatures (50, 80, 120 and 150 °C), whereby they found that extreme high temperatures caused inconsistent layer cross-sectional dimension during deposition. With the difficulty of continuous forming of the aluminum alloy due to thermal accumulation, the control of heat dissipation is a scheme to ensure the depositing quality of WAAM. Relevant scholars have conducted some research on this drawback. For example, Henckell et al. [11] used a mixture of nitrogen and hydrogen at a flow rate of 30 L/min to continuously cool the surface of the deposits during S355 steel depositing to enhance the heat dissipation ability and improve the forming ability of WAAM. The cooling nitrogen and hydrogen are expensive but do not refrigerate effectively in a flow rate of only 30 L/min; thus, the property of the deposits improved unsatisfactorily. Henckell et al. [12] also studied the energy input in WAAM with Gas Metal Arc Welding (GMAW), whereby they reduced welding energy input to lead to an adaptation of geometrical and microstructural features of additively manufactured work pieces. Su et al. [13] adopted WAAM to deposit Al-Mg alloy, which, based on cold metal transfer (CMT) technology, is where they used to manufacture Al-Mg alloys with different heat inputs by adjusting wire feed speed (WFS) and travel speed (TS). However, properties show no significant change with different TS. Wu et al. [14,15] adopted gas tungsten arc welding based WAAM (GT-WAAM) depositing TC4, where they set a gas cover on the surface of the deposits and achieved an accelerated deposition efficiency and a reduction in the oxidation of titanium alloy. Another study by Li et al. [16] assessed the influence of different cooling processes on geometric accuracy using a thermoelectric cooling device to cool the wall. Usually, the thermoelectric cooling device relies on conduction heat dissipation, and though the cooling effect in their experiment was satisfactory, the flexibility and adaptability were poor, it was unsuitable for different deposits accumulated layer by layer.

In this study, to refine the grain, ensure the forming quality and improve the forming efficiency, vortex tube and water-cooling substrate were performed in WAAM. The geometric shape, microstructure, and mechanical properties of deposits were investigated. We analyzed the influence on the results and assessed the reliability of the control WAAM.

2. Experiment

2.1. Experiment Setup

The apparatus comprised a FUNIUS TPS4000 power source and a welding torch (FUNIUS Inc., Pettenbach, Austria), water cooling unit (SHENY Inc., Shenyang, China), and welding robot (KUKA Inc., Augsburg, Germany), as shown in Figure 1. An infrared thermometer was used to measure the interlayer temperature between the deposition of layers. The feeding material used in this study included ER5356 aluminium alloy welding wire with a diameter of 1.2 mm, deposited on a 5052 aluminium alloy substrate with dimensions of 200 mm × 200 mm × 20 mm. The chemical compositions of feeding wire and structure are highlighted in Table 1.

Figure 1. Wire Arc Additive Manufacturing (WAAM) deposition system: (a) schematic diagram, (b) physical diagram.

Table 1. Chemical composition of ER5356 and 5052Al (wt. %).

Composition	Mg	Mn	Cr	Cu	Zn	Fe + Si	Al
ER5356	4.5–5.5	0.05–0.2	0.15–0.35	0.1	0.1	0.5	Bal.
5052	2.2–2.8	0.1	0.05–0.2	0.1	0.1	0.45	Bal.

Four straight wall structures were as-deposited at interlayer time 90 s using the WAAM system with four different cooling processes. The welding parameters are shown in Table 2. Welding-grade argon (99.995% purity) was applied as the shielding gas for the WAAM torch.

Table 2. Process parameters for WAAM deposition.

Process Parameters	Details	Parameter
Deposition power	Current	160 A
	Arc voltage	23 V
Speed	Travel speed	0.022 m/s
	Wire feed speed	11 m/min
Distance and angle	Contact tube to workpiece	12 mm
	filler wire to workpiece	90°
Flow rate(argon)	WAAM torch	24 L/min
Deposition power setting	MIG/MAG	-

2.2. Fast Cooling System

To evaluate the effectiveness of the fast cooling system, a commercially available vortex tube and a water-cooling base were fitted to the WAAM torch as shown in Figure 2. With an air compressor (YOU MAN Inc., Shanghai, China), a stable pressure-air stream was delivered in the vortex tube, and the high pressure airflow in the vortex tube generated heat transfer, instantly producing low-pressure low-temperature gas, as the low-pressure high-temperature gas is ejected from both ends of the vortex tube. The vortex tube parameters are highlighted in Table 3.

Figure 2. Vortex cooling system: (a) physical diagram, (b) schematic diagram.

Table 3. Process parameters of the vortex tube.

Parameters Unit	Air Pressure Bar	Gas Consumption Standard Liter Per Minute (SLPM)	Cooling Capacity Kcal/hr
Details	6.9	1410	857

Cooling air was oriented to the side of the as-deposited wall while the stable 20 °C water flowed through the water-cooled constant. Four cooling processes were specifically evaluated, that is, turning on the vortex tube after the complete deposition of each layer (interlayer cooling (IC)), keeping the vortex tube opened during the deposition process (on-line cooling (OL)), keeping the cooling water flowing inside the water-cooled constant during the deposition process (substrate cooling (SC)), depositing with natural cooling (NC). Whilst the compressed air enters the vortex chamber, the compressed air at room temperature transforms into high-temperature gas and cooling gas. The cooling gas is transferred to the shunt. The shunt fractionates the cooling gas into four proportions and squirts in parallel from the duck-billed nozzles of the four branches. A 10-mm distance between each nozzle maintains continuous cooling to prevent excessive arc heat from causing surface ablation of the nozzles during the additive process. The linear distance between nozzles and the as-deposited wall is 20 mm and the angle between the nozzles and the depositing wall is 20°. Besides, to avoid the negative effects of cooling gas on the OL process, the OL process was conducted after 6 layers. The previous 6 layers were deposited via the NC process. To verify that the cooling air did not negatively impact on the molten pool, the molten pool was monitored using a high-speed camera. From Figure 3, it is evident that there was no wandering of the arc, and the molten drop transited smoothly, the transition mode of the droplet was projected transfer and the droplet transferred to the molten pool without spatter. At the same time, the “bell jar” shaped arc remained stable. Furthermore, the WAAM process was not affected by cooling air via the OL process.

Figure 3. Image of the molten pool taken by a high-speed camera during the on-line cooling (OL) process.

2.3. Porosity

The WAAM 5356 aluminum alloy is mainly associated with porosity defects. The forming quality of the deposits could be effectively distinguished by examining the number and size of pores. In this study, we used the Olympus GX51 metallographic microscope (Olympus Inc., Shinjuku, Tokyo, Japan) to observe pore distribution, while Image-Pro Plus software (MEDIA CYBERNETICS Inc., Rockville, MD, USA) was used to measure the pore size and number in metallographic specimens; thus, the porosity of the sample was calculated.

2.4. Material Characterization Techniques

Both metallographic specimens and mechanical test specimens were extracted from the as-deposited 5356 walls for further assessment. The metallographic specimens were ground and polished to a mirror finish. The polish process was conducted mechanically and electrochemically. The specimens were etched for 10 s in an electrolyte containing 10% vol HClO_4 with balance $\text{C}_2\text{H}_5\text{OH}$, with the decomposition voltage 25 V. Then, the specimens were etched in an anode coating solution containing 1000 mL H_2O , 30 mL HF, and 11 g H_3BO_3 . The anode coating process lasted for 90 s with the decomposition at 20 V. The anode coated specimens were observed using an OLYMPUS GX71 optical microscope (OM) (Olympus Inc., Shinjuku, Tokyo, Japan), while the electrochemically polished specimens were observed using a ZESSIS SEGMA500 field emission scanning electron microscope (SEM) (ZESSIS Inc., Jena, Thuringia, Germany) and the electrochemically polished specimens in the SEM device were observed by an OXFORD energy dispersive spectrometer (EDS) (OXFORD INSTRUMENTS Inc., Oxford, UK). The polished specimens were observed using a MPDDY2049 X-ray diffractometer (XRD) (PANalytical B.V., Almoro, ND, USA). Vickers hardness testing was performed with an HVS—1000A automatic hardness tester (HUAYIN Inc., Laizhou, Shandong, China) using a test load of 200 g and a dwell time of 15 s. Tensile tests were performed at room temperature, with a constant crosshead displacement rate of 0.2 mm/min using an MTS370 universal testing machine (MTS Inc., Eden Prairie, MN, USA). The tensile test specimens and metallographic specimens extracted from the as-deposited walls are shown in Figure 4. According to ISO6892-1:2009, the blank parts in the size of Figure 4 was cut from the thin wall by wire-electrode cutting, and then the tensile specimen meeting the tolerance requirements of ISO6892-1:2009 was obtained by milling and grinding. ISO6892-1:2009 was used to stretch the specimen.

Figure 4. Schematic diagram of metallographic specimens and mechanical test specimens in as-deposited walls.

3. Results and Discussion

3.1. Geometry

At an interlayer dwell time of 90 s, the surface temperature was measured before depositing the next layer. The light spot size of the infrared thermometer we used was 5 mm. We timely measured the current layer before depositing the next layer. During the measurement, we measured the temperature central position of each layer, the temperature curves obtained are shown in Figure 5. With IC, SC, and OL cooling processes, the surface temperature of the as-deposited walls tended to be stable as the number of layers increased. However, the surface temperature increased continuously with the NC process. Furthermore, with the OL process, the temperature stabilized to 30 °C (Figure 5), this was the first cooling process to achieve temperature stability, and the temperature dropped to a minimum compared to other cooling processes. The lower interlayer temperature implies that the deposition process can easily be controlled. Notably, the temperature was recorded after 6 layers because these layers were close to the substrate and heat mainly dissipated as a result of the substrate.

Figure 5. Temperature of each layer deposited by different cooling processes.

The dimensions of WAAM 5356 walls fabricated by different cooling processes are shown in Figure 6. Partial pits were observed at both ends of as-deposited walls and were deposited by IC, SC, and NC processes. Because heat accumulates at the ends of as-deposited walls and significantly surpasses the melting point of the previous layers, there was less support at the ends than the middle position. During the WAAM process, when the wire feed stops, an arc pit is created at the end of the motion path because the arc extinguishes without the feeding wire. No obvious arc pits were observed at either ends of the as-deposited wall that was deposited by the OL process.

Figure 6. Geometry of WAAM as-deposited 5356 walls in different cooling processes.

The thin wall section outline obtained by Adobe PhotoShop is highlighted in Figure 7. Of note, the section is located in the middle of the as-deposited walls. The waviness of the walled side can roughly be seen in Figure 7; however, the waviness of the as-deposited wall deposited by the OL process was significantly smaller than others, and the average thickness was smaller than with other cooling processes. The specific value of thickness change in the height direction is shown in Figure 8. From the analysis, the waviness was 0.5 mm via the OL process and the average thickness was 9.4 mm, which was close to 9 mm of model thickness. The thickness of the as-deposited wall with the NC process did not attain a stable height direction because the temperature was continuously rising as the layer deposited. Heat accumulation was greater than heat dissipation, and the aluminium alloy was spread along via the NC process; thus, the thickness reached 10.8 mm from 9.4 mm. With the increase in the layers, the temperature of the interpass with the NC process continues to rise. According to the solidification characteristics of aluminum alloy, high interlayer temperature cause a low temperature gradient, which was conducive to columnar crystal growth. At the same time, with high interlayer temperature, the aluminum alloy metal has strong fluidity and can easily produce the fusion defect at the edge of the interlayer, while the OL cooling rate is not easy to form with columnar crystal, and, because of the rapid cooling, the defective fusion cannot easily occur at the edge of interlayers.

Figure 7. Section outline of as-deposited 5356 walls by different cooling processes.

Figure 8. Thickness of as-deposited walls in the direction of height deposited by different cooling processes.

3.2. Microstructure

The pore defects on the cross-section of as-deposited walls are shown in Figure 9. The pores are generally distributed near the fusion line in Figure 9a. During the solidification of the molten pool, the solubility limit of the hydrogen atom decreases, H atoms migrate to the high solubility region because the bottom of the molten pool is the previous layer in the WAAM process. The metal close to the bottom of the molten pool solidifies faster than that at top of the molten pool, and H atoms escape to the top layer. As the top molten pool solidifies, H atoms coalesce and form hydrogen. A proportion of hydrogen evaporates into air, while a small proportion remains in the top region of the molten pool; this indicates that pores are mainly entrapped at the top of the layer [17]. Extreme pore coarsening could be observed at the edge of the as-deposited wall by the NC process (Figure 9b). Poor fusion was found at the edge of the wall, pore coarsening was caused by both hydrogen and poor fusion

defect. In Figure 10, the porosity and maximum size of a single pore are highlighted. Pores from the NC process were exposed to higher temperatures due to the poor heat dissipation compared to the other three cooling processes; this potentially allowed hydrogen diffusion and pore coarsening. Besides, the porosity generated by the NC process was almost 8 folds higher than that generated by the OL process (the porosity via the OL process was 0.1%). Rapid cooling promoted the diffusion rate of hydrogen atoms while it reduced the generation of hydrogen pores.

Figure 9. Defects in macroscopic morphology: (a) the pores near the fusion line; (b) pores at the edge of the wall deposited by the natural cooling (NC) process.

Figure 10. Porosity and maximum size of a single pore in the as-deposited specimens deposited by different cooling processes.

The interlayer grains of Al5356 alloy deposited by WAAM were observed by an optical microscope (Figure 11). The solid lines signed the fusion line, the region between two solid lines is a deposited layer, and the dotted lines represent the boundary line between columnar grains and equiaxed grains. Columnar and equiaxed structures in the interlayer volumes are destroyed by recasting, and structures that are so clean and homogeneous from the metallographic point of view are difficult to create. During the solidifying process of WAAM aluminum alloy, temperature gradient (G) and solidifying rate (R) significantly impact the microstructure of solidification. G and R jointly determine the microstructure of solidification, G/R determines the solidification mode, whereas $G \times R$ determines the grain size [18]. As the molten pool solidifies, there was high G at the bottom of the molten pool, and the undercooling decreased, the grain morphology near the bottom edge of the molten pool was columnar, while an epitaxial growth occurred at the bottom of the molten pool. Notably, at the bottom of the molten pool, the grain of 5356 aluminum alloy (fcc), which, in the direction of $\langle 100 \rangle$, grew rapidly into columnar grains along the direction of the temperature gradient. As the molten pool solidified, R -value increased while G/R value decreased, and the equiaxed grain region appeared. At a stronger cooling effect, a larger R -value was obtained, and the G -value close to the bottom edge of the molten pool significantly decreased. Subsequently, the G/R decreased, and the undercooling increases; thus, a large range of

equiaxed grains are more likely to occur in an interlayer. From Figure 11c, the equiaxed grain region by the OL process was the widest 805 μm among the four cooling processes. However, regional columnar grains were found (Figure 11a), which passed through the layer due to the accumulated heat via the NC process.

Figure 11. Optical microscope (OM) photographs of samples deposited by the (a) NC process, (b) interlayer cooling (IC) process, (c) OL process, and (d) substrate cooling (SC) process.

From XRD (Figure 12), the second phase of the deposited 5356 aluminum alloy is mainly β (Mg_2Al_3) phase. The microstructure morphology close to the fusion line observed by SEM is shown in Figure 13, while components of the structure obtained by EDS composition analysis are shown in Figure 14. The skeletal structure (a) was identified as the β phase (Al_2Mg_3) [19]. The β phase was found near the fusion line. The non-equilibrium solidification of aluminum WAAM inevitably caused the non-uniformity of microstructure and composition. This inhomogeneity is characterized by the appearance of the non-equilibrium second phase. When the grains at the bottom of the molten pool grow into columnar grains, the first crystallized columnar grains exhibit a low solute concentration, while the later crystallized dendrites have a slightly higher solute concentration. The second phase and impurities gathered easily on the top of the molten pool [20]. Since the specimen obtained via the OL process have the fastest cooling rate, the equilibrium distribution coefficient k_0 is the least. The microsegregation of OL-deposited ER5356 alloy was the most severe among the four cooling processes, but, on the whole, the distribution of elements and phases is relatively uniform between and within layers (Figure 15).

Figure 12. XRD of as-deposited walls.

Figure 13. Scanning electron microscope (SEM) of the β (Mg_2Al_3) phase in thin walls deposited by the (a) NC process, (b) IC process (c), OL process, and (d) SC process.

Figure 14. Energy Dispersive Spectrometer (EDS) results of the microstructure characteristics observed via a scanning electron microscope (SEM): (a) skeletal structure; (b) feeding wire; (c) specimen by NC process.

Figure 15. EDS linear scanning through layers with length of 3 mm.

Through the EDS point scanning of samples, the Mg content of the specimen adopted via the NC process (Figure 14c) was significantly lower than that of the feeding wire materials in Figure 14b. This is attributed to the melting loss of the Mg element during the WAAM process, which consequently reduces the segregation degree.

3.3. Mechanical Properties

Stress–strain curves were observed during the tensile process (Figure 16). The tensile properties of the as-deposited walls are highlighted in Figure 17. The yield strength (131.8 Mpa) via the OL process slightly improved and was 12 Mpa higher than the NC process. The strength of the alloy relies on the stresses required to overcome obstacles in dislocation motion, these obstacles are defects such as solute atoms, precipitates, other dislocations, and grain boundaries. Thus, the yield strength (σ_y) of the as-deposited specimen defines pure aluminum (σ_{y1}) and different defects ($\Delta\sigma_i$) to its mechanical property, as follows:

$$\sigma_y = \sigma_{y1} + \Delta\sigma \quad (1)$$

$$\Delta\sigma = \Delta\sigma_{hp} + \Delta\sigma_s + \dots \quad (2)$$

Figure 16. Stress–strain curves of the fourth cooling process.

Figure 17. Yield strength (Mpa), tensile strength (Mpa), and elongation (%) of as-deposited specimens by different cooling processes.

In this study, due to the significant effect of the fast cooling rate on grain size by the OL process, as shown in Figure 11, a large equiaxed grains area occurred. Thus, a small average grain size of the as-deposited specimen was present. The reduction in grain size increases the yield strength according to Hall–Petch relation, as follows:

$$\Delta\sigma_{hp} = Kd^{-0.5} \quad (3)$$

where d is the average grain diameter and K is a constant [21]. Based on Equation (3), increasement of $\Delta\sigma_{hp}$ was predicted owing to the effects of grain refinement.

Due to the high cooling rate produced by the OL process, as-deposited Al material in most cases does not contain coarse intermetallic particles. However, we found a small amount of Mg-rich precipitate in the specimens (Figure 13), although the volume fraction was slightly low. This was an indication that Mg mainly exists in the molten pool as a supersaturated molten solution, and the as-deposited specimens were not strengthened by the Orowan process. Mg atoms dissolved in the matrix Al form point defects that impede dislocation motion. Moreover, Mg atoms diffuse near the moving dislocation during loading [22,23]. Thus, dislocation motion was suppressed during loading and $\Delta\sigma_s$ was increased which is defined as follows:

$$\Delta\sigma_s = HC^\alpha \quad (4)$$

where C is the Mg concentration (wt. %) and α and H are constants [24]. The α and H measured the strain rate. Therefore, the $\Delta\sigma_s$ was determined by the amount (C) of Mg in the matrix Al (Figure 14). Based on Equation (4), because of the melting loss of the Mg element was extreme during the NC process, the decrease was predicted owing to the effects of the amount (C) of Mg in the matrix Al with the NC process. Of note, the strength of as-deposited specimens was partly attributable to grain refinement and Mg concentration. Because tensile strength is affected by yield strength, the variation between them is the same, as shown in Figure 17.

The highest elongation occurred by the OL process (Figure 17); this can be explained by the grain refinement. That is, in a definite volume of an as-deposited specimen and under the same amount of plastic deformation, with the formation of equiaxed grains, the defective state was restored by the OL process; thus, cracking caused by stress at concentration opportunity was less and showed higher plasticity than other cooling processes. Similarly, in Figure 18, the highest microhardness through the OL process is depicted; this can further be explained by the grain refinement, which caused a significant increase in grain boundaries. The array of atoms in the interface was uneven and a considerable amount of defects occurred, this obstructed the dislocations; thus, the hardness increased.

Figure 18. Microhardness (HV) of as-deposited specimens via different cooling processes.

The fracture section morphology of the tensile specimens is shown in Figure 19. All the tensile samples fractured within the gauge length, and the obvious necking occurred in fracture position. Figure 19a shows the fracture morphology of the specimen deposited by the NC process, while Figure 19b shows the fracture morphology by the OL process. Several dimples and microscopic holes are distributed in the (b3) fracture, which reveals obvious ductile fracture characteristics. The existence of the second phase particles led to the final fracture, which showed a typical transgranular fracture and formed a dimple fracture morphology. The intergranular fracture characteristic (a2) was found in (a1) and was closely associated with the NC process, resulting in the columnar grain size. Moreover, the obvious intergranular fracture is caused by coarse grains, affecting the plasticity and other mechanical properties of the formed sample.

Figure 19. Tensile fracture at room temperature: (a) deposited by the NC process; (b) deposited by the OL process.

4. Conclusions

In this study, three different cooling processes were conducted in the WAAM 5356 alloys. The effects of different cooling processes on the geometry forming, defect, metallography, and properties were investigated. The major findings include:

In the WAAM process for the Al–Mg alloy, the adoption of the OL process maintained a minimum interlayer temperature, and the as-deposited wall thickness was the easiest to control to a specified size; in particular, the arc pit at the end of the wall was not visible which improved the stability of the WAAM process.

The porosity rate was reduced to 0.1% with the OL process, and a decrease in the interpass temperature during the WAAM process had a significant impact on grain refinement. The EDS and XRD analysis of the microstructure revealed that the β phase mainly appeared near the fusion line, and the interlayer segregation was most serious with the OL process, and the magnesium element that burned out occurred via the NC process.

It was evident that cooling regulation improved the yield strength, plasticity, and hardness, among which, the OL process had the best effect. Fine-grained strengthening plays an important role in the OL method. Although the tensile specimens prepared via different cooling processes were plastic fracture, the NC specimen shows the possibility of partial cleavage fracture.

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