Effect of transverse magnetic field on microstructure and mechanical properties of Ti-6Al-4 V manufactured by Laser-MIG hybrid welding

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Abstract
In this study, an external transverse magnetic field was used to assist laser-metal inert gas (MIG) hybrid welding of Ti-6Al-4 V (TC4). The microstructure and mechanical properties, such as microhardness and tensile properties of the weld joints, under an external magnetic field (EMF) and a reference weld joint without the EMF were investigated. The results showed that the laser-arc interaction zone (LAIZ) and heat-affected zone (HAZ) expanded, and the laser fusion zone (LFZ) and fusion zone (FZ) changed with the application of the EMF, resulting in a change in the distribution of microhardness. The microstructure of the FZ was comprised of Widmanstatten α′-grains and primary boundary β-grains without the EMF. After applying the EMF, the diameter of the primary globular β-grains in the FZ decreased by 38.6%, from 78.4 to 48.1 μm, and a small amount of residual dot-like β-grains appeared in the FZ of the joint. The mechanical properties of the weld seam improved significantly under the EMF; the average microhardness in the FZ increased by 9.7%, and the failure strain and stress of the tensile specimens improved by 45.6% and 6.6%, respectively, owing to the solid solution strengthening of the residual dot-like β-grains and finer primary β-grain boundary. In addition, a mixed fracture mode was observed on the fracture surfaces. This study revealed that the elementary microstructure of laser-MIG hybrid welded TC4 was correlated with the welding heat input of the EMF.

Keywords External magnetic field · Laser-MIG hybrid welding · TC4 alloy · Microstructure · Mechanical properties

1 Introduction
Ti-6Al-4 V (TC4) alloys are widely used in aerospace and astronautics owing to their excellent mechanical properties, high specific strengths, and long service lives [1, 2]. Recently, laser-metal inert gas (MIG) hybrid welding of titanium alloys has attracted the attention of researchers seeking to improve the microstructure and strength of welded joints [3, 4] because this technique offers an excellent combination of the advantages of both laser and arc welding, such as good joint bridging performance, deep weld ability, and strong weld joint strength [5].

The mechanical properties of weld joints depend on the microstructure of the weld seam, especially the weakest zones such as the fusion zone (FZ) and heat-affected zone (HAZ) [6, 7]. In general, the FZ indicates changes that arise in the machining structure due to temperature variations [8]. Wu et al. [9] indicated that a high heat input and quick cooling rate during the welding process lead to high temperature gradients and unsatisfactory microstructures. Abu et al. [10] investigated the tensile properties of TC4 during laser welding and determined that the HAZ has poor mechanical properties. Su et al. [11] studied the microstructure and tensile properties of the TC4 joint during the laser-MIG hybrid welding process; large grains, thick β phases, and high dislocation density caused by the high heat input lead to a low crack resistance in the entire weld seam. Thus, an effective method is needed to improve the laser-MIG hybrid welding process, which resolves the unsatisfactory microstructure and improves the mechanical properties in the weakest zones.

Therefore, an external magnetic field (EMF) has been applied to assist the welding process and improve the welding
structure and properties [12–18]. In laser welding, the welding process is affected when the magnetic field threshold exceeds 200 mT. Chen et al. [12] investigated the influence of a 240 mT magnetic field for the laser welding of dissimilar materials. A magnetic field of 240 mT is beneficial for the weld seam appearance and microstructure when welding dissimilar materials. Avilov et al. [13] observed the appearance of welding beads under different alternating magnetic fields during laser beam welding. An alternating magnetic field produces a beneficial effect by inhibiting gravitational dropout and sagging of the weld. Chen et al. [14] studied the influence of a magnetic field on the weld appearance, interfacial microstructure, and mechanical properties during high-speed gas metal arc welding, and the results show that the magnetic field efficiently softens the weld seam, decreases the undercut defect, and improves the weld mechanical properties. However, the welding process is considerably affected when the magnetic field threshold is less than 50 mT for arc welding. Guan et al. [15] indicated that the EMF (12.5–17 mT) frequency has a significant impact on the arc shape and arc movement, which has a considerable influence on the arc temperature gradient, undercooling degree, weld appearance, and mechanical properties in gas metal arc welding (GMAW). Chen et al. [16] studied the influence of a compound magnetic field on the temperature distribution, showing that the heat of the molten pool is transmitted to the edges of the weld pool under the action of a compound magnetic field. Moreover, laser-MIG hybrid welding is profoundly influenced by the EMF when the magnetic field threshold is less than 50 mT. Zhang et al. [17] showed that an external longitudinal magnetic field (16 mT) increases the stability of the welding arc and affects the force on droplet transfer during laser-MIG hybrid welding of 316L welding rods. Zhu et al. [18] indicated that the combined effect of a laser and magnetic field (16 mT) induces grain refinement and orientation alternation during laser-MIG hybrid welding of 316L welding rods. The literature described above confirms that the appropriate magnetic field has a considerable impact on the microstructure and mechanical properties during laser-MIG hybrid welding.

However, there are only a few systematic studies on the microstructure and mechanical properties during transverse magnetic field-assisted laser-MIG hybrid welding of TC4 titanium alloys. In addition, the authors’ previous work [19] reveals that a 24 mT EMF effectively improves plasma stability, droplet transfer frequency, and seam formation in the TC4 laser-arc hybrid welding process and is considered an appropriate EMF. In this study, the influence of the EMF on the microstructure and mechanical properties of laser-MIG hybrid welded TC4 was investigated. The microstructure and properties of the welded joint with the optimal EMF (24 mT) and without the EMF were compared in this study, which can provide further guidance for controlling the microstructure and mechanical properties of the weld seam of TC4 manufactured by laser-MIG hybrid welding.

## 2 Experimental procedures

### 2.1 Material

In this experimental setup, the base metal and filler wire were respectively 4-mm-thick TC4 and 1.0 mm in diameter TC4 filled wire, and the chemical compositions of these specimens are listed in Table 1. Physical cleaning and laser cleaning were performed on the surface of the samples to eliminate the effect of surface pollution on the experimental results. Pure argon with a flow rate of 2 m³/h was used as the shielding atmosphere to protect the molten pool from oxidation.

### 2.2 Welding procedure

A 4-kW fiber laser (YLR-4000, IPG Photonics, Oxford, MA, USA), which can provide a continuous wave output with a 1.07 μm laser emission wavelength, was used to weld 4 mm Ti-6Al-4 V plates via bead welding. A 0.3-mm-diameter laser spot was focused on the center of the sheet. A Fronius TPS 4000 (Wels, Austria) digital power controller was used as the MIG welding power controller. The transverse EMF was provided by permanent magnet blocks, which can provide a maximum magnetic field of 50 mT. The sheets were placed between two magnetic blocks, as shown in Fig. 1. The magnetic blocks remained stationary during welding, and the distribution of the magnetic induction lines was assumed to be parallel and perpendicular to the welding direction. The magnetic field intensity was adjusted by changing the distance between the magnetic blocks and welded plate.

### 2.3 Mechanical testing

The Vickers microhardness of the cross sections of the welding bead with an interval of 0.15 mm was tested by a Vickers microhardness indentation machine at a test load of 200 g

| Table 1 | Chemical composition of the base metal and the wire (wt. %) |
|--------|----------------------------------------------------------|
| Element | Al | V | N | H | O | Fe | C | Ti |
| Base metal | 5.5–6.8 | 3.5–4.5 | 0.05 | 0.015 | 0.20 | 0.30 | 0.10 | Bal |
| Weld wire | 5.8 | 4.2 | 0.05 | 0.015 | 0.20 | 0.30 | 0.10 | Bal |

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and dwell time of 10 s. Tensile tests were performed on a computer-controlled electronic universal testing (CEUT) machine. The tensile specimens were cut using a wire-cut electrical discharge machine, and the welding bead was in the center of the tensile specimens, as shown in Fig. 2. Tensile tests were conducted on the CEUT machine with a loading rate of 2 mm/min.

### 2.4 Microstructure characterization

The analysis of the microstructure was divided into two parts: one part was the microstructural variations that occurred in the weld under the EMF, and the other part was the tensile failure test specimen. The metallographic specimens were cut from the welded bead perpendicular to the welding direction, as shown in Fig. 2, ground with abrasive paper down to a 1200 mesh size, polished by a 0.5 μm Al₂O₃ slurry, and chemically etched by a solution of Kroll’s reagent (2 ml HF + 6 ml HNO₃ + 92 ml H₂O). The welding morphology was observed using a stereoscopic optical microscope (OM), and the microstructure of the metallographic specimens was characterized by the ultra-depth of field OM. The details of the microstructure were captured by scanning electron microscopy (SEM, SU3900, Hitachi, Tokyo, Japan). Fractured specimens were cut from the tensile failure specimens and examined using SEM to determine the failure modes and mechanisms of fracture initiation and propagation.

### 3 Results

#### 3.1 Weld seam morphology

The morphology of the weld seam is an important factor that indicates the weld seam bridging ability and melt-metal distribution. The shapes of the laser-MIG welding specimen under the EMF and the reference specimen without the EMF are shown in Fig. 3. The welding appearances of the specimens were uniform. However, identifying the quality of the weld bead from its appearance was difficult. Therefore, the influence of the magnetic field on the weld microstructure and properties was investigated.
3.2 Cross-sections of the weld appearance

Figure 4 shows cross-sections of the weld under the EMF and the reference specimen without the EMF. According to the thermal history of these specimens, the welding region was divided into distinctive microstructural zones, including the laser-arc interaction zone (LAIZ) on the top of the fusion region composed of needle-like Martensite $\alpha'$-grains, laser fusion zone (LFZ) on the bottom of the fusion region composed of basketweave Widmanstätten $\alpha'$-grains, FZ composed of coarse Widmanstätten $\alpha'$-grains, and HAZ composed of $\beta$-grains and fine Widmanstätten $\alpha'$-grains.

To investigate the influence of the EMF on the microstructure distribution and analyze the characteristics of the molten pool, the dimensions of the weld region, that is, the height of the LAIZ and LFZ ($H$, mm), width of the FZ and HAZ at a depth of 1 mm from the base metal surface ($W$, mm), and reinforcement angle ($\alpha$, °) were measured on the cross section, as shown in Fig. 5. For the reference specimen, the FZ consisted of a short LAIZ and long LFZ, whereas the transform layer (TL) exhibited a significant partition of the FZ and HAZ. After the addition of the EMF, the values of $H_{LAIZ}$ increased from 2581 to 3346 $\mu$m, whereas those of $H_{LFZ}$ decreased from 3750 to 3295 $\mu$m; in addition, the main weld zone transferred from a dominant LFZ (59%) to a combination of the LFZ (50%) and LAIZ (50%). Furthermore, $W_{HAZ}$ increased from 337 to 500 $\mu$m, $\alpha$ increased from 28.4° to 48.4°, $W_{FZ}$ decreased from 267 to 113 $\mu$m, and the TL changed from a combination of FZ (44%) and HAZ (56%) to overwhelmingly HAZ (81.5%).

3.3 Metallographic structure

3.3.1 Macrostructure

As shown in Fig. 6, the molten zone (MZ) of the reference specimen without the EMF, which included the LAIZ and LFZ, exhibited the nucleation of columnar primary $\beta$-grains at the solid–liquid interface at the edge of the LAIZ and grew toward the center of the molten pool against the direction of heat flow during the laser-MIG hybrid welding process. In addition, needle-like martensite $\alpha'$-grains nucleated at the boundary of the columnar $\beta$-grains and grew parallel to the solidified columnar $\beta$ boundary with a decrease in the temperature in the LAIZ [20]. Furthermore, primary globular $\beta$-grains nucleated in the LFZ, and $\alpha'$-grains grew in the globular $\beta$-grains and were interlaced to form a basketweave...
Widmanstatten microstructure. For the specimen under the EMF, the grains in the LAIZ were mainly primary β-grains oriented from the solid–liquid interface, and the growth direction was similar to that of primary β-grains in the LAIZ of the reference specimen. Globular β-grains and Widmanstatten α′-grains were also observed in the LFZ.

The TL of the specimens with and without the EMF are shown in Fig. 7. For the reference specimen, primary globular β-grains nucleated at the FZ, α′-grains grew in the globular β-grains and interlaced to form a basketweave microstructure, which was the typical Widmanstatten structure. In addition, fine β-grains were formed at lower peak temperatures in the HAZ. For the specimen with the EMF, primary globular β-grain boundaries and Widmanstatten α′-grains were also observed in the FZ, and the boundaries of the FZ and HAZ became blurred.

The average width and height of the columnar β-grains in the LAIZ and the grain size of coarse Widmanstatten in the TL were specifically analyzed to study the influence of the EMF on the macrostructure during TC4 laser-MIG hybrid welding, as shown in Fig. 8. The width of the columnar β-grains in the LAIZ decreased from 231.7 to 180.5 μm upon application of the EMF; however, the height of these grains increased by 44.6%, from 1075.5 to 1555.4 μm, upon application of the EMF. Finally, the diameter of the primary globular β-grain boundary in the FZ decreased by 38.6%, from 78.4 to 48.1 μm, with the application of the EMF.

### 3.3.2 Microstructure

The microstructures of the TL for the reference specimen without the EMF and the specimen with the EMF were investigated using SEM, and the observed positions of the specimen are marked in Fig. 4. The microstructure of the HAZ for TC4 resulting from β phase transition mainly included residual dot-like β-grains, grain boundary αm, acicular martensite α′-grains, and basketweave Widmanstatten structures [21].

Figure 9a depicts the microstructures of the HAZ for the reference specimen without the EMF using SEM, which varied from grain to grain, and generally consisted of a principal basketweave Widmanstatten structure and other β phase transition structures with different shapes and sizes. As shown in Fig. 9b, the FZ was composed of coarse grain basketweave Widmanstatten α′-grains with an average length of 56.0 μm and average width of 2.4 μm, as well as primary globular β-grains. The primary acicular α′-grains with an average length of 18.2 μm and average width of 0.6 μm are displayed in Fig. 9d. As shown in Fig. 9c, primary acicular α′-grains nucleated from the prior β-grain boundary, penetrated the whole β-grain, intersected among themselves, and formed basketweave Widmanstatten α′-grains [22]. The fine grain Widmanstatten α′-grains shown in Fig. 9e had an average length of 18.3 μm and average width of 1.0 μm, as well as residual dot-like β-grains. Figure 9f shows residual dot-like β-grains distributed between the fine secondary α′-grains, which was similar in appearance to that of the base metal.

The microstructure of the TL for TC4 under the EMF was also comprised of a principal basketweave Widmanstatten structure and other phase β transition structures with different shapes and sizes, as shown in Fig. 10a, in agreement with that of the reference specimen without the EMF. As exhibited in Fig. 10b, the FZ was composed of fine grain
basketweave Widmanstatten $\alpha'$-grains and boundary $\alpha_m$-grains with an average length of 44.3 μm and average width of 1.3 μm, which indicated that the composition of the FZ was identical in both specimens; nevertheless, the grain boundary exhibited a smaller length and width compared with that of the reference specimen without the EMF. The Widmanstatten structure was composed of intersected basketweave $\alpha'$-grains with an average length of 18.2 μm and average width of 0.6 μm, which was smaller and denser than that of the reference specimen without the EMF. Acicular martensite $\alpha'$-grains and residual dot-like $\beta$-grains were observed between the intersected basketweave $\alpha'$-grains, as shown in Fig. 10d. Figure 10c and e shows the microstructures of the HAZ that had fine Widmanstatten $\alpha'$-grains with an average length of 11.6 μm and average width of 1.5 μm and some residual dot-like $\beta$-grains, and shorter lengths and longer widths were observed in the specimen with the EMF than those of the reference specimen without the EMF. In addition, magnified images of the Widmanstatten $\alpha'$-grains showed that tertiary $\alpha'$-grains were between the secondary $\alpha'$-grains, and residual dot-like $\beta$-grains were distributed among these grains also.

3.4 Microhardness and tensile properties

The microhardness distributions along the horizontal directions of the cross sections of the reference specimen and the specimen under the EMF are respectively labeled $L_1$ and $L_2$ in Fig. 4. Both weld specimens showed similar microhardness distribution trends from the base metal to the fusion zone; specifically, the microhardness in the molten zone was larger than that of the base metal, and a low microhardness region existed in the FZ. Five microhardness tests were performed for each weld specimen at each condition, and the average of the results is presented in Fig. 11. The average microhardness of the base metal (BM), HAZ, and weld center (WC) along the horizontal direction of the reference specimen were 320, 375, and 400 HV, respectively. The microhardness value of the reference specimen in the FZ was 310 HV, which was even lower than that of the base metal. The average microhardness of the BM, HAZ, WC of the specimen under the EMF were 320, 360, and 375 HV, respectively, which were similar to those of the reference specimen; however, the lowest microhardness of the FZ increased from 310 to 340 HV after application of the
EMF, an increase of 9.7% compared with that of the reference specimen.

Two tensile experiments were performed for the weld specimen in each condition, and the average of the results is presented in Fig. 12; the failure locations of the specimens with and without the EMF are also shown in Fig. 12. The application of the 24 mT EMF improved the failure strain of the tested specimens by 45.6%, from 5.7 to 8.3%; in addition, the failure stress of the tested specimens also improved by 6.6% after the application of the EMF, from 976 to 1041 MPa.

Figure 13 shows the tensile fractographic images of the reference specimen and the specimen under the EMF. As shown in Fig. 13a, the fracture surface of the reference specimen without the EMF was divided into two sections. Equiaxed dimples of different sizes and depths covered the area shown in Fig. 13c of the reference specimen without the EMF, and the area shown in Fig. 13d exhibited a quasi-cleavage fracture characterized by a cleavage surface with a wide area, demonstrating that the fracture mode of the reference specimen was a mixed fracture. Figure 13b indicates that the fracture surface of the specimen under the EMF promoted a fractional microstructure similar to that of the reference specimen. Figure 13e exhibits ductile fracture characterized by dimples with a deep depth and large area, demonstrating the better performance of the fracture interface. Finally, Fig. 13f displays a quasi-cleavage
fracture, which indicated a mixed fracture mode on the fracture surfaces.

4 Discussion

4.1 The influence of the EMF on laser-MIG energy and molten pool

In laser-MIG hybrid welding, the performance of the weld cross-section is mainly influenced by the heat resource, weld pool flow, and droplet transfer. The heat input \( Q \) is the sum of the laser heat input \( Q_{\text{laser}} \) and arc heat input \( Q_{\text{arc}} \). \( Q \) can be expressed as follows [19]:

\[
Q = Q_{\text{laser}} + Q_{\text{arc}} = \frac{\eta_1 P_{\text{laser}} + \eta_2 UI}{v}
\]

where \( P \) is the power of the laser, \( U \) is the arc voltage, \( I \) is the welding current, \( \eta_1 \) is the efficiency of the laser beam, \( \eta_2 \) is the efficiency of the arc power, and \( v \) is the welding speed.

The arc distribution of the laser-MIG hybrid welding process directly affects the efficiency of the arc power and flow of the molten pool, ultimately affecting the microstructure and performance. An arc is a strong, lasting discharge phenomenon in the gas medium between two electrodes or between electrodes and the base metal at a certain voltage. The physical nature of the arc is the charged particles and current are regarded as the aggregation of an electric streamline [4], which is deeply affected by an EMF. The ampere force caused by the EMF on the front electric streamline is backward and downward, and the ampere force caused by the EMF on the tail electric streamline is upward and downward. The arc is attracted by the laser in the laser-MIG hybrid welding process, which leads to the predominance of the forward arc. The forward arc is compressed under the filled...
wire by the EMF, and the heat input is concentrated in the arc under the wire [20]. Thus, the heat input is concentrated owing to the influence of the EMF. The arc voltage, \( U \), and welding current are constant values; thus, the total heat input is unchanged in this process, which leads to a constant width of the TL, as well as penetration. The distribution of heat is concentrated on the surface of the LAIZ, which leads to an increase in \( H_{LAIZ} \) and a decrease in \( H_{LFZ} \).

Electrons entering the molten pool move in a spiral under the influence of the Lorentz force [23]. The positive particles in the front of the molten pool move backward and downward under the action of the EMF, which causes backward and downward flow in the front half of the molten pool; and the positive particles in the tail end of the molten pool move backward and upward under the action of the EMF, which causes backward and upward flow in the back half of the molten pool. Ultimately, vertical circulation flow was produced under the influence of the transverse EMF, as shown in Fig. 14. Thus, agitation occurred in the molten pool, which led to a decrease in the temperature gradient and cooling rate. Furthermore, the area of the bottom collapse decreased under the influence of the EMF.

### 4.2 Grain morphology

The grain morphology and microstructure evolution of the TC4 laser-MIG hybrid weld seam depended on the heat input of the welding heat sources, the cooling rate of the weld molten pool, and the distribution of the temperature gradient in the temperature field. The grains of TC4 have a preferential growth direction: \(<1010>\) in hexagonal close packed (HCP) and \(<100>\) in body centered cubic (BCC). During the cooling process of the molten metal, the molten metal begins to solidify into a BCC structure when the temperature reaches the solid temperature, and the growth of grains whose preferential growth direction is consistent with...
the peak temperature gradient is accelerated, and the growth of grains whose preferential growth direction is consistent with a smaller temperature gradient is inhibited [24].

For the LAIZ, the heat input from the laser and arc gathered on top of the molten pool, heat distribution decreased from the top of the molten pool to the substrate, and primary β-grains tended to grow from the fusion line and interface between the LAIZ and LFZ to the top of the weld reinforcement with a columnar shape in the direction opposite to that of the heat flow, which provided the maximum driving force during the solidification process. The heat of the molten metal mainly dissipated along the negative Z direction to the solidified β-grains, not only from the base metal by heat conduction but also at the interface between the air and molten pool through convection and radiation. Consequently, the strongly directed heat flux formed from the fusion line to the top of the molten pool, generating long, narrow columnar grains that penetrated the entire LAIZ [25]. Owing to the high heat input and heat dissipation from the hybrid laser-MIG heat resource and narrow molten pool width, the columnar β-grains were characterized by a long grain length and low aspect ratio. Owing to the influence of the EMF, the temperature in the weld center caused by the arc shrinkage, temperature gradient, and cooling rate decreased owing to the agitation in the molten pool. The improvement in the heat input in the weld center increased the height of the LAIZ. For the primary β-grains in the LAIZ, the temperature gradient and cooling rate decrease improved the growth time of the columnar grains, which led to an increase in the β-grain boundary area. For the LFZ, heat from the arc struggled to reach the bottom of the welding pool, which led to a lower heat input compared with that of the LAIZ, eventually resulting in the formation of a large number of globular β-grains. The grains in the LFZ were hardly affected by the EMF.
The microstructure of TC4 comprised of the α and β phases possessed reversible transformation properties. When the temperature of the molten pool reaches the β transus temperature $T_{\beta}$ (980 °C) [20], the β phase is transformed into the α phase, and when the temperature of the specimen reaches the martensite starting temperature $M_S$ (575 °C), the α phase is transformed into the α' phase. During the laser-MIG hybrid welding process of TC4, the maximum temperature in the molten pool was above the liquid temperature, $T_L$ (1655 °C), using the hybrid heat source, and the α phase transformed into the β phase with an increase in temperature and then melted into a liquid phase when the temperature was above the liquid temperature, $T_L$ (1655 °C). At the beginning of cooling, the liquid phase rapidly transformed into the β phase. With a further decrease in temperature, the β phase will transform into α, $\alpha_m$, and α' phases, according to the cooling rate of the microstructure. As shown in Fig. 15, the remnant α phases were formed by the diffusive transformation from the β phase to the α phase for cooling rates between 1.5 and 20 °C/s, and the boundary $\alpha_m$ phase was obtained when the cooling rate was between 20 and 410 °C/s. Martensite and Widmanstatten structures were observed when the β phase transformed into α' phases at cooling rates above 410 °C/s. For the LAIZ, acicular Martensite α' and grain boundary β phases were observed in Fig. 6, indicating the cooling rate of the transformation process was above 410 °C/s due to the high heat input under the laser-arc hybrid heat resource. However, for the LFZ, the arc hardly affected this area, and the laser dominated as the heat resource, causing the maximum temperature and cooling rate of the LFZ to decrease from the $M_S$. Thus, Widmanstatten α' with boundary $\alpha_m$ phases precipitated more easily.
The HAZ, the zone nearest to the base metal, was composed of fine Widmanstätten α’-grains, boundary α_m-grains, and residual dot-like β-grains. This area was far from the MZ and affected by the small amount of heat input from the hybrid welding resource by heat conduction; in addition, the peak temperature was below the melting point, and the diffusive phase transition from the α phase to the β phase was incomplete during the temperature-rise period [20], which lead to the remnant dot-like β-grains. In addition, α_m grain boundaries were observed in Figs. 9 and 10, which indicated the cooling rate was between 20 and 410 °C/s. The FZ was the zone nearest to the MZ, which contained the mushy and high temperature heat-affected zones, causing the higher peak temperature compared with that of HAZ, and thus Widmanstätten α’-grains in the forms of coarse grains and boundary α_m-grains were also observed in the FZ. During laser-MIG hybrid welding, the morphological instability of dendrites is attributed to the EMF caused by the welding current and magnetic field intensity [14]. As shown in Fig. 16, when the 24 mT EMF was applied to laser-MIG hybrid welding, hot metal liquid flow produced in the arc affected the molten pool, which caused the primary β-grains in the front of the pool to remelt and prevented the primary β-grains from growing from the mushy zone into the liquid zone [18], which decreased the growth time of the Widmanstätten α’ phase growing in the primary β-grains and caused a decrease in the α’ phase length and primary boundary β-grain diameter. Furthermore, as the peak temperature of the FZ decreased, incomplete diffusive phase transition from the α phase to the β phase occurred during the temperature-rise period [20], which lead to a remnant of the dot-like β-grain in the FZ and a decrease in the Widmanstätten α’-grain growth and TL width [26]. Consequently, residual dot-like β-grains, smaller primary β-grain boundaries, and fine Widmanstätten α’-grains were observed in the FZ of the specimen under the EMF.

4.4 Mechanical properties

In addition, the higher dislocation density in the martensite phase than in the α and β phases could give rise to dislocation hardening, and boundary hardening would occur when acicular α’ was formed, especially the secondary and tertiary α’ phases with smaller grain sizes [6]. The microhardness of the HAZ and WC was larger than that of the base metal owing to dislocations and boundary hardening [27], and the
The microhardness of the FZ was lower than that of the MZ and HAZ owing to the coarse Widmanstatten \(\alpha'\)-grains. Owing to the finer Widmanstatten \(\alpha'\)-grains and smaller primary \(\beta\)-grain boundary diameters, the hardness of the FZ increased under the influence of the EMF. The tensile properties of the tested specimens depended on the grain size and dislocation conditions between the large grains. The solid solution strengthening of residual dot-like \(\beta\)-grains decreased the dislocation density, and a finer \(\beta\)-grain boundary reduced the possibility of slip band formation; these two factors increased the resistance of the material to deformation under an external force, thereby improving the tensile properties.

5 Conclusions

The influence of the EMF during laser-MIG hybrid welding on the morphology, microstructure, microhardness, and tensile properties was investigated and discussed in this study. The main results are summarized as follows:

1. With the addition of the EMF, specimens subjected to laser-MIG hybrid welding exhibited constant widths in the MZ and TL, decreased FZ width, increased HAZ width, increased LAIZ length, and decreased LFZ length; these specimens exhibited obviously better quality than the that of the reference specimen without the EMF, which was attributed to the heat input concentration caused by the compression of the forward arc and agitation in the molten pool.

2. The heat input concentration by the compression of the forward arc increased the maximum temperature, and the agitation of the hot liquid flow in the molten pool decreased the cooling rate, causing an increase in the columnar grain growth, which lead to an increase in the \(\beta\)-grain boundary diameter in the LAIZ.

3. The microstructure of the LAIZ consisted of a columnar \(\beta\) boundary and needle-like martensite \(\alpha'\)-grains, and that of the HAZ consisted of a global \(\beta\)-grain boundary and Widmanstatten \(\alpha'\)-grains. The EMF increased the area of the columnar \(\beta\)-grains in the LAIZ and decreased the diameter of the coarse Widmanstatten grains. The microhardness of the HAZ and WC were larger than that of the base metal owing to dislocations and boundary hardening.

4. The microstructure of the HAZ consisted of the Widmanstatten \(\alpha\) phase, residual dot-like \(\beta\)-grains, and grain boundary \(\alpha_m\), and that of the FZ was comprised of the Widmanstatten \(\alpha\) phase and primary \(\beta\)-grain boundaries. The FZ exhibited the lowest microhardness and tensile specimen failure in this zone, which indicated that the FZ was the weakest zone in the weld seam. For the FZ, the EMF decreased the coarse Widmanstatten \(\alpha'\)-grain length and diameter of the primary \(\beta\)-grain boundary, and residual dot-like \(\beta\)-grains caused solid solution strengthening, resulting in an improvement in the tensile properties. Compared with those of the reference specimen, the average microhardness of the FZ increased by 9.3%, and the failure strain and stress of the tensile specimens improved by 45.6% and 6.6% with the EMF; in addition, a mixed fracture mode was evident on the fracture surfaces.

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Declarations

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