Microstructure and mechanical properties of rhenium and GH3128 superalloy dissimilar welded joints by electron beam welding

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Abstract

This research aims to study the dissimilar welded joint of rhenium and GH3128 made by electron beam welding. The microstructure, solidification behavior, mechanical properties, and Re distribution behavior of the weld were systematically researched. The results showed that an excellently joint is obtained, and no defects such as cracks and pores were found in the weld. But thermal cracks occur on the Re base material under the influence of heat input. The fusion zone was mainly composed of columnar dendrites, while the plane grains grow near the Re-side fusion line (reaction layer), which was always enriched in the Re element. Re is evenly distributed in the Ni matrix as a solid solution element so that the hardness of the joint is always higher than that of the GH3128 base material. The tensile strength of the joint reached 498 Mpa and the elongation was about 3%. The tensile fracture occurred at the position of the reaction layer. The high hardness and brittleness of the reaction layer are attributed to the high content of Re and a lot of low-angle grain boundaries.

1. Introduction

Re has great potential for application in engine thrust chambers, strong thermal shock components, thermocouple element, and emitter materials due to its high melting point, high strength, good plasticity, excellent creep resistance, and no ductile-to-brittle transition temperature [1, 2]. However, extensive applications of Re have been limited due to its scarcity, high cost and high density. Therefore, it is necessary to connect Re and other materials. GH3128 is a nickel-based superalloy, which has been widely applied to the aerospace industry, turbine engine blades, and aero-engine combustion chambers, and its working temperature is generally around 950 °C [3]. Both rhenium and GH3128 materials can be used to manufacture hot section components of aero engines. These two materials need to be connected in the manufacture of engine nozzles, combustion chambers and other components. Realizing effective welding between Re and GH3128 can not only improve the performance of parts to meet the needs of certain high-temperature working environments of aero engines, but also reduce the use of Re, thereby reducing costs.

The welding between Re and GH3128 belongs to dissimilar metal welding. Various methods have been used to connect refractory metals and dissimilar materials, such as friction stir welding, diffusion welding, laser beam welding (LBW), and electron beam welding (EBW). Avettand-fenoël et al [4] investigated friction stir welding of WC-12Co alloy and S45C steel. The shear strength of the joint reached about 750 MPa, which was much higher than the joint strength obtained by other methods. Du et al [5] studied the diffusion bonding of Al3(TiZrHfNb)95 alloy and Ti3AlNb alloy. The shear strength of the joint was 427 MPa, due to the Al3Zr5 compound disappeared in the weld at a bonding temperature of 1100 °C. The fractured form of the joint changed from brittle intergranular fracture to toughness with the bonding temperature increases. Zhang et al [6] examined the LBW of Mo and Ti with a Zr filler. The tensile strength of the joint increased from 350 MPa to 470 MPa, which was more than 90% of the Ti base material (BM). Hajitabar [7] investigated the influence of welding current on microstructure and mechanical properties of EBW joint of Nb-1Zr alloy. The excellent joint was obtained at a welding current of 30 mA, and its tensile strength was 281 MPa. Zhang et al [8] conducted the EBW of titanium-
zirconium-molybdenum (TZM) alloys with the addition of Re interlayer. As the content of Re increased, the grain size and grain boundary angle of the fusion zone (FZ) decreased, the tensile strength of the joint increased, and its fracture position migrated from the FZ to the heat-affected zone (HAZ). The performance of the joint is affected by the different characteristics of dissimilar materials, especially Re is very easy oxidized during the heating process, thereby forming Re$_2$O$_7$, ReO$_3$, ReO$_2$ and other oxides which affect the performance of the joint [9]. To prevent joint embrittlement due to oxidation of FZ and HAZ, welding must be performed in a protective gas or vacuum environment. High-energy beam welding has a large depth-width ratio, small heat-affected zone, and small deformation, so it is more suitable for the connection of refractory metals and high-temperature alloys. Liu et al [10] studied the mechanical properties of laser lap-welded joint of 50Mo–50Re thin sheets. The presence of compounds with Re and O was found in the joint interface area, which leads to embrittlement and intergranular fracture of the joint. Compared with other welding methods, the vacuum protection of EBW is more suitable for welding refractory metals and dissimilar materials [11]. Kramer et al [12] investigated the effects of EBW and LBW methods on molybdenum-44.5% rhenium (Mo-44.5% Re) thin sheet. The results suggested that the EBW of Mo-44.5% Re has better performance, while the LBW samples generated surface and sub-surface cracks, and exhibited brittle fracture characteristics in the tensile testing. Therefore, the EBW method was used to connect Re and GH3128 in this experiment.

Refractory metals such as W, Ta, Re, and Ru elements are usually added to nickel-base superalloys as reinforcing elements to improve the strength of materials. However, due to the impact of cost, the amount of refractory metal elements is minimized while improving material performance. When the second generation of nickel-based superalloys started, Re is an important strengthening element. During the high-temperature creep process, the diffusion of Re element from the $\gamma'$ phase to the $\gamma$ phase leads to its enrichment at the $\gamma'$/$\gamma$ interface to strengthen the material, thereby increasing the creep strength of the alloy. This phenomenon has been observed in the studies of Shu et al [13] and Dubiel et al [14]. Some researchers have proposed that the excessive addition of the Re element would promote the formation of TCP phases in superalloy. Dubiel et al [14] found that the TCP phase formed during high-temperature creep process is rich in Re, W, Cr elements. Long et al [15] also found that excessive Re broke the solid solution equilibrium in the local regions and promoted the formation of TCP phases, thus significantly reduced the strength of the superalloy. At present, the role of Re in the weld during welding is unclear, and the dissimilar welding of Re and GH3128 is also blank. Therefore, the research of Re and GH3128 welding has great practical significance and application potential.

Table 1. Chemical compositions of the GH3128 alloy.

| Chemical element | C  | Cr             | Ni | W  | Mo  | Al  | Ti          |
|------------------|----|----------------|----|----|-----|-----|-------------|
| wt%              |    | 0.005          |    | 19.0~22.0 | Balance | 7.5~9.0 | 7.5~9.0 | 0.4~0.80 | 0.4~0.80 |
| Fe               |    | 0.2            |    |    |     |     |            |
| B                | 0.005 | 0.006     | 0.005 | 0.50 |     | 0.13       | 0.13       |

Table 2. The welding process parameters.

| Welding parameters | Acceleration voltage(kV) | Focus current(mA) | Welding current(mA) | Welding speed(mm/s) |
|--------------------|--------------------------|-------------------|---------------------|---------------------|
|                    | 120                      | 2460              | 5                   | 10                  |

Figure 1. The schematic diagram of electron beam welding Re and GH3128.
In this experiment, EBW was realized the connection of Re and GH3128 superalloy for the first time. The macro and microstructures, defects, distribution of Re, and mechanical properties of the joint were systematically investigated. The influence of Re was discussed on the weld and the fracture mechanism was also given, which was expected to provide the theoretical basis for practical applications.

2. Experiment

In this experiment, the 2 mm thick plates of Re and GH3128 were used. The BM is a pure Re plate with content of Re exceeds 99%. Table 1 shows the composition of GH3128 alloy. A vacuum EBW machine (Pro Beam K110) was used in this experiment, and its vacuum environment was $5 \times 10^{-3}$ Pa. The detailed welding parameters are shown in table 2. The oxides and stains on the surface of base materials were cleaned by diamond papers with the 400-grit, and then washed with alcohol and dried before assembly and welding. When assembling, install the welding plate tightly on the welding fixture to ensure that the gap is ignored, as shown in figure 1. A butt joint was used as a connection method for dissimilar materials. To match the actual project, a piece of GH3128 plate was added below the weld. The upper part of the weld was studied in this article.

After welding, the joint was cut along a cross-section perpendicular to the welding direction and the metallographic specimen were prepared, and then the samples were sequentially subjected to mechanical ground, polished, and chemical etched for microstructure analysis. The mixed solution of HCl and HNO$_3$ with the ratio of 4:1 was used as an etchant to etch the welds in the experiment. Optical microscopy (VHX-600 k) and scanning electron microscopy (SEM) (Hitachi S-3400N) were used to characterize the microstructure and fracture of the weld. The composition of the microstructure was characterized by energy dispersive spectroscopy (EDS) equipment. To further research the microstructure, the field emission electron probe device (JXA-8530F Plus) was used to observe the electron backscatter diffraction (EBSD) of the sample. Besides, the joint was cut by electrical discharge machining, the test surface was finely ground, and then the samples were prepared by electropolishing in a mixed solution of perchloric acid and glacial acetic acid (3:97). Electron probe microanalysis (EPMA) X-ray mapping was performed by wavelength dispersive spectroscopy (WDS) accessories to semi-quantitatively measure element segregation and homogenization in three different regions of the weld.
The microhardness tester (HXD-1000TMSC/LCD) was used to measure the microhardness of the joint, with a load of 200 g and a loading time of 15 s. The electronic universal tensile tester (Zwick/Roell Z100) was used to measure the mechanical properties of the joint, with a tensile speed of 1 mm min⁻¹. Figure 2 shows a schematic diagram of a tensile test sample. Subsequently, SEM was used to characterize the tensile fracture.

3. Results

3.1. Macrostructure of the welded joint

Figure 3 shows the macrostructure of dissimilar EBW joint of Re and GH3128. Figure 3(a) indicates that a smooth and uniform weld without surface defects such as splash, cracks and pores is observed from the surface of the joint. However, noticeable asymmetry is observed in the upper half of the weld, as shown in figure 3(b). A right-angle fusion line (FL) appears on the Re-side, while the funnel-shaped FL is observed on the GH3128-side.
The amount of GH3128 BM melted in the weld is significantly larger than that of Re due to the melting point of Re is much higher than that of GH3128 superalloy.

3.2. Microstructure of the welded joint
3.2.1. Fusion zone
Figures 4 and 5 show the microstructures of the joint. The fusion zone (FZ) mainly shows the nucleation mechanism of dendritic grains, and only a small amount of equiaxed crystals is observed in the center of the weld. There are two reasons for this phenomenon. Firstly, there is the highest temperature in the center of the molten pool during the welding process, resulting in a relatively high nucleation rate. The growth space of grains was extremely narrow in the initial stage of solidification. Secondly, under the electron beam stirring in the center of the weld, the dendritic grains formed in the early stage of solidification would be destroyed. The amount of crystal nucleation rapidly increased due to the grains were broken, resulting in the formation of small equiaxed grains. Comparing figures 4(a) and (c), figures 5(a) and (c) respectively, it suggests that the grains on the Re-side are finer than the grains on the GH3128 side. This is because the heat transfer rate of Re BM is higher, which leads to the faster cooling rate on the Re-side. Besides, figure 4(b) shows a clear solidified grain boundary marked with a red curve, which is the result of competitive growth along the trailing edge of the molten pool. The EDS results (table 3) of the dendrite and the XRD analysis (figure 6) of the joint suggest that the dendrites are the Ni-based solid solution rich in Cr and W formed during rapid solidification. The eutectic phase and short dendrites distributed on the eutectic form the microstructure of the Re-side. Combined with the previous research, it can be inferred that the short dendrites are pre-eutectic Ni-based solid solutions. A special reaction layer has appeared on the Re-side FL that would be analyzed in detail in the following chapters.

| Element | Re | Ni | Cr | Mo | W |
|---------|----|----|----|----|---|
| wt%     | 7.62 | 49.74 | 15.79 | 6.00 | 20.86 |

Figures 7. The EBSD results: (a) IPF map of the weld; (b) misorientation angle.
Figure 8. The microstructure of the reaction layer near the top Re-side fusion line.

Table 4. EDS result of the corresponding microstructure in the reaction layer.

| Element | wt% | Re  | Ni  | Cr  | Mo  | W   | Area |
|---------|-----|-----|-----|-----|-----|-----|------|
|         |     | 16.17| 44.07| 15.36| 6.11| 18.30| P1   |
|         |     | 9.17 | 47.74| 16.50| 7.24| 19.35| P2   |
|         |     | 5.47 | 53.39| 17.30| 6.45| 17.39| P3   |

Figure 9. The EBSD results: (a) IPF map for Re-side interface at the top; (b) grain boundary; (c) misorientation angle.
Figure 7 shows the EBSD results of the weld microstructure. Figure 7(a) suggests that the columnar crystals exhibit strong grain orientation from both sides of the weld to the center of the weld, and the growth of the grain follows the direction opposite to the cooling direction. But the GH3128-side grains in the weld are more elongated than the Re-side. The microstructure of GH3128 BM consists of austenite grains and twins. The grains on the GH3128-side boundary are epitaxially grown, while the boundary of the FL of the Re-side is relatively smooth due to the different crystal structures between FZ and Re BM. The mechanical properties of the joint would be reduced owing to the difference in the interface microstructure. What’s more, the aggregation of solute elements near the FL of Re-side lead to the formation of stray grains [17]. Figure 7(b) shows that the misorientation angle of the weld grains is mainly composed of high-angle grain boundaries (HAGBs) and a small amount of low-angle grain boundaries (LAGBs).

3.2.2. Reaction layer
Considering that the tensile fracture position was located in the Re-side FL, the microstructure of this zone is analyzed. The red dashed zone in figure 8 is a thin reaction layer like a brazing layer formed between the Re BM and FZ. According to the research of G. Chen [18], the content of Mo in the reaction layer was significantly higher than that of FZ in the Mo/Kovar joint. Table 4 shows the EDS results of the reaction layer in the Re/GH3128 joint. The element content at the interface is quite different compared to other areas in the FZ. The content of the Re element increased from 5.47 wt% to 16.17 wt%, which indicates the segregation occurred in the reaction layer, resulting in the distortion energy of grains increased. Although the solubility of Re in Ni is limited, the rapid cooling process causes more Re to be dissolved in the Ni matrix. S. Chen found that the segregation characteristics would lead to the loss of weld toughness [19], which is detrimental to the mechanical properties of the joint.

The EBSD result of the Re/FZ interface is shown in figure 9. First, the reaction layer grows as a planar crystal near Re BM, and then dendrites grow in the FZ direction on the basis of the planar crystal (figure 9(a)). Figures 9(b), (c) suggest that LAGBs are formed in the Re-enriched reaction layer, whereas, HAGBs are observed in the FZ with lower Re content. Combined with the previous analysis, it could be considered that the enrichment of Re would promote the formation of LAGBs. Newell [20] found that the LAGBs formed have a large shrinkage stress during the solidification process. Moreover, the LAGBs are caused by the dislocation accumulation but the atoms on the grain boundary are placed at the normal node position [21], which exhibits poor crack propagation resistance. Therefore, the crack always develops along the existing direction of LAGBs, showing a straight crack propagation path, thus reducing the mechanical properties of the joint. In the present research, the tensile fracture behavior of the weld also occurred in the LAGBs interface area, as shown in figure 14. Due to the influence of heat input, recrystallization occurs in the coarse-grained heat-affected zone in
the Re BM with a width of about 0.1 mm, and grains tended to grow in the orientation of \(\langle 10\bar{1}0 \rangle\) which is the easiest growth direction.

### 3.2.3. Element distribution

To understand the element distribution and segregation behavior in the joint, EPMA characterization was performed on different zones of the joint. Figures 10(a)–(i) provides the EPMA X-ray map, which shows a semiquantitative measurement of element microsegregation in the Re/weld interface. It can be seen that Re is enriched in the reaction layer near the FL of Re-side while Ni, Cr and Mo are poor in the zone. With the change of the Re element, the width of the reaction layer gradually decreases from top to bottom of the weld. However,
the Re is evenly distributed in other zones of the weld and dendritic structure cannot be distinguished. Ni and Cr elements are slightly segregated in the weld, and their content in the dendritic core zone is slightly higher than that in the interdendritic zone. But, the Mo, C, Ti and Si elements have serious microsegregation in the weld. They are always enriched in the interdendritic zone in the weld, and the dendrite structure can be observed. What’s more, W and Al are distributed evenly in the weld.

Figures 11(a)–(i) provides an EPMA X-ray image of elemental microsegregation in the top center zone of the weld. This shows that Re and W are relatively evenly distributed in the center of the weld and slightly enriched in dendrites core. The segregation behavior of Ni and Cr in the center of the weld is the same as that on the Re side, but the segregation characteristics in the center of the weld are more obvious. Mo, C, Ti, and Si are enriched in the interdendritic zone, while Al is still evenly distributed in the dendrites. The enrichment of Mo and Ti in the interdendritic zone is always accompanied by the impoverishment of Ni and Cr, while the enrichment of C in the interdendritic is always followed by the enrichment of Ti and the impoverishment of Mo. It can be speculated that Ti is a carbide forming element during the solidification of the weld.

Figures 12(a)–(i) shows the EPMA X-ray map near the FL of the GH3128-side. The distribution of elements in the zone is consistent with figures 10 and 11. Re is relatively evenly distributed in the FZ and weakly segregated
in the dendrite structure, as shown in table 5. It suggests that the positive segregation tendency of Re leads to a slight enrichment of Re in the dendrite core during the solidification process. However, carbides rich in Mo and W are formed at the grain boundaries of the GH3128 BM, which is different from the carbides in the weld. Ti is enriched in the GH3128 BM grain boundary to form the TiN compound.

Figure 13 compares the changes of Re content in different zones of the weld, and the EDS results are shown in table 5. It suggests that the dendrite core contains more Re content due to the positive segregation tendency of Re, which accumulates in the dendrite core zone during the solidification process. The lowest Re content is in the GH3128 BM-side, and the highest Re content is in the center weld. This is closely related to the flow of the molten pool in the weld. The flow of the pool brings the melted BM into the central weld to participate in the nucleation.

**Table 5.** EDS results of the corresponding Re-rich phase.

| Element | Re  | Ni  | Cr  | Mo  | W   | C    |
|---------|-----|-----|-----|-----|-----|------|
| wt%     | 17.96 | 23.80 | 23.68 | 29.64 | —   | 4.92 |

**Table 6.** The EDS results at cracks.

| Element | Re  | Ni  | Cr  | Mo  | W   | Area |
|---------|-----|-----|-----|-----|-----|------|
| wt%     | 51.36 | 20.88 | 9.35 | 3.68 | 14.73 | P_1  |
| wt%     | 98.21 | 0.36  | 0.50  | 0.94  | 1.76  | P_2  |

Figure 16. Microhardness curve of the welded joint.
Therefore, there is more Re in the center of the weld, which has the same conclusion with M.J. Torkamany’s research \[22\]. Of course, this is also related to the extremely low diffusion behavior of Re, but the solid-state diffusion behavior of the elements is insignificant during the welding process. Moreover, precipitation phase is also formed in the weld, as shown in figure 14. Table 6 suggests that the existence of the precipitation phase is accompanied by the enrichment of Re and Mo and the impoverishment of W.

### 3.2.4. Defects

Figure 15 shows a microcrack at the BM near the FL of the Re-side. It can be observed that the crack starts from the Re/weld interface and extends to Re BM along the grain boundary. Some ‘island-like’ phases formed in the crack. The EDS results of the crack are shown in table 7. According to the Ni–Re phase diagram \[23\], there is no intermetallic compound phase formed between Ni–Re. It suggests that the ‘islands’ are special Re-based solid solutions that rich in W and Ni due to rapid solidification, and a small amount of Ni atoms diffuses into the Re.
matrix. It can be considered that the crack is a ductile immersion crack formed by the thermal stress on the Re BM. G. Chen proposed the two stages of this kind of crack formation [18]. The first step is the crack growth stage. Affected by the welding thermal stress, Re BM is prone to cracks and growth due to its large number of voids, dislocations and other defects. The liquid metal in the weld would immerse into the cracks due to capillary action [24]. Besides, part of the Re BM would be melted due to heat conduction, which not only promotes the further propagation the cracks, but also leads to the formation of Re-based solid solution phase. The second step is the crack termination stage. The crack sensitivity is reduced by the recrystallization process, and the development of cracks is hindered until the growth stops.

### 3.3. Mechanical properties of welded joints

Figure 16 depicts a microhardness curve of the weld cross-section. The microhardness of the joint is gradually increase from the GH3128 BM to the Re BM. GH3128 BM has the lowest microhardness with an average hardness of 252 HV, and Re BM has the highest microhardness with an average hardness of 335 HV, while FZ has a medium level of microhardness with an average hardness of 277 HV. The highest hardness value of FZ is obtained at the FL of the Re-side, which is associated with stronger solid solution strengthening due to the higher content of Re [8]. When Re is solid-dissolved into the Ni matrix, the degree of lattice distortion of the matrix is improved due to the Re atoms and Ni atoms have different crystal structures and atomic radii, thereby achieving a strengthening effect. Additionally, the formation of a large number of LAGBs at the FL of Re-side further improves the microhardness [25]. Moreover, HAZ has an increase of microhardness due to the recrystallization process of the BMs caused by the heat input.

The distribution of Re in the weld also has a certain effect on the tensile properties of the joint. The fracture of the tensile specimen is shown in figure 17. Figures 17(a), (b) show the fractures morphology on the GH3128-side, and figures 17(c), (d) show the fractures morphology on the Re-side. The tensile strength of the joint reaches 498 Mpa and the elongation is about 3%. Figure 18 suggests that a slight necking occurred at the fracture. The fracture of the GH3128-side is composed of a mass of flaky fracture textures and cleavage steps, showing cleavage characteristics. The characteristics of grain boundaries are observed on the fracture, which is the characteristic of intergranular fracture. Moreover, irregular lumps can be observed on the GH3128-side fracture. The EDS results (table 8) illustrate that the content of Re is consistent with the Re BM. Therefore, it can be inferred that the lumps come from the HAZ of Re BM. The fracture surface of the joint is neat and flat, so its fracture form is brittle. According to the EDS results (table 8) of the fracture and figure 18, it can be determined that the fracture position of the joint is in the zone of the reaction layer formed by the supersaturated solid solution phase.

### 4. Discuss

#### 4.1. Solidification behavior in the fusion zone

The phase transformations in the FZ during the solidification are attributed to the chemical composition and the distribution behavior of solute elements. Affected by the thermal conductivity of the base materials, the dendrite sizes on both sides are not consistent. The GH3128-side grains are epitaxial growth, and the growth direction is opposite to the cooling direction. However, the nucleation of the Re-side weld would appear in the heterogeneous area on the semi-melted base metal at the boundary of the molten pool on the interface [26]. The planar crystal growth at the Re-side molten pool boundary in figure 9(a) can be explained by the highest temperature gradient and lower crystal growth rate at the molten pool boundary. In addition, the growth of planar grains is limited because the crystallization process is gradually advancing into higher temperature liquid metal. And the interface is flat, as shown in figure 8. Subsequently, the dendrite grows toward the center of the weld in a competitive growth manner on the substrate of the plane grain.

The solidification of the FZ starts from Ni-rich γ dendrites, which are accompanied by the distribution of solutes, and the solutes with a distribution coefficient k < 1 are continuously enriched in the liquid phase. Subsequently, MoTi-rich carbides are formed in the interdendritic zone, and its ultra-fine size can be explained.
by the ultra-fast cooling rate of EBW, as shown in figure 11. The composition of the carbides is different from that of the GH3128 BM [27] due to the segregation behavior of elements is limited by the extremely fast solidification process. Therefore, it can be inferred that the solidification sequence is $L \rightarrow L + \gamma \rightarrow \gamma + MoTi$-rich carbides. Besides, the Re-rich phase was formed by the residual liquid phase at the end of the solidification, as shown in figure 14. The Re-rich phase is enriched with unconsumed Re element and Mo element which are always segregated into the liquid phase during the solidification process, while the W element has been consumed in the previous solidification.

4.2. Re distribution behavior in the weld
As shown in figure 10, although most of the molten Re is brought into the weld and mixed during the flow of the molten pool, whereas, a small amount of Re is still retained in the reaction layer zone. As the molten pool flows, the width of the reaction layer gradually decreases from the top to the middle of the weld. Meanwhile, the semi-melted Re HAZ would also diffuse a trace amount of Re element into the weld, which is also the reason for the enrichment of Re in the reaction layer. The same phenomenon was found in the research of M.J. Torkamany [22]. Additionally, figures 9(a), (b) suggest that the LAGBs in the reaction layer are completely located inside the plane grains due to the Re segregation into the dislocations of the matrix during the solidification process. The hardening of the material is also related to the segregation of Re, which can pin LAGB dislocations and slow down the dislocation reaction [28]. Thus, the highest value of hardness of the FZ appears on the Re-side FL, as shown in figure 16.

According to figures 10–12 and table 5, Re exists in the Ni matrix in the form of solid solution in the FZ, and the solute element concentration distribution in the weld is relatively uniform. It suggests that despite the solidification time in EBW is extremely short, the elements have been completely mixed in the molten pool due to the strong stirring effect of the electron beam. The presence of Re in the solidification process not only effectively hinders the segregation behavior of Al [29], but also reduces the degree of segregation of W and Re [30], which explains the uniform distribution of Re, W and Al elements in the weld. Moreover, Re is not involved in the formation of carbides.

4.3. The effects of microstructure on mechanical properties of the joint
In the study, the hardness of the joint was improved by the solid solution strengthening effect of Re. In particular, the reaction layer near FL on the Re side contains a higher content of Re, which leads to the maximum hardness in FZ. The higher hardness value of the Re-side is also attributed to the fine grain strengthening of the smaller dendritic structure. Besides, the fracture of the joint occurs at the position of the reaction layer of the FL on the Re-side, which is not only related to the supersaturated solid solution strengthening of Re in the reaction layer, but also affected by the LAGBs. The LAGBs energy comes from the dislocation energy. But the dislocation density is determined by the misorientation angle between the grains, and LAGBs are formed by the dislocation stacking. According to the Read-Shockley formula:

$$\gamma = \gamma_0 \theta (A - \ln \theta)$$  \hspace{1cm} (1)

$$\gamma_0 = \frac{Gb}{4\pi(1 - \nu)}$$  \hspace{1cm} (2)

where $\gamma_0$ is a constant that depends on the material’s trimming modulus $G$, $b$ is the Burst vector $v$, is Poisson’s ratio, and the integral constant $A$ is determined by the atomic dislocation energy of the dislocation center. The formula (1) suggests that the grain boundary energy of LAGB increases with the increase of the misorientation angle. The strengthening of LAGB [31] of the reaction layer increases the microhardness and reduces the toughness of the joint. Furthermore, the formation of HAGBs during solidification can release intragranular and intergranular stress [32], which is beneficial to improve the deformability of the joint. Therefore, the fracture position of the joint always appears in the reaction layer and exhibits hard and brittle characteristics.

5. Conclusion
The microstructure and mechanical properties of Re and GH3128 EBW joints are affected by the distribution of Re in the weld. The main conclusions of this paper are as follows:

(1) Affected by the reaction layer, the microstructure on the Re-side transitions from plane grains to dendrites. The grains on both sides grow to the center weld in a competitive growth manner. Moreover, the solidification sequence is $L \rightarrow L + \gamma \rightarrow \gamma + MoTi$-rich carbides.

(2) In the FZ, Re is uniformly distributed in the Ni matrix with the form of solid solution elements, and the segregation is not obvious.
(3) Affected by Re solid solution strengthening, the hardness of the weld is improved compared with GH3128 BM. The higher hardness of the Re fusion line is attributed to the higher Re content in the reaction layer and the LAGB strengthening. However, this makes the joint exhibit brittle fracture characteristics.

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Data availability statement

The data that support the findings of this study are available upon reasonable request from the authors.

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