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Effect of solution treatment on the microstructure of dissimilar welding joints of Ta-10W and GH3128

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Abstract

Electron beam welding (EBW) process is employed to weld Ta-10W and GH3128 plates of 2 mm thick, which is very beneficial to the development of aerospace engine combustor. The microstructure of the as-welded (AW) and solution-treated (ST) samples are characterized. The phase transformation program of the fusion zone (FZ) of GH3128 and Ta-10W joints is $L \rightarrow L + M_6C \rightarrow L + \gamma + M_6C \rightarrow L + \gamma + M_4C + M_{23}C_6 \rightarrow \gamma + M_4C + M_{23}C_6 \rightarrow \gamma + M_{23}C_6 + P$. Whether in the AW or ST state, the reaction layers formed by hard intermediate phases are found between the FZ and Ta-10W. Comparing the two states, voids zone is formed at the interface of the reaction layer near the FZ in the ST state. Additionally, it is also found that the growth of the reaction layer was promoted at the solution temperature. It is certain that the solution treatment keeps many positive significances. First of all, the banded P phase worsening the properties of materials has not been formed. Secondly, the performance of the FZ is strengthened by the dispersed granular $M_6C$. Thirdly, the distribution of coincident-site lattice (CSL) grain boundaries (GBs) in the FZ is more uniform than that in the AW state.

1. Introduction

Ta-10W alloy is widely utilized in aerospace, weapons, missiles, shipbuilding industry and other fields, due to its high temperature strength, good mechanical properties, good weldability, excellent corrosion resistance and strong impact resistant capacity [1–3]. For example, the combustion chamber of American Agena spaceship is made of this alloy [4]. Due to Ta and W are scarce in nature and expensive, it is particularly important to realize the connection between Ta-10W and other metals or alloys. Ni-base alloy GH3128 is also widely applied in aero-engine combustor because of its excellent performance [5]. If Ta-10W and GH3128 are welded together and applied to the combustion chamber of aero-engine, the thrust-to-weight ratio can be increased and the material cost can be extremely saved.

The melting point of Ta-10W is 3035 °C, but its oxidation resistance at high temperature is very poor, which is due to the fact that oxidation will occur at about 500 °C without any protection [4, 6]. However, the melting point of GH3128 is only 1380 °C, which is quite different from that of Ta-10W. When two kinds of materials with different physical properties are welded, the melting ratio can be adjusted and the metallurgical conditions of welds can be improved, due to the controllable energy density and strong penetration of EBW. Therefore, the grains of welds are fine and the properties of joints are good [7, 8]. Moreover, the vacuum environment can purify molten metal. Therefore, it is the preferred method for the welding of Ta-10W and GH3128.

GH3128 is a kind of solid solution strengthened Ni-base superalloy, which is single-phase austenite with a small amount of fine and uniformly distributed TiN and $M_6C$ [9]. The addition of alloying elements will lead to the expansion of the Ni-rich FCC lattice, resulting in the pure strengthening of the austenite phase. Ta and W are the common solid solution strengthening elements in Ni based superalloy, but the higher content of them will increase the tendency of topologically close-packed (TCP) phase and worsen the properties of the alloy [10–12].

The FZ was formed by partially melted Ta-10W and GH3128 during the EBW process. The solidification sequence of the FZ was changed due to the entry of a large number of Ta and W elements into the FZ.
Perricone et al. [13] studied the effect of element composition on the solidification behavior of Ni–Cr–Mo and Fe–Ni–Cr–Mo systems, and considered that the solidification transformation program of Ni-20Cr-12Mo was $g \rightarrow + \rightarrow LL$, while those of Ni-20Cr-24Mo and Ni-44Fe-20Cr12Mo were $g \rightarrow + \rightarrow LL$ and $g \rightarrow + \rightarrow PP$ and $g \rightarrow + \rightarrow LL$, respectively. Thus, it could be seen that the element composition obviously exerted a huge impact on the solidification sequence. In electron beam welded joints, the effect of a large number of Ta and W elements on solidification behavior urgently needs to be revealed. Up to now, there are still many problems to be solved for the welding of refractory metals or alloys with other metals of lower melting point. EBW of molybdenum and Kovar alloy, and that of tantalum and GH3128 had been studied, and both found that the fracture occurred in the reaction layer during tensile tests [9, 14]. However, these studies focused on the welding process and did not explain the behavior of the reaction layer in detail. Therefore, the problems of reaction layer in the welding process need to be discussed further.

In addition, the molten pool metal was remelted and solidified during welding, and a large amount of Ta and W were melted into the undesirable phase, which led to the deterioration of the properties of the joint [15–17]. Generally speaking, the post-weld heat treatment (PWHT) of superalloy weldments is carried out for the following reasons, including the elimination of residual stress, homogenization, dissolution of unwanted secondary phases that may be formed in the FZ or heat affected zone, stable size and improvement of corrosion resistance. Xing et al. [18] investigated the PWHT of Inconel625 deposited metal. It was considered that the segregation behavior decreased with the increase of PWHT temperature, and the composition inhomogeneity disappeared at 950 °C. Li et al. [19] found that the grains grew randomly with dislocation annihilation and twins after high temperature annealing of selective laser melting Inconel625. Hu et al. [20] conducted solution treatment of laser solid forming Inconel625. It was found that recrystallization occurred completely when the solution temperature exceeded 1200 °C, the average diffusion coefficient and dissolution rate increased obviously with the increase of solution temperature, and the volume fraction and size of Laves phase and the dislocation density were decreased obviously. However, the related reports regarding the PWHT on the weld joint of GH3128 and Ta-10W through the EBW process are still negligible in the literature. Although Inconel625 and GH3128 are similar in chemical composition, the solidification conditions of EBW process and the characteristics of a large number of Ta and W melting into the FZ should be fully considered. Therefore, the problems related to the PWHT of the joint also need to be revealed.

### Table 1. Chemical compositions of GH3128 (in wt%).

| Chemical element | Ni    | Cr     | W      | Mo    | Al    | Ti    | C     |
|------------------|-------|--------|--------|-------|-------|-------|-------|
| wt%              | Bal.  | 19.0–22.0 | 7.5–9.0 | 7.5–9.0 | 0.4–0.8 | 0.4–0.8 | ≤0.05 |
| Fe ≤0.2          | B     | ≤0.005 | ≤0.06  | ≤0.05 | ≤0.5  | ≤0.8  | ≤0.013 | ≤0.013 |
| Fe ≤0.2          | B     | ≤0.005 | ≤0.06  | ≤0.05 | ≤0.5  | ≤0.8  | ≤0.013 | ≤0.013 |

### Table 2. Chemical compositions of Ta-10W (in wt%).

| Chemical element | Ta    | Mo    | W    | Ti | Ni | Si | H | N | O | C |
|------------------|-------|-------|------|----|----|----|---|---|---|---|
| wt%              | Bal.  | 0.02  | 11.0 | 0.1 | 0.1 | 0.005 | 0.015 | 0.015 | 0.01 |
In this paper, the solidification transformation procedure of the FZ of electron beam welded Ta-10W and GH3128 joint, and the formation and element diffusion of reaction layer in AW and ST states are discussed for the first time. In addition, the reasons for the differences in microstructures and mechanical properties between the two states are explained. It is expected to provide theoretical support for the engineering application of EBW of Ta-10W and GH3128.

2. Experiment

K110 electron beam welding machine with a vacuum of $2.3 \times 10^{-4}$ mbar was employed to weld 2 mm thick plates of GH3128 and Ta-10W. Before welding, sandpaper was used to wear off the marks left by wire cut, and then acetone was used to clean the workpiece to be welded. The chemical composition of GH3128 and Ta-10W are given in tables 1 and 2. The schematic diagrams of welding assembly and formation of the FZ are shown in figure 1. The EBW parameters are presented in table 3. It is worth noting that the rational use of fixture is a favorable way to prevent welding deformation.

After welding, one of the workpieces was polished and ultrasonic cleaned to remove the oil stain and oxide film on the surface, then it was put into the vacuum heat treatment furnace for solution treatment. The temperature curve of solution treatment is shown in figure 2. The temperature in the furnace was reduced to 300 °C in about 6 min when argon quenching was used. The AW and ST workpieces were cut by wire cut and made into two groups of 4 mm × 4 mm × 10 mm metallographic samples. The metallographic samples were polished to the surface without scratches, and both of them were corroded with prepared aqua regia (the ratio of hydrochloric acid to nitric acid is 3:1).

Table 3. Process parameters in the electron beam welding.

| Acceleration voltage/kV | Beam current/mA | Welding speed/mm/s |
|-------------------------|-----------------|--------------------|
| 120                     | 6               | 10                 |

Figure 2. Temperature-time curve of heat treatment.

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The characteristic microstructures and morphologies of the reaction layer and FZ were observed by field emission scanning electron microscope (FEI Nova Nano 450 SEM), and the chemical compositions of the corresponding microstructure were analyzed by energy dispersive spectrometer (EDS). The microhardness distributions of the joints were characterized by HDX-1000TMC/LED Vickers microhardness tester under the condition of 200 g load and 15 s dwell time. Electron backscatter diffraction (EBSD) samples were prepared by electrolytic polishing, and the grain orientation distribution and characteristics were observed by Oxford Nordly max3 EBSD.
Figure 3. Microstructures and (a) cross section of welded joint of AW state. FZs (b) reaction layer, (c) equiaxed dendrites, and (d) FZ/GH3128 interface. And (e) cross section of welded joint of ST state. FZs (f) reaction layer, (g) bulk γ matrix, and (h) FZ/GH3128 interface.
3. Results

3.1. Microstructure characterization

The metallographic images of AW state are presented in figures 3(a)–(d). It is observed that there are no solidification cracks in the welded joint, which reveals that promising welded joints can be obtained by EBW of GH3128 and Ta-10W. The microstructure of the FZ shows substructural dendrites, and region c shows a mixture of equiaxed and dendritic crystals. The change of dendrite morphology during solidification is attributed to undercooling, i.e., the ratio of temperature gradient (G) to growth rate (R) \[21\]. Typically, G is relatively smaller and R is larger in the center of the weld, which results in a smaller G/R ratio and a larger undercooling, so that the equiaxed grains can be observed in the weld. In addition, a reaction layer was observed near the fusion line on the Ta-10W side, while it was not found in that near the GH3128 side. The metallographic images of ST state are presented in figures 3(e)–(h). Compared with the microstructure morphology of AW state, it is found that the microstructure of the FZ is mainly massive austenite, which is because the solution treatment gives sufficient diffusion time for each element. And it makes the original dendrites grow fully until the interdendritic region disappears and eventually forms massive grains.

Table 4. EDS results of the points marked in figure 4 (at %, wt %).

| Chemical element | Ni    | Cr    | Ta    | W     | Mo    |
|------------------|-------|-------|-------|-------|-------|
| P1               | 50.78/27.61 | 7.25/3.49 | 34.21/37.33 | 5.74/9.77 | 2.02/1.8 |
| P2               | 66.08/53.7 | 19.32/13.9 | 7.87/19.7 | 3.10/7.88 | 3.63/4.82 |
| P3               | 55.40/40.68 | 21.90/14.24 | 12.72/28.78 | 3.95/9.07 | 6.03/7.23 |
| P4               | —     | —     | 86.67/86.48 | 13.33/13.52 | —     |
| P5               | 49.42/26.35 | 7.66/3.62 | 38.46/63.19 | 3.70/6.17 | 0.76/0.67 |
| P6               | 31.72/15.43 | 15.02/6.47 | 44.69/67 | 5.89/8.97 | 2.68/2.13 |
| P7               | 40.29/26.23 | 27.51/15.86 | 17.92/35.96 | 6.94/14.15 | 7.34/7.8 |
| P8               | 52.12/35.61 | 18.05/10.92 | 13.30/28.01 | 6.88/14.71 | 9.65/10.75 |
| P9               | 72.23/50.99 | 4.35/2.72 | 19.40/42.25 | 1.47/3.24 | 2.55/0.8 |

Figure 4. (a) and (b) SEM photos of the reaction layer of AW state, and (c) crack and (d) voids of ST state.
Figure 4 shows field emission SEM images of the reaction layer in the two states. For the AW state, it is found that the morphology of the reaction layer near the Ta-10W side is a single phase, while that near the FZ side is a typical eutectic structure, which is similar to that of electron beam welded joints of molybdenum and Kovar alloys [14]. For the ST state, it is found that the reaction layer is obviously delaminated, and accompanied by the trace of diffusion. Additionally, cracks and voids are observed at the boundary of the reaction layer near the FZ. EDS was employed to analyze the chemical composition of these characteristic morphologies, and the results are shown in table 4. Considering the complexity of the issues related to the reaction layer, it will be discussed in detail in section 4.1.

Figure 5 shows the characteristic morphology of the FZ in AW state. The bright bands and dispersed particles were observed, and the chemical compositions of the bands are shown in table 5. Compared with the constituent elements of matrix P11, the contents of W and Mo in the bright bands of P10 are much higher. Although W and Mo are solid solution strengthening elements in superalloys, they are also the main elements for the formation of TCP brittle phase. In general, the solid solution strengthening elements W and Mo are consumed in large amounts to stabilize the P phase. Therefore, it can be determined that the white banded morphology is P phase. The fine and dispersed particles precipitated between dendrites were preliminarily inferred to be nano-sized carbides, which were also observed in the FZ during welding of Hastelloy N superalloy [22]. Similarly, figure 6 shows the characteristic morphology of the FZ in ST state. The massive precipitates in both the grain boundary and the intra-crystalline, and fine granular precipitates similar to those in the AW state are still observed, and the chemical constituents of the massive precipitates are shown in table 6. P12 and P13 are similar in chemical composition with higher contents of W and Mo, which indicates that the tendency of formation of M6C type carbide is more obvious [23]. In addition, dispersed nano-sized carbides which have been found in the two states

| Chemical element | Ni       | Cr       | Ta       | W       | Mo       | Al       | Ti       |
|------------------|----------|----------|----------|---------|----------|----------|----------|
| P10              | 37.53    | 18.86    | 14.38    | 13.73   | 14.70    | 0.27     | 0.52     |
| P11              | 56.26    | 19.68    | 9.96     | 6.65    | 6.24     | 0.61     | 0.71     |

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will be determined by using thermodynamic calculation in 4.2. By comparing the typical morphologies of the FZ in two states, it is found that the P phase disappears after solution treatment, the higher contents of W and Mo are enriched in M₆C.

Figure 7 shows the EBSD pole maps of the joints of AW and ST states. Ta-10W is the crystal structure of BCC, while the matrixes in GH3128 and FZ are FCC. It is worth noting that the phases in the reaction layer, carbides and P phase with complex crystal structure are removed by noise reduction. Figure 7 is the mapping of the detection region X₀, Y₀, Z₀, which can reflect the spatial distribution of orientation, i.e., it can reflect the parallel crystal direction (corresponding to the color in IPF color code) of the sample characteristic direction (X₀, Y₀, Z₀) of each pixel. It can be seen that most of the grains of Ta-10W base metal show ⟨001⟩ orientation in Y₀ direction and ⟨101⟩ direction in Z₀ direction in the two states. For GH3128 base metal, the color distribution is more uniform along the direction of X₀ and Y₀ and Z₀, which means that there is no obvious preferred orientation of grains in GH3128. As far as the FZ is concerned, the grains near the Ta-10W side show the ⟨101⟩ orientation in the Y₀ direction in the AW state, while the ⟨001⟩ orientation in the X₀, Y₀, Z₀ direction is observed in the ST state. However, the grains near the GH3128 side have no obvious preferred orientation in both states. In addition, the grains grow along both sides of the base metal to the centerline in the FZ, which is usually called epitaxial growth [24].

Figure 8 shows the distribution of grain misorientations of FCC and BCC crystal structures in the field of view of figure 7, respectively. The dotted lines represent the standard Mackenzie distribution which is texture-free in polycrystals with completely random orientation. Thus, it can be seen that the statistical grain misorientation is different from the completely random distribution. The grain boundary with a rotation angle of 60° is often called Σ3 boundary, which belongs to a type of CSL grain boundary. The volume free energy of Σ3

### Table 6. EDS results of the points marked in figure 6 (wt %).

| Chemical element | Ni   | Cr   | Ta   | W    | Mo   | Al  | Ti  |
|------------------|------|------|------|------|------|-----|-----|
| P₁₂              | 30.30| 15.83| 9.96 | 26.81| 16.27| 0.39| 0.44|
| P₁₃              | 29.04| 16.10| 9.16 | 26.76| 18.26| 0.40| 0.26|
grain boundary is lower than that of the other CSL grain boundaries, and $\Sigma 3$ grain boundary is less prone to corrosion and intergranular cracks than other high angle GBs (misorientation degree is more than $15^\circ$) [25, 26]. The toughness of the materials can be improved when the appropriate process is adopted to increase the number of CSL grain boundaries [27]. In order to clearly present the distribution of CSL grain boundaries, the results obtained by analysis software are shown in figure 9, in which the red part represents the distribution of $\Sigma 3$ grain boundaries and other colors represent other CSL grain boundaries. It is found that $\Sigma 3$ grain boundaries are mainly distributed in the base metal GH3128, while only a few other CSL grain boundaries are distributed in the FZ, which predicts that the joint is more prone to corrosion and intergranular cracks than GH3128. In addition, the CSL grain boundaries in the AW state are obviously concentrated on the GH3128 side, which predicts that the creep property and intergranular corrosion resistance of the FZ near the GH3128 side is better than those near the Ta-10W side. While the CSL grain boundaries in the ST state are more dispersed, which predicts that the creep and intergranular corrosion resistance are more uniform than those in the AW state.

### 3.2. Microhardness testing

The hardness distribution is shown in figure 10. The microhardness of the reaction layers in the AW and ST states are 549.4 HV and 750.3 HV, respectively. It can be seen from table 4 that the content of Cr in the reaction layer of the ST state is higher than that of the AW state, which causes an increase in the microhardness of the reaction layer. In addition, the average microhardness of the FZ in the AW state is 272.2 HV and that in the ST state is 341.2 HV. The increase of microhardness is due to the formation of diffusely distributed $M_6C$ in the FZ after solution treatment.
Figure 8. Measured misorientation distribution functions (MDFs).

Figure 9. Distribution of CSL GBs of the joint: (a) AW and (b) ST.

Figure 10. Microhardness distribution profile of the joints: (a) AW and (b) ST. HV: Hardness value.
4. Discussion

4.1. Reaction layer

The generation of reaction layer is a common problem in dissimilar metal welding. Chen et al [28] investigated the EBW of tantalum and austenitic stainless steel and found that intermetallic compounds such as brittle ε phase (Fe₂Ta) and μ phase (FeTa or Fe₇Ta₆) distributed in layers between tantalum and the FZ. Similarly, they also found the formation of reaction layer in subsequent EBW of molybdenum and Kovar alloys [14].

In this study, it is found that the structures with different morphologies of the reaction layer mainly contain elements such as Ni, Cr, Ta, W and Mo. Due to the difference of crystal structure between Ni (FCC) and other four solute metal elements (BCC), it is inevitable that the solvent atom Ni cannot infinitely dissolve the solute atoms such as Cr, Ta, W and Mo. The atomic radius differences between Cr, Ta, W, Mo and Ni and their solubility in Ni at 1000 °C are listed, as shown in table 7 is given out [29]. It is found that the solubility of Ta in Ni is much lower than that of other elements at 1000 °C, and the difference of atomic radius between Ta and Ni is much larger than that of other atoms. A large number of experiments show that when other conditions are
similar, when the atomic radius difference $\Delta r \geq 15\%$, the solubility decreases with the increase of $\Delta r$ [30]. In addition, the electronegativity difference between Ta and Ni is larger than that of the other three elements and Ni.

Compared with table 5, it is $\gamma$ matrix when the content of Ta is 9.96%wt in the chemical composition of P11, while Ta is easy to combine with W and Mo to form intermetallic compound P phase when the content of Ta is 14.38%wt in that of P12. Compared with table 4, it can be judged that the reaction layer is the intermediate phase formed by Ni, Cr, Ta, W and Mo, due to the Ta content in the characteristic region of the reaction layer is very high. Usually, the intermediate phase can be a compound or a compound-based solid solution (secondary solid solution). On the other hand, the microhardness of the reaction layer is much larger than that of the base metals, and from this result, hard intermediate phases are formed in the reaction layer. The schematic diagram is shown in figure 11 to facilitate the understanding of the reaction layer in the two states. It is considered that the reaction layer near the Ta−10W side is the same intermediate phase because P1 and P5 are similar in chemical compositions, and the other characteristic regions with great differences in components are considered to be different intermediate phases. The typical eutectic structure formed for the reaction layer of AW joint which is observed in figure 4 consists of two different intermediate phases (intermediate phase B and intermediate phase C). In addition, it is found in figure 4 that the reaction layer in the AW state is thinner than that in the ST state, which is due to the fact that the growth of the reaction layer is promoted at higher temperatures [31].

As mentioned above, voids and cracks occur at the interface between the reaction layer and the FZ in the ST state. When the voids are dense, they will join together to form cracks. However, there are no voids and cracks in the AW state, which is mainly due to the diffusion of a large number of elements at solution temperature. Figure 12 shows the element changes of characteristic components such as P4, P5, P6, P7, P8 and P9. It is found that a large number of Ni and Cr in the FZ diffuses into the reaction layer. Although Ta also diffuses into the FZ, this is not enough to make up for the vacancy left by the diffusion of elements such as Ni and Cr. This results in the formation of voids, which are called Kirkendall voids [32, 33].

In addition, it is found that the atomic proportion trend from P5 to P6 is quite different from the overall trend. The atomic composition of the intermediate phase A is relatively fixed, then the surplus Ta atoms are

Table 7. Summary of approximate atomic diameters and solubility data for Ta, W, Cr and Mo in Ni at 1000 °C [29].

| Solute | Approximate atomic size difference compared to nickel, % | Approximate solubility in nickel at 1000 °C, wt% |
|--------|-------------------------------------------------------|-----------------------------------------------|
| Ta     | −15                                                   | 14                                            |
| W      | −10                                                   | 38                                            |
| Cr     | −0.3                                                  | 40                                            |
| Mo     | −9                                                    | 34                                            |
squeezed to the P6 position during the formation of the stable intermediate phase A, resulting in the increase of the atomic proportion of Ta elements in the P6 position. Moreover, the atomic proportion of Cr hit a peak at P7 due to the change of Ni element content, which is found that the sum of atomic proportion of Ni and Cr at P7 and P8 is relatively close in table 4. Thus, it can be seen that the chemical composition of intermediate phase D is dynamic. When the diffusion of Ni element decreases, then Cr element is added in time to stabilize intermediate phase D.

4.2. Phase transformation in FZ

Many studies have used thermodynamic methods to calculate the solidification sequence to understand the solidification transformation program of the FZ in the welding process [34–36]. Similarly, the solidification phase diagram of the FZ of Ta-10W and GH3128 electron beam welded joints was calculated by thermodynamics, as shown in figure 13 is given out. γ, P and dispersed nano-carbides are found in the FZ in the AW state, and γ, M₆C and dispersed nano-carbides are found in the ST state, which are consistent with the results of thermodynamic calculation. The nanoparticles are both dispersed in the FZ in the two states, and it can be inferred that it is M₂₃C₆ combined with thermodynamic calculation. The thermodynamic calculation results are in good agreement with the actual FZ.

According to figure 13 and the above analysis, the crystallization process and reaction scheme in the FZ during the welding process is as follows:

(i) At 1389 °C, a large amount of refractory metallic elements (e.g., Ta, W and Mo) start to combine with C to form M₆C, resulting in the start of M₆C precipitation in the liquid phase.

(ii) At 1335 °C, the solidification of the liquid metal in the melt pool is accelerated and the Ni-rich γ matrix begins to precipitate in the FZ.

(iii) The M₆C precipitation peaks at 1319 °C. As the temperature decreases during solidification, M₆C is gradually converted to the stable M₂₃C₆. At 1314 °C, the solidification of the molten pool metal is complete.

(iv) At 791 °C, the P phase starts to precipitate. As the temperature decreases, the γ and M₆C are gradually reduced and a large amount of free Ta, W and Mo combine to form the P phase.

(v) At 650 °C, M₆C disappears completely. M₆C is completely converted to M₂₃C₆ and the surplus elements such as Ta, W and Mo combine to form P phase.

Hence it can be inferred that the phase transformation program of GH3128 and Ta-10W EBW joint is L → L + M₆C → L + γ + M₆C → L + γ + M₆C + M₂₃C₆ → γ + M₆C + M₂₃C₆ → γ + M₂₃C₆ + P.

Similarly, the P phase dissolves at 1180 °C after solid solution treatment, which results in a large amount of free W and Mo in the FZ. At the same time, a large number of C and Cr are found in the FZ because part of the M₂₃C₆ is dissolved. These free elements aggregate to form M₆C, and a large number of solid-solution strengthening elements such as W, Mo, Cr will combine with Ni to form γ matrix when C element is exhausted, which leads to the increase of γ after solution treatment.

5. Conclusions

1. The program of phase transformation in the FZ of GH3128 and Ta-10W EBW joint is L → L + M₆C → L + γ + M₆C → L + γ + M₆C + M₂₃C₆ → γ + M₆C + M₂₃C₆ → γ + M₂₃C₆ + P.

2. Whether in the AW state or ST state, the reaction layers formed by different intermediate phases were found between the FZ and Ta-10W. And in the ST state, voids zone is formed at the interface of the reaction layer near the FZ. This is mainly due to the diffusion of elements at solution temperature. At the same time, the reaction layer will grow at higher temperature.

3. In the ST state, the potential deterioration of the FZ may be reduced due to the banded P phase is not formed. And the dispersed granular M₆C strengthens the performance of the FZ. Similarly, the distribution of CSL grain boundaries in the FZ is more uniform than that in the AW state.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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