Effects of the beam offset on microstructure and properties of electron beam welded tantalum and Inconel 718 joints

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Abstract
In this paper, electron beam welding of tantalum and Inconel 718 superalloy was performed. The formability, microstructure, defect characteristics and mechanical properties of joints were investigated by controlling the position of the electron beam. The weld zone of tantalum and Inconel 718 joints was mainly composed of columnar crystals and dendrites during the welding of non-beam offset and 0.5 mm beam offset to tantalum. The reaction layer composed of a large number of intermetallic compounds was found on the tantalum side, and it was the place where the fracture occurred. Tensile strength of the joints was 313 MPa and 138 MPa, respectively, and the joints exhibited brittle fracture mode due to the formation of voids and cracks in the reaction layer. The microhardness of the weld zone was higher than that of the base metal due to the strengthening effect of tantalum. Fortunately, when the beam deviated by 0.5 mm to the Inconel 718 side, equiaxed grains formed in the weld zone, and the morphology of the reaction layer changed, which improved the toughness of the joint. The tensile strength of the joint reached 480 MPa under the condition of 0.5 mm beam deviated to the Inconel 718 side.

1. Introduction
Tantalum (Ta) is a refractory metal with a melting point of up to 2996 °C, that is widely used in products, including aero engines, capacitors, and corrosion resistance components, in a variety of fields, including the aerospace, electronics, and chemical processing fields, owing to its good wear resistance and corrosion resistance, excellent weldability, and high-temperature performance [1, 2]. The uneven distribution of Ta ores limits availability of Ta, and complicated processing and high extraction price, all of which seriously hinder the application of Ta [3]. Therefore, the dissimilar joining of Ta and other metals is required to reduce the cost of Ta. Inconel 718 alloy is one of the most widely used superalloys. It has excellent high-temperature strength, good weldability and machinability, and is usually used as an aero-engine part [4, 5]. Therefore, the connection of Ta and the Inconel 718 alloy can not only reduce the weight of the engine (density of 16.6 g cm⁻³ and 8.47 g cm⁻³, respectively) but also reduce the use of Ta. Although Ta is stable at room temperature, Ta will react with oxygen, nitrogen, and hydrogen at high temperatures and form brittle compounds, which will significantly affect the mechanical properties and corrosion resistance of Ta [1]. Therefore, the welding of Ta and Inconel 718 alloy must be carried out in a vacuum or in a protective inert gas environment.

To date, research on Ta and superalloy welding has been less reported, however research on Ta and other metals is more available and can be used as a reference. The common dissimilar welding of Ta and other metals is mainly carried out by the methods of brazing, arc welding, explosive welding, diffusion bonding and electron beam welding. Two series of filler metals Ni-Cr-Fe-Si-B and Ni-Cu-Ge-Si-B were designed by Ji [6], and were used to braze Ta1 and S30408 plates in a 10⁻² Pa vacuum. Both brazing filler metals could achieve effective brazing of Ta1 and S30408. The highest shear strength of the joint formed by Ni-Cr-Fe-Si-B filler metal acquired at 1030 °C for 2 h was 169 MPa. Zhou et al [7] carried out diffusion bonding of Ta and Cu from 800 °C–1200 °C, it was shown that the voids disappeared with the increase of diffusion time and holding time, and the maximum
shear strength of the joint reached 121.1 MPa at 1000 °C for a holding time of 60 min Xing et al [8] used direct current (DC) argon arc welding and alternating current (AC) pulsed argon arc welding to weld Ta steel composite plates. A narrower heat affected zone (HAZ) and better welding joints were obtained when using AC pulsed argon arc welding. However, the HAZ of DC argon arc welding joint was quite wide, and intermetallic compounds were found in the middle interlayer. Kosec et al [9] showed that island-like alloys were formed on the explosive welding interface between low-carbon steel and Ta. These alloys were composed of a mixture of Ta and steel that improved the strength of the joint. At the same time, some non-metallic oxides whose main component was Ta2O5 were also found in the joint. Chen et al [10] conducted electron beam welded Ta and austenitic stainless steel with different beam offset, it was presented that a well-formed joint was obtained when the beam deviated by 0.2 mm to the stainless steel, and lamellar distribution of Fe2Ta, Fe3Ta and Fe7Ta6 near the Ta side fusion line reduced the tensile strength of the joint, which was only 255 MPa.

Although brazing, explosive welding and diffusion bonding can achieve the welding of Ta and other metals, the welding process is complex, the welding time is longer, and the tensile strength of the joints is lower than that of the base metal. The fusion welding processes are widely applied for the connection of superalloy or refractory metal [11]. Compared with arc welding, high energy density beam welding is considered as a promising technology for welding dissimilar metals due to its small welding deformation, narrow heat-affected zone, and lower heat inputs [12–14]. The Ta and Inconel 718 welding process must be conducted in vacuum or inert gas environment, hence high vacuum electron beam welding (EBW) is a preferred method to solve these problems.

The butt joining of 2-mm-thick Ta and Inconel 718 alloy plates was conducted by EBW for the first time. The formability, microstructure, defect characteristics and mechanical properties of the joints are studied. Attempts are made to improve the joint performance through electron beam offset.

| Table 1. Chemical compositions of Ta (in wt%). |
| Ta | Si | Mo | W | Ti | Ni | H | N | O | C |
| Bal. | 0.005 | 0.02 | 0.05 | 0.1 | 0.1 | 0.0015 | 0.01 | 0.015 | 0.01 |

| Table 2. Chemical compositions of Inconel 718 (in wt%). |
| Ni | Cr | Fe | Mo | Nb | Co | Cu | C | Mn | Si | Al | Ti | S |
| 50–55 | 17–21 | Bal. | 2.8–3.3 | 4.75–5.5 | 1.0 | 0.3 | 0.08 | 0.35 | 0.35 | 0.2–0.8 | 0.65–1.15 | 0.01 |

| Table 3. Process parameters for electron beam welding. |
| Sample | Acceleration voltage/kV | Beam current/mA | Welding speed/mm/s | Beam offset/mm |
| SN1 | 120 | 7 | 5 | −0.5 |
| SN2 | 120 | 7 | 5 | 0 |
| SN3 | 120 | 7 | 5 | 0.5 |
2. Experiment

In this experiment, 2-mm-thick Ta and Inconel 718 alloy were used as the base metals. Their chemical compositions are listed in table 1 and table 2. Figure 1 shows the schematic diagram of the butt joint, and it was noted that Ta and Inconel 718 plates were fixed with clamps. Before welding, the oxide film, stain and other impurities were cleaned by physical and chemical methods. The experiments were performed on a vacuum electron beam welding machine (K110). According to preliminary experiments and related researches \[15, 16\], the beam offset value was set to 0.5 mm, and the process parameters are listed in table 3. Among them, the electron beam deviated by 0.5 mm to the Inconel 718 alloy side as sample SN1; sample SN2 was the joint made by central welding; and the electron beam deviated by 0.5 mm to the Ta side was sample SN3.

After welding, metallographic samples and tensile samples were produced by wire cutting. The tensile sample is shown in figure 2. The metallographic samples were ground, polished and etched. Aqua regia solution was used for the welded samples. The metallographic structure was observed with a VHX-600K optical microscope (OM) and S-3400N scanning electron microscope (SEM). The composition of the weld was analysed by energy dispersive spectroscopy (EDS). Elemental distributions were determined by an electron probe micro-analysis (EPMA). The tensile test was performed by an electronic universal testing machine (AG-IC) with a tensile speed of 1 mm min\(^{-1}\), and each type of joint was measured three times. A microhardness tester (HXD-1000 TMSC/LCD) was used to characterize the microhardness distribution of the joint with a 200 g load and 15 s dwell time, and the horizontal distance between adjacent indentations was 0.1 mm.

3. Results and discussion

3.1. Macrostructure

Figure 3 shows the cross-sectional morphology of samples SN1-SN3 of the Ta and Inconel 718 alloy welded joints. The macro morphology of the joint of sample SN2 is shown in figure 3(b). Both Ta and Inconel 718 melted, but the melting volume of Inconel 718 was significantly greater than that of Ta. According to table 4, the melting point and thermal conductivity of Ta were higher than those of Inconel 718, and more Inconel 718 entered the molten pool, making the morphology of the molten pool asymmetric. Figure 3(a) shows that the Ta...
hardly melted and the molten pool was mainly composed of Inconel 718 when the electron beam deviated to the Inconel 718 side by 0.5 mm, and the best joint formation was obtained under this condition. In contrast, as shown in figure 3(c), when the electron beam deviated by 0.5 mm to the Ta side, the melting amount of Ta increased significantly and the melting amount of Inconel 718 decreased. However, due to the high melting point of Ta, the base metal of Ta was not completely melted by the electron beam under the same parameters, forming a concave joint.

3.2. Microstructure

Figure 4 shows the most typical cross-section of the Ta and Inconel 718 alloy (SN2). Some coarsening columnar crystals and dendrites were observed on the Inconel 718 side; fine columnar crystals and dendrites were identified on the Ta side. In the middle of the weld zone, there were mainly dendrites. In addition, a few unmixed Ta ‘islands’ with the size of 10 μm were found in the weld zone. The grains had an obvious growth trend from both sides to the middle. During the solidification process, the variation in weld structure morphology was attributed to the relationship between the temperature gradient (G) and growth rate (R) [18]. Due to high cooling rate produced in the EBW process, the solidification time was very short. Near the fusion line, the temperature gradient G was larger than that at the centre of the weld. On the contrary, the growth rate R reached the maximum value at the centre of the weld. Therefore, from the fusion line to the centre of the weld, the G/R value decreased gradually. The solidification mode of the weld changed from cellular to columnar and then from columnar to dendritic. These dendrites grew to the centre of the weld along the opposite direction of the heat flow and were partially tilted under the influence of the welding movement speed, which might not be perpendicular to the fusion line. In addition, the thermal conductivity of the Ta base metal was higher than that of the Inconel 718 base metal, so the GR value of the Ta side was also greater than that of the Inconel 718 side. Therefore the microstructure on the Ta side was finer upon solidification.

XRD analysis results and EDS results are shown in figure 5 and table 5, respectively. The solidification of the molten pool started from the liquid → γ reaction. The first solidified dendrite core had a higher melting point and was rich in solute elements such as Ni, Fe and Cr, while the front of the dendrites had solute elements with low melting points such as Nb, Mo, Ti and Ta. Because the enrichment of low melting point alloy elements would form new constitutional supercooling, most of the elements such as Nb, Mo, Ti and Ta were enriched in
the interdendritic area. The $\gamma$ phase was dominated by Ni, Fe, Cr and Ta and became the main phase of the weld. As the temperature decreased, the liquid $\rightarrow \gamma$/MC eutectic reaction occurred between the interdendritic area, consuming a large amount of C. Finally, the liquid $\rightarrow \gamma$/Laves reaction occurred. The solidification sequence can be expressed as follows: $L \rightarrow L + \gamma \rightarrow L + \gamma + MC \rightarrow \gamma + MC + \text{Laves}$ [19]. Therefore, the dendrite core was mainly the $\gamma$ phase, and the interdendritic area was the Nb-rich $\gamma$/Laves + $\gamma$/MC phase. The harmful Laves phase was the inevitable final solidification phase. Due to the extremely high cooling rate in electron beam welding, the time required for solute redistribution was insufficient, and fewer Laves phases were obtained.

The cross-sectional microstructure of sample SN1 is shown in figure 6. Compared with sample SN2, the width of the weld increased, and the fusion line on the Ta side was almost straight, which was mainly related to the position of the electron beam. When the electron beam deviated by 0.5 mm to the Inconel 718 side, the melting amount of the Inconel 718 base metal that has a low melting point, increased. The microstructure of the weld zone of sample SN1 was similar to that of sample SN2, mainly composed of columnar crystals and dendrites. However, the grain size in the weld zone of sample SN1 was finer than that in sample SN2. In addition, a few equiaxed grains were found in the middle–upper part of the weld zone, as shown in figure 6(d). Unmixed Ta was not found in the weld zone. The width of the weld was positively correlated with the strength of the convection in the molten pool [20]. Therefore, from the macro morphology of the weld cross-section, it could be inferred that the convection strength of sample SN1 was greater than that of sample SN2, and the convection above the molten pool was stronger than that at the bottom. The convection effect of the molten pool on the upper part of the weld was very strong, and the dendrites in the weld were broken and brought into the centre of the molten pool. Most of them melted under the action of high temperature, and a small amount of residue became crystal nuclei, forming fine equiaxed grains and dendrites. In the upper part of the weld, the action time of the electron beam was longer, the $G/R$ value was smaller than the edge of the molten pool, and fine equiaxed grains more easily formed.

Figure 7 shows the microstructure of sample SN3. Compared with sample SN2, the degree of depression of the welded joint increased. Only the upper part of the Ta base metal melted, and the lower part hardly melted. The melting amount of Ta significantly increased under the condition of 0.5 mm beam offset to the Ta side. So the fusion line on the Ta side was peninsula-shaped and penetrated the weld zone. The composition of the molten pool was different from the base metal, and the liquidus temperature was also different. Ta would reduce the liquidus temperature of the molten pool [21]. So the liquidus temperature of the molten pool was lower than that of the Ta base metal. The front of liquid Ta solidified quickly when it was carried into the molten pool under the action of convection, forming a peninsula of base metal composition [22]. The weld zone of sample SN3 was

| Point               | Ta   | Nb  | Ni   | Cr   | Fe   | Mo  | Ti  |
|---------------------|------|-----|------|------|------|-----|-----|
| Dendrite core       | 3.25 | 3.22| 53.14| 18.76| 18.95| 2.05| 0.63|
| Interdendritic area | 8.05 | 6.55| 47.20| 17.80| 15.85| 3.50| 1.05|

Table 5. Chemical composition of dendrites (in wt%).
Figure 6. Microstructure of the dissimilar electron beam welding of sample SN1: (a) cross-section; (b) Inconel 718 side; (c) Ta side and (d) middle-upper part of the weld.

Figure 7. Microstructure of the dissimilar electron beam welding of sample SN3: (a) cross-section; (b) Inconel 718 side; (c) Ta side and (d) middle-upper part of the weld.
mainly composed of dendrites and columnar crystals, and no equiaxed grains were found. However, compared with sample SN2, the size of unmixed Ta 'islands' increased to 25 μm.

3.3. Weld defects

3.3.1. Reaction layer

Figure 8 shows the microstructure of the reaction layer of the welded joint. Clearly, there was a narrow light-coloured layer I and a mixed layer II between the Ta base metal and the weld zone, which was called the reaction layer.

![Figure 8. Microstructures of the reaction layer on the Ta side.](image)

![Figure 9. Element map scanning results of the reaction layer measured by EPMA.](image)

| Table 6. Chemical composition of points 1, 2 and 3 in figure 8 (in wt%). |
|-----------------|-----|-----|-----|-----|-----|-----|-----|
| Point | Ta  | Nb  | Ni  | Cr  | Fe  | Mo  | Ti  |
|-------|-----|-----|-----|-----|-----|-----|-----|
| 1     | 63.21 | 1.76 | 22.00 | 4.94 | 6.05 | 1.36 | 0.68 |
| 2     | 24.89 | 4.49 | 41.81 | 13.30 | 12.44 | 2.35 | 0.72 |
| 3     | 16.10 | 2.21 | 47.75 | 15.28 | 15.47 | 2.43 | 0.76 |

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layer. Among them, reaction layer II was formed by a mixture of dark columnar crystals and a net-shaped structure. The EPMA analysis of the reaction layer is shown in figure 9. Ta accumulated in reaction layer I and the net-shaped structure of reaction layer II. Nb, Mo and Ti elements were simultaneously enriched in the net-shape structure. The EDS results are shown in table 6. Compared with the composition in the weld zone, the content of Ta in the reaction layer increased significantly, and the content of Ta in reaction layer I was much greater than that in reaction layer II. While the content of Ni, Fe and Cr decreased significantly from the weld zone to the Ta base metal. The closer the distance was to the Ta base metal side, the higher content of Ta. The solubility of Ni, Fe, and Cr in Ta was extremely low, and the diffusion and mixing of these atoms would cause severe metallurgical reaction in the reaction layer, causing the formation of complex compounds. The XRD analysis results also verified the formation of intermetallic compounds. It can be inferred that the reaction layer I might be Ta-rich intermetallic compounds or secondary solid solution [23]. A liquid unmixed base metal layer could form due to the decrease in fluid flow near the boundary of the molten pool [22]. The composition of this layer was the same as the composition of the Ta base metal. Because Ta decreased the liquidus temperature of the molten pool, the solidification temperature of the molten pool was lower than that of the reaction layer. In the rapid solidification process, a high temperature diffusion zone with a wide temperature range formed before the solidification of the molten pool, so that Ni, Fe, Cr and other elements diffused into the layer, forming reaction layer I. Reaction layer I was the fragile zone of the joint, and it was the place where the fracture occurred. As the reaction progresses, dark columnar crystals nucleated on reaction layer I and grew towards the centre of the weld. Under the action of rapid cooling, the columnar crystals could not fully grow, forming a net-shaped structure, and the two structures mixed to form reaction layer II. The structure formed by the mixture of columnar crystals and net-shaped structure reduced the brittleness of the reaction layer II.

As shown in figure 10, the position of the electron beam significantly affected the microstructure of the reaction layer. The thickness of reaction layer II of sample SN1 was the largest, followed by sample SN2, and the thickness of sample SN3 was the thinnest, while the thickness of reaction layer I was just the opposite. Because reaction layer I formed by a variety of Ta-rich intermetallic compounds was the fragile zone of the joint, and all fractures of the tensile specimen occurred here. Reaction layer I of sample SN3 was the thickest, and had the highest brittleness, so sample SN3 had the worst tensile strength; notably, sample SN1 had the highest tensile strength.

Figure 11 shows the morphology of liquation cracks on the Inconel 718 side. The metallographic analysis of the joint showed that liquation cracks were found in the HAZ. Liquation cracks usually extended to the Inconel 718 base metal along the direction perpendicular to the fusion line in the HAZ. As shown in figure 11(b), the continuous distribution of fillers and intermittent precipitates were observed in the liquation cracks. Spheres with larger volumes were found in the partially melted zone (PMZ) at the bottom of the cracks. The EDS results are shown in table 7, and compared with the composition of the weld zone (table 5). The content of Ni, Fe, and Cr in the crack were much lower than those in the weld zone, while the content of Ti and Nb were higher. Additionally, a small amount of Ta was found. Therefore, it can be inferred that the filler in the liquation cracks that extended into the HAZ was a mechanical mixture composed of liquid metal in the molten pool and interdendritic low melting phase (MC/Laves phase). Under the action of the thermal cycle, the low melting point MC/Laves phase distributed in the PMZ first melted, and the grain boundaries cracked due to tensile stress during cooling. The cracks formed on the liquated sphere in the interdendritic area of the PMZ and then propagated into the HAZ [24]. The liquid in the molten pool permeated and expanded to the intergranular area, until the grain boundary was filled, thus forming liquation cracks.
Voids and cracks were found in the reaction layer on the Ta side, as shown in figure 12. With the increase in the thickness of the reaction layer, the size of the voids increased, and some adjacent voids were connected by cracks. This phenomenon indicated the brittleness of the reaction layer. The appearance of the voids might be related to the Kirkendall effect [25, 26]. The reaction layer was consisted of different compounds which were produced by the reaction of Ni, Fe, and Cr with Ta. Vacancies generated in the reaction layer due to the unbalanced effect of atomic diffusion. When the concentration of vacancies reached a certain level, they would be transformed into Kirkendall voids. With the growth of the reaction layer, the voids coalesced into larger cavities. Stress along the interface was generated, due to the significant difference in the thermal expansion coefficients between the reaction layer and Ta, and cracks started to initiate and expand from the voids. Voids and cracks on the reaction layer containing a large amount of brittle intermetallic compounds further reduced the mechanical properties of the joint. The greater the thickness of the reaction layer, the greater the size of the voids and cracks, and the worse the mechanical properties of the joint.

3.3.3. Macrosegregation
As shown in figure 13, obvious macrosegregation characteristics such as ‘island’ and ‘peninsula’ were found in the joint [22, 27, 28]. Table 8 shows the chemical composition of the points in figure 13. The white island whose main component was Ta had a regular surface, which could be regarded as a Ta-base solid solution. The peninsular structure was divided into two parts: the composition of the outer dark part was similar to that of the reaction layer I, which could be considered a part of reaction layer I; the internal composition was similar to that of the island, which was a Ta-base solid solution. As shown in figure 13(c), the cracks penetrated into the base metal from the Ta side fusion line. Point 2 in table 8 was the EDS analysis of the cracks. Compared with the composition of the reaction layer (point 1 in table 6), the penetrating cracks were also contained a large amount of brittle intermetallic compounds. Because the cracks penetrated more into the inner part of the Ta base metal, the content of Ni, Fe, Cr and other elements were lower than those in the reaction layer. Under the influence of thermal stress, cracks easily initiated at the defects (such as pores, impurities, etc) of the Ta base metal. So the liquid metal in the molten pool entered into the cracks, and further penetrated into the Ta base metal under the action of capillary. The Ta base metal near the crack was heated and melted, which not only enlarged the crack size, but also increased the Ta content. More Ta-rich phases were produced in the cracks, which affected the mechanical properties of the joint. The Ta was carried into the molten pool due to some cracks contacted each other during penetration. If Ta was brought into the cooler molten pool near solidification front, that is, the liquidus temperature of the molten pool was lower than that of Ta base metal, and it would rapidly solidify into a peninsula or island before it had time to mix with the surrounding molten pool.

Table 7. Chemical composition of the liquation crack in figure 11 (in wt%).

| Point | Ta  | Nb  | Ni  | Cr  | Fe  | Mo  | Ti  |
|-------|-----|-----|-----|-----|-----|-----|-----|
| 1     | 1.47| 64.61| 11.12| 4.36| 4.33| 6.95| 7.16|
| 2     | 4.81| 44.32| 23.19| 8.77| 7.96| 7.41| 3.54|

Figure 11. Liquation cracks on the Inconel 718 side: (a) OM image and (b) high magnification SEM image of the crack.
Figure 12. Voids of the reaction layer on the Ta side: (a) high magnification SEM image of the voids and (b) OM image of the cavities.

Figure 13. Macrosegregation of the joint: (a) unmixed Ta island; (b) high magnification SEM image of an unmixed Ta island; (c) peninsula and cracks on the Ta side; and (d) high magnification SEM image of the peninsula.

| Point | Ta   | Nb  | Ni  | Cr  | Fe  | Mo  | Ti  |
|-------|------|-----|-----|-----|-----|-----|-----|
| 1     | 96.46| 0   | 2.04| 0.34| 0.96| 0   | 0.21|
| 2     | 75.27| 0.55| 14.56| 4.33| 5.02| 0   | 0.28|
| 3     | 63.25| 0.59| 22.72| 5.34| 6.95| 0.63| 0.51|
| 4     | 97.82| 0   | 0.71| 0.43| 0.65| 0   | 0.39|
3.4. Mechanical properties

The distribution of microhardness along the cross-section of the welded joint in samples SN1-SN3 is shown in figure 14. The electron beam position had little effect on the microhardness of the weld zone in samples SN1-SN3. The microhardness of the weld zone ($\approx 328$ HV) was significantly higher than that of the Ta ($\approx 251$ HV) and Inconel 718 ($\approx 223$ HV). The highest value of microhardness was obtained on the reaction layer, especially the value of microhardness of sample SN2 reached 891 HV. The closer the distance was to the Inconel 718 side, the lower the microhardness of the weld zone. The increase in microhardness of the weld zone was mainly related to the hardening effect of Ta: (1) Ta entered the $\gamma$ matrix, which increased the solubility and improved the solid solution strengthening effect. At the same time, Ta was also a solid solution strengthening element, and the effect of strengthening the $\gamma$ matrix was stronger than that of Ni, Fe, and Cr [1]. (2) Ta entered carbides and formed Ta-rich MC-type carbides, which replaced Ti in MC and improved the stability of MC-type carbides. Moreover, Ta-rich MC-type carbides had greater microhardness, which significantly improved the microhardness of the weld zone [29]. Generally, the hardness of the material was defined as its resistance to plastic deformation, so the reaction layer containing a large amount of brittle intermetallic compounds had the highest value of microhardness compared with the weld zone. Regardless of whether the electron beam deviated, the content of Ta in the weld zone was not much different, so there was no significant difference in the microhardness of the

![Figure 14. Microhardness distribution profiles of the joint with different parameters.](image1)

![Figure 15. Tensile test of electron beam welded Ta and Inconel 718 with different parameters.](image2)
weld zone of the three samples. Only the Ta content on the Inconel 718 side far from the Ta base metal was significantly reduced and the microhardness also showed a decreasing trend.

Figure 15 shows the stress-strain curves of the joints of the Ta and Inconel 718 alloys. The tensile strength and elongation of the welded joint of sample SN1 reached the maximum value of 480 MPa and 2.8%, respectively; the lowest tensile strength of sample SN3 was 138 MPa; and that of sample SN2 was 313 MPa. The fracture morphology of the three samples is shown in figure 16. A large number of cleavage steps and river patterns were observed in the fracture surface, which was characteristic of a typical brittle fracture. Regardless of whether the electron beam deviated, fracture occurred in reaction layer I, as shown in table 9. The fracture of

Table 9. Chemical composition of points 1, 2 and 3 in figure 16 (in wt%).

| Point | Ta   | Nb  | Ni   | Cr  | Fe  | Mo  | Ti  |
|-------|------|-----|------|-----|-----|-----|-----|
| 1     | 69.37| 0.42| 19.36| 4.09| 5.76| 0.68| 0.32|
| 2     | 84.68| 0   | 10.03| 1.80| 2.72| 0.56| 0.21|
| 3     | 59.26| 1.31| 23.10| 6.87| 7.64| 1.59| 0.23|

Figure 16. Fracture morphology of the joint with different parameters: (a) and (b) sample SN1; (c) and (d) sample SN2; (e) and (f) sample SN3.
sample SN1 was relatively smooth and Ta islands were also found on the fracture. Compared with sample SN1, cracks and voids were found on the fracture of sample SN2 and sample SN3, which not only reduced the effective bearing area of the joint, but also became the starting point of fracture. In addition, reaction layer I of sample SN2 and sample SN3 was thicker than that of sample SN1, so the fragile zone was larger, which seriously reduced the tensile strength of the joint. And the connection of the sample SN3 joint was incomplete and the strength was the lowest. The reasons for the highest tensile strength of sample SN1 were as follows: (1) a small amount of Ta was melted, and the thickness of reaction layer I was the thinnest, so the brittleness of the reaction layer reduced. The size and number of voids decreased, which further reduced the occurrence of cracks. (2) The melting amount of the Inconel 718 alloy clearly increased due to the influence of electron beam offset, forming a structure similar to welding-brazing as shown in figure 3(a), in which Ta was wrapped.

4. Conclusions

The effect of electron beam offset on the dissimilar metal butt welding of Ta and Inconel 718 superalloy was studied by EBW. The main conclusions were as follows:

(1) The microstructure of the weld zone was mainly composed of columnar crystals and dendrites. When the electron beam deviated by 0.5 mm to the Inconel 718 alloy side, a small number of equiaxed crystals were found in the weld zone.

(2) The microhardness of the weld zone was greater than that of the two base metals. The thickness and morphology of the reaction layer changed with the position of the electron beam, and the maximum microhardness appeared on the reaction layer.

(3) Tensile fracturing occurred in reaction layer I, which was a typical brittle fracture. The voids and cracks on the reaction layer would seriously reduce the strength of the joint. When the electron beam deviated to the Inconel 718 alloy side, the tensile strength of the joint was 480 MPa, which was higher than the tensile strength of the beam offset to the Ta side and non-beam offset.

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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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