Effect of interlayer material on microstructure and mechanical properties of diffusion brazed IC10 superalloy

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Abstract. The diffusion brazed test of IC10 superalloy was carried out by using Russia solder M3 interlayer powder. The wetting quality of the high-temperature alloy was evaluated by the contact angle of wetting at various temperatures. The mechanical properties of the joints were determined at operating temperatures. The result shows that at the temperature of 1250°C, the contact angles of interlayer powder M3 alloy do not exceed 4-5°, it has high wettability and spread ability over the surface of the IC10 alloy. The average test time before failure at the temperature of 900°C and the stress of 320 MPa was 38 hours, and at the stress of 350 MPa was 9.6 hours, all both higher than 80% of the base material under the same experimental conditions. The structure of the joints has a smooth transition from the base metal to the soldered joint. In the structure of the joint, the γ'-phase with a large volume concentration in the base metal. IC10 superalloy has excellent high temperature properties when the M3 interlayer was used.

1. Introduction

Composite and intermetallic materials, including those based on Ni3Al, are increasingly used in modern gas turbine construction[1,2]. In the Ni-Al system, there are phases of the γ, γ' and β-phases[2]. Phase γ' is an ordered intermetallide with a face-centered lattice, in which Cr, Mo, W are bound to dissolve with decreasing solubility in this row. Metals such as Ti, Ta and Nb dissolve in the γ'-phase and strengthen it[3]. Hardening occurs due to the formation of a hardened structure during crystallization, which is stable up to the temperature of 1250°C. Due to alloying, the intermetallide Ni3Al acquires high mechanical properties and stability in oxidizing environment and combustion products of fuel[4]. The most universal and promising way to bond the IC10 alloy is welding, but the welding problem appears when the intermediate solder with a lower melting point than the base metal but was used, it is a weak link in the joint[5].

Diffusion brazing, also known as transient liquid phase (TLP) bonding, was initially developed by paulonis et al[6], is currently an optimized technology utilized for jointing Ni-based superalloys. Studies have shown that an effective way to increase the strength of the joint is to provide the closest properties of the interlayer and the base metal, for which it is necessary to have a close chemical composition, which can ensured both during TLP-diffusion welding and subsequent heat treatment, also with the
composition of the solder\cite{5,7}. It is this principle adopted in the development of solder SBM-3 (M3+1.2% B). The composition of the solder is selected taking into account the solubility of the elements strengthening the solution and the efficiency of the reinforcing elements in the γ'-phase. Therefore, tungsten and molybdenum were partially replaced by rhenium, and tantalum was also introduced into the solder.

In this paper, we describe the research methodology, the materials and solders, the interaction of solders with the base metal, the brazing alloys selected for the alloy IC10. The results of the wetting and flowing of solders, the filling of gaps, the effect of the chemical composition of solders on the temperature range of fusion of solders and the soldering temperature, the formation of the structure, the chemical composition of the solder weld metal, the diffusion zone, and the properties of the joints are presented. An important factor affecting the structure and mechanical properties of the joints, as well as determining the technology of joining the IC10 alloy, is thermal treatment, the development of which is the subject of further research, which may introduce some adjustments to the technology.

2. Experimental

2.1. Experimental materials

The IC10 alloy used in the test material is a directionally solidified columnar superalloy based on the intermetallic compound Ni3Al. It has good oxidation resistance and corrosion resistance, can withstand the service temperature of 1100°C, and the microstructure is stable at high temperature\cite{8}. As its high temperature durable performance and good resistance to cold and thermal fatigue, it has been used to manufacture gas turbine guide vane of aero-engine. The high-temperature alloy IC10 investigated in the work has the following chemical composition (% mass)\cite{5,9}:

| Alloys | C    | Cr   | Co   | Mo   | W    | Al   | Ta   | Hf   | B    | Ni   |
|--------|------|------|------|------|------|------|------|------|------|------|
| IC10   | ≦0.01| 6.5~7.5| 11.5~12.5| 1.0~2.0| 4.8~5.2| 5.6~6.2| 6.5~7.5| 1.3~1.7| ≦0.02| base |

For the development of solders (interlayers), CrNi61CoWTiAlMoTaRe alloys, nominally designated M3, containing tantalum and rhenium, were fused, in contrast to the most widely sold in Russia solder VPr36\cite{10}. It is known that the alloying of heat-resisting nickel alloys with rhenium, which dissolves in the alloy matrix and reduces the rate of diffusion processes, contributes to the increase in the long-term strength of the metal at high temperatures. As the concentration of Re increases from 1.0 to 4.0% by weight, the diffusion coefficient of Ni atoms decreases from $5 \times 10^{-14}$ to $2 \times 10^{-14}$ cm$^2$/s. The two main systems (M3 and VPr 36) has the following chemical composition (% mass):

| Alloys | C    | Cr   | Co   | Mo   | W    | Al   | Ti   | Nb   | Ta   | Re   | B    | Ni   | TL-°C |
|--------|------|------|------|------|------|------|------|------|------|------|------|------|-------|
| M3     | 0.1  | 12.5 | 7.0  | 1.0  | 4.5  | 3.0  | 4.8  | 0.2  | 3.5  | 2.6  | –    | –    | 1297  |
| VPr 36 | –    | 9.0  | 9.0  | 1.7  | 4.0  | 4.3  | –    | 4.0  | –    | –    | 1.1  | base | 1098  |

2.2. Experimental methods

At the first stage of the research, an analog of the IC10 alloy was used in the form of square plates 25×25 mm with a thickness of 3~5 mm, on the surface of which 100 mg of solder preform was applied. After heating in a vacuum furnace to a certain temperature and cooling. The flowing of the solder was determined from the specific flowing area. After that, the samples were cut along the center of solder drop and the contact angle was determined from the photo of the macrosection. At the final stage of the research, cylindrical washers with a diameter of 16 mm from the IC10 alloy, obtained from cylindrical finger specimens of the same diameter, were used. To study the flowing of the analogue solder IC10
into the gap, wedge-shaped samples measuring 20×12×3 mm in the bottom and 20×6×3 mm in the upper were used. The solder was installed on the lower plate on one side of the gap. Similar samples were used to study the formation of a soldered joint with a constant gap of 0.08 mm along the length. The gap was fixed with a tungsten filament of the same diameter. To study the filling of the gap, wedges were used, in which the gap varied from zero to 0.6 mm or from zero to 0.3 mm by installing a wire of tungsten with a diameter of 0.6 and 0.3 mm, respectively. The corresponding samples are shown in Figure 1.

![Figure 1. View of the surface of the alloy IC10 flowing into the wedge gap (a) of the experimental solder.](image)

Before TLP-diffusion welding, the joined surfaces were grinded in the direction of the solder wicking. Polishing surfaces is excluded. Before assembly, the parts of the surface to be soldered are rubbed with alcohol. To form the clearance value under TLP-diffusion welding, a wire of tungsten of a certain diameter is used, which is fixed by means of capacitor welding. The formation of joints of butt welds was studied on cylindrical samples with a diameter of 15 mm and a length of 35 mm. When the joint was made in the vertical position of the samples, the upper sample was made with an oblique cut at the joint to place the solder on the bottom. To fix the gaps, a tungsten filament of a certain diameter was used, and the butacrylic resin AST-T was used to retain the solder.

Differential thermal analysis was carried out on a thermal analyzer VDTA-8M with simultaneous measurement of the temperature of the sample and the standard under heating and cooling in an electric resistance furnace. Structural studies were carried out using optical metallography, translucent and scanning electron microscopy, as well as X-ray structural analysis. The determination of mechanical characteristics was carried out with long-term tensile tests of cylindrical samples.

Since the wetting and spreading of the solder depends not only on the chemical composition of the solder and the base metal, but also on the external conditions, the processes were studied when soldering in a vacuum of 3~5×10^{-3} Pa, which guarantees activation of the surface of the high-temperature alloy at a temperature above 1150°C.

3. Results and discussion

3.1. The thermogram of Intermediate layer

It is known that in the case of high-temperature nickel alloys with multicomponent alloying, the presence of many phases of which the number increases substantially with the introduction of boron is characteristic. First of all, we note the strengthening γ’-phase of the Ni3Al, Ni3 (Al,Ti) and other intermetallide η-phase Ni3Ti phases, the γ”-phase Ni3Nb, the MC carbons, M23C6, M6C, M7C3(NbC, HfC, Cr23C6(Cr,W,Mo)23C6, Nb3Co3C, Ta3Co3C, hexagonal carbide Cr7C3(Cr7C3)), tetragonal borides M3B2(Ta3B2, Nb3B2, (Mo,Ti,Cr,Ni)3B2), topologically close-packed (TCP) Laves phases Co2Ta, Co2Ti, and in the presence of iron Fe2Nb, Fe2Ti, Fe2Mo, μ- phases of Co2W6, Co7(Mo,W)6, σ-phases of CrCo, CrNiMo, and in the presence of iron FeCr, FeCrMo, CrFeMoNi[5,10,11]. The isolation of these phases occurs in different temperature ranges and conditions and is accompanied by many thermal effects both during heating and cooling. The thermogram of the solder VPr36 is shown in Figure 2.
In the analogue of the alloy IC10, the main alloying elements, hardening the $\gamma$-solid solution include Co, Cr, W, Mo, and the dispersion hardening is provided by Al, Ti, Ta, Hf. The carbide forming elements include Ti, Ta, Hf, W, Mo, Cr. The total content of alloying elements reaches more than 40% by weight. From the known triple and quadruple phase diagrams, the phase $\gamma'$ borders on one of the following two-phase regions: $\gamma'+\sigma$, $\gamma'+\mu$, $\gamma'+R$, $\gamma'+P$, where $\sigma$, $\mu$, $R$ and $P$ are TCP phases[5,9]. These phases bind a significant number of basic alloying refractory elements and impoverish them with the $\gamma$ phase. The region of precipitation of TCP phases is in the temperature range 1000~1150℃. Such phases can also be in the base metal. The thermogram of the IC10 alloy analog is shown in Figure 3.

Different elements of the alloying complex have an ambiguous effect on the temperatures of phase transformations in complex-alloyed high-temperature nickel alloys. Thus, it has been experimentally established that Ti, Al, Ta, Hf, Zr, W, Mo, have a positive effect on $T_{\gamma'}$, i.e. increase the thermal stability of the $\gamma'$-phase and the alloy as a whole, and Co, Cr, V and C reduce $T_{\gamma'}$. All alloying elements, except Re, W and Co, reduce $T_s$, especially significantly - carbon and boron[12]. Therefore, the difference in the values of the temperatures of the phase transformations can be explained by the non-coincidence of the ratio of the content of the elements of the alloying complex, however, the order of melting and crystallization of the phases is conserved.

The thermograms of M3 alloys without boron and with boron shows the same regularities in the influence of its concentration on the number of phases and the melting and crystallization temperature. The results obtained make it possible to choose a solder based on the M3 alloy. The thermograms of M3 alloys without boron and with boron (1.0% ~1.2% B) alloys is shown in Figure 4.
With the increase of the content of B, the melting temperature of the interlayer material decreased obviously. At the same time, it can be found that the melting process of the interlayer material fluctuates to a certain extent, which is due to the segregation of dendrite during the solidification and crystallization of the alloy, so that the γ' phase first precipitates from the interdendrite region and then precipitates in the dry dendrite region. The reaction heat effect can not be concentrated due to the long precipitation process, so the solidification range of the alloy increases. It can be seen that as the main precipitation strengthening phase of Ni3Al-based superalloy, the precipitation of γ’ phase will not only improve the mechanical properties of the alloy, but also affect its melting characteristics[13].

3.2. The spreading performance of Intermediate layer

Spreading of solders for the IC10 alloy and its analogue was studied in the temperature range from 1235℃ to 1265℃. The residence time at the spreading temperature both on the surface of the planar samples and on the wedge was 20 min in a vacuum of 3×10⁻³ Pa. Appearance of the samples after the spreading of the solder is shown in Figure 5, it shows the spreading of the solder VPr36.

Solder VPr36 is characterized by spreading with a large aureole outside the bulk of the drop. This is a very thin layer of solder penetrating into very narrow gaps, close to zero. At a lower temperature (1235℃), the spreading of the solder based on the solder VPr36 (Figure 5(a)). At a temperature of 1265 ℃ the area of spreading of solder VPr36 somewhat decreases (Figure 5(c)) due to more intensive dissolution of the base metal. According to the spreading characteristic for solder VPr36, the optimum temperature is 1250℃.

Studies of wetting and spreading of solder based on the M3+1.2% B alloy showed that at the temperature of 1235℃, its specific spreading area is 1.1~1.25 mm²/mg. It increases with increasing temperature. At the temperature of 1250℃, it is about 1.5 mm²/mg, at 1265℃ it is 1.7~1.9 mm²/mg. The regularities of the effect of the surface treatment method, established for solders with rhenium, were confirmed using a solder based on the M3+1.2% B alloy, in which the same spreading areas were obtained with increasing temperature. As in the case of solders VPr36 at temperatures of 1250 and 1260℃ on the surface of the alloy IC10, when spreading the solder on the basis of alloy M3+ 1.2% B, a halo is formed over a large area behind a distinct drop contour. The general view of the samples after the spreading of the solder is shown in Figure 6.

Figure 5. Spreading of the VPr36 solder 100 mg for the alloy IC10 at 1235℃(a);1250℃(b);1265℃(c).

Figure 6. Spreading of alloy M3+1.2% B on the alloy IC10 at a temperature of 1235℃(a) ;1250℃(b) and 1265℃(c).
The contact angle of wetting were determined from the macrosections obtained after cutting the plates along the diameter of the spreading drop. The marginal angles of wetting of the IC10 alloy with VPr36 solder are shown in Figure 7, and on the basis of the M3+1.2% B alloy in Figure 8. The angles of wetting with VPr36 solder are reached at a temperature of 1235℃-10°, at 1250℃ -7°, 1265℃-3°.

Figure 7. The Contact wetting angles of the IC 10 alloy with an alloy based on VPr36 solder at temperatures 1235℃ (a), 1250℃ (b) and 1265℃ (c).

The alloy based on the M3 + 1.2% B alloy also has high wettability and spreadability over the surface of the alloy IC10 (see Figure 8). At the temperature of 1235℃and 1265℃,the contact angles do not exceed 8°. At the temperature of 1250℃,the contact angles do not exceed 4~5°.

Figure 8. The Contact wetting angles of the IC 10 alloy with an alloy based on an M3+1.2% B alloy at temperatures 1235℃ (a), 1250℃ (b) and 1265℃ (c).

The characterization of wettability of materials still follows the solid-liquid-gas three-phase equilibrium equation proposed by ThomaS Young in 1804[14]:

$$\cos \theta = \frac{\sigma_{GS} - \sigma_{LS}}{\sigma_{GL}}$$

(1)

The surface tension of solid / gas, the surface tension of solid / liquid and the surface tension of liquid / vapor are represented by $\sigma_{GS}$, $\sigma_{LS}$, $\sigma_{GL}$ respectively, and $\theta$ denotes the contact angle of solid / liquid interface in equilibrium state, that is, wetting angle. The wetting angle is used as a marker to measure the wetting degree of liquid to solid interface. When the wetting angle is less than 90°, and the non-wetting angle is greater than 90°. According to formula (1), the lower the surface tension between solid and liquid, the better wettability between solid and liquid. According to the above results, both VPr36 solder and M3+1.2% B alloy has good wettability on the IC10 base metal.

3.3. Experimental methods

Similar alloys with boron were smelted on the basis of the heat resistance nickel alloys M3. The concentration of boron varied within 1.0~1.2% by weight. Studies have shown that the metal structure based on IC10 and M3 at the same boron concentration is identical. Figure 9(a) shows the structure of the M3+1.2% B alloy, which is similar to the structure for an analog of the IC10 alloy and the same boron concentration. Concentrations of the alloying elements in the individual phases are significantly different, which is clearly seen from Figure 9(b) at high magnification. In the dark phase, the concentrations of the elements are close to the IC10 alloy.
In the light phase, the main carbide forming elements are Re (25.35% by weight), Cr (16.85), W (7.23), Mo (5.67). The uneven distribution of Re leads to the fact that in the main matrix of the alloy it can be absent as shown in Figure 10. In this case, the white phase in Figure 10(c) has a skeletal shape, this reduces the concentration of carbide-forming elements in it, in particular, the Re concentration up to 14.89%.

In the structure of the M3 alloy containing 1.2% B, there are inclusions of a square shape with elongated diagonals containing (% by weight): 46.16 Re; 15.16 Cr; 24.31 Ni; 3.22 Co; 3.19 W; 2.27 Mo; 1.68 Al; 1.33 Ti; 1.73 Zr; 0.83 Hf, an area of 100–160 μm². The microstructure of the M3 alloy containing 1.2% B and the chemical composition by area and at individual points are shown in Figure 9. The distribution of alloying elements over the area is close to the distribution in the base metal, but there are white inclusions with a high content of Re, Cr, W and Mo.

![Figure 9. Microstructure (a, b, c) (alloy M3+ 1.2% B) and the chemical composition (% wt) of the alloy along the horizontal and at points (1,2,3,4,5,6).](image-url)
3.4. Microstructure and chemical composition of IC10 Joints
At the temperature of 1250 ℃ and holding time for 30min, the microstructure and chemical composition of IC10 Joints was studied. In tests of some samples, regardless of the main metal, for short-term and long-term tests, the fracture was over the base metal, which allowed us to investigate TLP-diffusion welded joints after heat treatment, which was carried out in vacuum with a reduction after soldering the temperature to 1170 ℃ with a holding time of 2.5 hour, then the temperature was reduced to 1050 ℃ followed by a 2.5 hour exposure. Figure 10(a) and 10(b) shows the structure of the joint and the base metal, respectively. The width of the seam is approximately 0.08 mm. Dissolution of solder grain boundaries is practically absent. The structure of the joints has a smooth transition from the base metal to the soldered joint. In the structure of the bonding zone, the γ'-phase with a large volume concentration in the base metal is clearly discernible (Figure 10(a)), eutectic free bonding zone indicates the complete isothermal solidification of the joint. The microstructure of the base metal (Figure 10(b)) in some areas has a distinction of phases of various shapes, including lamellar. These phases are shown by arrows. It is possible to adjust the heat treatment modes to dissolve them.

Figure 10. General view of the sample (a) after the tests, microstructure of the joint (b), soldering of the IC10 alloy with SBM-3 solder and base metal (c).

The change in the width of the interlayer and the point of determination of the chemical composition of the metal in the seam are shown in Figure 11, and the spectra at different points are shown in Figure 12, the figure only shows the most characteristic spectra. Figure 11 shows that the structure of the TLP-diffusion-welded joint is fairly homogeneous. The spectra at points 10 (Figure 12(a)) (base metal) and along the horizontal (Figure 12(b)) are practically identical and the chemical composition of the metal varies little within the solubility of the elements. At the separation point of carbides, borides, carboriborides, the spectra differ sharply because of the high concentration of certain elements in carbides or other phases, for example, at points 3 (Figure 12(c)). Except for a small amount of granular white compounds precipitated in the isothermal solidified zone of the joint, it is basically composed of γ' double-phase reticulate structure similar to that of the base metal IC10 alloy matrix, in which γ phase is matrix phase, and the γ' phase is dendriform around the γ' phase. It can be seen from the analysis of white compounds, point 3, there are main elements of Ta,Ti,W,Nb,Hf which should be carbides rich in Ta and Ti. It is known that in superalloys with Zr and Hf contents exceeding the trace addition range, they are known to be carbides rich in Zr and Hf. Carbides are divided into two: metal elements with Ti,V,Nb and MC carbides dominated by Ta are called MC\(_{\text{(1)}}\) carbides, MC carbides with metal elements mainly Hf or Zr are usually called MC\(_{\text{(2)}}\) carbides\[15\].
| No   | Cr  | Co  | Mo  | W   | Al  | Ti  | Nb  | Re  | Zr  | Ta  | Hf  | Fe  | Ni  |
|------|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|-----|
| Point1 | 7.71 | 9.27 | 0.4 | 1.28 | 10.8 | 2.29 | 0.74 | -   | 0.01 | 1.46 | -   | 0.83 | base |
| Point2 | 6.66 | 8.8  | 0.51 | 1.13 | 12.8 | 1.88 | 0.28 | -   | 0.03 | 0.5  | -   | 0.76 | base |
| Point3 | 3.59 | 4.55 | 0.36 | 5.4  | 2.76 | 13.54 | 1.75 | -   | 0.04 | 28.3 | 0.9  | 0.79 | base |
| Point4 | 6.4  | 8.83 | 0.56 | 0.9  | 9.4  | 3.17 | 1.03 | 0.08 | -   | 2.46 | 0.53 | 1.25 | base |
| Point5 | 9.96 | 9.39 | 0.61 | 0.48 | 7.92 | 2.6  | -   | -   | 0.06 | 1.35 | 0.82 | 0.51 | base |
| Point6 | 6.91 | 8.58 | 0.75 | 0.93 | 11.5 | 2.42 | 0.01 | -   | 0.01 | 1.33 | 0.32 | 0.56 | base |
| Point7 | 4.45 | 4.49 | 0.78 | 5.14 | 2.32 | 13.2 | 1.16 | -   | -   | 32.6 | 0.88 | 0.87 | 34.2 |
| Point8 | 5.79 | 9.3  | 0.73 | 1.06 | 12.3 | 0.71 | -   | -   | 0.03 | 2.68 | 0.44 | 0.9  | base |
| Point9 | 26.5 | 8.08 | 6.78 | 4.92 | 4.29 | 3.12 | -   | 1.58 | 0.21 | 2.56 | 0.78 | 0.52 | base |
| Point10 | 7.71 | 12.9 | 1.1  | 1.2  | 11.2 | 0.3  | -   | -   | 0.05 | 0.99 | 0.18 | 0.7  | base |

**Figure 11.** Microstructure of the joint in the TLP diffusion welding of the IC10 alloy with SBM-3 solder and the chemical composition of different points.

**Figure 12.** Spectra at point10(a), along the horizontal(b), point3(c) for the TLP-bonding of the IC10 alloy with SBM-3 solder.
Local X-ray spectral analysis by points and along the horizontal (by area) showed that the chemical composition of the weld metal and the base metal after the thermal treatment is fairly close, not only with a 0.07~0.08 mm thick but also larger (see Figure 11(a)). The use of SBM-3 solder makes it possible to obtain strong joints with gaps greater than 0.1 mm. In the soldered seam there are no continuous layers of eutectic or brittle phases and an equiaxial structure is formed.

3.5. Strength Properties and fracture morphology of IC10 Joints

The IC10 alloy is designed for operation at temperatures up to 1100 °C. Its important characteristic is long tensile strength. The results of tests for the long-term strength of the IC10 alloy after standard heat treatment (average values) in the longitudinal direction at temperatures of 900, 1000 and 1100 °C on a time base of 50, 100, 200, 300, 500 and 1000 hours are shown in Figure 13.

The average test time before failure at a temperature of 900°C and a stress of 320 MPa was 38 hours, and at a voltage of 352 MPa, 9.6 hours. The test base of 35 hours corresponds to the strength of the alloy IC10 with directional crystallization equal to 390 MPa and then the strength of the joint is 320/390=0.82 of the strength of the alloy with directional crystallization. The test base of 9.0 hours corresponds to the long-term strength of IC10 alloy with directional crystallization of 435 MPa, and then the bond strength is 352/435=0.81 of the strength of IC10 alloy with directional crystallization.

![Figure 13. Long tensile strength of alloy IC10 (average values) in the longitudinal direction of the crystal at temperatures of 900, 1000 and 1100 °C.](image)

Mottura et al detect the existence of Re cluster by means of three-dimensional atomic probe[16], extended X-ray fine absorption structure and first-principle calculation. The first-principle calculation shows that the diffusion activation energy of Re in Ni is the largest of all elements. It is considered that the essence of the "Re effect" is that the very low diffusion coefficient of Re inhibits the diffusion process in the alloy and thus improves the high temperature strength of the alloy. Re has many beneficial effects in Ni-based superalloys. For example, the growth rate of γ'-enhanced phase decreased significantly; Segregate on the γ matrix so that the γ/γ' mismatch becomes more negative, which is conducive to the formation of high-density dislocation networks, and the formation of short-range ordered atomic clusters in the matrix hinders the dislocation movement, resulting in a more obvious strengthening effect than the traditional solid solution strengthening[17].

Investigations of the fracture surface of long-term strength samples showed that the destruction occurs along the relief grain boundary (Figure 14(a)). The result shows that the grain boundaries participate in deformation at high temperature and contribute to the total deformation. The mechanism of grain boundary deformation is through grain boundary slip and migration. It is easy to form cracks at the grain boundary due to the stress concentration caused by the incoordination of grain boundary slip and intra-grain deformation. According to the analysis of the forming process of the TLP joint, the microstructure of the weld is formed by epitaxial growth from both sides of the interface to the center of the weld at the joining temperature, and the atom arrangement is disordered at the intersection. There is inevitably a transverse interface perpendicular to the direction of the applied stress and is difficult to be eliminated by diffusion. If the diffusion time is longer, the continuous transverse interface is formed in the center of the weld and several continuous transverse interfaces are formed in the weld when the...
The diffusion time is short. In addition, due to the effect of surface tension, the grooves between the cells grown by epitaxial growth are increased, and the longitudinal interface parallel to the direction of stress becomes the easy part of crack initiation under high temperature and long-lasting conditions[8].

Figure 14. The microstructure of the metal in the fracture zone after testing the joint of IC10 alloy

Under the action of persistent stress, the boride phase, as the obstacle of dislocation movement, causes dislocation plug product. As a result of dislocation plug product, tensile stress occurs at the interface between boride and base metal, which makes the crack initiation and propagation (Figure 14(b)). Under the condition of lower stress at high temperature, the ability of atomic activity increases, and the long-term action provides the possibility for the diffusion process of atom movement. At the same time, as the lower stress level, the envelope around the boride can hinder the propagation of the crack. Therefore, the dislocation of the plug in the borides can contribute to the deformation through the atom diffusion climbing to another slip surface to make the dislocation plug relax and make the interface in the weld become a weaker link[19]. The fracture surface is intergranular fracture at 900 ℃.

4. Conclusion

The correlation between microstructure and mechanical properties of TLP bonded IC10 is investigated. The following conclusions can be drawn from this study:

(1) Studies of capillary properties have shown that all two systems have good wetting of the intermetallic alloy and spreading over it. At a soldering temperature of 1250 ℃, the contact angles of the heat-resistant alloy IC10 with SBM-3 solder do not exceed 7°, the structure and chemical composition of the weld metal are close to the base metal. Short-term and long-term strength of the joint is not less than 0.8 from the strength of the base metal at a gap size of 0.08 mm and slightly higher at smaller gaps.

(2) The average test time before failure at a temperature of 900 ℃ and a stress of 320 MPa was 38 hours, and at a voltage of 352 MPa, 9.6 hours. The strength of the joint is 320/390 = 0.82 of the strength of the alloy with directional crystallization. The test base of 9.0 hours corresponds to the long-term strength of IC10 alloy with directional crystallization of 435 MPa, and then the bond strength is 352/435 = 0.81 of the strength of IC10 alloy.

(3) Under the action of high temperature and high stress, boride, as the phase that hinders the movement of dislocation, leads to dislocation accumulation, and finally leads to the initiation and propagation of cracks, which leads to the final failure of weld.

Acknowledgments

The authors gratefully acknowledge the financial support by the GDAS' Project of Science and Technology Development(No. 2018GDASCX-0113), Key Program for International Cooperation of Science and Technology(No. 2015DFR50310), GDAS' Project of Science and Technology Development (No. 2018GDASCX-1005). The authors thanks Professor Viktor.Biktop for experimental assistance and useful discussions.

References

[1] Gorbovets M A , Bazyleva O A , Belyaev M S , et al. Low-Cycle Fatigue of VKNA Type Single-Crystal Intermetallic Alloy Under “Hard” Loading Conditions[J]. Metallurgist, 2014, 58(7-
8):724-728.

[2] Mishin Y . Atomistic modeling of the γ and γ’-phases of the Ni–Al system[J]. Acta Materialia, 2004, 52(6):1451-1467.

[3] Lucaci M , Orban R L , Patroi D , et al. Some Aspects Regarding the Complex Alloying of the Ni3Al Intermetallic Compound with Substitutional and Interstitial Elements[J]. Advanced Materials Research, 2007, 23:67-70.

[4] Ovcharenko, Vladimir E, Boyangin, E.N, Pshenichnikov, A.P, et al. Structural-Phase State and Strength Properties of Pressure-Synthesized Ni3Al Intermetallic Compound[J]. Materials Science Forum, 2017, 906:95-100.

[5] Xiaona L , Jinhe L , Wenjun K , et al. Welding Research Progress of Novel Superalloy IC10[J]. Hot Working Technology, 2008.

[6] Shirzadi A A , Wallach E R . Analytical modelling of transient liquid phase (TLP) diffusion bonding when a temperature gradient is imposed[J]. Acta Materialia, 1999, 47(13):3551-3560.

[7] Idowu O A , Richards N L , Chaturvedi M C . Effect of bonding temperature on isothermal solidification rate during transient liquid phase bonding of Inconel 738LC superalloy[J]. Materials Science & Engineering A (Structural Materials:, Properties, Microstructure and Processing), 2005, 397(1-2):98-112.

[8] Liu Jide, Jin Tao, Zhao Nairen, et al. Influence of TLP Bonding on the Tensile Properties for a Kind of Nickel-Based Single Crystal Superalloy[J]. Rare Metal Materials and Engineering, 2007,36 (2).

[9] Lei Y E , Xiaohong L I , Wei M , et al. Brazing of IC10 superalloy with Ni-based brazing fillers using Hf and Zr as melting-point depressants[J]. Transactions of the China Welding Institution, 2009.

[10] Evgenov A G , Afanas’ev-Khodykin, A. N, Nerush S V , et al. Metallurgical Aspects of Production of Solder Powders for Vacuum Diffusion Brazing[J]. Metallurgist, 2014, 57(11-12):1120-1125.

[11] Malashenko, V.V. Kurenkova, A.F. Belyavin, V.V. Short-term strength and microstructure of soldered joints of alloy VZHL 12U, obtained using boron-containing solders with covetous silicon[J].Modern electrometallurgy,2006(3):26 - 42

[12] Huo Jiajie. Effect of Co, Cr, Mo, Ru Additions on TCP Phase Evolution and Creep Behavior at 950°C in 4thGeneration Ni-base Single Crystal Superalloys[D]. Beijing University of Science and Technology, 2018.

[13] Zhang Cheng. Research on the Precipitation And Coarsening Behavior of γ” Phase in Alloy GH4169 [D]. Tianjin University, 2014.

[14] Sheng Zunyou. Investigation of the Effects of Alloying Elements on the Wettability of Cu/W and Study of the Wettability between NdFeB and Sn-Zn-Bi Alloys[D]. Xi’an University of Technology, 2007.

[15] Xing-Fu Y U, Hong-Qiang D U, Tian S G, et al. Creep deformation mechanism in Re free second generation nickel-base single crystal superalloy during medium temperature and high stress[J]. Chinese Journal of Nonferrous Metals, 2012, 22(7):1921-1928.

[16] Zheng Yunrong, Zhang Detang. Color metallographic study of superalloys and steels [M]. Beijing: national Defense Industry Press, 1999:153-170.

[17] Giameli A F, Anton D L. Rhenium additions to a nickel-based superalloy: effect on microstructure [J].Metall. Trans. A, 1985(16): 1997-2004.

[18] Liu J L, Jin-Jiang Y U, Jin T, et al. Influence of temperature on tensile behavior and deformation mechanism of Re-containing single crystal superalloy[J]. Transactions of Nonferrous Metals Society of China, 2011, 21(7):1518-1523.

[19] Xiao Liyuan. Research on Process and Mechanism of TLP Bonded IC10 Single Crystal Superalloy [D]. Harbin Institute of Technology, 2018.