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On Post-Weld Heat Treatment of a Single Crystal Nickel-Based Superalloy Joint by Linear Friction Welding

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Abstract

Three types of post-weld heat treatment (PWHT), i.e. solution treatment + primary aging + secondary aging (**I**), secondary aging (**II**), and primary aging + secondary aging (**III**), were applied to a single crystal nickel-based superalloy joint made with linear friction welding (LFW). The results show that the grains in the thermomechanically affected zone (TMAZ) coarsen seriously and the primary γ' phase in the TMAZ precipitates unevenly after PWHT **I**. The primary γ' phase in the TMAZ and weld zone (WZ) precipitates insufficiently and fine granular secondary γ' phase is observed in the matrix after PWHT **II**. After PWHT **III**, the primary γ' phase precipitates more sufficiently and evenly compared to PWHTs **I** and **II**. Moreover, the grains in the TMAZ have not coarsened seriously and fine granular secondary γ' phase is not found after PWHT **III**. PWHT **III** seems more suitable to the LFWed single crystal nickel-based superalloy joints when performing PWHT.

Key words: Linear friction welding; Nickel-based single crystal superalloy; Post-weld heat treatment

1. Introduction

Linear friction welding (LFW) is one of the key technologies to manufacture and repair blisks of high thrust-weight ratio aeroengines. Up to now, the research and application of LFW have been focused on titanium alloys [1], [2], [3]. In order to improve the thrust-weight ratio of aeroengines, LFWed superalloy blisks would be used in high pressure compressors and turbines. Experimental studies on several kinds of superalloys have been conducted [4], [5]. Karadge et al. [4] investigated the weldability of a single crystal to a polycrystalline nickel-based superalloy with LFW and concluded that a favorable orientation of the single crystal is critical for a good weld. Mary et al. [5] studied the LFWed IN-718 joint, to find that the flash is composed of a superposition of two types of oxides: an Al oxide layer in the inner part and an outer layer rich in Nb oxide.

Our research group preliminarily studied LFW of DD6 single crystal superalloy (comparable with CMSX-4) [6]. The result shows that the microstructure in the WZ is fine polycrystal instead of a single crystal. Partial dynamic recrystallization occurs in the TMAZ during LFW. The average tensile strength of the joints is comparable to that of the parent metal (PM), while the γ' phase dissolves in the matrix but it has insufficient time to precipitate in the WZ and TMAZ. This may affect in a negative way the creep and fatigue properties of DD6 single crystal superalloy. Hence, post-weld heat treatment (PWHT) may be necessary for the joint. Up to now, however, there is no report on the PWHT of the superalloy joint. Therefore, this study mainly explores the feasibility of PWHT to a single crystal nickel-based superalloy joint.

2. Experimental procedure

6.25 x 19.2 x 50 mm DD6 blocks were LFWed with a custom made machine at NPU and the details have been reported in [6]. Fig. 1 shows a typical cross section of the joint. In the weld zone (WZ) evident recrystallization can be observed.

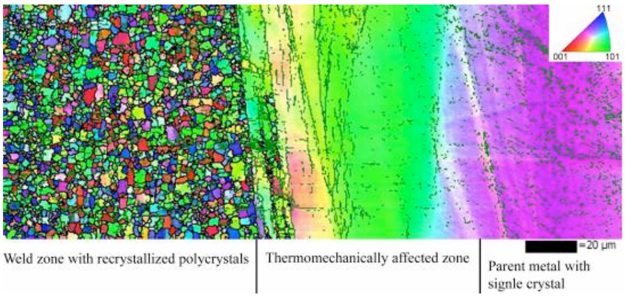

Fig. 1. Typical EBSD map (inverse pole figure) of the cross section of an LFWed DD6 joint [6].

Generally speaking, the main objective of heat treatments for nickel-based superalloys is to produce high volume fraction and better morphology of the γ' phase [7]. For DD6 single crystal, the standard heat treatment suggested by Yu et al. [8] consists of solution treatment and aging treatment. The solution heat treatment, at 1315 °C for 4 h, is intended to dissolve the γ' phase for subsequent re-precipitation in an optimized morphology and size. In addition, the solution heat treatment also results in

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elimination or reduction of the segregation to produce a more uniform, homogeneous microstructure. Aging treatment is intended to get certain amount and size of γ' phase from supersaturated solid solution, and thus the largest reinforcement effect. Aging treatment is usually divided into two steps. The primary aging, at 1120 °C for 4 h, has a great influence on the size, morphology and amount of γ' phases. The secondary aging, at 870 °C for 32 h, has a little influence on the size, morphology and amount of the γ' phase, and is mainly used to increase the squareness of the γ' phase [9]. Based on the standard heat treatment of DD6, three kinds of heat treatment conditions were conducted in the present study: (**I**) two preheating + solution treatment + primary aging + secondary aging (1290 °C/1h/air cooling (AC) + 1300 °C/2h/AC + 1315 °C/4h/AC + 1120 °C/4h/AC + 870 °C/32h/AC); (**II**) only secondary aging (870 °C/32h/AC); (**III**) primary aging + secondary aging (1120 °C/4h/AC + 870 □/32h/AC).

The heat-treated samples were polished and etched with a solution of 8 ml HCl + 12 ml HCO_3 + 5 ml H_2O_2 to prepare them for the scanning electron microscope (JEOL JSM-5800).

3. Results and discussion

3.1 PWHT I

As shown in Fig. 2a, , the grains have grown up significantly in the WZ, and cracks can be found after PWHT **I**. This may be due to the curled flash causing root cracks extend into the samples. During heat treatment, along with oxygen entering into the weld, cracks extend to the internal weld and the weld is oxidized at a high temperature.

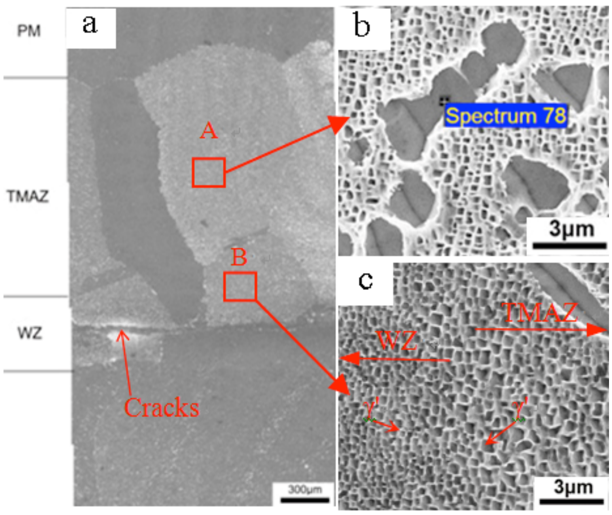

Fig. 2. Microstructures of a joint after PWHT Ⅰ: (a) low-magnification of the joint, (b and c) correspond to A and B zones marked in (a), respectively.

Although the WZ grains are larger than those in the as-welded state (Fig. 1), they are smaller than the heat-treated TMAZ grains, and coalesce. It is mainly because the microstructure in the WZ has fully recrystallized during LFW and the distortion energy caused by deformation has been released to a great extent.

Fig. 2c shows that the distribution of the γ' phase near the WZ is more even than that near the TMAZ, while the size of the γ' phase near the WZ is smaller than that near the TMAZ. In addition, the γ' phase is approximately aligned cubical.

The microstructure in the TMAZ is polycrystal instead of a single crystal and the grains in the TMAZ coarsen and become larger than 0.25mm, as shown in Fig. 2a. Due to the combined effects of high distortion energy and high temperature during solution treatment, a secondary recrystallization may occur near the WZ. In this process, high-angle boundaries grow rapidly towards low-angle boundaries and the single crystal side. Moreover, the γ' phase was completely dissolved in the matrix. Thus, there is hardly any obstacle to hinder the growth of grains. The occurrence of secondary recrystallization may be because most grains have low-angle boundaries while few are high-angle boundaries in the TMAZ after welding. High-angle boundaries have a higher grain boundary energy and this causes higher mobility of high-angle boundaries at elevated temperatures. Subsequently, the large grains consume small grains producing a few unusually large grains.

Large volume "holes" can be found in Fig. 2b. EDS test results (Table 1) show that no Al was found. However, the γ' phase is Ni_3Al. Thus, it may be speculated that these holes belong to γ matrix indicating the γ' phase precipitates unevenly in the TMAZ.

Table 1. EDS results of the hole in Fig. 2b.

Element	Cr	Co	Ni	W
wt.%	**5.57**	**10.56**	**64.45**	**14.44**
at.%	**6.45**	**10.77**	**66.96**	**4.72**

3.2 PWHT II

As shown in Fig. 3, the WZ grains after single aging heat treatment are slightly larger than those in the as-welded state (Fig.1). Due to aging for a long time, recrystallized grains become slightly larger. However, the aging temperature is low as well as atomic diffusion. In addition, the γ' phase precipitates and hinders the migration of grain boundaries. As a result, the grains grow slowly. Moreover, the γ' phase precipitates and grow up to an irregular shape, as shown in Fig. 3c.

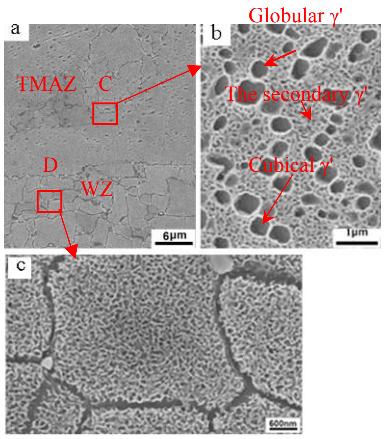

Fig. 3. Microstructures of a joint after PWHT II: (a) low magnification of the joint, (b and c) correspond to C and D zones marked in (a), respectively.

From Fig. 3a one can find that TMAZ grains hardly grow up because the aging temperature is low. Fig. 3b shows that the precipitation of the primary γ' phase is not sufficient and many fine secondary γ' particles are observed in the matrix. According to Gibbs-Thomson formula, if to be heated subsequently, these fine secondary γ' particles will dissolve into the primary γ' phase.

In addition, a few of globular primary γ' particles are found in Fig. 3b. The morphology evolution of γ' phase is driven by the decrease of the total energy of the system and it can be explained by the competitive mechanism of interfacial energy and elastic energy caused by lattice misfit. Elastic energy is proportional to the volume fraction of γ' particles and the square of misfit degree. Interfacial energy is proportional to the surface area of γ' particles. At the beginning of aging, the volume fraction of γ' particles and the misfit degree are both relatively small, while the surface area of γ' particles is relatively large. Thus elastic energy is relatively low and interfacial energy is relatively high. Accordingly, the decrease of interfacial energy is helpful to the decrease of total energy of the system. Therefore, spherical γ' phases which have minimum interfacial energy are easier to precipitate. As aging continues, diffusion of alloy elements makes the γ' particles grow up. The volume fraction of γ' particles and the misfit degree both continue to increase. Subsequently, the elastic energy also increases. In order to reduce the elastic energy of the system, globular γ' particles continue growing along the habit direction <100> of γ' phase and cubical γ' phase precipitates ultimately [10]. It can be derived from the above analysis that the secondary aging temperature seems to be relatively low and the atomic activity is low. So the γ' phase grows slowly. As a result, the volume fraction of the γ' phase is low and its size is small and the morphology is almost globular.

It should be pointed out that the secondary γ' phase in the matrix prohibits a/2<011> dislocation gliding in the matrix and promotes the reaction of a/2<011> dislocation to form a/3<112> dislocation which will shear the primary γ' phase; therefore the secondary γ' phase can decrease the stress rupture life of DD6 alloy.

3.3 PWHT **III**

It can be seen in Fig. 4a that the grains of the WZ and TMAZ have not grown up significantly after PWHT **III**.

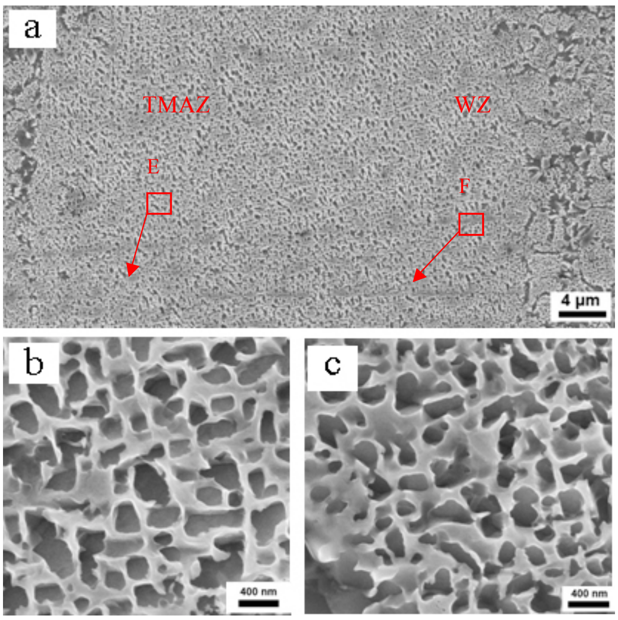

Fig. 4. Microstructures of joint after PWHT III: (a) low magnification of the joint, (b, c) correspond to E and F zones marked in (a), respectively.

In the TMAZ (Fig. 4b), the γ' phase precipitates relatively sufficiently and no secondary γ' phase is found. Similarly, the primary γ' phase precipitates relatively sufficiently and evenly in the WZ (Fig. 4c).

Yu et al. [8] found that the DD6 samples are composed of cubical primary γ' phase and fine granular secondary γ' phase after the primary aging. These secondary γ' particles increase the interfacial energy. To decrease the interfacial energy, Ostwald ripening (i.e. larger particles coarsen and smaller particles dissolve) will happen to the γ' phase during aging at 870 °C for a long time. The coarsening process is associated with Gibbs-Thomson effect. According to Gibbs-Thomson formula [11], the solubility *C(r)* is expressed as:

$$C(r) = C_a\left(1 + \frac{2\gamma V_m}{RTr}\right)$$

r is particle radius, C_a is the matrix solute concentration in the interface when r is ∞, γ is the interfacial energy, V_m is molar volume, R is volume fraction, and T is temperature.

According to the above formula, the concentration of the solute in the matrix adjacent to particles varies as the particle radius changes. Larger particles have smaller solubility than smaller particles. Therefore, concentration gradient can be formed in the matrix between the primary γ' phase whose average size is 0.3 μm and the secondary γ' phase whose average size is 0.036μm. Solute atoms flow from the fine secondary γ' phase to the periphery of the primary γ' phase, which promotes the grow-up of the primary γ' phase and the dissolution of secondary γ' phase. Finally, the secondary γ' phase gradually dissolves, the primary γ' phase grows up and the matrix channel becomes narrow with the long holding time at 870 °C.

After staying aging at 870 °C for a long time, the degree of supersaturation of the γ' phase forming elements is low, and the driving force for the γ' phase precipitation decreases. Therefore, no secondary γ' phase is found in the matrix after cooling to room temperature.

To sum up, PWHT III is relatively suitable for the LFWed single crystal nickel-based superalloy joints compared to PWHT **I** and PWHT **II** when performing PWHT.

4. Conclusions

(1) After PWHT **I**, the grains in the WZ grow to a certain degree, while the grains in the TMAZ coarsen. The γ' phase precipitates unevenly in the TMAZ.

(2) After PWHT **II**, the primary γ' phase precipitates insufficiently in the WZ and TMAZ. The secondary γ' phase which can decrease the stress rupture life of DD6 alloy precipitates in the matrix.

(3) After PWHT **III**, the grains in the TMAZ have not coarsened. The primary γ' phase precipitates to a sufficient degree and evenly with no secondary γ' phase precipitates in the matrix.

(4) Due to the increased friction heat accumulating during LFW, the γ' phase dissolves after welding. Thus, solution treatment may not be needed in PWHT of DD6 single crystal nickel-based superalloy joints. PWHT **I** is

relatively suitable for LFWed DD6 compared to PWHTs **I** and **II**.

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