The change of microstructures in Ni-base superalloy base metal during TLP diffusion bonding thermal cycle

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1. Introduction

The Ni-base superalloy GTD-111 is widely used in high power industrial gas turbine as the first stage blades. This alloy has a multiphase structure consisting of matrix, γ′precipitates, carbides, γ+γ′ eutectic. To obtain its high temperature strength, superalloy GTD-111 is strengthened by many kinds of strengthening mechanisms, mainly including precipitation hardening, solid solution hardening and grain boundary hardening. Elements Ni, Ti and Al are γ′ formers and together with significant amounts of Mo and W strengthen the alloy through a precipitation hardening mechanism. W, Mo, Ta, Ti and Cr are the most potent solid solution strengtheners[1]. Carbides and γ+γ′ eutectic are distributed during the grain boundaries. The serrated grain boundaries increase creep life and creep plasticity through of the tertiary stage of creep and preventing grain boundary sliding.

2. Experimental Procedures

The material used in this study was polycrystalline GTD-111 superalloy. Two states of superalloy, as-cast rods and ones with standard heat treatment, were used for bonding testing. The standard heat treatment for GTD-111 Ni-base superalloy includes a partial solution heat treatment at 1150℃/4hrs and an aging treatment at 845℃/24hrs. Bonding test was carried out with vacuum induction braze system.

The examination of the alloy microstructures was performed using optical microscopy (OM) and scanning electron microscopy(SEM). The specimens were etched chemically with an etchant composed of 1.2g $CuCl_2$ + 10ml HCl + 10ml ethanol, and electrolytically with a water (90ml) solution of chromic acid (Cr_2O_3 1g). An energy dispersive spectroscopy (EDS) was used to determine the elemental distribution of various phases.

3. Results and discussion

In as-cast GTD-111, beside the other phases, there are a small amount of deleterious phase such as η and TCP in grain boundaries and dendrite boundaries, and often near eutectic phases. These phases and the large clusters of γ+γ′ eutectic that had formed during solidification, had low melting temperature and poor mechanical property. From the EDS analysis, we also found the component of γ+γ′ eutectic particle in as-cast specimens was asymmetrical.

Figure 1 shows the optical micrograph of base metal before and after bonding test.

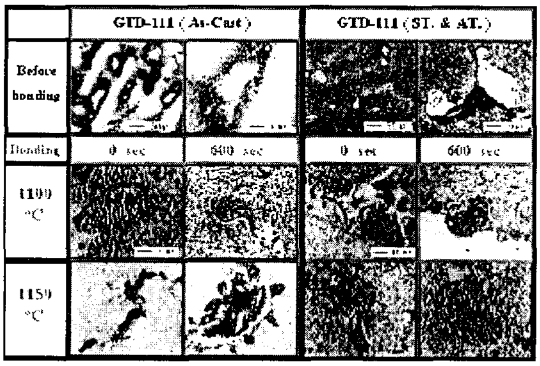
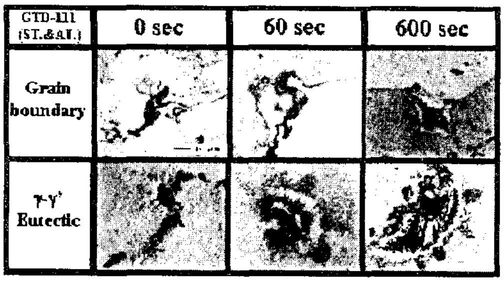


Fig.1 optical micrograph of $\gamma+\gamma'$ eutectic

The bulk structures with half-round shape in microstructure are $\gamma+\gamma'$eutectic. In the figure (left bottom), some dark and complex phases are observed, which are liquid phases after solidification. Using the as-cast specimen, in the bonding process of low temperature (1100℃ & below), there was few change in the microstructures of base metal, including grain size, dendrite size, carbide and $\gamma+\gamma'$ eutectic. However, when the bonding temperature was as high as 1150℃ or over, the liquid phase was found in some places of grain boundaries and dendrite boundaries and increased with prolonged holding. In the bonding process held at 1150℃ / 600sec, extensive formation of liquid phase was found to take place in the some places of dendrite boundaries and grain boundaries far from the bonding interface.

Figure 2a is the $\gamma+\gamma'$eutectic change with the bonding time. Compared with microstructures of the base metal before bonding, the liquid phase came from the melted eutectic phase, which mainly composes with $\gamma+\gamma'$ eutectic and a small amount of deleterious phases. During bonding process, the deleterious phases first melt and liquid phase occurred beside $\gamma+\gamma'$eutectic particles. With prolonged holding, liquid phase increased earlier, and then decreased gradually after isothermal solidification occurred, but did not completely disappeared after holding for 5.4ks at 1150℃(the bonded interlayer with 50um thickness isothermal solidification had finished after holding 5.4ks at 1150℃ during the TLP bonding.). The isothermal solidification caused by diffusion of Ti and Cr took place around the liquid phase. However the diffusivity of Ti and Cr in the matrix and γ' precipitates is very low compared with boron which is contained in insert metal for TLP bonding. This process can be shown in sketch of the microstructures change (Fig.2b). After solidification with a fast cooling rate (10℃/sec), the melt phases had a complex microstructures, where the primary MC carbides, eutectic phases and some deleterious phases would be observed. And a layer of phase with few γ'precipitates, which came from isothermal solidification, existed around the melt phases.



(a) Optical micrograph

(b) Sketch of the microstructures during bonding

Fig.2 Microstructures of as-cast GTD-111 alloy after bonding at 1150℃

In the bonding test, we found the high temperature property of GTD-111 notably improved after standard heat treatment. The micrographs in figure 3 are about the specimens with heat treatment bonded in different temperature. It shows that, in the bonding process held at 1150℃/600sec there was no liquid phase been found in base metal. Even prolonged holding to 1.8ksec or 3.6ksec at this temperature, the microstructures also did not change obviously. Bonding with the base metal after standard heat treatment, in the condition of 1200℃ and holding some time, the different phenomenon was observed in base metal. In the condition of 1200℃/0sec bonding, the microstructures seemed like the ones bonding in lower temperature. But with prolonged holding, the liquid phase took place around the carbides and $\gamma+\gamma'$ eutectic became difficult to make in focus with the optical microscope. After bonding at 1200℃/600sec, a lot of liquid phase filled during all the grain boundaries, and the MC carbide badly decomposed into small parts.

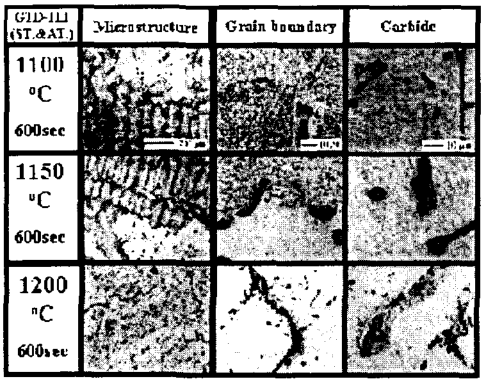
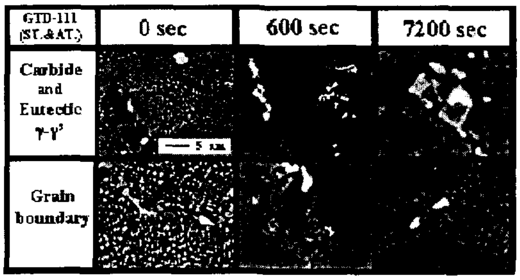
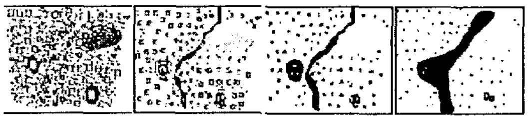


Fig.3 Optical micrograph of base metal with heat treatment after bonding

Observing by SEM(Fig.4a), in the bonding condition of 1200℃, solution of $\gamma+\gamma'$ eutectic phase took place and liquation, caused by carbides decomposition, was first found during the grain boundaries. With prolonged holding, liquation increased and also occurred around the MC carbides. After 7.2 ksec holding, a lot of liquation came out during the grain boundary and dendrite boundary. This process was different to the eutectic melting in the bonding condition of low temperature with as-cast specimens. The former was that carbides decomposed and the active elements diffused to base metal matrix, which caused liquation occurred (Fig.4b). The latter was eutectic phase melt at high temperature because of the deleterious phases and the asymmetrical component of $\gamma+\gamma'$eutectic in as-cast alloy (Fig.2b). This difference also caused the different melting places in the base metal. Bonding with as-cast alloy, the melt phase was eutectic, which has a melting temperature in the range of 1150℃ to 1160℃[2]. But after standard heat treatment, the melting temperature of $\gamma+\gamma'$eutectic was higher than γ' precipitate fully solution temperature (1180℃)[3]. So bonding with the specimen with heat treatment, $\gamma+\gamma'$ eutectic did not melt but was solutioned in high temperature.



(a) SEM micrograph



(b) Sketch of the microstructure during bonding

Fig.4 Microstructure of GTD-111 alloy with heat treatment after bonding at 1200℃

4. Conclusion

In the bonding process heat at low temperature (1100℃ & below), there was few change in the microstructure of base metal, including grain size, dendrite size, carbide and $\gamma+\gamma'$ eutectic. Bonding with as-cast base metal at 1150℃(& over), $\gamma+\gamma'$ eutectic and deleterious phases melted, and liquid took place in grain boundaries and dendrite boundaries. Bonding with alloy after heat treatment, melt phenomena was not observed until $\gamma+\gamma'$ eutectic solution occurred at temperature 1200℃.

References

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