Materials Science & Engineering A 691 (2017) 71–80

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

The effect of preheating on microstructure and mechanical properties of laser solid forming IN-738LC alloy

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ARTICLE INFO

Keywords:
Laser solid forming
IN-738LC
Preheating
Microstructure
Mechanical properties

ABSTRACT

The microstructure and microhardness of Ni-base superalloy IN-738LC fabricated by laser solid forming (LSF) with the preheating was investigated. The results showed that with the increase of the preheating temperature (T_0) (25 °C, 800 °C, 900 °C and 1050 °C), the total crack length decreased significantly, the segregation of Ti and Al worsened, and the volume fraction of equiaxed grain increased, as well as the size of blocky carbides, γ-γ′ eutectic and γ′ particles. The high-angle grain boundaries, at which the continuous liquid film can be easily formed, were found to be susceptible to cracking. Both the concentration of Ti and Al and the increase of dendrite spacing contribute to the growth of carbides and γ-γ′ eutectic in the interdendrite while the size of the γ′ phases was largely affected by preheating. It was interesting to find that γ′ phase exhibited a bimodal distribution in the bottom of the deposit with the preheating at 1050 °C. The average microhardness of the deposits without preheating and with preheating at 800 °C and 900 °C is about at the same level, while the deposit with preheating at 1050 °C has the lowest average value. The crack-free deposit can be obtained when T_0 was up to 1050 °C, and its room temperature tensile properties are superior to cast IN-738LC alloy.

1. Introduction

IN-738LC, as a typical cast Nickel-base superalloy, exhibits excellent high temperature creep properties due to the precipitation of the high volume fraction of ordered $Ni_3(Al,Ti)$-γ′ phase. In addition, it also has remarkable resistance to hot corrosion. As a result, it is widely used in the production of hot section components of both land-based and aero gas turbine engine [1,2]. But the coarse microstructure, porosity, microshrinkages and other inhomogeneities in casting parts have been reported to limit their high temperature mechanical properties [3–5]. Laser Solid Forming (LSF), developing on the basis of Laser Cladding technology and Rapid Prototyping technology, provides several advantages over conventional cast in fabricating IN-738LC alloy, such as without dies, no size restrictions, full-dense structure and high performance [6].

Many researchers have carried out LSF experiments on different superalloys and found they presented excellent mechanical properties. But cracking is frequently found for LSF superalloy with high content of Al+Ti) [7,8], since it results in the more severe grain boundary (GB) liquation, caused by nonequilibrium melting of the more low-melting-point phases. Meanwhile, there exist the large tensile thermal stresses generated by the rapid thermal cycling in LSF [9–11]. Many methods have been taken to reduce the GB liquation, including adjusting the heat input [12,13], reducing the volume fraction of low melting point eutectic at the grain boundary [14] and adjusting the alloy composition [15,16]. But the above methods have their own advantages and limitations, such as, to lower the heat input can only reduce the GB liquation partly, to reduce the volume faction of residual eutectic by increasing the cooling rate would also result in large stress. Meanwhile, preheating is also considered to be effective in reducing cracks in high energy beam welding and LSF, because preheating can reduce the stress obviously, which has been simulated and experimentally proved [2,17–19].

In fact, when fabricating IN-738LC samples by LSF, the laser deposited layer will experience lots of thermal cycles and the γ′ precipitate particles would undergo the repeated solid-state dissolution and precipitation, which would produce large thermal stress and phase transformation stress, respectively [20,21]. The preheating can reduce

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http://dx.doi.org/10.1016/j.msea.2017.03.046
Received 22 December 2016; Received in revised form 9 March 2017; Accepted 10 March 2017
Available online 12 March 2017

these stresses to a certain extent according to the preheating temperature. With the preheating, the cooling rate of molten pool would be considerably decreased, which may also result in the change of the solutal microsegregation behavior with concomitant modification of the primary solidification path [22]. In addition, the thermal gradient of the molten pool, which could also be changed by preheating, together with the suitable adjusting laser processing parameters, would affect the grain morphology by contributing to the columnar to equiaxed transition (CET) [23]. The high preheating temperature can also enhance the diffusion of elements from the precipitated phase into the matrix around it, which could result in the change of microhardness [2,19,24]. For the precipitation-strengthening superalloy IN-738LC, the formation and growth of precipitation phases can be certainly affected by the preheating temperature during the LSF process.

On considering the obtaining of a high performance crack-free IN-738LC alloy components by LSF using the preheating method, the effect of preheating on the microstructure and mechanical properties must be clearly understood. In this paper, the effects of preheating on cracking and microstructure evolution in LSF IN-738LC alloy were investigated. The potential impact of preheating on microhardness and mechanical properties is also discussed, which can provide a theoretical support for the future extensive application of LSF technology in superalloys with high Al+Ti) content.

2. Experimental

2.1. Materials

The deposition material used for the experiment was gas-atomized spherical IN-738LC alloy powder. The nominal composition of the alloy powders is listed in Table 1, and its size ranges from 53 to 150 μm in diameter. The substrate material used for the experiment was 304L stainless steel, the surface of which was sanded with sandpaper and cleaned with acetone prior to LSF.

2.2. LSF process

The LSF experiment was carried out in Xi'an Bright Laser Technologies LTD on a laser solid forming system consisted of a 1 kW YAG laser, a four-axis numerical control working table and a powder feeding system with a coaxial nozzle. The LSF process was carried out in a glove box filled with Ar gas, which was also used to transport alloy powders into molten pool and to protect the powers from oxidation. The LSF processing parameters are listed in Table 2. A middle frequency inductive power was used to heat the substrate and an infrared thermometer was used to real-time monitor the temperature of the substrate, which is controlled through a negative feedback control. The schematic diagram of induction-assisted LSF equipment is shown in Fig. 1.

2.3. Characterization

The LSF IN-738LC alloy samples are shown in Fig. 2a and the three sampled areas in vertical section are shown in Fig. 2b. The LSF samples were sectioned transverse to the laser scan direction by a wire electro discharge machining device. The specimens were chemically etched (5 g $FeCl_3$, 20 ml HCl+100 ml C_2H_5OH) to reveal the microstructure. They were further etched electrolytically in a solution of 12 ml H_3PO_4+48 ml H_2SO_4+40 ml HNO_3 to reveal the morphology of γ′ particles and carbides. The microstructure was studied by an OLYMPUS-GX71 optical microscope (OM) and a TESCAN VEGA II-LMH scanning electron microscope (SEM) equipped with an Oxford Nordlys Max2 electron backscatter diffraction (EBSD). Phase analysis was carried out by a TESCAN MIRA3 XMU field emission scanning electron microscope (FE-SEM) and a TECNai F30 G^2 transmission electron microscope (TEM). The microsegregation was analyzed by a SHIMADZU-1720 electron probe microanalysis technique (EPMA). Microhardness was tested on a Struers Duramin-A300M microhardness tester with a load of 200 g and a dwell time of 15 s.

The rectangular block LSF IN-738LC sample, shown in Fig. 2c, is machined to a standard tensile specimen with 22 mm gauge length and 4 mm gauge diameter. The room temperature tensile test was performed on an INSTRON11-3382 tensile testing machine. The tensile test was controlled by displacement and the displacement rate was 2 mm/min. The fracture surface was characterized by the SEM.

3. Results and discussion

3.1. Cracks and as-deposited microstructure

Fig. 3 shows the OM micrograph of transverse sections in the LSF IN-738LC alloy. The cracks, which can be observed in Fig. 3a-c, extend along the grain boundaries through multiple cladding layers. With the increase of preheating temperature, the number and width of the cracks are both decreased. In order to reduce the effect of the cutting position on cracks, each deposit was cut into four pieces evenly in the direction perpendicular to the scanning direction. So three transverse sections for each deposit can be obtained and the total crack length on each section was calculated. The error of the total crack length on three transverse sections was ± 3 mm, ± 2 mm, ± 2 mm and 0 for the deposits without preheating and with preheating at 800 °C, 900 °C and 1050 °C, respectively. Since the total crack length on the three sections for each deposit was similar, the middle transverse section of each deposit was selected as Fig. 3 and the total crack length on it was listed in Fig. 4 to ensure that the stress state of the position is as similar as possible. It is obvious from Fig. 4 that the total crack length decreases with the increase of preheating temperature and the crack-free deposit can be obtained with preheating at 1050 °C.

It is well known that cracking during LSF is the result of the competition between mechanical driving force for cracking and the intrinsic resistance to cracking of the material [9]. The former mainly refers to the large stress/strain produced by the rapid thermal cycling and the latter will be weakened by the γ-γ′ eutectic liquation in IN-738LC alloy. During LSF IN-738LC alloy, the rapid heating of the laser would lead to the incomplete solid-state dissolved γ′ phase reacting with the surrounding γ matrix by a eutectic-type reaction, i.e. constitutional liquation [25]. Fig. 5 shows the solidification sequence of IN-738LC alloy calculated by the Scheil model. It can be seen that the formation of γ-γ′ eutectic at 1152.9 °C after the formation of MC carbide at 1330.7 °C during the solidification, which is near the melting point of the alloy (1345 °C). Obviously, the constitutional liquation in the form of γ-γ′ eutectic reaction is more likely to occur than that of carbide during the rapid heating process due to the lower reaction temperature of γ-γ′ eutectic. This is consistent with the conclusion of Ojo in electron beam welding nickel base superalloy TMS-75 [26]. It has been reported that a continuous liquid film , caused by constitutional liquation, is more likely to form along the grain boundary than the interdendritic region, which is the key factor producing a crack

Table 1
The chemical composition of the powders (wt%).

C	Cr	Al	Ti	Mo	W	Co	Nb	Ta	Si	Fe	B	Ni
0.11	15.79	3.4	3.4	1.77	2.54	8.25	0.8	1.74	0.049	0.078	0.008	Bal.