Materials Characterization 142 (2018) 550–559

Contents lists available at ScienceDirect

Materials Characterization

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

Microstructural characterization of laser metal powder deposited Alloy 718

Andreas Segerstark[a,*], Joel Andersson[a], Lars-Erik Svensson[a], Olanrewaju Ojo[b]

[a] Department of Engineering Science, University West, Trollhättan SE-461 86, Sweden
[b] Department of Mechanical Engineering, University of Manitoba, Winnipeg R3T 5V6, Canada

ARTICLE INFO

Keywords:
Laser metal deposition
Additive manufacturing
Powder
Superalloy
Microstructure

ABSTRACT

A microstructural study of Laser Metal Powder Deposition (LMPD) of Alloy 718, using a low (40 J/mm) and high (100 J/mm) heat inputs (HIs) was performed. The microstructure was characterized in as-deposited condition as well as after a standard heat-treatment, using optical microscope (OM), scanning electron microscope (SEM) and Transmission Electron Microscope (TEM). Laves, MC-carbides, γ′ and γ″ are observed in the interdendritic areas of both conditions. However, the dendritic core only consists of γ-matrix. The high HI condition shows a slightly larger Primary Dendrite Arm Spacing (PDAS) as compared to the low HI condition. Additionally, the particle size of the Nb-rich constituents in the interdendritic regions (Laves-phase and Niobium carbide) is larger in the high HI sample. After heat-treatment, the Laves phase dissolves and is replaced by δ-phase in the interdendritic regions, while γ′, γ″ and MC-carbide remain in the interdendritic regions. However, the γ″ precipitates seems to be less developed in the dendritic core as compared to the interdendritic regions, especially in the high HI sample. This can be attributed to a heterogeneous distribution of Nb in the microstructure, with a lower Nb content in the dendritic core as compared to close to the interdendritic regions.

1. Introduction

Laser Metal Powder Deposition (LMPD) is an Additive manufacturing (AM) method that deposits material by blowing powder into a melt pool, created by a laser, on a metallic substrate. The method has shown potential as a repair method for components with columnar or single crystal structures, as these structures are attainable with this method, as shown by [1, 2]. Other applications for the method is to add features to simpler castings, refurbish end of life components, and deposit corrosion resistant or wear resistant coatings on components [3].

Pinkerton et al. [4] investigated the effect of the melt pool shape and thermal history in laser metal deposition of Waspaloy powder. The thermal history of various deposition conditions was correlated to the microstructure and grain structure of the material. They found that the grain structure of the deposits changed significantly with varying laser power and powder feeding rate in the range of 350–560 W, for the laser power and 0.21 g/s–0.63 g/s, for the powder feeding rate. Columnar grains generally formed at lower powder feeding rates and formed in the (001) orientations.

Zhang et al. [5] deposited thin walled Alloy 718 sample, 12 layers high, using a heat input of approximately 190 J/mm. It was found that Laves phase formed in the interdendritic regions of the deposit. The Laves phase area fraction was found to increase with an increased distance away from the substrate. This was attributed to the increased number of thermal cycles in the bottom of the sample, which resulted in diffusion of Nb into the γ-matrix and a successive reduction of Laves phase. Additionally, γ″ precipitates are formed at the bottom of the sample.

Tian et al. [6] have reported γ″ precipitation in LMPD of Alloy 718 powder when depositing on a thin wall substrate with the dimension 50.8 mm, 2.26 mm and 50.8 mm (length, width and height, respectively). The layers were built by first depositing a border which then was filled using a cross hatched scanning pattern. Temperature measurements showed that the temperature was in a temperature range of 600–700 °C in the bottom of the deposit, throughout the whole build (~20 min). In the bottom part of the deposit, γ″ precipitates are found in the dendritic core as well as in the interdendritic regions. However, in the top the γ″ was only found in the interdendritic regions, close to the Nb-rich constituents.

In this study, an investigation of how the formation of phases in Alloy 718 can be controlled by changing the heat input (HI) has been carried out. To do this a lower (40 J/mm) and higher HI (100 J/mm), henceforth referred to low HI and high HI respectively, was used to prepare thin walled samples. The parameter settings, although referred to low and high HI, both have a quite low energy input with the main focus of being used for repair of aerospace components. The built material was evaluated both in the as-deposited condition and after a heat treatment cycle originally developed for repair welding [7]. The built

* Corresponding author.
E-mail address: andreas.segerstark@hv.se (A. Segerstark).

https://doi.org/10.1016/j.matchar.2018.06.020
Received 19 September 2017; Received in revised form 6 June 2018; Accepted 13 June 2018
Available online 18 June 2018
1044-5803/

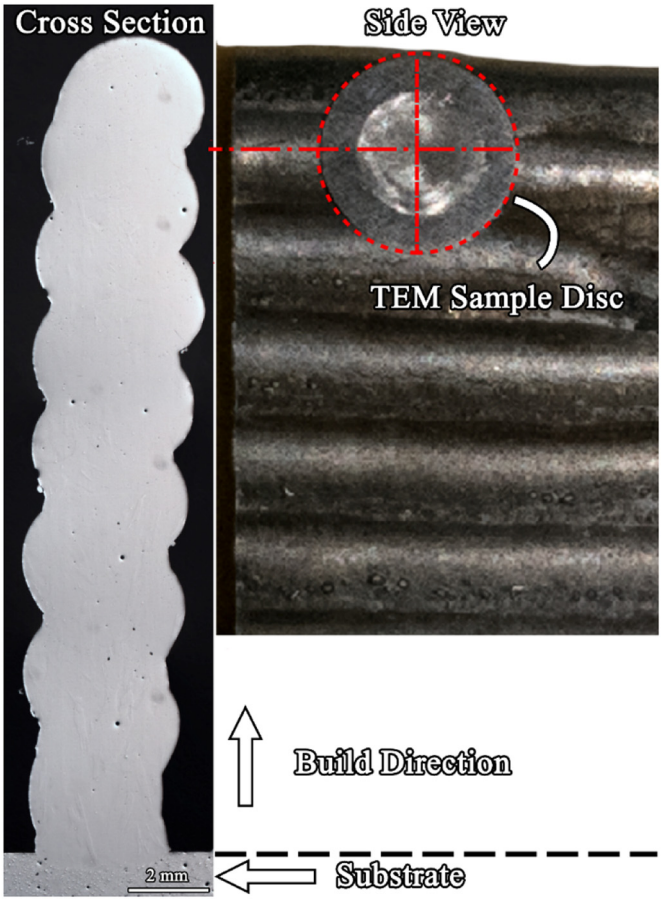

Fig. 1. A cross sectional view and a side view of the high HI sample. The position of a TEM sample disc that was excised from the top of the sample is put as an overlay on the side view image to the right.

samples were microstructurally evaluated to find HI dependent differences in the material prior to, and post heat treatment.

2. Method and Experiment

The samples were prepared by depositing gas atomized Alloy 718 powder onto an as-cast Alloy 718 substrate. Thin walled samples were deposited, with a single deposit in the width direction and 15 layers in the height direction, with a length of ~35 mm. In Fig. 1 a cross-section and a side view of the high HI sample can be seen. The nominal composition of the powder and substrate is shown in Table 1. The powder had a particle size distribution of 20–75 μm. The powder was spherically shaped, had a low porosity and few satellites as previously reported in [8]. The powder was supplied to a coaxial powder nozzle, with an annular outlet, using a volumetric powder feeding system, and Argon as carrier gas. Additionally, the melt pool was locally shielded trough the powder nozzle with Argon as shielding gas.

The laser used as energy source was an IPG 6 kW Ytterbium fibre laser with a 0.8 mm optical fibre connected to the laser optics. The laser optics was mounted on an IRB-4400 ABB robot which was used to perform the motion during deposition. The process parameters for producing the samples are shown in Table 2. The laser power for the low and high HI sample was 400 W and 1000 W, respectively, while the rest of the process parameters were kept constant.

The samples were transversely sectioned and mounted in non-conductive Bakelite before microstructural evaluation. The samples were analyzed using Optical Microscope (OM), Scanning Electron Microscope (SEM) and Transmission electron microscope (TEM) to determine and quantify the phases in the material.

Quantification of secondary phases was performed on 10 images for each position (bottom, middle and top) for the two deposition conditions with a total of 30 images per condition. The images were taken using a Hitachi TM3000 SEM with a Back-Scatter Electron (BSE) detector at 2000 times magnification. The area fraction of the Nb-rich constituents was quantified using automatic image quantification in the open source software ImageJ-Fiji, in accordance with ASTM Standard E1245-2003 [9]. Closer investigation of the phases was performed using a FEI Nova NanoSEM 450 with a high-resolution field emission gun (FEG).

The phases were identified using FEI Talos F200X TEM coupled with a Super-X Energy Dispersive Spectroscopy (EDS) system with 4 silicon drift detectors. Selected area electron diffraction patterns (SADP) obtained in the TEM were used for identifying the phases formed in the deposited material. Additionally, the chemical composition of the phases was quantified using the EDS.

Hardness measurements of the deposits were performed from the top to the bottom with the use of a Vickers micro-hardness testing system. The indents were evenly distributed along the height of the deposits with a load of 0.5 N and a dwell time of 10 s.

JMatPro simulation software was used to estimate the chemical composition of the γ-matrix at the last stage of solidification. Using TEM-EDS measurements from the interdendritic regions and the JMatPro simulated values, a Time Temperature Precipitation (TTP) diagram for the γ′/γ″ phases were plotted in JMatPro.

Temperature measurements, using type-K thermocouples, have been taken in the high and low HI samples at substrate level and after 10 depositions, using the method previously described in Segerstark et al. [10]. Matlab has been used to extrapolate the temperature measurements up to liquation temperatures using line fitting of the cooling curves.

The samples were heat-treated using an air furnace at 954 °C/1 h - air cooled +760 °C/5 h-furnace cooled +650 °C/1 h – air-cooled as described by Barron [7]. This heat treatment cycle is a common practice heat treatment originally developed for repaired jet engine components. The main purpose of the cycle was to prevent coarsening of γ″ precipitates in the components.

Table 1
Chemical composition of the GA Alloy 718 powder and the Alloy 718 substrate.

Alloy 718	Fe	Ni	Cr	Nb	Mo	Ta	Ti	Mn	C	B
GA powder	Bal.	52.7	17.5	5.0	3.17	0.003	1.07	0.065	0.031	0.003
As-cast substrate	Bal.	53.4	18.35	5.24	3.02	–	0.92	0.04	0.046	0.003

Alloy 718	Al	Co	Si	Cu	P	S	Se	Pb	Bi
GA powder	0.68	0.2	0.088	0.048	0.006	4.5 (ppm)	1.4 (ppm)	0.09 (ppm)	< 0.05 (ppm)
As-cast substrate	0.48	0.08	0.08	0.02	0.009	0.002	–	–	–