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On the role of melt flow into the surface structure and porosity development during selective laser melting

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ABSTRACT

In this study, the development of surface structure and porosity of Ti–6Al–4V samples fabricated by selective laser melting under different laser scanning speeds and powder layer thicknesses has been studied and correlated with the melt flow behaviour through both experimental and modelling approaches. The as-fabricated samples were investigated using optical microscopy (OM) and scanning electron microscopy (SEM). The interaction between laser beam and powder particles was studied by both high speed imaging observation and computational fluid dynamics (CFD) calculation. It was found that at a high laser power and a fixed powder layer thickness (20 μm), the samples contain particularly low porosity when the laser scanning speeds are below 2700 mm/s. Further increase of scanning speed led to increase of porosity but not significantly. The porosity is even more sensitive to powder layer thickness with the use of thick powder layers (above 40 μm) leading to significant porosity. The increase of porosity with laser scanning speed and powder layer thickness is not inconsistent with the observed increase in surface roughness complicated by increasingly irregular-shaped laser scanned tracks and an increased number of discontinuity and cave-like pores on the top surfaces. The formation of pores and development of rough surfaces were found by both high speed imaging and modelling, to be strongly associated with unstable melt flow and splashing of molten material.
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1. Introduction

Selective laser melting (SLM), due to its capacity to fabricate complex freeform geometries directly from computer-aided design (CAD) models, has been hailed as one of the most promising manufacturing technologies for net shape industrial scale production. So far, extensive studies have been performed aimed at defining the influence of the processing conditions on the microstructure and defect densities as well as mechanical properties. These studies have been carried out on a number of materials such as Ni-based superalloys [1,2], Ti-based alloys [3–5], Al-based alloys [6,7], steels [8] and composites [9,10]. It is clear from this work that there are also several concerns associated with this process that need to be addressed. These include residual stress development [11], cracking (particularly for certain materials such as nickel-based superalloys) [2], porosity [5,7] and mechanical anisotropy [5]. Among these concerns, porosity development is one of the most common issues that has been observed in almost all of the metallic materials processed by SLM and obviously is one of the major factors that could affect build quality and performance. However, so far, the reports on the mechanism of formation of porosity during SLM are rather limited with most of the explanations relying on assumptions. Thijs et al. [3] assumed that the formation of pores in a Ti-based alloy is due to powder denudation around the melt pool within a layer and an accumulation of the surface roughness across the layers but suggested that the formation of pores in an aluminium alloy could be due to the collapse of key holes [6]. Vilaro et al. [12] suspected that the development of pores could be the result of gas entrapment during melting and rapid solidification of the SLM process. Qiu et al. [5] directly observed open pores on the top surfaces of selectively laser melted samples and argued that the spherical pores could be due to incomplete re-melting of some localised sites on the previous layer and to the insufficient feeding of molten material to these sites. All these suggested mechanisms for the development of porosity, are based on unsubstantiated assumptions concerning the detailed mechanisms occurring during melting and solidification during SLM. Recently, Qiu et al. [13] observed the splashing of molten

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material and the evolution of the melt pool during SLM of an aluminium alloy and suggested that the formation of pores in as-SLMed samples could be associated with melt pool instability and splashing. Panwisawas et al. [14] developed a melt flow dynamics model for SLM most recently and successfully used it to explain the morphological development of pores that were formed under different processing conditions. However, studies on the influence of melt flow behaviour on surface structure and porosity development, are generally lacking.

On the other hand, it is noted that the building rate of the current SLM process is generally low. At present, it typically processes only around a 20–30 μm thick powder layer [3–5]. The building rate could be improved by increasing powder layer thickness and laser scanning speed as the laser power is increased. However, while increasing the building rate, low levels of defects such as porosity also need to be guaranteed to ensure the required structural integrity and properties. It is therefore necessary to conduct a thorough investigation into the influence of powder layer thickness and laser scanning speed on porosity level. Previously, Ma et al. [8] studied the influence of powder layer thickness on relative density of stainless steel and indicated that increasing powder layer thickness to a certain level did not affect the relative density significantly.

In this paper, we perform a systematic parametric study to investigate the influence of laser scanning speed and powder layer thickness on porosity development and correlate the porosity development with the top sample surface structures (which could be considered as the traces of melt flow in SLM) and to the melt pool and flow behaviour studied by both high speed imaging and CFD simulation.

2. Experimental and modelling methods

The material used in this study was gas atomised Ti–6Al–4V powder supplied by TLS in the size range of 20–50 μm. A Concept Laser M2 Cusing SLM system which employs an Nd:YAG laser with a wavelength of 1075 nm and a maximum laser output power of 400 W measured in continuous wave mode was used to prepare samples for microstructural characterisation. Cubic samples with a dimension of 10 × 10 × 10 mm were fabricated at a fixed laser power of 400 W and a powder layer thickness of 20 μm but at different laser scanning speeds ranging from 2000 mm/s up to 4000 mm/s. Samples were also fabricated in argon at 400 W and at both 2400 mm/s but with different powder layer thicknesses ranging from 20 μm up to 100 μm to study the influence of powder layer thickness on porosity development. An island scanning strategy which has been detailed elsewhere [5] was used to fabricate the current samples.

Metallographic specimens were mechanically prepared using conventional methods and examined using an optical microscope (OM) and a JEOL 7000 FEG-SEM (scanning electron microscope) to reveal the size, distribution and morphology of pores. The porosity distribution was examined using tessellated micrographs to study large areas and the porosity area fraction (A_f) quantified using ImageJ. The top surfaces of the as-fabricated samples as a result of laser melting over the final layer of powder were also investigated using SEM. The roughness of the top surfaces was measured by surface profilometry in a Surfcorder SE 1700 machine.

High speed imaging was conducted using a Photron FASTCAM Mini UX100 high speed camera to develop understanding of the interaction between laser beam and powder materials as well as the influence of laser processing condition and layer thickness. The imaging was performed on SLM of 20 μm thick powder layers at a laser power of 400 W and under two laser scanning speeds including 2300 mm/s and 3500 mm/s, and also on powder with different layer thicknesses from 20 μm up to 100 μm at 400 W and 2400 mm/s. The images were taken at 10,000 frames/s.

To further investigate and gain better insight into the melt flow in association with surface structure development, a computational fluid dynamics (CFD) calculation, using the C++ open source CFD toolbox called Open Field Operation and Manipulation (OpenFOAM®) has been carried out to simulate the interaction between the laser heat source and the Ti–6Al–4V powder materials. All interfacial forces present in the process have been taken into account and simulated; these include surface tension (capillary force), Marangoni's flow (thermo-capillary force), recoil pressure, drag force due to solid/liquid transition via Darcy's term, and buoyancy force. The energy dissipation in the mushy zone during melting, and heat loss due to evaporation, conduction, convection and radiation have been simulated in this work. However due to the additional computational requirements, reflections of radiation have not been simulated in this work. In order to rationalise the thermal fluid flow, one needs to solve the coupling of the Navier–Stokes equation, energy conservation, continuity equation and volume-of-fluid equation. The volume-of-fluid methodology has been used to simulate the evolution of the liquid/gas interface. In this work, the summation of metallic α_1 and gaseous phases α_2 is always unity:

$$\alpha_1 + \alpha_2 = 1 \tag{1}$$

In addition, a weight function of any parameter x is used to smear out the effect of metallic and gaseous phases, defined as

$$\bar{x} = x_1\alpha_1 + x_2\alpha_2 \tag{2}$$

Assuming that the molten Ti–6Al–4V is incompressible so that the continuity condition is

$$\vec{\nabla}\cdot\vec{u} = 0 \tag{3}$$

Here $\vec{u}$ is flow velocity. To compute the evolution of the liquid/gas interface, the volume-of-fluid equation is used

$$\frac{\partial\alpha_1}{\partial t} + \vec{\nabla}\cdot(\alpha_1\vec{u}) = -\frac{\dot{m}_V}{\rho_2} \tag{4}$$

where t is time, and the sink term in the right hand side describes the loss of metallic phase due to evaporation when the evaporation temperature T_V is reached. ρ_2 is the density of metal vapour. The mass evaporation rate $\dot{m}_V$ is expressed as $\dot{m}_V = p_V\sqrt{\frac{m}{2\pi k_B T}}$ and the recoil pressure p_V is defined by $p_V(T) = p_0\exp\left\{\frac{\Delta H_V}{R}\left(\frac{1}{T_V}-\frac{1}{T}\right)\right\}$, where p_0, ΔH_V and R are atmospheric pressure, enthalpy change due to evaporation and universal gas constant, respectively [15–18]. In order to predict the evolution of melt pool at the beginning of interaction between the heat source and the materials during SLM, one needs to identify all forces presented during the process via the conservation of momentum or Navier–Stokes equation, which was written as

$$\frac{\partial\bar{\rho}\vec{u}}{\partial t} + \vec{\nabla}\cdot(\bar{\rho}\vec{u}\otimes\vec{u}) = -\vec{\nabla}p + \vec{\nabla}\cdot\bar{\bar{T}} + \bar{\rho}g\hat{e}_z\beta(T-T_{ref}) - K_C\left(\frac{(1-f_L)^2}{f_L^3+C_K}\right)\vec{u} + \left[\sigma\kappa\hat{n} + \frac{d\sigma}{dT}(\vec{\nabla}T - \hat{n}(\hat{n}\cdot\vec{\nabla}T)) + p_V\right]|\vec{\nabla}\alpha_1|\frac{2\bar{\rho}}{(\rho_1+\rho_2)} \tag{5}$$

Eq. (5) indicates that the rate of change of fluid momentum on the left hand side (LHS) is driven by all interfacial forces on the right hand side (RHS). Divergence of stresses is divided into pressure, p, and viscous deviatoric stress tensor, $\bar{\bar{T}}$:

$$\bar{\bar{T}} = 2\bar{\mu}\left[\left(\frac{1}{2}\vec{\nabla}\vec{u} + \frac{1}{2}(\vec{\nabla}\vec{u})^T\right) - \frac{1}{3}(\vec{\nabla}\cdot\vec{u})\mathbb{I}\right] \tag{6}$$

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