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Experimental Studies on Process-Induced Morphological Characteristics of Macro- and Micro-structures in Laser Consolidated (LC) Alloys

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Abstract

Laser consolidation (LC) developed by National Research Council's Industrial Materials Institute (NRC-IMI-London) since mid-1990's, is a laser cladding based rapid manufacturing and material additive process that could fabricate a "net-shape" functional metallic shape through a "layer-upon-layer" deposition directly from a CAD model without using molds or dies. In order to evaluate the LC processability of different materials, some representative nickel-based superalloys (IN-625, IN-718, IN-738 and Waspaloy), stainless steels (austenitic SS316L and martensitic SS420) and lightweight alloys (Ti-6Al-4V titanium alloy and Al-4047 aluminum alloy) have been investigated. Like other laser cladding based processes, due to process-induced rapid directional solidification, the LC alloys have demonstrated certain unique morphological characteristics. Moreover, the "as-consolidated" LC alloys, in nature, are in "as-quenched" state, and some precipitation processes from their matrices, which are sometimes critical to the development of mechanical performance of the materials, could be effectively suppressed. Post heat treatments, therefore, could necessarily facilitate the process of achieving their required operational microstructures. In this paper, a comprehensive investigation was performed including metallurgical soundness and process-induced morphological characteristics and residual stresses of the LC materials, and microstructure development brought by post heat treatments using optical microscope (OM), scanning electronic microscope (SEM), X-ray diffraction (XRD) and hole-drilling method. The implications on the mechanical performance of the LC materials were discussed as well in order to provide basic information for potential industrial applications of the LC materials.

Keywords

Laser consolidation; Freeform fabrication; Rapid directional solidification; Microstructure; Residual stress; Nickel-base alloy; Stainless steel; Titanium alloy; Aluminum alloy;

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1. Introduction

Several rapid manufacturing processes using laser powder deposition have been independently developed by various organizations across the world, including laser engineered net shaping (LENSTM) (Sandia National Laboratory) [1,2], direct light fabrication (DLF) (Los Alamos National Laboratory) [3], direct laser deposition (DMD) (The University of Michigan) [4-6], laser direct casting (LDC) (The University of Liverpool) [7], and others since the mid-90's of the last century. The similarities of these processes are that all of them are laser cladding based material additive processes [8] whereas their differences lie in each individual process with its own unique system configuration and process conditions, which further differentiates the process capability and productivity, surface quality and dimensional accuracy for the shape (components, parts and tooling) thus fabricated, morphologies of macro- and micro-structures, and mechanical performance of the processed materials. Consequently, each of the processes might address different industrial demands [9-16]. Nonetheless, the majority of these processes might only produce "near-net" shape with rough surface finish and low dimensional accuracy, and additional machining could be a must to achieve the desired surface finish and dimensional accuracy for the final products [17], which, more or less, could reduce the attractiveness of these processes.

In contrast, laser consolidation (LC) process, being developed by NRC-IMI-London since mid-1990's, utilizes unique process conditions and system configuration that enable fabrication of "net-shape" functional shape with excellent surface finish and dimensional accuracy in one step [18]. This material additive process, in essence, can directly fabricate metallic shape through computer aided design (CAD) model. The shape thus obtained does not require further surface machining [19], which, therefore, offers unique opportunities to reduce manufacturing time in many potential applications such as complex 3D shapes that are difficult to fabricate by conventional manufacturing processes [20]. Moreover, this process offers other advantages including reduced cost, minimal wastage of materials and flexibility for design changes, etc. The LC process can also be used to repair/refurbish damaged high-value components, parts or tooling to restore their original geometries and functionalities, and extend their service life with affordable cost. The LC components, parts or tooling can minimize or even eliminate the usages of conventional welding processes that potentially induce hot cracks in the weld metal and liquidation cracks in the heat-affected zone (HAZ) for certain alloys. Recently, the LC process has been used to build functional parts (or structures) for mold and die applications such as plastic injection nozzles with embedded multiple cooling channels [21]. A series of literature is reported in the public domain by Xue-led research group about applying LC process to directly manufacture various industrial components, parts and tooling [18-26]; Figure 1 presents some of demonstrated pieces produced in NRC-IMI-London.

Within a span of more than past ten years, much work has been done on metallurgical soundness and morphological characteristics of numerous LC alloys including nickel-base high-temperature alloys (IN-625, IN-718, IN-738 and Waspalay), stainless steels

(austenitic SS316L and martensitic SS420) and lightweight alloys (Ti-6Al-4V titanium alloy and Al-4047 aluminum alloy) which have shown a range of potential applications in aerospace, defense, automotive, medical devices and other manufacturing industries. Some of these materials have been investigated by other laser-aided materials additive processes as well [1-16, 27-32]. However, since most of these processes only produce shape in “near-net”, processing conditions and system configurations could be quite different from the LC process. These differences could bring certain different macro- and micro-structures to LC alloys, which might affect the resulting mechanical performance of the materials. For example, due to process-induced rapid directional solidification, some nickel-based LC alloys exhibit coarse columnar grains in combination with refined directionally solidified dendrites inside; moreover, the LC alloys in “as-consolidated” condition are naturally in “as-quenched” state, with a consequence that, in some cases, the LC alloys cannot directly be utilized without further post heat treatments to achieve their required operational microstructures [33].

Therefore, the present study is an attempt to have a comprehensive understanding and to explore potential possibilities of controlling the macro- and micro-structures of the LC materials in order to achieve their required mechanical performance at desired service conditions, and to provide more process confidences and potential industrial applications in terms of LC-produced shape (such as components, parts and tooling). Furthermore, the similarities and differences in the LC macro- and micro-structures, and residual stresses amongst different material groups and their evolution brought by post-LC heat treatments have been studied, and potential impacts on mechanical performance of those LC materials are discussed as well.

2. Experimental

2.1. Materials

The Table 1 summarizes the selected materials and their characteristics used for the current LC study. The commercial spherically shaped gas-atomized IN-625 (commercial name: NI-328), IN-718 (NI-202), IN-738 (NI-284) and Waspaloy (NI-492) alloy powders (Praxair Surface Technologies, Indianapolis, IN), SS316L (Diamalloy 1003) alloy powder (Sulzer Metco Inc., Westbury, NY), SS420 (FE-211-1) alloy powder (Praxair Surface Technologies, Indianapolis, IN), Ti-6Al-4V alloy powder (Advanced Powders & Coatings Inc., Boisbriand, PQ), and Al-4047 (Al-Si) alloy powder (Valimet, Inc., Stockton, CA) were used as LC materials. The size of these powders range between 10 ~ 55 μm in diameter and their chemical compositions are summarized in Table 2. Since the carbon content in Praxair's FE-211-1 powder (nominally referred as to AISI SS410 steel by the manufacturer) was high (~ 0.5wt.%) and actually within the nominal carbon content of AISI SS420 steel, the authors regarded this powder as AISI SS420 steel instead.

2.2. Laser consolidation (LC) process

LC process with blown powder feeding was used for the present study. In the process setup illustrated in Fig. 2a, a 0.5~1 kW Nd:YAG laser operated in pulse mode in

combination with a precision powder feeder and a computer numerically controlled (CNC) motion system was utilized. The laser focusing optics and the powder feeding nozzle were installed on the Z-axis of the CNC motion system while one substrate was mounted on the X-Y motion table. The LC process was carried out in a glove box filled with argon to maintain a low oxygen level (less than 50 ppm) and in addition, the argon was also used as carrying gas for the delivery of alloy powder and shielding gas for the protection of the laser optics. During the LC process, the laser beam was focused onto the substrate to create a melt pool (less than 1 mm in diameter) while the alloy powder was simultaneously injected into the pool. As the laser beam (or substrate) moved following a CAD model through a pre-designed “laser tool” path, the molten material was re-solidified to form the 1st layer of the shape, which was metallurgically bonded to the substrate. The successive layer was, then, deposited on the top of the preceding layer through re-melting of a thin portion of that layer and re-solidified together with the injected powder. This “layer-upon-layer” deposition was repeated until the desired shape was formed (such as the ones shown in Fig. 2b).

2.3. Post heat treatments

During the LC process, the size of the melt pool (normally less than 1 mm in diameter) was significantly smaller than the dimensions of the underneath substrate and the wall being built that consequently acted as a huge “heat sink” to the melt, leading to a rapid quenching of the LC material immediately after its solidification. Hence, for some LC materials, proper post heat treatments might be essential to achieve their required operational microstructures. Table 3 summarizes the post heat treatments used in the current investigation.

2.4. Characterization

After the process, the LC specimens were mechanically cross-sectioned and polished down to less than 1 μm for metallographic examination. The macro- and micro-structures of the LC alloys were revealed using proper chemical etchants. An Olympus optical microscope (OM) was used to study overall macro- and micro-structures of the LC specimens. A Hitachi variable pressure scanning electron microscope along with an Oxford energy dispersive spectrometer (VP-SEM/EDS) was used to investigate the microstructures under high resolution. A Philips X’Pert X-ray diffractometer (XRD) was employed to identify phases presented in the LC specimens. The microhardness of the LC specimens was measured on a Buehler microhardness tester using a Vickers indenter. And a 100 kN Instron mechanical testing system was used to evaluate tensile properties of the LC materials as per ASTM E8-04.

2.5. Residual stresses

Hole-drilling strain-gage method as per ASTM E837-95 was used to determine average residual stresses through the thin wall of a tubular LC specimen. The measurements were conducted by using an RS-200 milling guide and a strain measurement system (model: 5000, Vishay Intertechnology, Inc., Raleigh, NC). A special

strain gage rosette (type: TEA-06-062RK-120) was glued on the surface of the tubular LC thin-walled specimen where the relieved strains were to be measured. A high-speed air turbine mill with a carbide cutter was used to drill a flat-bottom hole at the centre of the rosette, while the relieved strains in the vicinity of the hole were measured by the strain gage rosette. The residual stresses were then deduced using ReStress® software supplied with the system. The substrate which supported the LC specimen was removed prior to the measurements.

3. Results and Discussion

3.1. Macro- and micro-structures in ‘as-consolidated’ LC alloys

3.1.1 Morphological characteristics of solidified structures

It was found that all the thin-walled LC specimens sectioned along their wall build-up directions exhibit layer-wise morphologies with no exception (Figs. 3-6), which truly reflect the nature of the “layer-upon-layer” based LC process. Nevertheless, there are still some different details in their solidified structures amongst the different LC material groups, which, in certain way, also reflect different solidification behaviors of the materials in LC-process-induced thermal conditions.

For the nickel-base LC alloys (IN-625, IN-718, IN-738 and Waspaloy), both grains and dendrites, which spanned multiple deposited layers, grew almost parallel to their wall build-up directions. For instance, Figures 3a, 3c and 3e present the solidified structures of the “as-consolidated” LC IN-718, LC IN-738 and LC Waspaloy sectioned along their wall build-up directions: non-uniform coarse columnar grains with dominant fine directionally solidified dendrites (including a few cellular colonies) inside. These solidified structures reflect a kind of epitaxial growth of dendrites during a rapid directional solidification. The XRD results of the “as-consolidated” LC IN-625, LC IN-718 (Fig.3b), LC IN-738 (Fig. 3d) and LC Waspaloy (Fig. 3f) reveal the presence of a directionally solidified structure aligned with $\langle 100 \rangle$ crystallographic direction in their f.c.c.-structured γ matrices, which is almost parallel to the wall build-up direction as well. Table 4 summarizes the ratios of the XRD integral intensities of selective diffraction lines of $I_{\gamma(200)}/I_{\gamma(100)}$ to characterize the degree of preferred orientation presented in the LC materials, in which the XRD patterns of the respective powders were treated as a baseline for the respective homogenous and randomly orientated polycrystalline materials. Both the optical morphologies and XRD measurements of the nickel-base LC alloys are fully in agreement with the presence of preferred orientation.

For the LC stainless steels including austenitic SS316L and martensitic SS420, their chemical compositions and thermal conditions (such as undercooling, temperature gradient, cooling rate, solidification rate, etc.) have led to different avenues of liquid-solid phase transformation during their solidification, resulting in different solidified structures. The “as-consolidated” LC SS316L sectioned along the wall build-up direction exhibits coarse columnar grains with dominant refined cellular structure (plus a few fine directional dendrite colonies) inside, which span several deposited layers (Fig. 4a). The

XRD pattern (Fig. 4b) indicates that the major phase presented in the “as-consolidated” LC SS316L is a slightly preferred orientated austenitic phase (γ) since the XRD pattern of the LC alloy has shown a slightly higher ratio of the integral intensities of the diffraction lines of $I_{\gamma(200)}/I_{\gamma(111)}$ with about 0.7 as compared to the respective powder (about 0.5) (see Table 4). Comparatively, the “as-consolidated” LC SS420 sectioned along the wall build-up direction, on the other hand, shows that each deposited layer actually contains two sub-layers: one with dominated directionally solidified dendritic structure (a major portion of the layer) along the wall build-up direction and another with fine equiaxial cellular structure (a small portion of the layer) (Fig. 4c). Moreover, the continuity of the solidified structure has been broken between two adjective layers. Unlike the nickel-base LC alloys and LC SS316L, it is hard to recognize any grain boundaries in the “as-consolidated” LC SS420. The XRD analysis (Fig. 4d) reveals that the phases presented in the “as-consolidated” LC SS420 are martensitic α phase along with a substantially high amount of retained austenitic γ' phase (about 53.9 vol.%). Nevertheless, there is still a slightly preferred orientation in the α phase as well as γ' phase since the LC SS420 exhibits slightly higher ratios of the integral intensities of the diffraction lines of $I_{\alpha(110)}/I_{\alpha(002)}$ and $I_{\gamma(200)}/I_{\gamma(111)}$ with about 15.7 and 0.6, respectively, as compared to the corresponding powder (about 13.6 and 0.4, respectively) (see Table 4).

The morphology of solidified structure in the LC Al-4047 aluminum alloy sectioned along the wall build-up direction (Fig. 5a) is similar to the “as-consolidated” LC SS420 (Fig. 4c). The XRD pattern of the LC AL-4047 (Fig. 5b) reveals the presence of a majority of f.c.c.-structured α -Al phase plus a small amount of Si phase. The α -Al phase shows a high degree of preferred orientation because the ratio of the integral intensities of the selective diffracted lines of $I_{\alpha(200)}/I_{\alpha(111)}$ is about 3.7, which is much higher than the respective powder (about 0.5) (see Table 4).

The “as-consolidated” LC Ti-6Al-4V titanium alloy shows quite different solidified structure: although the morphology of the LC alloy sectioned along the wall build-up direction still reveals a layer-wise structure in combined with across-multiple-layer coarse columnar grains (Fig. 6a) with acicular phase inside (Fig. 6b), there is no presence of fine directionally solidified dendrites. Figure 6c exhibits the same coarse columnar grains obtained by sectioning transversely to the wall build-up direction, which appear as equiaxed grains, but actually are the cross-sectioned view of the columnar grains. The XRD pattern of the LC Ti-6Al-4V (Fig. 6d) indicates the main phase presented in the “as-consolidated” LC Ti-6Al-4V is h.p.t.-structured martensitic α phase. As compared to the wrought material, Ti-6Al-4V in both gas-atomized powder and “as-consolidated” form show a single martensitic α phase without any β phase, presumably due to the rapid solidification and quenching induced by both processes. As expected, there is a preferred orientation in the martensitic α matrix in the LC Ti-6Al-4V, which was inherent from the prior- β columnar grains, since the LC alloy exhibits a higher ratio of the integral intensities of the selective diffraction lines of $I_{\alpha(002)}/I_{\alpha(100)}$ with about 2.04 as compared to the respective powder (about 0.98) (see Table 4).

Table 5 summarizes the primary/secondary dendritic arm spacing (PDAS/SDAS) and cell spacing (CS) of some LC materials, which indicate that the LC materials experienced a rapid solidification, leading the refined solidified structures to most of them.

3.1.2. Phases and microstructures

With combined analyses of OM, SEM/EDS and XRD, the major phases presented in the “as-consolidated” LC materials can be identified.

For the nickel-base LC alloys including IN-625, IN-718, IN-738 and Waspaloy, the major phase in “as-consolidated” condition is γ matrix. It is also revealed that some primary carbide precipitated along the interdendritic and intradendritic regions (such as the cases shown in Fig. 7a and 7b). On the other hand, the rapid self-quenching after solidification could result in a supersaturated γ solid solution, which has been proven for the LC IN-718, LC IN-738 and LC Waspaloy since the high-resolution SEM images of those “as-consolidated” LC alloys show little evidence of the presence of any γ' (or γ'') particles in their γ matrices. Post heat treatments, hence, are required to ensure these age-hardenable LC alloys to achieve their full precipitations.

For austenitic SS316L, Figure 7c shows that the “as-consolidated” LC stainless steel has a mixed fine cellular/dendritic γ structure with a dominance of cellular structure. For martensitic SS420, Figure 7d shows that the “as-consolidated” LC steel has a dendritic morphology with no obvious martensitic characteristics, and carbide particles exist at both interdendritic and intradendritic regions. As mentioned, a considerable amount of retained austenite (γ') remains in the “as-consolidated” LC SS420 (about 53.9 vol.%). This might be attributed to rapid solidification and dissolution of a large amount of carbides into the steel during the molten stage, which caused the depression of M_s , leading to a less amount of martensite (α) in resultant duplex phases ($\alpha + \gamma'$) than that would be predicted for conventionally treated material. The process-induced high amount of retained austenite in the LC SS420 is usually unstable and may be transformed during service, causing dimensional change, brittleness or even cracking. Consequently, it is desirable to eliminate it through post heat treatments before usage.

For the “as-consolidated” LC Al-4047 alloy, as shown in Figure 7e, fine dendritic structure exists with the primary α -Al phase as well as laminar-type eutectic (α -Al+ Si) structure at the interdendrites.

The “as-consolidated” LC Ti-6Al-4V microstructure shows acicular martensitic α phase (Fig. 7e), and no other phase can be observed along its grain boundaries. In general, the microstructure of Ti-6Al-4V alloy depends upon the cooling conditions employed upon the high-temperature β form of the alloy. A typical “as-cast” Ti-6Al-4V microstructure consists of transformed β containing acicular α as well as α at prior- β grain boundaries, while an annealed wrought Ti-6Al-4V bar typically consists of equiaxed α grain plus intergranular β [36]. The lack of β phase in the “as-consolidated” LC Ti-6Al-4V is attributed to the high cooling rate (rapid quenching) inherent to the LC process.

3.1.3. LC solidification process

It is well known that laser-cladding based processes (including the LC process) induce rapid directional solidification as suggested by the morphological characteristics of the solidified structures in the materials thus produced. More specifically regarding the LC process, a melt created by a pulsed laser beam only has a diameter of less than 1 mm, dimension-wise much larger substrate and the wall being built would act effectively as a huge “heat sink” to rapidly extract the latent heat generated during the solidification of the melt and dissipate through the solidified phase(s) parallel but opposite to the wall build-up direction (which was the maximum temperature gradient as well). This would lead to a rapid directional solidification and subsequent rapid quenching for the melt. In addition, as compared to other laser-cladding based rapid manufacturing processes (such as LENS and DMD) in which continuous wave (CW) lasers are frequently used as their heat sources [1-7], the pulsed laser used in the LC process [18, 19] is associated with relatively high peak power but with low heat input, which induces different thermal patterns on the solidification of LC materials [39]. That is, the melt of a LC material could be created within an extremely short period of pulse duration, and there is no heat input between two consecutive pulses, which could result in a much faster cooling rate for the melt. For example, based on the measured PDAS/SDAS and CS of the microstructure, one could estimate cooling rate of a material during their solidification [40]. Marsden *et al* [41] had measured dendritic CS produced by laser melting of SS420 steel and concluded that with an order of $3 \sim 4 \mu\text{m}$ the melt was solidified at a cooling rate of more than $10^3 \text{ }^\circ\text{C/s}$ from the liquid phase. For the present LC SS420 steel, the CS is about $\sim 3.7 \mu\text{m}$, which, therefore, appears to have a comparable cooling rate of $10^3 \text{ }^\circ\text{C/s}$ as well.

Like any “as-cast” materials, the morphologies of the LC-solidified structures embed thermal histories of the solidification process. For instance, the nickel-base LC alloys (IN-625, IN-718, IN-738 and Waspaloy) occur with the dominated across-multiple-layer epitaxial dendrites; austenitic LC SS316L is dominated with extremely fine cellular structure; martensitic LC SS420 and LC Al-4047 aluminum alloy show that, within each deposited layer, there is a thin layer of fine equiaxial cells plus a large amount of directionally solidified dendrites; and LC Ti-6Al-4V titanium alloy shows coarse directional columnar structure with no presence of cells/dendrites inside. These morphological characteristics, more or less, reflect their actual thermal conditions during their solidification. In the present study, the authors attempted to use phenomenological approach to explain what might have happened in the LC solidification process although a full understanding of the solidification behaviors in these LC materials induced by the laser in pulse mode might need an in-depth experimental study combined with solidification modeling/simulation.

In most cases, for the deposition of the first layer, the LC materials would go through a process of nucleation and directional growth of crystallites. When a pulsed laser beam melted a portion of substrate in combination with injected powder (i.e., LC material) to form a melt, a substantially high thermal undercooled layer would be created in the melt ahead of solid/liquid interface due to the substrate-induced “heat sink” effect, which

would promote to form a small layer of fine nuclei at the interface. With the nucleation and growth, the latent heat of solidification (or the heat of fusion) would be generated at the interface and dissipated rapidly and directionally through the solidified phase (opposite to the wall build-up direction) to create a rapid directional solidification condition for the growth of the nuclei. Simultaneously supersaturated solute atoms would be rejected from the phase just being solidified into the melt ahead of the solid/liquid interface if the solute partition coefficient (k_0) was less than unity (most of the LC materials satisfy this condition except LC Ti-6Al-4V alloy [30]) and created a constitutionally undercooled melt layer ahead of the advancing solid/liquid interface, which could pose a perturbation at the prior stable planar interface and potentially lead to formation of cellular or columnar dendrites in the melt depending on the solidification rate [42].

Mostly, the nuclei just formed ahead of the interface would rapidly grow into columnar dendrites (such as LC IN-625, LC IN-718, LC IN-738 and Waspaloy) almost parallel but opposite to direction of the heat flow (maximum temperature gradient) which was the direction of preferred crystallographic growth of the dendrites ($\langle 100 \rangle$ for f.c.c.-structured crystal) as well. However, when the solidification rate is very high, a transition from the growth of columnar dendrites to fine cellular structure could occur in the melt [42]. The morphology of solidified structure in the LC SS316L has shown such a pattern. This also resonates with the study of Lee et al [43] in the microstructure map of directionally solidified Fe-18Cr stainless steels as function of solidification rate that the appearance of cellular structure could be either with relatively low or extremely high solidification rate (such as laser deposition) while the presence of dendritic structures could be with the solidification rate in-between. Antonsson and Fredriksson [44] had a similar observation on the solidification of IN-718 alloy as well. In the present case, pulsed laser might have introduced a much higher cooling rate (or high solidification rate) in the SS316L melt, which has pushed the solidification mode toward the region from fine dendritic to fine cellular structure.

In the subsequent layer deposition, there could be some different growth modes related to the epitaxial growth based on existing dendrites or re-nucleation and growth, depending on the nature of LC materials (for instance, the chemical and crystallographic compatibility between the new and the existing deposited layer) in addition to their thermal conditions. For the nickel-base LC alloys (IN-625, IN-718, IN-738 and Waspaloy), the primary solidification γ phase in the first layer once solidified would be rapidly quenched and remained as the same crystal structure at room temperature. When the second or subsequent layers were deposited onto the top of the first (or preceding) layer, a thin top portion of the first (or preceding) layer would be re-melted in combined with the injected powder to form a melt. The columnar dendrites in the new layer could have grown epitaxially from the columnar dendrites in the first (or preceding) layer, since the primary solidification phase (that was the γ phase) of the new layer being deposited has an identical crystal structure (probably chemical composition as well) with the first (or preceding) layer and thus, seemingly, the required free energy for the continued growth instead of new nucleation was minimal and the continued epitaxial growth could prevail. This was exactly what happened in the LC of the nickel-base alloys. In the end,

the newly deposited layer would maintain the continuation and orientation of the prior columnar dendrites, which were almost aligned with the direction of the wall build-up direction, to form the directionally solidified structure that could span several deposited layers. The characteristics of the preferred orientation in the LC nickel-base alloys might be favored in certain aero-engine applications, such as for the repair and restoration of directionally solidified turbine blades.

Stainless steels (austenitic SS316L and martensitic SS420), however, could face multiple choices of primary solidification mode (either stable δ ferrite or metastable γ austenite), depending on chemical compositions of the alloys and their solidification conditions, which makes the solidification process of steels more complicated. Vilar and Colaco [45-47] found that primary γ solidification mode is kinetically favored for higher solidification rate (such as laser deposition) whereas primary δ solidification mode predominates at very low solidification rate (such as equilibrium solidification). Lin and Yue [48] and Vilar and Colaco [45-47] had studied the solidification modes of SS316L in the laser rapid forming and DIN X42Cr13 (0.47 wt.%C, 12.8 wt.%Cr, Fe bal.) martensitic stainless steel (similar to SS420) in laser melting, respectively, and concluded that the growth of single metastable γ phase was the most favorable in both cases. It could be said, hence, that the primary solidification mode for both SS316L and SS420 in the LC process was the metastable γ phase. After solidification, the LC SS316L would retain the high-temperature γ phase at the room temperature; whereas the LC SS420 would experience a martensitic transformation ($\gamma \rightarrow \alpha + \gamma'$), and present duplex phases ($\alpha + \gamma'$) at room temperature.

Since the primary solidification phase (γ phase) of the LC SS316L would remain at room temperature, which should have a thermodynamic advantage to the epitaxial growth for the subsequent deposition of the new layers. However, the LC SS316L along the wall build-up direction has, in fact, shown the cross-multiple-layer coarse columnar grains with dominated extremely fine cellular structure inside because of the extremely rapid solidification rate, which favored to form a large amount of new nuclei. Only limited epitaxially grown colonies of directional solidified dendrites spanning several deposited layers could be observed, which contributed a weak preferred orientation parallel to the wall build-up direction in resulting LC SS316L microstructure.

In contrast, when depositing a new layer of the SS420, the primary solidification phase (γ phase) was not fully crystallographically compatible with the first (or preceding) layer (duplex phases $\alpha + \gamma'$ instead). Thus the continued epitaxial growth of the columnar dendrites from the first (or preceding) layer seemingly has been disrupted, and re-nucleation ahead of the solid/liquid interface in the melt prevailed and some of them subsequently rapidly grew into the melt to form directional columnar dendrites along the wall build-up direction but completely constrained within the layer. The resulting solidified morphology of the LC SS420 shows a large amount of directionally solidified dendrites adjective to a thin layer of fine cells within each layer.

For solidification process of the LC Al-4047 aluminum alloy, f.c.c.-structured α -Al phase would be the primary solidification phase which remained at room temperature.

During the continued cooling down, the residual liquid between the primary α -Al dendrites would be solidified to form eutectic phase (α -Al+Si). For the deposition of subsequent layer, seemingly, the formation of new nuclei at the highly undercooled melt ahead of the solid/liquid interface could be more favored than continually epitaxial growth of the dendrites from the first (or preceding) layer, possibly due to the chemical and crystallographical incompatibility between the new and the first (or preceding) layer (with primary α -Al phase plus eutectic α -Al+Si phases). Thus a thin layer of cells was formed ahead of the interface in the melt, and similar to the LC SS410, some of the newly formed nuclei could rapidly grow preferably along the wall build-up direction to form the directionally solidified dendrites but constrained within the layer.

The solidification process of the LC Ti-6Al-4V titanium alloy was slightly different, and the morphology of the solidified structure along the wall build-up direction shows across-multiple-layer coarse columnar prior- β grains with acicular martensitic α phase inside. There is no presence of fine dendritic structure like other LC materials. Kobryn and Semiatin [49] had suggested that the thermal gradient parallel to the wall build-up direction during solidification of the laser additive manufactured Ti-6Al-4V material was very high relative to the solidification rates, and consequently the formation of new nuclei ahead of the advancing solid/liquid interface could be much less possible. In addition, Wu et al [30] and S. Bontha et al [50] found experimentally and theoretically that laser deposited Ti-6Al-4V preferred to grow into columnar grains under virtually all laser processing conditions (different cooling rates) rather than nucleation, and suggested that since the primary solidified phase (b.c.c.-structured β phase) in the Ti-6Al-4V showed Ti-depletion, and the corresponding liquid phase showed Ti-enrichment as per its liquid-solid equilibrium (i.e., the solute partition coefficient (k_0) was higher than unity [30]) as indicated in the phase diagram [30] and therefore, in order to form of new nuclei, the liquid phase needed significant high undercooling, but the growth of the columnar grains needed minimal continued undercooling instead [30]. For the LC Ti-6Al-4V, it could be concluded that epitaxial growth to form coarse columnar β phase preferably along the direction of the maximum temperature gradient (parallel to the wall build-up direction), on the previously deposited layers due to similarities in the chemical composition and surface energies of the metal [30, 49]. Subsequently, the solidified melt was transformed h.p.t.-structured martensitic α phase after a rapid quenching, but retaining the prior- β columnar grain boundaries. Moreover, the martensitic α phase also inherited the preferred orientation of the parent β columnar grains.

The formation of the grain boundaries in the LC alloys could be ascribed to the inevitable misaligned orientation between some neighboring dendritic/cellular colonies since the direction of the preferred crystallographic growth and maximum temperature gradient might not be perfectly aligned one another during the directional solidification in the LC process. Consequently, inside each colony, almost parallel columnar-like dendrites constitute small-angle boundaries; the orientation difference amongst the colonies was relatively large and when the growing colonies (towards their own directions of the preferred growth under the maximum temperature gradient) met each other, large-angle boundaries formed.

Process-induced rapidly directionally solidified microstructures in LC materials could offer some unique characteristics of tensile mechanical properties such as: (1) LC materials have demonstrated almost comparable or even better hardness, yield and tensile strengths than the corresponding ones in wrought and conventional “as-cast” condition, even though the LC materials actually should be considered as “as-cast” materials [18-25]; (2) LC materials, more or less, have exhibited some anisotropic behavior in their tensile properties, particularly in their ductility. For instance, LC IN-625 alloy shows the tensile elongations of 48% and 31% along and transverse to the wall build-up direction [18]; LC IN-738 alloy shows the tensile elongations of 18% and 6.7% along and transverse to wall build-up direction [22], and LC 316L alloys shows the tensile elongations of 43% and 35% along and transverse to wall build-up direction [18].

3.2. *Effects of heat treatments in LC alloys*

As it has mentioned, due to the substrate-induced “heat sink” effect, a rapid quenching occurs in LC materials immediately after their solidification, that is, the “as-consolidated” LC materials could be considered as in “as-quenched” state; therefore, for some precipitation hardened nickel-base alloys (such as IN-718, IN-738 and Waspaloy), any precipitation of γ' (or γ'') particles from their γ matrices may be effectively suppressed or retarded. Post heat treatments (HTs), such as the recommended standard heat treatments (SHTs) in industry [38] which have been applied for the corresponding cast or wrought nickel-base alloys, are required to ensure the LC alloys to achieve their full operational microstructures. On the other aspects, for heat-treatable LC steels (such as SS420), post heat treatments could be used to control the amounts of retained austenite (γ') and carbide precipitates in order to control the dimension stability, relieve residual stresses and adjust mechanical properties such as hardness, toughness and strength in the LC steels.

3.2.1. *Nickel-base LC alloys*

It is known that directionally solidified microstructures are not only likely to impart better creep rupture properties of nickel-based alloys, but their high-temperature ductility and thermal fatigue resistance would also improve further due to the reduction of the grain boundaries perpendicular to the applied stresses. The nickel-based LC alloys possess such kind of directionally solidified microstructure to provide above benefits. However, their thermal stability was a concern for practical applications. In order to assess this particular behavior of the alloys, LC IN-625 and LC IN-738 specimens have been exposed to elevated temperatures for 3 to 6 days to assess any evolution in their microstructures and preferred orientations.

The XRD analysis reveals that after exposure to the elevated temperatures in the range of 600°C to 900°C for up to 72 h, its $\gamma(200)$ diffracted line of the LC IN-625 alloy still remained as the strongest one, and the ratio of integral intensities of the diffraction lines of $I_{\gamma(200)}/I_{\gamma(111)}$ was gradually reduced from 3.7 to 1.7 (Fig. 8a). For the LC IN-738 alloy, the preferred orientation exhibits a reasonably good thermal stability: the preferred orientation still retained after exposure to the elevated temperatures (845°C) for more

than 152 h, and the $\gamma(200)$ diffraction line was still the highest one although the ratio of $I_{\gamma(200)}/I_{\gamma(111)}$ was down from 15.8 to 4.1 (Fig. 8b).

Optical observation of microstructures also indicates that the heat-treated LC IN-625 alloy could remain the directionally solidified microstructure intact up to 700°C for 72 h (Fig. 9a); afterward, the columnar dendrites gradually disappeared; at 1100°C the γ matrix was completely replaced by coarse columnar grains aligned with the wall build-up direction, but its preferred orientation along $\langle 100 \rangle$ still reasonably retained ($I_{\gamma(200)}/I_{\gamma(111)}$ was 1.7). Figure 9b shows SEM microstructure of the LC IN-718 alloy sectioned along the wall build-up direction after the recommended SHT. The prior directionally solidified dendrites in the LC alloy completely disappeared; while the carbides (white) presented at prior interdendritic regions and grain boundaries. Similar phenomena were observed in the LC IN-738 (Fig. 9c) and LC Waspaloy (Fig. 9d) as well. After their respective SHTs, dendrites in both LC alloys were replaced by either coarse columnar grains (LC IN-738) or equiaxial grains with twins inside (LC Waspaloy). Even so, all the nickel-base LC alloys still retained certain level of preferred orientation in their respective γ matrices even after their SHTs (see Table 5).

In a short term, the process-induced preferred orientation in the nickel-base LC alloys keeps relatively stable, even though it is not naturally thermodynamic equilibrium microstructure. This knowledge presents a promise in repairing/restoring damaged aerospace components since some of the parts were made by directionally solidified cast alloys to achieve better high-temperature mechanical performance, and any repaired/restored portion preserving this characteristic is desirable.

On the other hand, after the SHTs of precipitation-hardenable nickel-base LC alloys (IN-718, IN-738 and Waspaloy), γ' (or γ'') particles should have precipitated in their respective γ matrix, which reflected in the significant increases in their hardness, yield and tensile strengths (Table 6) since the precipitation of γ' (or γ'') particles in the γ matrix was the primary strengthening mechanism for these alloys. For the heat-treated LC IN-738 alloy, bi-modal γ' particles in the γ matrix can be observed along with fine carbides precipitated at prior interdendritic and grain boundaries with proper chemical reagent (Fig. 9c). However, the authors have failed to chemically reveal morphologies of any γ' (or γ'') particles under SEM with the current chemical reagent (Fig. 9b and 9d).

3.2.2. LC stainless steels

It should be noted that although “as-consolidated” alloys (considered as “as-cast”) have comparable and even better hardness, yield and tensile strengths than the wrought and conventional “as-cast” materials, LC materials exhibit sometimes relatively lower ductilities, especially, as far as anisotropic behavior is concerned. The post heat treatments might provide an opportunity to adjust mechanical properties of the LC materials. In this paper, the austenitic LC SS316L, which is non-thermal-treatable, has been heat treated for the purpose of the investigation of the stability of refined microstructure; while the martensitic LC SS420 is thermally treatable, effect of heat

treatments on the hardness and tensile mechanical properties of the LC steel has been fully explored, and the relevance to the evolution of microstructure was studied as well.

Figure 10a shows the optical microstructures of the LC SS316L treated at 538°C and 871°C for 24 h, respectively. The former still retained fine cellular structure (with a few of columnar dendritic colonies); the latter has been replaced by non-uniform elongated coarse grains. Similar to the LC IN-625, the LC-induced refined microstructure in the LC 316L is relatively stable (at least up to 538°C), which subsequently could be beneficial to the improvement of high-temperature mechanical properties of the material thus obtained.

Comparatively, the evolution of microstructure in the heat-treated LC SS420 inevitably affects the mechanical performance of the material. Table 7 indicates that the “as-consolidated” LC SS420 has almost comparable tensile strength (σ_{UTS}) and hardness as compared to the respective wrought one, but had much lower yield strength ($\sigma_{0.2}$) and elongation (δ). After directly tempered at 316°C, the yield strength of the LC SS420 shows certain level of improvement at the expense of hardness, ultimate tensile strength and elongation. The optical and SEM microstructures of the LC SS420 tempered at 316°C (Fig. 10b) showed the prior dendritic structure started to dissolve, revealing characteristics of the tempered martensitic α' and retained austenitic γ' phase (about 31.8 vol.%) in the matrix. It is hard to find any presence of carbides.

It is believed that the mechanical characteristics of the LC SS420 might be related with the inherent process-induced microstructure. In this case, the “as-consolidated” LC SS420 has been re-austenitized at 1008°C and quenched in oil in order to eliminate the presence of the dendritic/cellular structure. The prior dendritic/cellular structure was completely disappeared, and the fine carbides presented at prior inter-dendritic/cellular sites (Fig. 10c). Then the “as-quenched” LC SS420 has been tempered at 316°C and 600°C, respectively, and their tensile mechanical testing has been performed. The results indicate that the tensile strengths of the LC SS420 austenitized at 1008°C and tempered at 316°C was significantly improved as compared to the respective one in “as-consolidated” condition (even higher than the respective wrought one) without compromising its hardness and elongation. However, the elongation of the above heat-treated LC SS420 was still lower than the respective wrought one (2.9% vs. 8%). On the other hand, the elongation of the LC SS420 austenitized at 1008°C and tempered at 600°C was significantly improved as compared to the respective one in “as-consolidated” condition (5.3% vs. 2.9%) at expense of the hardness and tensile strengths, but still lower than the respective one in wrought condition (5.3% vs. 8%). The improvement of elongation in the heat-treated LC SS420 could be ascribed to precipitation of fine carbides and continued coarsening at both prior inter- and intra-dendritic/cellular regions with increasing tempering temperature. Figure 10d shows the morphology of precipitated carbides in the LC SS420 re-austenitized at 1008°C and tempered at 457°C.

Figure 11 presents the variations of the volume fraction of the retained austenite (γ') and microhardness of the LC SS420 steel as a function of the tempering temperature. “As-quenched” LC SS420 after re-austenitized at 1008°C reveals a much less proportion of retained austenite (about 18.9 vol.%) as compared to in “as-consolidated” condition

(about 53.9 vol.%). The volume fraction of retained austenite in the LC material kept almost unchanged up to 477°C, and completely disappeared at 497°C. Meanwhile, a secondary hardening region occurred at the tempering temperature of about 427°C ~ 500°C with the maximum microhardness about 620 HV. From SEM microstructure observation and XRD phase analysis, the occurrence of the secondary hardening in the LC SS420 material during tempering may primarily be attributed to the precipitation of fine carbides. Afterward, the microhardness of the LC SS420 significantly dropped to about 380 HV, which could be caused by the continued coarsening of the carbides and disappearance of retained austenite. In short, post heat treatments can be used to adjust microstructure and mechanical performance of the LC SS420 to meet different needs.

3.3.3. *Residual stresses*

Some rectangular-shaped thin-walled tubes were made for this study. The state of residual stresses through the thickness of the walls was determined using the hole-drilling strain-gauge method. Table 8 summarizes residual stresses in some LC materials. It is interesting to note that most of the LC materials frequently showed relatively high magnitude of compressive residual stresses along the transverse to the wall build-up direction, while the residual stresses along the wall build-up direction were either in compression or even in tension.

The state and magnitude of the residual stresses could significantly affect mechanical performance of the LC-produced shape. The knowledge of the residual stresses in the LC alloys is therefore very helpful for the understanding of the functional properties. Some modeling on development of residual stresses induced by LENS or DMD process have been reported [39, 52-55]. However, the generation of process-induced residual stresses is complex process, which is closely related to not only thermal-mechanical and thermal-physical behaviors of the material being processed, but also process configuration (such as continuous laser [1-7] or pulsed laser [18, 19]) and processing parameters (solidification conditions) as well. Thus, the applicability of existing model(s) to the LC process is still questionable, and new model(s) or modified existing model(s) might be needed in order to explain the mechanism of formation of residual stresses in LC materials.

4. **Conclusions**

The laser consolidation (LC) process can produce metallurgically sound materials, which are free from cracks or porosities. Due to “heat sink” effect caused by the substrate, the process is inherent with rapid directional solidification and rapid quenching towards the LC materials, leading to some unique morphologies in macro- and micro-structures:

- (1) All the “as-consolidated” LC alloys showed layer-wise morphologies of solidified structures sectioned along the wall build-up direction, which reflect the “layer-up-layer” deposition.
- (2) Some morphologies of solidified structures in the LC materials are different. All “as-consolidated” nickel-base LC alloys showed coarse columnar grains in combination with dominant fine directionally solidified dendrites inside, which could span several deposited layers; “as-consolidated” austenitic LC SS316L

shows across-multiple-layer coarse columnar grains in combination with dominant fine cellular structure inside, whereas for “as-consolidated” martensitic SS420 and Al-4047 (Al-Si) aluminum alloy each deposited layer contains two sub-layers: one with dominated directionally solidified dendrites along the wall build-up direction and another with fine equiaxial cellular structure; and “as-consolidated” LC Ti-6Al-4V titanium alloy showed across-multiple-layer coarse columnar grains with no presence of any fine cellular/dendritic structure inside.

- (3) The “as-consolidated” LC alloys sectioned along the wall build-up direction showed more or less preferred orientation.
- (4) The γ' (or γ'') precipitation process in the nickel-base precipitation hardenable LC alloys such as LC IN-718, LC IN-738 and LC Waspaloy has been suppressed or retarded because of a rapid quenching immediately after re-solidification. Post heat treatments are required to achieve their operational microstructures for the high-temperature performance. Moreover, for all the nickel-based LC alloys, post heat treatments might eliminate their dendritic structures from the γ matrix, but the preferred orientation, more or less, has been preserved.
- (5) For heat-treatable martensitic LC SS420, post heat treatments could adjust its mechanical performance (tensile properties and hardness). The strengths and hardness of the LC SS420 after proper heat treatment can be equivalent to the respective wrought one. However, the elongation is inferior to the respective wrought one.
- (6) The residual stresses in the present LC alloys (IN-625, IN-718, SS316L and Al-6Al-4V) showed compression along the transverse to the wall build-up direction; whereas the stresses parallel to the wall build-up direction could be either tension or compression.

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Figures

Fig. 1: Examples of laser consolidated (LC) components, parts and tooling
(<https://www.imi.nrc.gc.ca>).

(a)

(b)

Fig. 2: (a) An illustration of the LC process, and (b) a LC thin-walled structure (IN-738) where “L” is the wall build-up direction whereas “T” is transverse to the wall build-up direction.

(a)

(b)

(c)

(d)

(e)

(f)

Fig. 3: (a)(c)(e) Layer-wise morphologies of solidified structures in “as-consolidated” LC IN-718, LC IN-738 and LC Waspaloy; and (b)(d)(f) XRD patterns of powdered, “as-consolidated” LC, and standard-heat-treated (SHT) LC IN-718 ($\text{CuK}\alpha$), LC IN-738 ($\text{MoK}\alpha$) and LC Waspaloy ($\text{CuK}\alpha$). All the LC specimens were sectioned along the wall build-up direction, showing non-uniform coarse columnar grains with dominant fine directionally solidified dendrites inside; whereas their major phase is f.c.c.-structured γ phase with preferred $\langle 100 \rangle$ orientation.

(a)

(b)

(c)

(d)

Fig. 4: Layer-wise morphologies of solidified structures in “as-consolidated” (a) austenitic LC SS316L: coarse columnar grains with dominant fine equiaxial cellular structure (a few directional dendrite colonies) inside, and (c) martensitic LC SS420: two sub-layers within one deposited layer: one with predominated directionally solidified dendrites and another with fine equiaxial cellular structure. Both LC specimens were sectioned along the wall build-up direction. The XRD patterns of (b) powdered and “as-consolidated” LC SS316L and (d) powdered, “as-consolidated” and “tempered at 316°C” LC SS420 (CuK_α).

(a)

Fig. 5: (a) A layer-wise morphology of solidified structure in “as-consolidated” LC Al-4047 alloy sectioned along the wall build-up direction. Two sub-layers within each deposited layer: one with predominated directionally solidified dendritic structure and another with fine equiaxed cellular structure. (b) XRD patterns of powdered and “as-consolidated” LC Al-4047 alloy ($\text{CuK}\alpha$).

(a)

(b)

(c)

(d)

Fig. 6: (a) (b) A layer-wise morphology of solidified structure in “as-consolidated” LC Ti-6Al-4V alloy sectioned along the wall build-up direction: across-multiple-layer coarse columnar grains with acicular α phase inside; (c) the morphology of “as-consolidated” LC alloy sectioned transversely to the wall build-up direction, and (d) XRD patterns of wrought, powdered, and “as-consolidated” LC Ti-6Al-4V alloy ($\text{CuK}\alpha$).

(a)

(b)

(c)

(d)

(e)

(f)

Fig. 7: SEM microstructures of (a) LC IN-718, (b) LC IN-738, (c) LC SS316L, (d) LC SS420, (e) LC Al-4047, and (f) LC Ti-6Al-4V alloy sectioned along the wall build-up direction.

(a)

(b)

Fig. 8: The effect of heat treatment conditions on preferred orientation in (a) LC IN-625 and (b) LC IN-738 alloy.

(a)

(b)

(c)

(d)

Fig. 9: Mirostructures of heat-treated (HT) (a) LC IN-625 (700°C/72 h), (b) LC IN-718 (980°C/1 h + 720°C/8 h + 620°C/8 h), (c) LC IN-738 (1120°C/2 h + 845°C/24 h) and (d) LC Waspaloy (1079°C/4 h + 843°C/4 h + 760°C/16 h). All the specimens were sectioned along the wall build-up direction.

(a)

(b)

(c)

(d)

Fig. 10: Mirostructures of heat-treated (HT) (a) LC SS316L (538°C/24 h and 871°C/24 h), (b) LC SS420 (316°C/2 h), (c) LC SS420 (1008°C/0.5 h), and (d) LC SS420 (1008°C/0.5 h + 457°C/2 h). All the specimens were sectioned along the wall build-up direction.

Fig. 11: The variations of volume fraction of retained austenite and microhardness of LC SS420 steel with tempering temperature (re-austenitizing temperature was 1008°C).

Tables

Table 1: Selective alloys and their characteristics for the current LC study [34-37]

Material	Characteristics
IN-625	Ni-based solid solution strengthened superalloy that has excellent corrosion and oxidation resistance, and outstanding strength and toughness in the temperature range up to 1093°C.
IN-718	Ni-based age-hardenable superalloy with corrosion resistance, high strength, excellent high-temperature creep-rupture strength and outstanding weldability.
IN-738	Ni-based age-hardened superalloy with excellent creep strength and hot corrosion resistance.
Waspaloy	Ni-based age-hardenable superalloy with high temperature strength and good corrosion resistance.
SS316L	Low carbon molybdenum-bearing austenitic stainless steel which is immune from carbide precipitation at grain boundaries and has a better weldability; the alloy has better toughness, high creep, stress to rupture and tensile strength at elevated temperature as well.
SS420	Martensitic stainless steel with high strength and corrosion resistance.
Ti-6Al-4V	A α - β type Ti-base alloy with an excellent combination of strength and toughness along with excellent corrosion resistance.
Al-4047	Aluminum-silicon (Al-Si) alloy with excellent weldability.

Table 2: Chemical compositions of the alloy powders provided by the manufacturers

Alloy	IN-625	IN-718	IN-738	Waspaloy	SS316L	SS420	Ti-6Al-4V	Al-4047
Chromium	22.0	19.0	16.1	19.1	16.4	14.0	0.02	-
Molybdenum	9.0	3.0	1.77	4.2	2.23	-	-	-
Tungsten			2.59					
Tantalum	3.7	5.0	1.74	-	-	-	-	-
+Niobium			+0.86					
Zirconium	0.03	-	0.06	-	-	-	-	-
Manganese	-	-	-	-	1.42	0.37	-	-
Vanadium	-	-	-	-	-	-	3.94	-
Copper	-	-	-	-	-	-	-	0.01
Silicon	-	-	-	-	0.45	1.0	-	11.41
Magnesium	-	-	-	-	-	-	-	0.02
Boron			0.011					
Carbon	-	0.05	0.10	-	0.024	0.5	0.07	-
Nitrogen	-	-	-	-	-	-	0.01	-
Oxygen	-	-	-	-	-	-	0.13	-
Aluminum	-	-	3.27	1.1	-	-	6.18	Bal.
Cobalt	-	-	8.34	13.1	-	-	-	-
Titanium	-	-	3.38	2.9	-	-	Bal.	-
Iron	0.4	17.0	-	0.049	-	Bal.	0.10	0.17
Nickel	Bal.	Bal.	Bal.	Bal.	10.4	-	0.02	-

Table 3: Post heat treatments (HTs) for the LC materials (where the SHTs are referred to recommended standard heat treatments employed in industry [38])

LC Material	Heat treatment (HT) schedule
IN-625	600°C/72 h, 700°C/72 h, 800°C/72 h, 900°C/72 h, and 1100°C/2 h with air cooling, respectively.
IN-718	980°C/1 h + 720°C/8 h + 620°C/8 h with air cooling (SHT)
IN-738	1120°C/2 h + 845°C/24 h with air cooling (SHT)
Waspaloy	1079°C/4 h + 843°C/4 h + 760°C/16 h with air cooling (SHT)
SS316L	Annealed at (a) 538°C/24 h and (b) 871°C/24 h in air.
SS420	(a) Tempered at 316°C/2 h, (b) annealed at 890°C/1 h, and (c) re-austenitized at 1008°C/0.5 h and quenched in oil, and then tempered at 316°C, 427°C, 457°C, 477°C, 497°C, 538°C and 600°C for 2 h, respectively.

Table 4: Summary of XRD integral intensities of selective diffraction lines to characterize the degree of preferred orientation in LC materials sectioned along the wall build-up direction.

LC Material	Condition	$I_{\gamma(200)}/I_{\gamma(111)}$ (γ phase)	$I_{\alpha(110)}/I_{\alpha(200)}$ (α phase)	$I_{\alpha(200)}/I_{\alpha(111)}$ (α phase)	$I_{\alpha(002)}/I_{\alpha(100)}$ (α phase)
IN-625	Powder	0.5			
	As consolidated	3.7			
	900°C/72 h	1.7			
IN-718	Powder	0.5			
	As consolidated	1.2			
	SHT	1.0			
IN-738	Powder	0.5			
	As consolidated	15.8			
	SHT	15.8			
Waspaloy	Powder	0.5			
	As consolidated	15.3			
	SHT	2.1			
SS316L	Powder	0.5			
	As consolidated	0.7			
SS420	Powder	0.4	13.6		
	As consolidated	0.6	15.7		
Ti-6Al-4V	Powder				0.9
	As consolidated				2.6
Al-4047	Powder			0.5	
	As consolidated			3.7	

Table 5: Average primary/secondary dendritic arm spacing (PDAS/SDAS) and cellular spacing (CS) of “as-consolidated” LC alloys

LC alloy	PDAS (μm)	SDAS (μm)	CS (μm)
IN-625	~ 2.1	~ 1.2	-
IN-718	~ 3.7	-	~ 3.8
IN-738	~ 3.1	~ 1.7	-
Waspaloy	~ 2.7	-	~ 2.3
SS316L	~ 2.5	-	~ 2.5
SS420	~ 3.6	~ 1.6	~ 3.7
Al-4047	~ 1.1	~ 0.9	~ 1.1

Table 6: Effect of SHTs on hardness and tensile mechanical properties of nickel-based LC IN-718, LC IN-738 and LC Waspaloy.

LC Material	Direction	$\sigma_{0.2}$ (MPa)	σ_{UTS} (MPa)	δ (%)	HV
As consolidated IN-718	L	432	802	39	257
SHT LC IN-718	L	1085	1238	21	445
As consolidated IN-738 [22]	L	869	1202	18	354
SHT LC IN-738 [22]	L	869	1269	17	413
As consolidated Waspaloy	L	508	797	40	276
SHT LC Waspaloy	L	769	1126	27	369

Table 7: Effect of different heat treatment on hardness and tensile mechanical properties of LC SS420.

LC Material	Direction	$\sigma_{0.2}$ (MPa)	σ_{UTS} (MPa)	δ (%)	HV
As consolidated SS420	L	833	1602	2.9	536 (~52 HRc)
Wrought SS420 bar (tempered at 204°C) [51]	n/a	1480	1720	8	52HRc
SS420 (tempered at 316°C)	L	1087	1394	1.6	458 (~46 HRc)
SS420 (austenitized at 1008°C and tempered at 316°C)	L	1550	1888	2.9	525 (~51 HRc)
SS420 (austenitized at 1008°C and tempered at 600°C)	L	977	1184	5.3	376 (~38 HRc)

Table 8: Residual Stresses in some LC alloys

LC Specimen	σ_T (MPa)	σ_L (MPa)
IN-625	-349	-42
IN-718	-297	121
SS316L	-107	-47
Ti-6Al-4V	-105	-43.5