

High Pressure Die Casting (HPDC) with Laser Additively Manufactured (LAM) Lattice Inserts: An Exploratory Study on Densification and Interfacial Microstructure

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Research Article

Keywords: High Pressure Die Casting, Additive manufacturing, Laser Powder Bed Fusion (PBF-LB/M), interface, microstructure, lattice structure

Posted Date: October 12th, 2026

DOI: <https://doi.org/10.21203/rs.3.rs-10919134/v1>

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Additional Declarations: No competing interests reported.

Abstract

The integration of additively manufactured (AM) inserts into High-Pressure Die Casting (HPDC) has to date focused predominantly on tailored conformal cooling channels of molds, while the use of AM lattice structures as functional inserts within HPDC components has received comparatively little attention. Such hybrid-manufactured parts are attractive for applications requiring low density, lightweight design, and high specific strength combined with energy-absorption capability. However, achieving adequate infiltration and densification of the lattice insert, together with robust mechanical bonding at the interface between the AM printed lattice insert and the HPDC alloy, remains an open challenge. In this study, AM-fabricated 316L stainless steel lattice structures of body-centered cubic (BCC) and Kelvin cell topologies, with different strut thicknesses (0.6 and 1.2 mm), were embedded as inserts in HPDC components cast from EN AC-43500 aluminum alloy, with and without intensification pressure to assess its effect on infiltration and densification behavior. The interfacial microstructure between the two dissimilar materials was characterized along with microhardness measurement across the interface. The results indicate that intensification pressure promotes high densification in both BCC and Kelvin hybrid composites, even at long flow lengths. Good and defect-free mechanical bonding was observed at the steel/aluminum interface while Fe-rich intermetallic phase formation was suppressed owing to the rapid solidification of conditions of the HPDC process.

Introduction

Laser based metal additive manufacturing such as the Laser Powder Bed Fusion (PBF-LB/M) process is popular for their design liberties and their ability to manufacture near-net shaped and complex geometrical components [1]. With metallic powder as feedstock, the PBF-LB/M has proven to have varying material applicability including alloys of aluminum, steel, nickel, Titanium etc [2]. High Pressure Die Casting (HPDC) on the other hand is a conventional manufacturing permanent mold casting process which is desired for the manufacturing of lightweight components not only for good dimensional accuracy but also for excellent surface finish [3]. HPDC mainly involves the dosing of the molten metal, a three-stage injection process; part solidification, part ejection and die spraying. Previous attempts have been made by several authors to harness the unique strength of both processes to achieve desired functional performance. Most notably, additive manufacturing (AM) processes have been severally used to manufacture mold with tailor-made conformal cooling channels to control thermal cycling and inhomogeneities [4–6]. The authors[7] using maraging steel PBF-LB/M printed HPDC mold inserts for the purpose of casting fatigue test samples. In another work by the same authors [8], maraging steel cores were directly printed on a die base using the PBF-LB/M process to instigate interfacial bonding. Others such as[9] also sought to improve thermal management of the HPDC die by redesigning conformal cooling channels of HPDC mold insert and additively manufacturing them using the PBF-LB/M process. Using different AM processes,[10] evaluated the performance of the AM manufactured inserts and concluded that some alloys met the requirement for their application as AM inserts for HPDC. More recently, lattice structures integrated in targeted regions of AM-built inserts have attracted attention due to the potential density reduction and the offer of more flexibility in thermal management control. Their periodic structure which constitutes nodes and struts create more pore spaces and structural stability [11]. The authors[12] incorporated lattice structures into AM printed molds and concluded that they helped temperature and cooling uniformity. The authors[13] after incorporating AM

printed self-supporting lattice structure cooling channels observed an improvement of heat transfer at circular channel regions which was attributed to increased interfacial surface areas and fluid vorticity.

[14], additively manufactured A357 cubes on a A356 die cast low control arm cast part (aluminium on aluminium). The authors achieved defect-free metallurgical bonding after sandblasting and remelting of the cast substrate. With focus on dissimilar materials, steel on aluminium,[15] performed hybrid casting where AlSi10MnMg was cast (HPDC) on cold rolled DP800 dual phase steel and achieved very good bonding strength at the interface of the hybrid cast parts.

In the current work, PBF-LB/M 316L printed lattice structures of two main designs, Body Centered Cubic (BCC) and Kelvin lattice structures of different strut thicknesses are incorporated as HPDC mold inserts. Kelvin lattice structure is a tetradehedron structure which is made up of 8 hexahedrons and 6 tetrahedrons, It is widely popular for its high energy absorbing properties [16, 17]. With similarly good energy absorbing properties is the bending dominated BCC lattice structure. The different lattice topologies such as BCC and Kelvin with varying strut thicknesses is expected to influence the extent of liquid melt permeability, pore distribution and flow resistance. Therefore, the influence of the different lattice topologies, strut thickness, flow length and filling conditions (intensification) on densification is investigated and discussed. The interfacial bonding between the HPDC aluminium alloy and the additively manufactured steel lattice structures is also investigated and characterized with Scanning Electron Microscopy (SEM) using both Energy Dispersive Spectroscopy (EDS) and Electron Backscatter Diffraction (EBSD). The characterized interfacial microstructure is discussed along with microhardness profiles across the interface.

Experimental Procedure

Materials and Methods

The gas atomized 316L powder used to additively manufacture the lattice structures was procured from Carpenter Additive GmbH. The powder was characterized using the CAMSIZER X2 to determine its d10, d50, and d90 values which were 10 μm , 35 μm and 60 μm respectively. The mean particle sphericity, aspect ratio and symmetry were equally obtained as 0.89, 0.90 and 0.95 respectively. For each print, the feedstock powder was dried for at least 6 hours at 80°C. The feedstock powder was printed on the Aconity Mini Laser Powder Bed Fusion (PBF-LB/M) additive manufacturing machine from Aconity GmbH, Herzogenrath – Germany. The Aconity Mini is equipped with an IPG photonics (Yttrium Fibre Optic Laser, 400W) with spot diameter of 80 μm . Cubic samples of dimensions 10 x 10 x 10 mm³ which were designed with support structures and sliced using Netfabb Autodesk Premium software were first printed to obtain the optimized process window. In the process window design, bidirectional scan strategy was used, keeping hatch distance and layer thickness constant at 140 μm and 30 μm respectively. Scan speed and laser power were varied between (800 mm/s – 1450 mm/s) and (110 W – 210W) respectively to obtain 16 cubes of unique parameter combination. Using Zeiss Light Optical Microscope, images were taken and the relative densities analyzed using FIJI IMAGE J software. The highest relative density of 99.9% was obtain for volumetric energy density of 56 J/mm³. This parameter was used to print the lattice structures. BCC and Kelvin lattice structures were designed using the ntopology software. The BCC and Kelvin unit cell strut thicknesses were

varied between 0.6 and 1.2 mm and cuboidal lattice structures of 45 x 22 x 6 mm³ dimensions were designed and printed. Figure 1 shows the unit cells and lattice structures of both BCC and Kelvin with different strut thicknesses (0.6 and 1.2 mm) and orthogonal view.

The AlSi10MnMg (EN AC-43500) was melted with the aid of a resistance furnace with 250 kg capacity at a temperature of 720°C. The HPDC of the AlSi10MnMg (EN AC-43500) alloy was performed using a H-630 SC cold-chamber HPDC machine with a 6900 kN nominal locking force manufactured by Bühler AG, Uzwil - Switzerland. The casting die was made of 1.2343 hot work tool steel (HWTS). Castings were carried out at both with and without intensification pressure at the end of mold filling while the AM printed lattice inserts were fitted in a DC04 cold rolled sheet. **Table 1** below shows the chemical composition of the cast alloy as analyzed with an optical emission spectrometer and the 316L alloy as supplied by the manufacturer.

For the casting process, the melt temperature in the holding furnace was set to 710°C which resulted in a temperature of 680°C for dosing into the shot sleeve. Average melt velocity at the ingate was 40 ms⁻¹ during the mold filling phase and intensification pressure was set to 400 bar to achieve a quasi-steady-state temperature in the shot sleeve and tool. The shot sleeve and the die were preheated using an oil-based thermoregulation fluid with a respective temperature of 250°C and 200°C. Additionally three (3) test shots were performed ahead of the main casting experiments. Each shot took ~ 1,7 seconds with a mold filling time of ~ 35 ms. A total of 12 castings were performed; six (6) with and 6 without intensification pressure. Two inserts, one directly in contact with the main ingate (short flow-length) and the other in direct contact with the overflows (long flow-length) were selected from each casting for metallographic analysis (embedding, grinding and polishing). Figure 2 below shows the position of inserts at short flow length and long flow length and the schematic diagrams of the lattice structures.

Results and Discussions

Influence of Intensification pressure, flow length and Strut thickness on densification

Light Optical Micrographs (LOM) obtained from metallographically prepared sample cross-sections are presented below, showing standard filling (0 bar - without intensification) and with intensification pressure (400 bar), (Figs. 3 and 4) for both short flow length (SF) and long flow length (LF). Kelvin and BCC lattice structures with different strut thicknesses (0.6 and 1.2 mm) were integrated as inserts (Fig. 2). Figures 3 and 4 show a comparatively poorer densification for both BCC and Kelvin lattice structure inserts with standard filling conditions. After the application of intensification pressure at the end of mold filling, it can be seen that for both BCC and Kelvin lattice structures, as compared to standard filling conditions, there is a significant improvement in densification. It can also be observed that, with intensification pressure, for both BCC and Kelvin lattice structure inserts, densification is much better for long flow length than short flow length regardless of the strut thickness. The quantified percentage densification, measured from samples shown in Figs. 3 and 4 were plotted, as shown below in Fig. 5a-d. The highest densification which was possibly achieved for BCC was both BCC 0.6 mm strut thickness with densification of 99% and BCC 1.2 mm strut thickness with densification of 99.5% using intensification pressure (long flow length). Similarly, for

Kelvin lattice structures with intensification pressure, highest percentage densification was 99.2% and 98.1% for long flow length with 1.2 mm strut thickness and 0.6 mm strut thickness respectively. Lattice topology and strut thickness have an influence on densification through their combined effect on open volume, pore size, permeability, surface area, and thermal extraction [18, 19]. In general, for a fixed lattice unit cell, an increase in the strut thickness increases the lattice relative density and the wetting surface area per volume but reduces pore volume, and raises flow resistance and pressure drop of the liquid melt through the lattice [20]. In HPDC, during the primary injection (filling) phase, cavity filling occurs over a characteristically short period, typically in the order of tens of milliseconds governed by high gate velocities [21]. As this phase is optimized for rapid die filling rather than shrinkage compensation, a secondary (intensification) pressure, generated by the hydraulic system, is subsequently applied after cavity filling to promote feeding of the remaining liquid metal and to compensate for solidification shrinkage. From Fig. 3 and Fig. 4, the poor densification with standard filling conditions could be attributed to insufficient infiltration of the lattice structure by the aluminium alloy liquid melt. The infiltration pressure (as given by Darcy's equation below) needed to completely infiltrate the porous network of the lattice structure inserts was limited after the initial filling of the die cavity. From the densification plots in Fig. 5a and 5c, It can be seen that without intensification, short flow length favors higher densification for both BCC and Kelvin lattice structures, independent of the strut thickness. Similarly, as shown on Fig. 5b and 5d, with intensification, densification is favored by long flow length for both BCC and Kelvin lattice structures regardless of strut thickness. Numerous studies have demonstrated that increased intensification pressure typically leads to higher densification and reduced porosity levels [22]. There is the suppression of pore nucleation and growth since elevated intensification pressure also increases the local hydrostatic pressure acting on the solidifying melt, thereby raising the pressure threshold required for pore nucleation and restricting the subsequent growth of any nucleated pores.

$$\Delta P \propto \frac{\mu LU}{K}$$

Where ΔP is the pressure required for infiltration, μ is the melt viscosity, L is the flow length, U is the superficial melt velocity, and K is lattice permeability [23, 24].

In general, the effectiveness of intensification pressure in promoting densification is also influenced by the flow length, as pressure transmission from the gate to distal regions of the casting relies on maintaining an open, unsolidified feeding channel. At short flow lengths near the gate, the feeding path remains liquid for a longer duration, allowing intensification pressure to be transmitted with minimal loss and effectively suppressing porosity.

At long flow lengths, however, the melt front may lose superheat during filling and begin solidifying earlier, so the feeding channel could close before or shortly after intensification is applied. As a result, the pressure that does propagate is further attenuated by frictional losses along the increasingly narrow and partially solidified path. Consequently, regions far from the gate typically receive a reduced effective pressure relative to the nominal intensification pressure set at the machine, resulting in coarser shrinkage porosity and lower densification [25]. With this background, without intensification pressure, and for long flow length, densification is expected to be poorer as compared to short flow length. This is clearly confirmed in Figs. 3

and 4 as well as in Fig. 5a and 5c where long flow length had poorer densification under standard filling conditions. The micrographs in Figs. 3 and 4 as well as the densification plot in Fig. 5b and 5d further show that, long flow length promoted better densification as compared to short flow length when intensification was applied. It is plausible to suggest that for short flow length, infiltration was limited presumably because, at the end of the mold filling under standard conditions, after the application of intensification pressure, there was a poor continuous and non-uniform pressure on the liquid to filtrate the lattice structure. This might be due to proximity to the gate and the lattice struts acting as obstacles, forcing the liquid melt to impinged directly on the lattice structure insert or flown around it rather than uniformly through it, hence the poor densification. With better densification for long flow length, it is also plausible to suggest that, the surrounding cavity might have already been filled when intensification began which allowed pressure to act more uniformly and flown through the insert. Therefore, the dynamic impact at the end of mold filling during the transition from standard filling to intensification could be regarded as good enough for long flow length condition.

With intensification, and long flow length, BCC 0.6mm and 1.2mm strut thickness had good densification for presumably different reasons. It is plausible to suggest that while the thin 0.6 mm strut thickness offered a lower hydraulic resistance and easier infiltration pressure transmission through the lattice leading to good densification, for the 1.2 mm strut lattice structure, it could be due to a better and uniform surface area wetting contact with the melt or the presence of fewer or partially closed pore throats. From Figs. 3 and 4, two dense samples, BCC – 1.2 mm and Kelvin – 1.2 mm (long flow length and with intensification of 400 bar) were both selected for microscopic investigations as discussed below.

Microstructure Characterization – Scanning Electron Microscopy (SEM – EDS/EBSD)

Interfacial Characterization

In Fig. 6a, the interfacial microstructure between the 316L BCC – 1.2 mm strut (LF) AM manufactured lattice and the HPDC aluminium alloy can be observed in the SEM micrograph. Very good bonding can be observed between the two different materials made from the two different processes as no clear defects can be seen. This indicates a very good surface wetting of the 316L AM manufactured strut by the EN AC-43500 HPDC alloy, which improved densification as earlier observed. From the EDS line mapping (Fig. 6b), apart from Mn and Mg which are found to be slightly distributed in both alloys and are found to be distributed across the interface, the others are distinctively high in the respective alloys; such as Fe, Cr Mo, Ni for 316L, and for the EN AC-43500, Al and Si. The interfacial microstructure between the 316L Kelvin – 1.2 mm strut thickness (LF) AM lattice structure and that of the cast alloy is also shown in Fig. 7a to also have good interfacial bonding with no observed lack of fusion at the interface. The EDS line map (Fig. 7b) shows elemental distribution across the interface similar to that of the BCC in Fig. 6b. Reliable interfacial bonding is essential for producing high-integrity AM–HPDC hybrid and multi-material components capable of transferring static and dynamic loads. As shown in Figs. 6 and 7, the constituent elements remain largely confined to their respective alloy regions, with only the elements common to both alloys exhibiting continuity across the interface. The absence of a pronounced interdiffusion zone or chemically mixed reaction layer suggests that

bonding is governed predominantly by mechanical interlocking rather than by extensive metallurgical interaction. For dissimilar aluminium and iron-based alloys, interfacial bonding may occur through two principal mechanisms; metallurgical bonding and mechanical bonding [26]. Metallurgical bonding generally involves atomic diffusion, local melting, interfacial alloying, or the formation of intermetallic phases. It is therefore typically associated with a chemically graded interface, a fused transition zone, or a distinct reaction layer containing elements from both alloys, as reported by Rohlwing et al. [27]. In contrast, the sharp elemental segregation observed in the present case provides limited evidence for substantial interdiffusion or reaction-layer formation. Mechanical bonding is instead associated with physical interlocking and adhesion at the interface.

In insert-based casting, this mechanism can be enhanced when the solidification front initiates at or advances against the insert surface, allowing the solidifying aluminium alloy to conform to surface asperities, pores, or other microscale features of the AM insert. To further characterize the texture at the interface, EBSD characterization was performed for both 316L BCC – 1.2 mm and EN AC-43500 HPDC interface as well as the 316L Kelvin – 1.2 mm and AC-43500 HPDC interface. In Fig. 8, the BCC lattice structure shows a columnar grain morphology with random texture which grows along the build direction (BD) while for HPDC the alloy rather shows an equiaxed grain morphology also with a random texture. The morphology of grains is governed by the ratio of temperature gradient (G) to the rate of solidification (R), as a result, high temperature gradient drives the formation of columnar grains [28].

In AM processes such as the powder bed fusion (PBF-LB/M), as heat flux is conducted away from the build substrate due to the layer upon layer build approach, very high temperature gradient is generated in the direction of build with magnitude in the order of 10^7K/m [29]. Such high temperature gradient is responsible for the columnar grains as observed in the BCC lattice and generally fine due to the fast cooling rates of the PBF-LB/M process. Temperature gradient in HPDC is not as high as found in PBF-LB/M processes, and coupled with the high HPDC cooling rates (typically 10K/s), equiaxed grain morphology is favored. In the case of the Kelvin lattice structure as observed in Fig. 8, the microstructure is not significantly different. However, the AM grain morphology is not as columnar as previously observed in the BCC built lattice structure. This can be attributed to differences in the lattice topology of the unit cells between the BCC and Kelvin which introduces local disparity in the direction of heat flux and for that matter temperature gradient and grain morphology.

Mechanical Properties

Microvickers hardness measurement profile (HV0.3) was performed across the 316L AM lattice inserts (BCC and Kelvin – 1.2 mm strut thickness; LF) and the HPDC alloy interface. As shown in Fig. 9, For the AM-built lattice inserts, the recorded microhardness was high and ranged between 201 and 237 HV0.3. Across the interface, the microhardness declined to levels below 100 HV0.3 (60.2–97.8 HV0.3) into the cast material. As expected, the higher microhardness is from indentations from the 316L AM steel lattice inserts (BCC and Kelvin) while the regions of low hardness are from the HPDC aluminium alloy indentations. The sharp decline in microhardness rather than a gradient manner decline across the dissimilar materials at the interface affirms the nature of the bonding mechanism. Because the interfacial bonding is by means of mechanical

bonding other than metallurgical bonding, no fusion or reaction zone at the interface exists, which will have possibly led to a gradual microhardness decline across the dissimilar interfacial materials. As-built 316L microstructures are predominantly single-phase austenitic matrix with typically little or no carbides present, the recorded hardness may therefore be due to the strains from the supersaturated austenitic matrix and the fine grain structure in the as-built state. The EN AC-43500 hardness may also be attributed to the fine grain microstructure and strength contribution from the Si eutectic network which characterizes AC-43500 HPDC manufactured microstructures.

Conclusions

This work was set out to investigate densification of additively manufactured lattice inserts in HPDC molds and the investigation of the interface between the two dissimilar materials; 316L stainless steel lattice structure and the EN AC-43500 HPDC aluminium alloy. The influence of different lattice topology (BCC and Kelvin lattice structures), lattice structure strut thicknesses (0.6 mm and 1.2 mm) as well as flow length (short and long) and intensification pressure (0 bar and 400 bar) on densification was also studied.

The results show that embedding AM-fabricated 316L stainless steel lattice structures as functional inserts in HPDC components is feasible, provided that the process parameters are appropriately controlled. The application of intensification pressure proved to be critical for achieving relatively high densification for both BCC and Kelvin lattice structures. Standard filling conditions without intensification resulted in poor infiltration of the lattice structures. This led to low densification regardless of the lattice topology, lattice strut thickness or flow length. With intensification, long flow length had a comparatively higher densification to short flow length. This might be due to a very good dynamic transition from standard filling condition to intensification where possibly the surrounding cavity were already filled ahead of intensification which enabled the pressure to act more uniformly and flow through the insert. Close proximity of the lattice insert to the gate could have led to poor densification arising from the liquid melt's impingement on the lattice structure rather than a uniform flow through the lattice structure for short flow length. Microstructure observation at the interface revealed a continuous, defect-free interface between the steel insert and the aluminum matrix. The interfacial bonding is attributed to mechanical interlocking rather than a fused zone by metallurgical bonding. The rapid solidification inherent to HPDC effectively suppressed the formation of thick Fe-rich intermetallic layers, thereby preserving interfacial integrity while enabling strong load transfer between the dissimilar materials.

Declarations

Conflict of interest:

The authors declare no conflict of interest

Funding:

Author Contribution

F. A-K. - wrote the main manuscript , Experiments - (Printed AM lattice structures), analysis and preparation of figures
S. M.P.R. - Performed HPDC experiments
L.B - Performed HPDC experiments, Review and Supervision
U.V - Review and Supervision, Overall Project Coordinator

Acknowledgement

The authors thank the Deutsche Forschungsgemeinschaft (DFG, German Research Foundation) [Germany’ s Excellence Strategy-EXC-2023 Internet of Production-390621612] for funding this work

Table 1

Table 1: The composition of the cast aluminum alloy AlSi10MnMg and the 316L steel alloy

Chemical Composition – EN AC-43500 (wt.%)											
AlSi10MnMg	Al	Si	Cu	Mg	Mn	Fe	Ti	Sr			
wt.%	Bal.	10.3	<0.02	0.39	0.61	0.15	0.15	0.016			
Chemical Composition – 316L (wt. %)											
316L	Fe	Cr	Ni	Mo	Mn	Si	Ni	P	S	C	N
wt.%	Bal.	16-18	10-14	2-3	2.0	1.0	10-14	0.045	0.30	0.030	0.10

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Figures

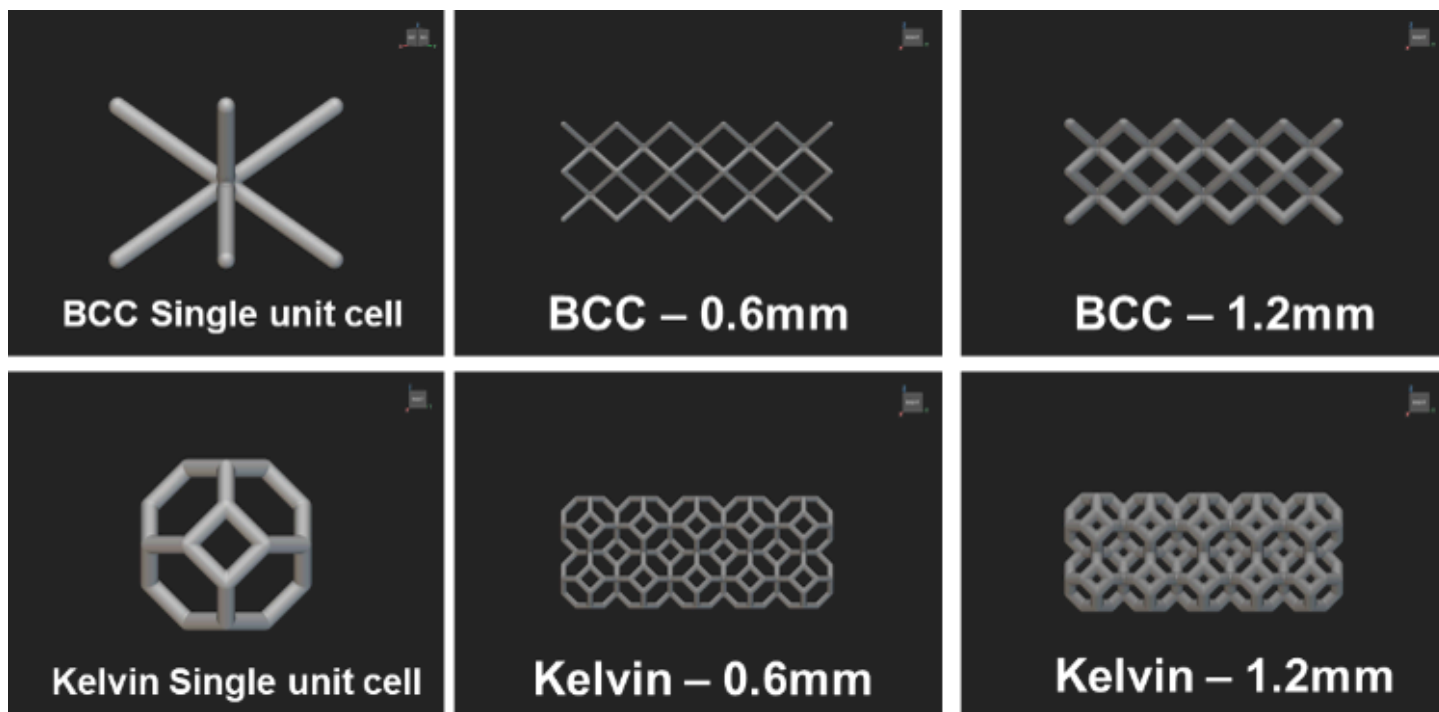


Figure 1

Shows images of the designed lattice structures: BCC and Kelvin of varying strut sizes, 0.6 and 1.2 mm and their orthogonal view.

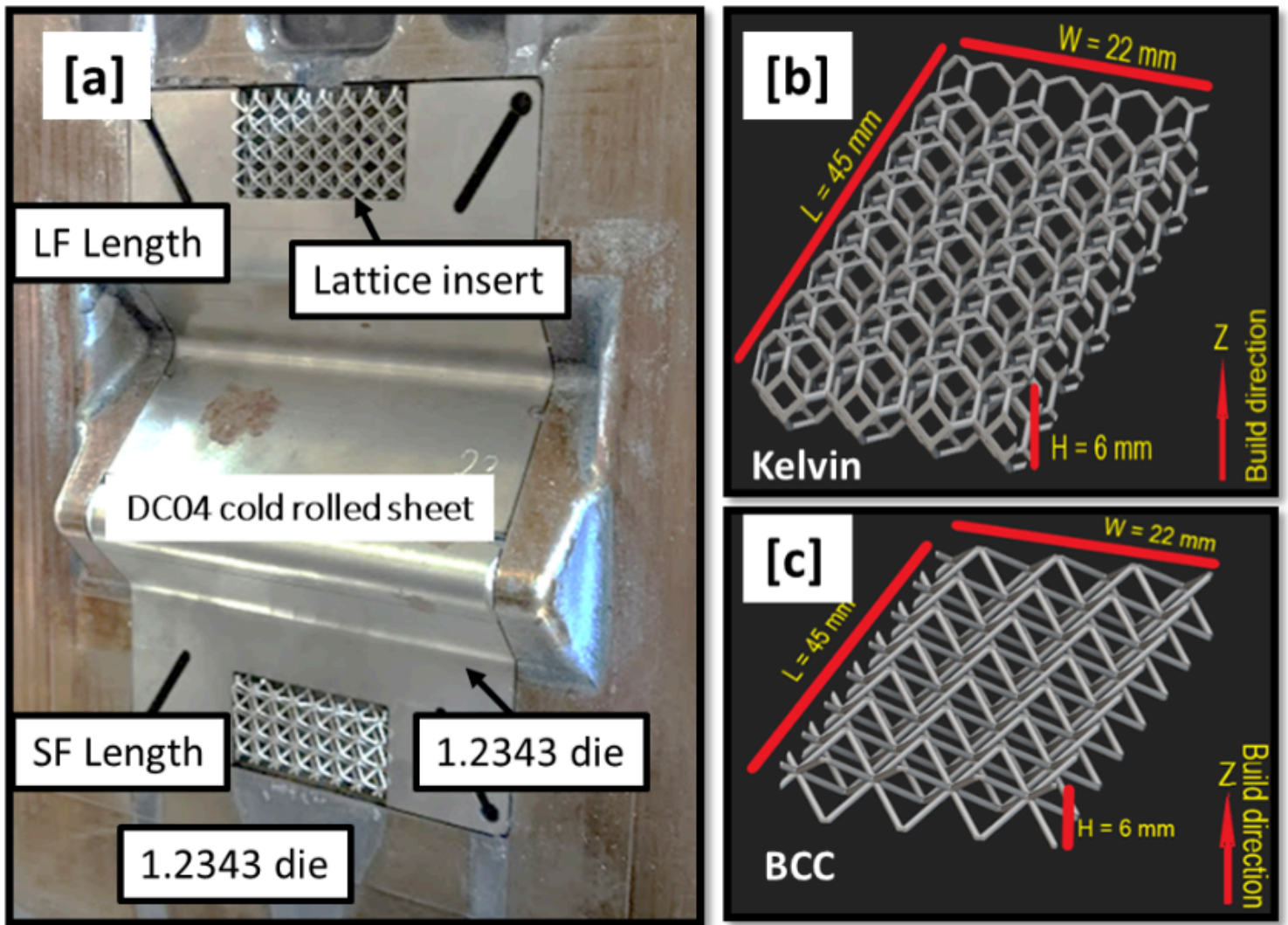


Figure 2

showing (a) inserts positioned at short flow (SF) and long flow length (LF) and fitted in a DC04 cold rolled sheet (b) Schematic of diagram of Kelvin lattice structure (c) Schematic of diagram of BCC lattice structure

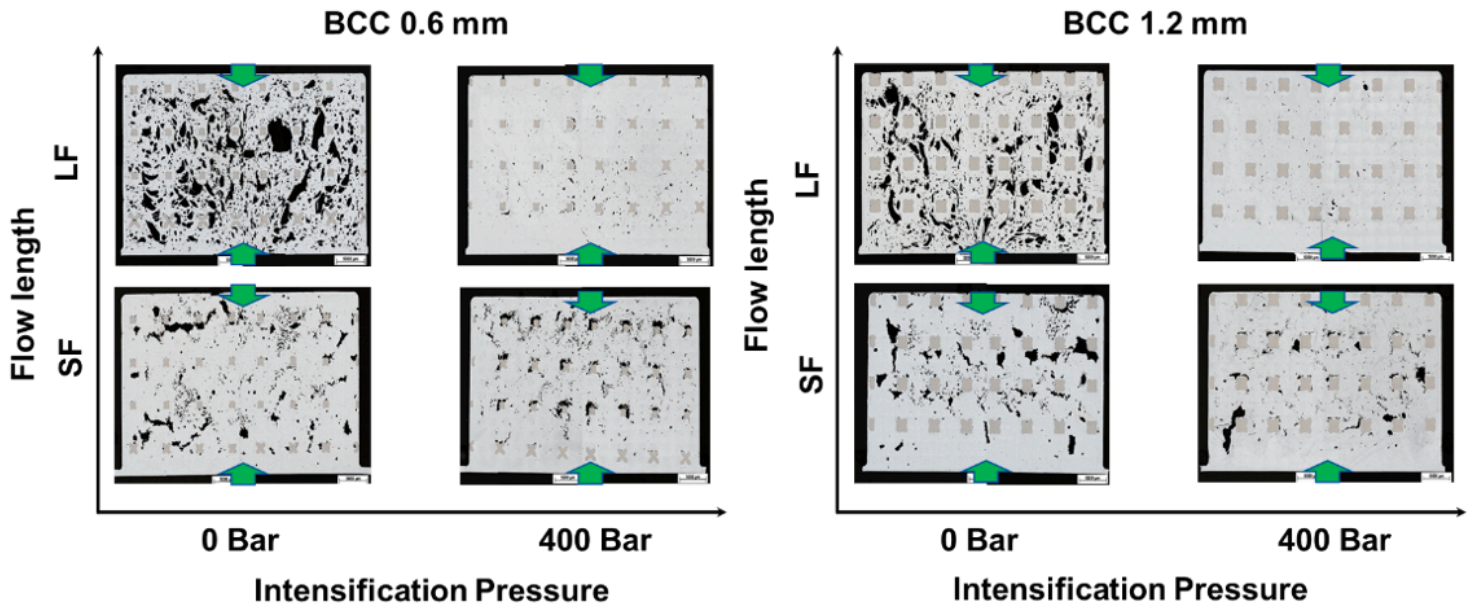


Figure 3

LOM micrographs of BCC lattice structures (0.6 and 1.2 mm strut thickness) with standard filling conditions (0 bar) and intensification pressure (400 bar) for Short Flow Length (SF) and Long Flow Length (LF)

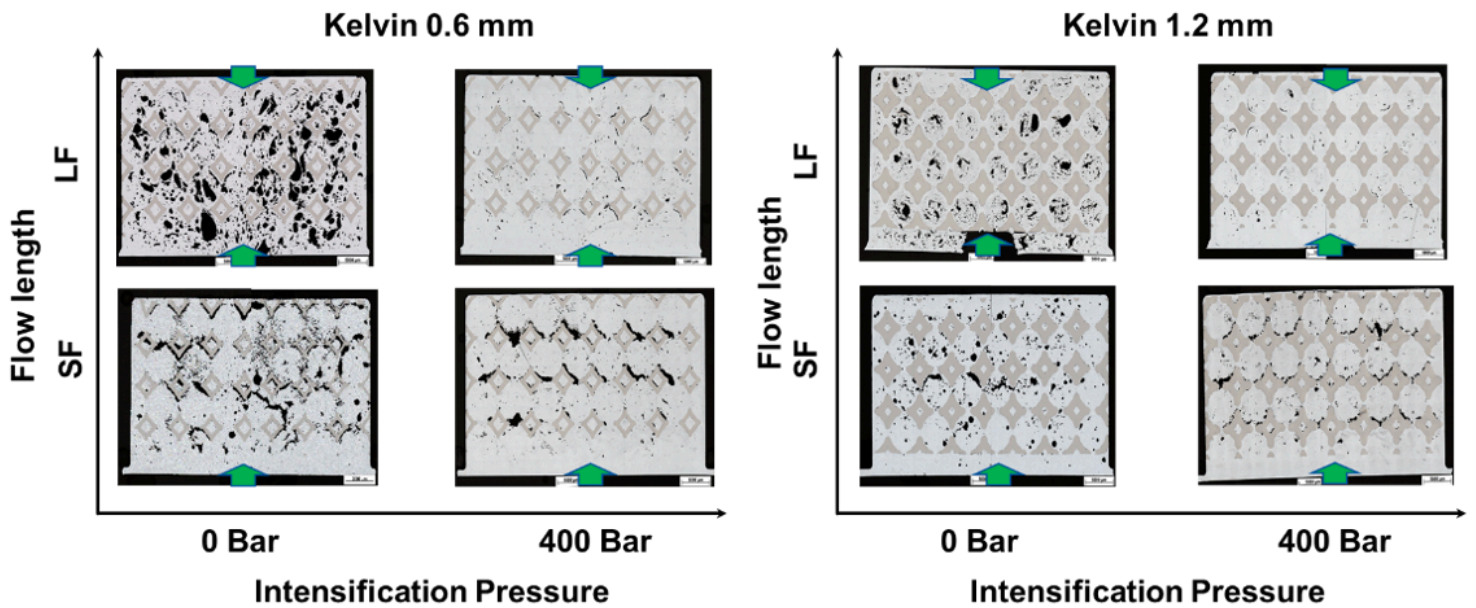


Figure 4

LOM micrographs of Kelvin lattice structures (0.6 and 1.2 mm strut thickness) with standard filling conditions (0 bar) and intensification pressure (400 bar) for Short Flow Length (SF) and Long Flow Length (LF)

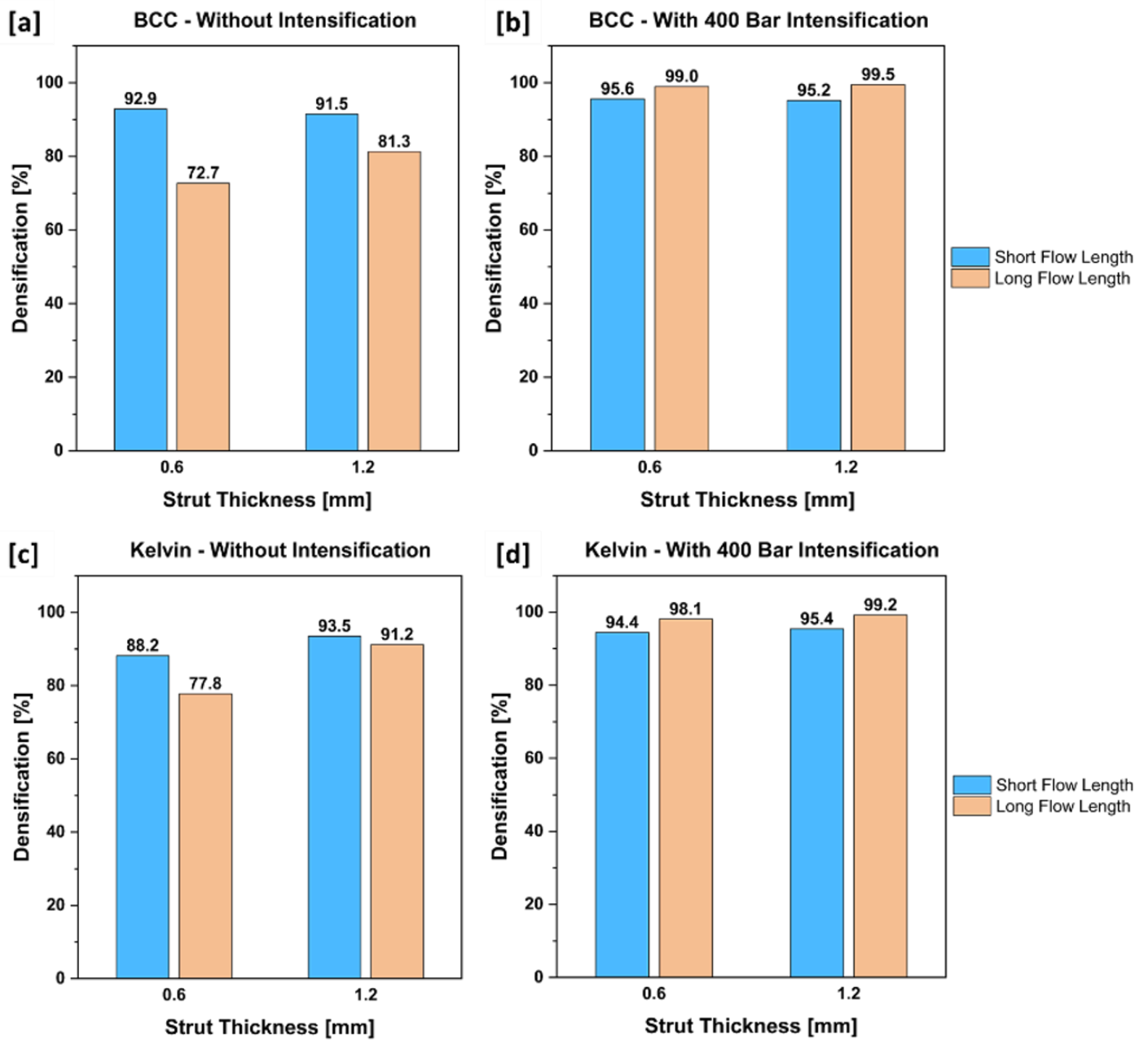
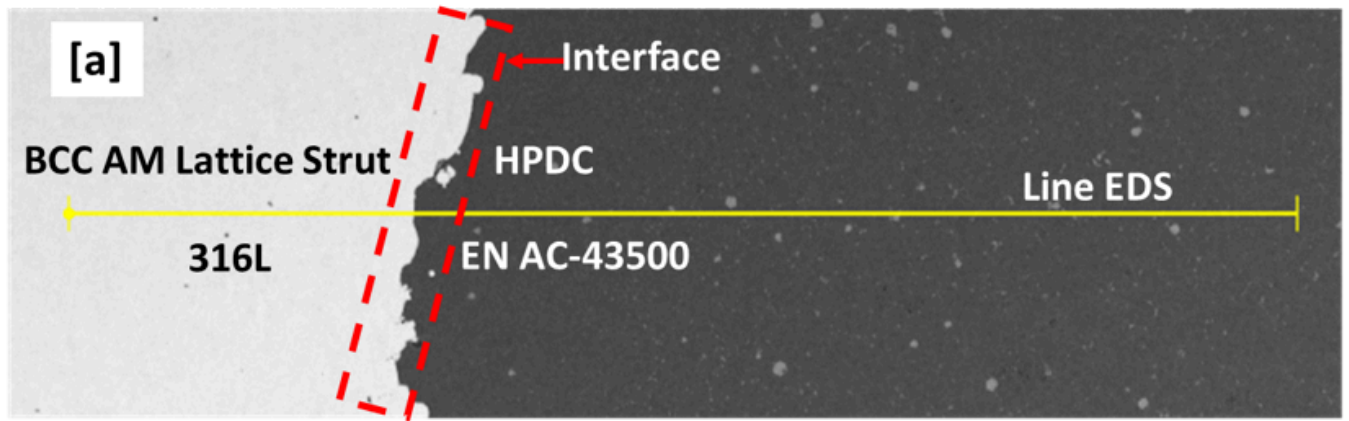


Figure 5

a-d: Lattice topology and strut thickness and intensification on densification



[b]

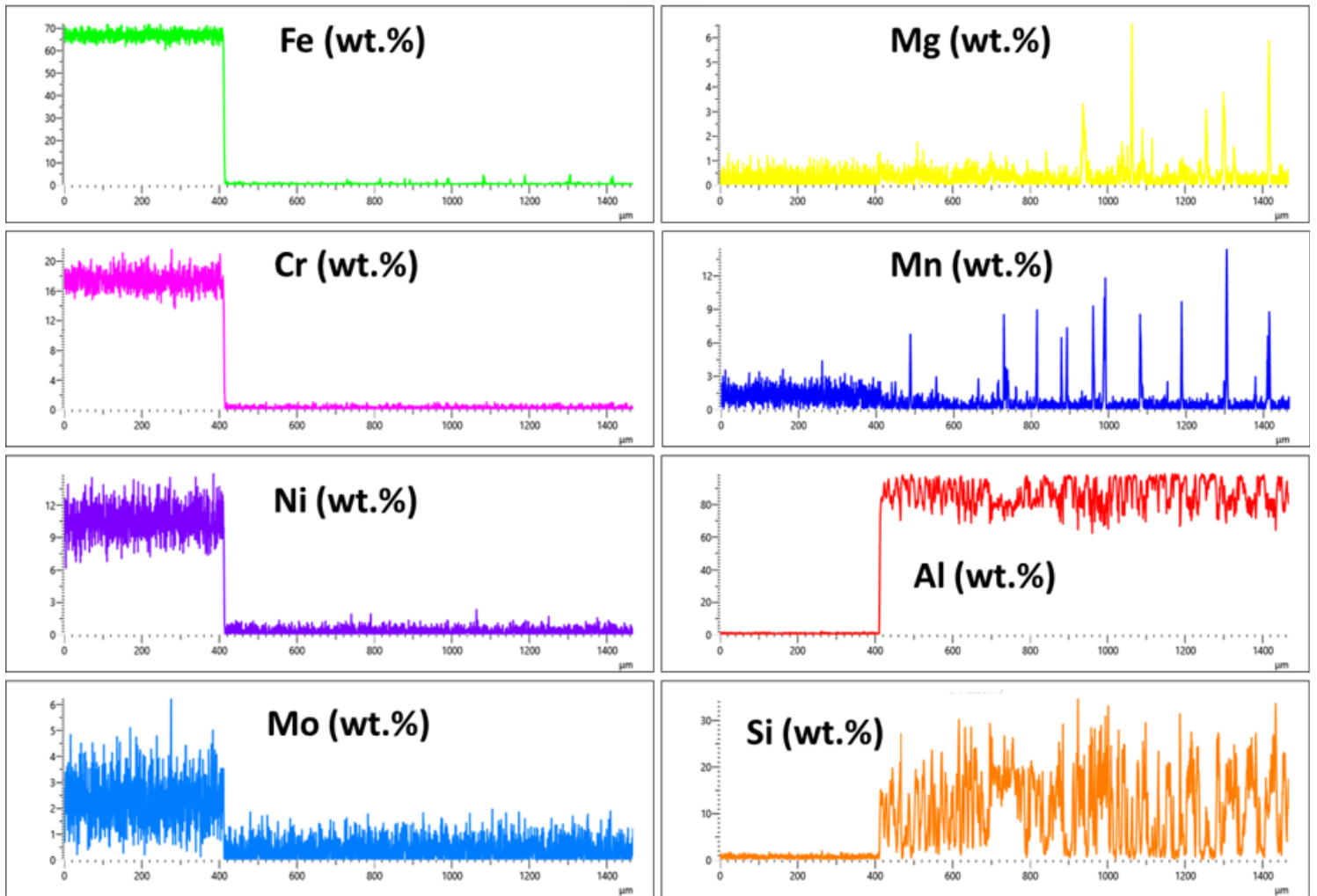


Figure 6

(a) SEM micrograph showing good interfacial bonding between the 316L BCC lattice with 1.2 mm strut and the AC-43500 HPDC aluminium alloy (Long flow length) (b) EDS line mapping across the interface.

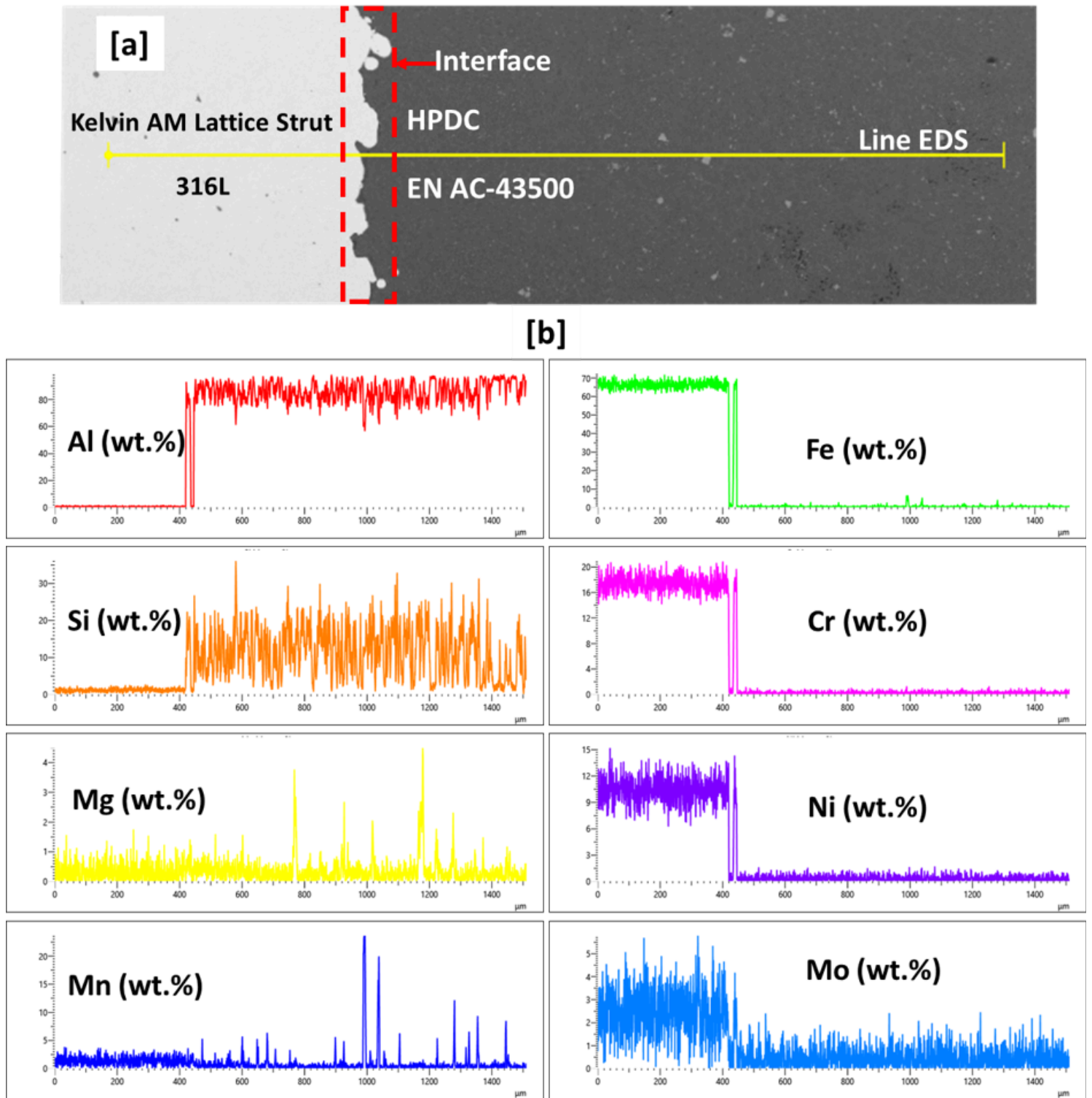
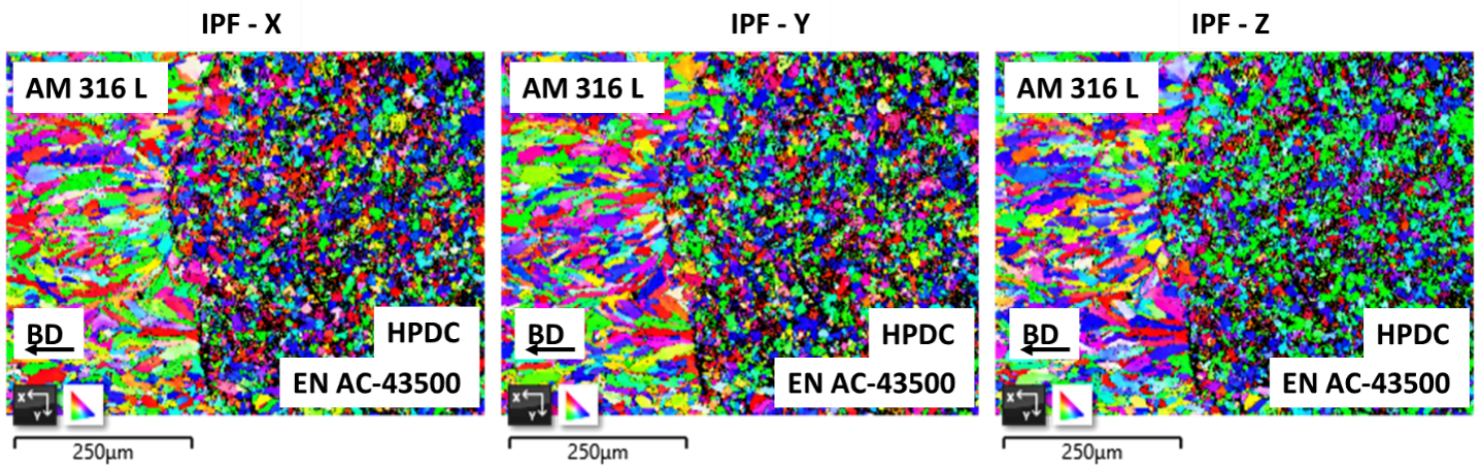


Figure 7

(a) SEM micrograph showing good interfacial bonding between the 316L Kelvin lattice with 1.2 mm strut and the AC-43500 HPDC aluminium alloy (Long flow length) (b) EDS line mapping across the interface.

BCC – 1.2 mm (LF)



Kelvin – 1.2 mm (LF)

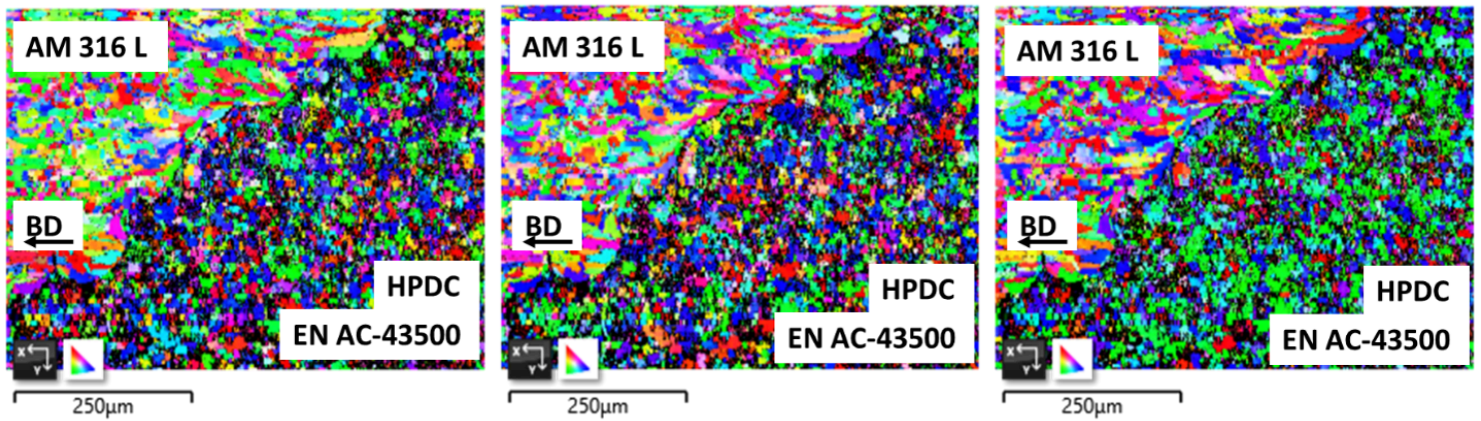


Figure 8

(a) EBSD – IPF X, Y and Z Maps showing grain morphology at the interfacial boundary between the additively manufactured lattice (316L BCC and Kelvin lattice with 1.2 mm strut thickness) and the EN AC-43500 HPDC aluminium alloy.

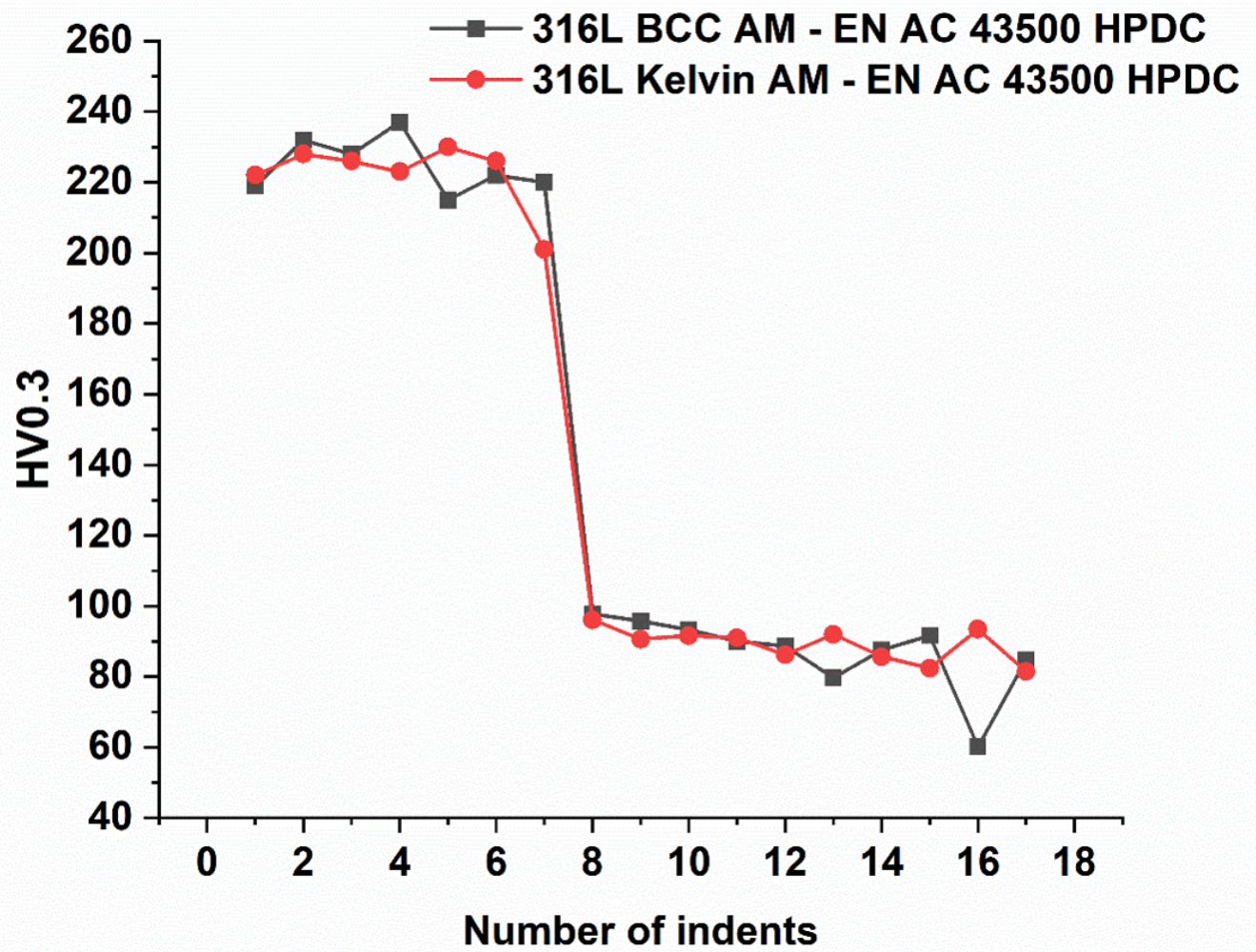


Figure 9

Hardness test across the interface of both 316L BCC AM - AC 43500 HPDC and 316L Kelvin AM – EN AC 43500 HPDC