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Modification of Electrical and Mechanical Properties of Selective Laser-Melted CuCr0.3 Alloy Using Carbon Nanoparticles

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Abstract

The current work investigates the influence of carbon nanoparticle addition on the selective laser melting behavior and resultant properties of gas atomized CuCr0.3 powder. The carbon addition does not affect the powder flowability, nor its packing density, but it significantly enhances the optical absorption for infra-red radiation. Despite the improved optical absorption of the powder bed after carbon addition, a comparable SLM behavior is observed for both virgin and carbon-mixed-CuCr0.3 powders. Furthermore, the addition of carbon nanoparticles facilitates in situ de-oxidation of copper, and chromium-oxides present in the CuCr0.3 powder. Accordingly, a larger fraction of chromium is obtained in its metallic form in the as-built carbon-mixed-CuCr0.3 part. The presence of metallic chromium allows tailoring of the mechanical and electrical properties by post-heat-treatments. The best combination of mechanical (tensile strength of 298 ± 19 MPa, yield strength of 208 ± 18 MPa, ductility of $21 \pm 4\%$) and electrical ($87 \pm 1\%$ IACS) properties are realized for the carbon-mixed-CuCr0.3 alloy after a direct-age hardening heat treatment. The improvement in the mechanical properties could be attributed to the precipitation of nanometer-sized, coherent, metallic chromium precipitates in the copper matrix.

1. Introduction

Selective Laser Melting (SLM) is a powder-based additive manufacturing (AM) process, which uses a high-density laser source to selectively melt and join successive powder layers.^{[1][2]} As SLM is a layer-by-layer powder-bed manufacturing process, it allows the

manufacturing of complex 3D parts with intricate features from a variety of potential metal powders.^{[3][4]}

Pure copper is a soft metal with a very high thermal ($400 \text{ W/(m}\cdot\text{K)}$), and electrical conductivity (IACS 100%).^[5] Therefore, it is used mainly for thermal and electrical transport applications. In its pure form, copper exhibits a low mechanical strength; hence, its use is limited in functional parts. The mechanical strength of copper can be enhanced by alloying with other elements such as nickel, tin, chromium, phosphorus, etc. However, alloying, especially with the alloying elements that form a solid-solution with copper such as nickel and phosphorus,^{[6][7]} drastically reduces the thermal and electrical conductivity. Therefore, to achieve a good combination of mechanical strength and electrical conductivity, precipitation hardenable copper alloys such as copper-chromium, are preferred.

Combining the ability to create geometrically complex metal parts by SLM, and the processability of high strength, high conductivity copper-based alloys, the applicability of SLM for thermal management applications could be broadened.^[8] In line with this approach, Neugebauer et al.^[9] demonstrated improved heat transfer efficiency for an air-water heat exchanger fabricated by SLM. Moreover, Moratalla Martinez et al.^[10] validated the improved performance of a magnetic inductor fabricated using a powder bed fusion additive manufacturing (PBF-AM) technique. Despite the potential, the SLM fabrication of the crack-free and dense parts, with a good combination of mechanical and electrical/thermal properties, starting from pure and/or low-alloyed copper powders, is extremely challenging. This is primarily due to (i) the very high copper laser reflectivity at the wavelengths (1060-1080 nm) of the most commercially used fiber lasers, and (ii) the high copper thermal conductivity, making that any generated heat is quickly dissipated leading to insufficient melting of the copper powder bed, especially when using the commercially available fiber lasers with a maximum power output of 400 W and a minimum beam diameter of $50 \mu\text{m}$.^{[11][12]}

Few researchers^{[10]–[13]} have overcome the issue of insufficient melting by employing a high power 1 kW fiber laser and produced dense and crack-free parts. However, Jadhav et al.^[13] reported the damage to the laser optics caused by the copper back-reflections. Heussen and Meiners^[14] used a green laser (which exhibits lower copper reflectivity at a wavelength of 532 nm) as a sustainable way to manufacture pure copper and low-alloyed copper parts. However, the electrical and mechanical properties of the copper parts manufactured using the green laser have not been reported. In another approach, the optical absorption of copper can slightly be increased by oxidizing or pre-alloying copper with a small amount of alloying elements such as chromium, zirconium, etc. Accordingly, the selective laser melting of oxidized Cu,^[15] Cu-Cr,^{[16]–[19]} and Cu-Cr-Zr^{[20]–[23]} alloys has been carried out successfully to fabricate dense parts. Recently, the use of a carbonaceous coating covering a metal powder^[24] and the addition of carbon nanoparticles to copper powder^[7] have been proposed to significantly increase the optical absorption of the highly reflective metallic powders at room temperature. Although the addition of carbon nanoparticles facilitated the fabrication of dense copper parts, segregation of carbon and phosphorus along the grain boundaries of the epitaxially grown grains resulted in poor mechanical properties. Furthermore, the authors postulated that the carbon segregation could be minimized by alloying copper with carbide forming elements such as chromium, titanium, and the mechanical properties could be improved.

Accordingly, this research aims to evaluate the effect of carbon addition on powder properties, SLM processability, electrical, and mechanical properties of the CuCr0.3 alloy. Therefore, the influence of carbon addition on different powder characteristics such as flowability and packing behavior, as well as the optical absorption, is evaluated. Subsequently, the influence of carbon addition on the SLM behavior of pre-alloyed CuCr0.3 is assessed and compared to that of the virgin carbon-free powder. Finally, the electrical and mechanical properties of the dense parts are determined in the as-built and optimally post-heat-treated conditions and linked to the microstructural development.

2. Materials and Methods

2.1. Starting Powders

Nitrogen gas atomized pre-alloyed CuCr0.3 powder (D50: 33.7 μm , Sandvik Osprey Ltd., UK) and nanometer-sized carbon black powder (D50: 25 nm, Printex® U, Dakota Coatings NV, Belgium) were used as the starting powders. Pre-alloyed CuCr0.3 powder was mechanically mixed with 0.05 wt.% of carbon powder inside a polyethylene bottle sealed with argon for 12h using a Turbula type multidirectional mechanical mixer. The amount of carbon was fixed at 0.05 wt.% because of three reasons: ‘(i) to reach the powder optical absorption $\geq 55\%$, which is generally comparable to the commercial laser processable alloys such as Ti-6Al-4V, and AlSi10Mg^{[25]–[27]} (the comparable optical absorption of the carbon-mixed-CuCr0.3 alloy powder to the commercial laser processable alloys was aimed to attain a safe SLM process from the laser back-reflection point of view), (ii) to have sufficient carbon for the de-oxidation reaction forming gaseous CO/CO₂ during the SLM process, and (iii) for the in situ reaction between carbon and chromium forming chromium carbide during the post-heat-treatments. It is important to note that, during SLM of carbon-mixed-copper powder,^[7] the reduction of oxides by carbon nanoparticles has been validated. Although this was logically realized by the formation of CO/CO₂ gases, no specific gas porosity was detected in the SLM parts. This implies that the potential CO/CO₂ gases diffuse out of the matrix and leave along with the inert gas flow moving above the powder bed. The pre-alloyed CuCr0.3 powder particles had an overall spherical shape, although few irregularly shaped particles along with some smaller satellite particles were observed, as shown in **Figure 1(a)**. A backscattered electron (BSE) image of a single powder particle is shown in **Figure 1(b)**, indicating the presence of a phase with dark contrast at the powder surface. EDS analysis revealed that these zones were rich in chromium and oxygen, indicating that chromium was at least partially present as chromium oxide at the powder surface. A carbon mixed CuCr0.3 powder particle is shown in **Figures 1(c-d)**, indicating a uniform distribution of the nanometer-sized carbon particles on CuCr0.3 alloy powder particle surface. The powder optical absorption was measured by a Lambda 950 spectrophotometer from PerkinElmer and a reference material, ‘Spectralon®’ was used with 100% reflectance within the measured wavelength range. All the optical absorption measurements were carried for wavelengths in the range of 400 nm to 1400 nm, with an interval of 2 nm, and in the reflectance mode. The optical absorption (%) was calculated using the formula ‘optical absorption (%) = 100 – reflection (%)’. Powder flowability was measured using a standard Hall flow test according to the ASTM B213 norm and using a dynamic angle of repose (AoR) test.^[28] A powder with an average angle below 40° is defined as good flowable, in the dynamic AoR technique. The apparent and tapped density of the powder was measured

as per the ASTM B212 and ASTM B527 norms, respectively. The amount of carbon and oxygen in the powder and the parts was measured using coulometry (Coulomat 702 SO/CS) and the inert gas fusion technique by LECO, respectively. The chemical composition of the powder was measured using an inductively coupled plasma optical emission spectroscopy (ICP-OES, Varian 720 ES) and is presented in **Table 1**. It can be noted that the virgin CuCr0.3 alloy powder exhibited an undesired high oxygen content of 0.14 wt.%. This was an indication that the virgin CuCr0.3 alloy powder was oxidized before performing the SLM process.

Figure 1: Scanning electron microscopy images of the CuCr0.3 alloy powder, (a-b) before, and (c-d) after, mixing with the carbon nanoparticles. The white arrows indicate the presence of uniformly distributed carbon nanoparticles (bright contrast SE images) on the surface of a micrometer sized CuCr0.3 particle.

Table 1: Chemical composition of the CuCr0.3 alloy powder.

Element	Cr	Fe	O	Cu
Amount (wt.%)	0.26	0.018	0.14	Balance

2.2. Selective Laser Melting (SLM)

Selective laser melting (SLM) was carried out for both virgin and carbon-mixed-CuCr0.3 powders on a copper base plate, inside an in-house built SLM machine^{[2][29]}. The SLM machine was upgraded with a continuous single-mode 1 kW, 1080 nm fiber laser having a beam diameter of 40 μm . The SLM process was performed in a protective argon atmosphere. The Magics software from Materialise, Belgium, was used to prepare the STL files. All the layers were scanned once in a bidirectional scanning mode, while a scan rotation of 90° was imposed between the successive scanning layers.

Initially, (10×10×7) mm³ cube-shaped parts were manufactured using various laser scan settings. During the parameter optimization, three different laser power values (P) of 350, 450, and 600 W were applied, in combination with six laser scan speeds (v) such as 300, 400, 600, 800, 1000, and 1200 mm/s. However, the layer thickness (t) and the hatch spacing (h) were kept constant at 0.03 mm and 0.09 mm, respectively. After part fabrication, the densities were measured using the Archimedes method in ethanol fluid. These were divided by the true densities of the virgin (8.91 g/cm³) and the carbon-mixed-CuCr0.3 alloy (8.80 g/cm³) powders, as measured using a Multipycnometer™ device, in order to calculate the relative densities. The laser volumetric energy density (E in J/mm³) was calculated using **Equation 1**.

$$E = \frac{P}{v \times h \times t} \quad (1)$$

Where P is laser power (W), h is hatch spacing (mm), v is scanning speed (mm/s), and t is powder layer thickness (mm). The optimum laser scan settings for both virgin and carbon-mixed-CuCr0.3 powders are summarized in **Table 2**. In both cases, a laser power of 600 W offered the highest part density. Consequently, these parameters were selected to manufacture more parts (fabricated horizontally) for tensile testing, electrical conductivity measurements, and microstructural analysis before and after post-heat treatments. Moreover, single-layer experiments were carried out to evaluate the relationship between melt pool depth and the laser scanning speed. Accordingly, a laser output power of 600 W, hatch spacing of 0.09 mm, and various laser scan speeds, ranging between 300 and 1000 mm/s, were combined to irradiate a 0.03 mm thick powder layer, which was deposited on a 2 mm rolled copper sheet. After the single layer experiments, the cross-sectional surface of the single layers was analyzed by light optical microscopy (3D Digital Microscope, VHX-6000, Keyence), and melt pool depths were determined as a function of laser scan speed.

Table 2: Optimized SLM parameters producing dense parts from both virgin and carbon-mixed-CuCr0.3 powders, when a fixed hatch spacing of 0.09 mm and a layer thickness of 0.03 mm was employed.

Powder	Laser Power (W)	Scan Speed (mm/s)
CuCr0.3	600	400
CuCr0.3+C0.05	600	300

2.3. Post-Heat-Treatments

The post-heat-treatments (HT) were performed inside a tube furnace under a protective argon atmosphere. After holding at elevated temperature, the parts were quenched in water.

The water quenching of copper parts resulted in the formation of a very thin surface oxide film. This thin surface oxide film was easily removed using an abrasive paper with a grit size of 800 before the subsequent aging treatment, and electrical conductivity measurements.

Two types of heat treatments were performed, as shown in **Table 3**:

- (i) Solution Annealing followed by Age Hardening (SA+AH)
- (ii) Direct-Age Hardening (D-AH)

When the SA+AH method was applied, the solution annealing step (SA) was performed to dissolve and homogeneously distribute the metallic chromium particles in the copper matrix. This was realized by holding the as-built part at a high temperature of 1015°C for 1.5h and subsequent quenching in water. As such, a saturated Cu-Cr solution is assumed to form during the SA step, and Cr precipitates should form during the subsequent age hardening (AH) step carried out conventionally at 400°C. In contrast, during the Direct-Age Hardening (D-AH) treatment, it is assumed that all of the metallic Cr is already present as a substitutional solid-solution alloying element in the supersaturated copper matrix, due to the very high SLM cooling rates. While the heat treatment conditions used for the SA+AH method were selected based on the equilibrium copper-chromium phase diagram, the D-AH treatment was performed at 500°C to achieve the best combination of mechanical and electrical properties.^{[17][30]} To explore the optimum heat treatment conditions, the hardness and electrical conductivity values were recorded at different time intervals until the electrical conductivity was saturated or until the hardness decreased. Finally, the post-heat-treatment condition offering the best combination of hardness and electrical conductivity was selected to produce parts for tensile testing. **Table 3** provides an overview of all heat treatment conditions and the optimum heat-treatment condition that were chosen for both SA+AH and D-AH treatments.

Table 3: Overview of the optimum heat treatment (HT) conditions for the SA+AH and D-AH treatments.

Heat treatment (HT)	Temperature (°C)	Dwelling Time (h)	Cooling type	Optimum HT time (h)
Solution Annealing (SA)	1015	1.5	Water quenching	NA
*Age Hardening after Solution Annealing (SA+AH)	400	(0.25 to 12)	Water quenching	3
Direct-Age Hardening (D-AH)	500	0.5 to 2.5	Water quenching	1.5

*SA+AH is a two-step heat treatment process, wherein the SA (first step) was carried out at 1015°C for a fixed time of 1.5h, and the AH (second step) was performed at 400°C for various time periods (from 0.25 to 12h).

2.4. Microstructural, Electrical Conductivity, and Mechanical Property Characterization

For the microstructural analysis and hardness measurements, the cube-shaped parts, embedded in Bakelite resin were ground using 320-4000 grit size grinding papers. Afterward, the embedded parts were polished using a 3 μm diamond slurry and etched with a solution comprised of iron(III) chloride (3.5 g), hydrochloric acid (2.5 ml) and Ethanol (75 ml) for 20 s. The microstructure of the parts was analyzed using light optical microscopy (3D Digital Microscope, VHX-6000, Keyence) and scanning electron microscopy (XL30 FEG, Philips, and Nanosem 450, FEI) having energy dispersive spectroscopy (EDS) detectors. Chemical composition of the precipitates and the matrix phase was measured using EDS, and an average of minimum of eight measurements is reported for the precipitates, after subtracting the corresponding chemical composition of the copper matrix. Hardness measurements were carried out using the Vickers micro-hardness indentation technique applying a load of 0.3 kg for 15 s. The bars for electrical conductivity measurements (60 mm $\times$ 5 mm $\times$ 2 mm) were surface ground on an abrasive paper with grit size 800 to remove the surface oxides, prior to the measurements. The electrical conductivity measurements were performed by a four-point method using the Resistomat® model 2301-V001, (Burster, Gernsbach, Germany). The electrical conductivity values are reported in % international annealed copper standards (% IACS), wherein the 100% IACS refers to the conductivity of pure copper in the annealed state i.e., 58×10^6 S/m. The thermal conductivity values were theoretically calculated using the Wiedemann- Franz law ^[31] and the Lorenz number was considered as $L=2.41 \times 10^{-8}$ W. Ω /K².
[5]

Tensile tests were performed on dog-bone shaped samples having dimensions of 60 mm $\times$ 4 mm $\times$ 2 mm (total length $\times$ width $\times$ thickness) using an Instron 4467 machine. An extensometer with a gauge length of 12.5 mm was used for the strain measurements. The tensile specimens were loaded perpendicular to the building direction (BD) during tensile testing. The mechanical properties such as yield strength (YS: at 0.2% strain), Ultimate tensile strength (UTS), and elongation at break (% EB) are reported for the as-built and optimum post-heat-treated conditions.

3. Results

3.1. Powder Characteristics

Figure 2(a) presents the optical absorption of the virgin and carbon-mixed-CuCr0.3 alloy powders along with a reference oxygen-free pure copper powder, for the wavelength range of 400-1400 nm at room temperature. The reference oxygen-free copper powder displays the lowest optical absorption of 26% at the wavelengths of the fiber lasers, generally between 1060-1080 nm (near-infrared region). Both CuCr0.3 alloy powders show higher optical absorption compared to the oxygen-free pure copper powder. The carbon-mixed-CuCr0.3 alloy powder exhibits the highest optical absorption of 56% at the fiber laser wavelength. In the visible region (400-600 nm), a sharp increase in the optical absorption is observed for all copper

alloys due to the shift of the optical absorption mechanism from intraband to interband electron transition.^[32] **Figure 2(b)** displays the average and avalanche angle for both virgin and carbon-mixed-CuCr0.3 alloy powders. Despite the slight increase in the average and avalanche angle for the carbon-mixed powder compared to the virgin CuCr0.3 alloy powder, both angles are well below the threshold value of 40°, indicative for their good flowability. **Table 4** summarizes the powder properties such as Hall flow time and apparent, and tapped density for both virgin and carbon-mixed-CuCr0.3 alloy powders. The addition of carbon slightly reduces the flowability and apparent, and tapped density. Nevertheless, the carbon-mixed-CuCr0.3 powder is still flowable and packs with an acceptable apparent and tapped density.

Figure 2: (a) The optical absorption for the virgin and carbon-mixed-CuCr0.3 alloy powders in comparison with the reference oxygen-free copper powder. A considerable increase in the optical absorption is observed for both CuCr0.3 alloy powders at the fiber laser wavelength. (b) The average and avalanche angles for the virgin and carbon-mixed-CuCr0.3 alloy powders measured during the dynamic angle of repose test. Since the measured average and avalanche angles are below 40°, both powders exhibit good flowability.

Table 4: The Hall flow time, apparent, and tapped density of the virgin and carbon-mixed-CuCr0.3 alloy powders.

Property	CuCr0.3	CuCr0.3+C0.05
Hall flow time (s)	13.6	16.1
Apparent density (%)	51	47
Tapped density (%)	59	53

3.2. SLM Parameter Optimization and Top Surface Observation

Figures 3(a) and 3(b) show the effect of the volumetric laser energy density (E) on the relative part density for the virgin and carbon-mixed-CuCr0.3 alloys, respectively. Dissimilar markers are used for the different laser powers, and the ranges in the applied laser scan speed are denoted using dashed blue rectangles for 300-400 mm/s, and dashed red ellipse for 600-1200 mm/s.

Both figures indicate a similar part density trend with respect to the applied laser energy density. Hence, no significant change in the SLM processability was observed between the

virgin and carbon-mixed-CuCr0.3 alloy powders. In general, the part density increases with an increase in the applied volumetric laser energy density (E). Dense parts were mainly obtained for laser energy densities above 300 J/mm³. However, it is important to note that all the dense parts (> 98.5%) were processed at the lower laser scan speeds (300 mm/s, and 400 mm/s), regardless of the employed laser power (350W, 450W, and 600W). Furthermore, the part produced from the CuCr0.3 alloy using a laser energy density of 370 J/mm³ (P of 600 W, v of 600 mm/s) shows substantially lower part density of 92.3% compared to the 98.6% dense part produced when a laser energy density of 324 J/mm³ (P of 350 W, v of 400 mm/s) is applied. The same trend is also seen for the carbon-mixed-CuCr0.3 alloy. This demonstrates a clear interrelationship between the part density and laser scan speed within the employed laser power range.

Figure 3: The influence of the laser volumetric energy density on the relative density of the parts for (a) CuCr0.3, and (b) CuCr0.3+C0.05 alloys. Data sets encircled using the blue dashed rectangle show high relative part densities obtained for the low scan speeds (300 mm/s, and 400 mm/s), whereas the red dashed ellipse shows lower part densities realized for the parts processed at high laser scan speeds, above 600 mm/s.

Figures 4(a) and 4(b) show top surfaces of the dense SLM parts for the virgin and carbon-mixed-CuCr0.3 alloys, respectively. Both top surfaces show stable melt tracks with an overlap of ~40% between neighboring melt tracks without any surface porosities. This indicates that the employed hatch spacing of 0.09 mm was optimum. The melt pool borders, as indicated by the dashed white curves, are mirrored over the melt pool fusion boundary. This is associated with the bi-directional (zig-zag) scanning strategy that was employed during the SLM process. Moreover, carbon segregation, as observed on the top surface of the carbon-mixed-CuCr0.3 parts (**Figure 4(b)**), occurs either at the melt pool center or between two adjacent melt pools, as indicated in the inset of **Figure 4(b)**.

Figure 4: Top surfaces of (a) CuCr0.3, and (b) carbon-mixed-CuCr0.3 alloy parts. Both parts show stable melt tracks (melt pools and scanning directions are indicated using white dashed curves and arrows, respectively) with no surface porosities. Carbon segregation is visible at the melt pool center and between the two adjacent melt tracks, as shown in the inset of Figure 4(b).

3.3. Heat Treatment Optimization

Two types of heat treatments (SA+AH and D-AH) were performed on both virgin and carbon-mixed-CuCr0.3 alloy parts. The optimum heat treatment conditions facilitating the best combination of hardness and electrical conductivity were determined.

3.3.1. Solution Annealing + Age Hardening (SA+AH)

Figure 5 shows the evolution of the Vickers micro-hardness and electrical conductivity values for the virgin (blue color) and carbon-mixed-CuCr0.3 (orange color) alloys in response to the solution annealing and age hardening (SA+AH) heat treatment.

In the as-built condition, the carbon-mixed-CuCr0.3 alloy shows higher electrical conductivity compared to the virgin CuCr0.3 alloy part, although the hardness values were comparable. Upon solution annealing (SA) treatment, an increase in the hardness, and a drastic reduction in the electrical conductivity is observed for the carbon-mixed-CuCr0.3 alloy. However, the opposite trend is observed for the virgin CuCr0.3 alloy. Furthermore, the age hardening of the carbon-mixed-CuCr0.3 alloy leads to an increase in the electrical conductivity due to precipitation, while this does not notably influence the electrical conductivity of the virgin CuCr0.3 alloy.

Nevertheless, the best combination of hardness and electrical conductivity for both alloys is obtained after age hardening for 3h, as indicated by a green dashed rectangle. Therefore, the tensile properties are investigated after applying this heat treatment condition.

Figure 5: The evolution of hardness (dashed lines) and electrical conductivity (solid lines) for the virgin (blue color) and carbon-mixed-CuCr0.3 (orange color) alloys during the solution annealing and the age hardening (SA+AH) heat treatment. The green dashed rectangular box indicates the selected optimum heat treatment condition for the evaluation of tensile properties.

3.3.2. Direct-Age Hardening (D-AH)

Figure 6 shows the evolution of hardness and electrical conductivity for the virgin (blue color) and carbon-mixed-CuCr0.3 (orange color) alloys as a result of the direct-age hardening heat treatment (D-AH).

As seen, the electrical conductivity values of both alloys increase upon the age-hardening treatment. However, a sharp increase in the hardness for the carbon-mixed-CuCr0.3 alloy and stable hardness trend for the virgin CuCr0.3 is observed. Overall, the carbon-mixed-CuCr0.3 alloy shows higher hardness and electrical conductivity in comparison with the virgin CuCr0.3 alloy when the same heat treatment conditions are applied. The optimum heat treatment condition yielding the best combination of hardness, and electrical conductivity is chosen as 1.5h at 500°C. This was set for the evaluation of tensile properties, as indicated in **Figure 6**, using a green dashed rectangle. Moreover, it could be seen that the D-AH parts demonstrate higher peak hardness values compared to the SA+AH parts, for both virgin and carbon-mixed-CuCr0.3 alloys.

Figure 6: The evolution of hardness (dashed lines) and electrical conductivity (solid lines) for the virgin (blue color) and carbon-mixed-CuCr0.3 (orange color) alloys during the direct-age hardening (D-AH) heat treatment. The green dashed rectangular box indicates the selected optimum heat treatment condition for the evaluation of tensile properties.

3.4. Microstructural Analysis

As shown in **Figures 7 and 8**, the microstructural analysis is performed on the dense parts, made of both virgin and carbon-mixed-CuCr0.3 alloy powders, both in as-built as well as for the two optimum heat-treated conditions.

Figure 7 shows the microstructures of the CuCr0.3 alloy parts in the AB, SA+AH, and D-AH conditions (see **Table 3**). No clear melt pool boundaries are observed in the as-built condition due to their disruption caused by the epitaxial growth of large columnar grains along the building direction (BD) and spanning several powder layers. After the heat treatments, no significant macroscopic changes appear in the microstructure. However, precipitation is evident for both AB and heat-treated parts (SA+AH and D-AH), as indicated in **Figure 7** using white arrows. Overall, the number of precipitates are higher for the heat-treated parts compared to the as-built part, and they are located inside the grains as well as at the grain boundaries.

Figure 7: The microstructures obtained for CuCr0.3 alloy parts in the (a-c) as-built (AB), (d-f) solution annealing + age hardening (SA+AH) and (g-i) direct-age hardening (D-AH) conditions. The microstructural images show an epitaxial growth of the columnar grains parallel to the BD. Precipitates can be observed in the as-built as well as post-heat-treated (SA+AH and D-AH) conditions, as indicated by the white arrows.

Figure 8 shows the microstructures of the carbon-mixed-CuCr0.3 alloy in the AB, SA+AH, and D-AH conditions. Generally, the microstructures are similar to the virgin CuCr0.3 alloy with respect to the heat treatment conditions. However, some carbon segregation defects (these defects can appear as voids after polishing, and etching of the parts, however, they were originated from the carbon segregation defects, as indicated by red dashed ellipse) are observed in the as-built, and D-AH conditions of the carbon-mixed-CuCr0.3 parts. The length of these segregation defects is generally $\leq 5 \mu\text{m}$, and they are observed at the grain boundaries of the as-built and D-AH parts, as shown in **Figures 8(b) and 8(h)**. Analogous carbon segregation defects were also observed in the research performed on the carbon-mixed with pure copper powder. In that case, however, the defects were very long ($\sim 2 \text{ mm}$) and densely present in the copper matrix ^[7]

Compared to the as-built parts, the D-AH part showed a reduction in the length of the carbon segregation defects, as shown in **Figure 8(h)**. However, the SA+AH heat-treated parts do not show any evidence of such carbon segregation defects. This is perhaps due to the

redistribution of carbon by reaction with chromium from the copper matrix during the high temperature (1015°C, 1.5h) solution annealing (SA) treatment.

Similar to the virgin CuCr0.3 alloy, the as-built as well as heat-treated parts from the carbon-mixed-CuCr0.3 alloy show precipitates in the microstructure. Similarly, the heat-treated parts (SA+AH and D-AH) exhibited a higher density of precipitates compared to the corresponding as-built part. Indeed, the precipitate composition, size, its distribution as well as microstructural features such as segregated carbon particles can directly influence the mechanical properties. However, this will be further discussed in section (4.4).

Figure 8: The microstructures obtained for the carbon-mixed-CuCr0.3 alloy parts in the (a-c) as-built (AB), (d-f) solution annealing + age hardening (SA+AH), and (g-i) direct-age hardening (D-AH) conditions. The microstructure shows an epitaxial growth of the columnar grains parallel to the BD, along with precipitates that are observed for the as-built as well as post-heat-treated (SA+AH and D-AH) parts, indicated by the white arrows. Defects that are created by carbon segregation at the grain boundaries are observed in the AB and D-AH parts. However, they are not visible for the SA+AH heat-treated parts.

3.5. EDS Analysis of the Precipitates and Oxygen, Carbon Content Analysis of the Parts

Figure 9 shows an overview of the EDS analysis results for the precipitates that are present in the as-built as well as heat-treated parts (SA+AH and D-AH) of the virgin and carbon-mixed-CuCr0.3 alloys. The results are presented as the fraction of one particular element (C, or O, or Cr) present in the precipitate. This was calculated by dividing the weight

percentage of one particular element by the sum of weight percentages of all other elements (C, O, and Cr) present in the precipitate.

As seen, the precipitates present in the virgin CuCr0.3 alloy are mainly composed of chromium and oxygen for both as-built and heat-treated parts (SA+AH and D-AH). This indicates that these precipitates are mainly chromium oxides (possibly Cr₂O).

In contrast, the EDS analysis of the precipitates in the as-built part of the carbon-mixed-CuCr0.3 alloy does not show any carbon nor oxygen. This indicates that the precipitates observed in the as-built condition are metallic chromium. Moreover, the precipitates obtained after the D-AH treatment are rich in chromium with a minor amount of oxygen (the minor amount of oxygen could have picked during the heat treatment). Finally, the precipitates obtained after SA+AH treatment show the presence of chromium, oxygen, and carbon.

Figure 9: Elemental chemical composition of the precipitates, presented in fractions, that are observed in the as-built as well as heat-treated parts (SA+AH, D-AH), for the virgin and carbon-mixed-CuCr0.3 alloys. The elemental fraction is calculated by dividing the weight percentage of one particular element by the sum of the weight percentages of all elements present in the precipitate. Inset shows a representative location of the EDS measurements on the precipitate, and in the neighboring matrix.

Table 5 shows the oxygen and carbon content for the virgin and carbon-mixed-CuCr0.3 alloy parts in the as-built and heat-treated (SA+AH and D-AH) conditions, measured using the inert gas fusion technique and coulometry, respectively. These results are in line with the EDS analysis results obtained for the precipitates and confirm that a substantially lower amount of oxygen (~86%) is observed for the carbon-mixed-CuCr0.3 alloy parts compared with the corresponding virgin CuCr0.3 alloy parts. Note that the EDS analysis does not show a drastic difference in the oxygen content between the virgin and carbon-mixed-CuCr0.3 alloy, as it appears in **Table 5** for the corresponding heat treatments. This is because the EDS analysis represents a local precipitate composition wherein the residual oxygen is mostly attached to

chromium. Whereas the inert gas fusion measurements show the global oxygen contents for both alloys.

Moreover, a drastic reduction of ~50% in the carbon content is also realized in the as-built carbon-mixed-CuCr0.3 parts (310 ppm) compared to the powder (611 ppm). Therefore, by combining the results of the oxygen and carbon content analysis, it can be stated that the addition of carbon resulted in an in situ de-oxidation of the (pre-oxidized) virgin CuCr0.3 alloy parts during the SLM process. Furthermore, it should also be noted that a slight reduction in the oxygen content (from 1409 ppm to 1100 ppm) between the powder and as-built part of the virgin CuCr0.3 alloy is observed. However, this reduction in the oxygen content is substantially lower compared to the reduction (from 1458 ppm to 149 ppm) that is observed between the carbon-mixed-CuCr0.3 powder and corresponding as-built part.

Table 5: The oxygen and carbon content in ppm, for the virgin and carbon-mixed-CuCr0.3 alloy parts measured in the as-built (AB) as well as heat-treated conditions (SA+AH and D-AH). NA: Not applicable.

Element	CuCr0.3			CuCr0.3+C0.05		
	AB	SA+AH	D-AH	AB	SA+AH	D-AH
Oxygen (ppm)	1100	1200	930	149	264	164
Carbon (ppm)	NA	NA	NA	310	NA	NA

Note: The carbon content is only measured for AB (carbon-mixed-CuCr0.3 alloy) condition, since the SA+AH and D-AH heat treatments would not alter the chemical composition of the AB part.

3.6. Mechanical and Electrical Properties

Figure 10 displays the summary of the mechanical properties obtained for the virgin and carbon-mixed-CuCr0.3 alloys in the as-built as well as the optimally heat-treated conditions (SA+AH and D-AH). In the as-built condition, the addition of carbon improves the yield strength (YS) and ultimate tensile strength (UTS), while it lowers the ductility (EB). Both SA+AH and D-AH heat treatments increase the mechanical strength for the carbon-mixed-CuCr0.3 alloy compared to the virgin CuCr0.3 alloy. Despite this positive influence, ductility (EB) is slightly reduced due to the addition of carbon nanoparticles to the CuCr0.3 alloy.

Nonetheless, samples that underwent a D-AH treatment demonstrate better yield and tensile strength values compared to the as-built samples. Moreover, the SA+AH heat-treated samples show the highest tensile elongation and the smallest standard deviation in tensile properties of all carbon-mixed-CuCr0.3 alloys.

Figure 10: The mechanical properties of the virgin and carbon-mixed-CuCr0.3 alloys in the as-built (AB), solution annealed + age hardening (SA+AH), and direct-age hardening (D-AH) conditions. Note that the values are written on top of the bars and they may possess different units.

Figure 11 shows the electrical (%IACS) and theoretically calculated thermal (λ) conductivity values in the AB, SA+AH, and D-AH conditions for the virgin and carbon-mixed-CuCr0.3 alloys. The highest electrical conductivity, reaching a value of $88 \pm 1\%$ IACS, is observed for the CuCr0.3 alloy in the SA+AH condition. Generally, both alloys show improvement in the electrical conductivity after heat treatment compared to the as-built condition.

Figure 11: The electrical (% IACS) and theoretically calculated thermal conductivity values (W/(m.K)) of the virgin and carbon-mixed-CuCr0.3 alloys in the as-built (AB), solution annealed + age hardening (SA+AH) and direct-age hardening (D-AH) conditions. Note that the values are written on top of the bars and they may possess different units.

4. Discussion

4.1. Influence of the Carbon Nanoparticle Addition on the Powder Properties

Higher optical absorption of 44% was observed for the virgin CuCr0.3 alloy powder compared to 26% optical absorption that was observed for the pure copper powder (oxygen content: 13 ppm) within a wavelength range of 1060-1080 nm (**Figure 2(a)**). The high optical absorption of the virgin CuCr0.3 pre-alloyed powder could be attributed to the combination of two parameters (i) alloying of copper with chromium, and (ii) surface oxidation of the virgin CuCr0.3 alloy powder (oxygen content: 1409 ppm) forming an optically absorptive oxide film on the powder surface.^{[15][33][34]} As the virgin CuCr0.3 powder exhibited higher optical absorption compared to pure copper powder, dense CuCr0.3 alloy parts could be fabricated using laser power below 600 W. On the other hand, the fabrication of dense pure copper part required a laser power of 800 W at a scanning speed of 400 mm/s.^[13] This indicates that the improved powder optical absorption due to pre-alloying with chromium and the powder surface oxidation is relevant for reducing the required laser powers in the SLM process.

In line with this approach, 0.05 wt.% of carbon nanoparticles were mixed with CuCr0.3 alloy. The addition of 0.05 wt.% of carbon nanoparticles increased the optical absorption of the virgin CuCr0.3 alloy powder from 44% to 56%, at room temperature, without any notable adverse effect on the powder flowability (**Figure 2**). This further enhancement in the optical absorption of the CuCr0.3 alloy powder could be attributed to the high optical absorptivity of carbon within the infrared region of the electromagnetic spectrum.

4.2. Influence of Carbon Nanoparticles and Laser Scan Speed on the Part Density

Despite the improved optical absorption of the carbon-mixed-CuCr0.3 alloy powder at room temperature, no significant variation in SLM processability was observed. This could be related to the poor wettability of solid carbon by liquid copper ^[35] as well as the very low solid solubility (up to 7 ppm at room temperature) ^[36] of carbon in copper. This combination leads to the segregation of carbon along the melt pool boundary (**Figure 4 (b)**) during the actual laser melting process. Additionally, the laser typically irradiates the liquid metal and not the powder except at very high scan speeds depending upon the metal thermal conductivity. Hence, the segregation of carbon along the melt pool boundary for carbon-mixed-CuCr0.3 alloy makes that the laser beam irradiates a liquid melt pool having a chemical composition similar to the virgin CuCr0.3 alloy. Therefore, a similar SLM behavior could be expected for both virgin and carbon-mixed-CuCr0.3 alloys.

Moreover, **Figure 3** showed that the laser scan speed (v) has a significant influence on the densification of both virgin and carbon-mixed-CuCr0.3 alloys within the selected range of laser powers. At low laser scan speeds (300 mm/s, and 400 mm/s), dense parts were achieved, whereas laser scan speeds above 600 mm/s resulted in poor part densities, regardless of the employed laser power. The poor densities obtained for parts processed with a laser speed above 600 mm/s are mainly due to the presence of lack-of-fusion pores. Lack-of-fusion defects appear when the effective laser energy density (dependent on the effective laser absorptivity) is not sufficient to fully melt the irradiated powder bed.^[1]

Trapp et al. ^[26] and Ye et al. ^[27] have demonstrated the influence of the laser scan speed on the melt pool shape and the effective laser absorptivity. They showed that at lower laser scan speeds, the effective laser absorptivity increases rapidly and reaches a saturation value. However, by using higher laser scan speeds, the laser absorptivity increases at a slower rate. Therefore, the peak of laser absorptivity may not be reached even after the use of a very high laser power such as 550 W. Moreover, Trapp et al. concluded that, above a threshold laser power, deeper melt pools or even melt pools with keyholes, were observed in the samples processed with lower scanning speeds. This increased the number of internal-back-reflections within the melt pools; hence, it resulted in a higher effective laser absorptivity (It is important to note that this laser absorptivity has a major contribution from the deep melt pool geometry and minor from the material composition) ^[26]. On the contrary, when higher laser scan speeds were applied, despite the fact that a sufficiently high laser power was used, the melt pool geometry remained semi-elliptical and shallow with less internal-back-reflection, and thus lowered the effective laser absorptivity (this laser absorptivity has a major contribution from the material composition and minor from the semi-elliptical/shallow melt pool geometry) . Therefore, it can be stated that the effective laser energy density (absorptivity) required for complete melting of the powder particles depends not only on the applied laser parameters and the alloy composition but also on the melt pool geometry that is generated during the melting process.

To determine the melt pool shape, single layer experiments were performed using the carbon-mixed-CuCr0.3 alloy powder (and virgin CuCr0.3 powder - not shown), applying different laser scan speeds of 300, 600, and 1000 mm/s, combined with a fixed hatch spacing of 0.09 mm and laser power of 600 W. **Figure 12** shows the cross-sectional surface of the

single-layer melt pools along with the corresponding parts produced at the same laser setting. The melt pool shapes are indicated using yellow-dashed curves, and the melt pool depths are denoted by white double-headed arrows. It can be seen that very deep, stable melt pools ($477 \pm 33 \mu\text{m}$) were obtained at a low scan speed of 300 mm/s (**Figure 12(a)**). However, unstable, relatively shallower, or semi-elliptical melt pools were produced at higher laser scan speeds such as 600 mm/s (**Figure 12(c)**) and 1000 mm/s (**Figure 12(e)**). Moreover, the cross-section surface of the parts produced using these laser scan speeds (300, 600, and 1000 mm/s) show that dense parts could only be produced at a low laser scan speed of 300 mm/s (**Figure 12(b)**). The parts fabricated using the laser scan speed of 600 mm/s (**Figure 12(d)**) and 1000 mm/s (**Figure 12(f)**) show lack-of-fusion pores.

Consequently, by combining the findings of the in situ absorptivity measurements ^{[26][27]} and the melt pool geometries obtained within this study, it could be stated that dense CuCr0.3 alloy parts (from both virgin and carbon-mixed-CuCr0.3 powder) were obtained at low laser scan speeds (300 mm/s, and 400 mm/s) due to the high effective laser absorptivity. This increased effective laser absorptivity could be attributed to the deep and stable melt pools offering the possibility of internal-back-reflections and subsequent laser absorption (**Figure 12(a)**). Conversely, the low-density of the parts processed at high laser scan speeds (600-1200 mm/s) could be ascribed to the low effective laser absorptivity caused by less-deep, semi-elliptical melt pools which do not allow multiple internal-back-reflections of the incoming laser light.

Furthermore, it is also important to note that low-density parts were obtained (at high scan speeds), even though the laser could penetrate much deeper (178 to 361 μm) than the applied powder layer thickness (30 μm). This observation highlights that the lack-of-fusion defects could not be associated with the laser penetration depths in the powder bed, rather they are related to the poor effective optical absorption characteristics of CuCr0.3 alloy (including the carbon-mixed-CuCr0.3 alloy powder, since carbon segregates along the melt pool boundary during the actual SLM process and does not contribute in enhancing the optical absorption) and the shallow melt pool geometry that is generated during the SLM process.

Figure 12: Cross-sectional light optical microscopy images for melt pool depths for single layer (a, c, & e), and the corresponding parts (b, d, & f), processed using different laser scan speeds (v 300, 600 and 1000 mm/s) and at a fixed laser power of 600 W and hatch spacing of 0.09 mm. The dense part was only obtained at a low laser scan speed of 300 mm/s, which exhibits very deep ($477 \pm 33 \mu\text{m}$) and stable melt pool geometry. Low-density parts were obtained at higher scan speeds (600 & 1000 mm/s), having unstable and less-deep melt pools. The melt pools are indicated using yellow dashed curves, while the melt pool depths are denoted using white double-headed arrows.

4.3. Evolution of Hardness and Electrical Conductivity in the As-built, and During Post-Heat-Treatments (SA+AH and D-AH)

4.3.1. As-built

In the as-built condition, the virgin CuCr0.3 alloy showed lower electrical conductivity compared to the carbon-mixed-CuCr0.3 alloy, while the hardness values were comparable. This indicates that the virgin CuCr0.3 alloy retained a larger fraction of chromium in the solid-solution with copper compared to the carbon-mixed-CuCr0.3 alloy after SLM^[19]. Furthermore, the EDS analysis of the precipitates (rich in chromium and oxygen) and the overall high oxygen content (1100 ppm) of the virgin CuCr0.3 alloy part suggest that the remaining chromium (which is not in solid-solution with copper) exist in the form of chromium oxide. On the other hand, the EDS analysis of the precipitates present in the carbon-mixed-CuCr0.3 alloy part suggests that the chromium is present in its metallic form. This was also supported by the overall low oxygen content of 149 ppm that was measured for the as-built carbon-mixed-CuCr0.3 alloy part.

4.3.2. SA+AH

After the solution annealing (SA) step, the virgin CuCr0.3 alloy parts showed a reduction in the hardness along with a sharp increase in the electrical conductivity.

Furthermore, the age hardening of these parts displayed no significant improvement in the hardness and electrical conductivity. This abnormal trend of the virgin CuCr0.3 alloy could be explained by the formation of chromium oxide precipitates during high-temperature solution annealing treatment (1015°C, 1.5h) and subsequent age hardening. As the chromium oxide was formed during the high-temperature solution annealing heat treatment, the copper matrix became purer, and a very high electrical conductivity is realized without much contribution to the mechanical strength. On the contrary, solution annealing of the carbon-mixed-CuCr0.3 alloy showed an increase in the hardness and a drastic reduction in electrical conductivity. Upon further age hardening of the solution annealed parts made of carbon-mixed-CuCr0.3 alloy, the typical precipitation strengthening behavior was observed i.e., a concomitant increase in the hardness and electrical conductivity^{[18][30]}. This hardness and electrical conductivity trend during the SA+AH heat treatment confirms that the as-built part made of the carbon-mixed-CuCr0.3 alloy was de-oxidized and that chromium is present in its metallic form. Furthermore, the solution annealing of the carbon-mixed-CuCr0.3 alloy parts lead to the formation of a copper-chromium solid-solution (by dissolving metallic chromium in copper) and; hence, a drastic reduction in electrical conductivity was recorded. The subsequent age hardening treatment of the solution annealed parts rendered a uniform distribution of nanometer-sized chromium precipitates within the copper matrix of those parts. This offered an improved combination of the hardness and electrical conductivity due to the precipitation strengthening effect.^[20] Moreover, the EDS analysis of these precipitates also suggests that some chromium-carbides precipitates also exist along with the metallic chromium precipitates. However, it proves to be very difficult to quantitatively define the ratio of chromium that is present in its pure metallic form to the amount of chromium that is present as a carbide.

4.3.3. D-AH

During the D-AH heat treatment, both virgin and carbon-mixed-CuCr0.3 alloys showed a concomitant increase in electrical conductivity upon age hardening. However, the carbon-mixed-CuCr0.3 alloy showed a rapid increase in the hardness compared to the virgin CuCr0.3 alloy. In general, the hardness and electrical conductivity of the carbon-mixed-CuCr0.3 alloy are higher, which could be associated with the coherent precipitation of fine and uniform metallic chromium during the D-AH heat treatment.^[37] This result highlights the positive influence of the in situ de-oxidation process that takes place during SLM due to the mixing of carbon with CuCr0.3 alloy. In contrast, a stable trend (or slower rise) of the hardness was observed for the virgin CuCr0.3 alloy. This could be credited to the formation of chromium-oxide precipitates and their inefficient strengthening effect compared to the metallic chromium precipitates in the copper matrix.^[17] Furthermore, the peak hardness values recorded for the D-AH condition were higher than the SA+AH condition, for both virgin and carbon-mixed-CuCr0.3 parts. This also indicates that the precipitation strengthening effect offered by the metallic Cr precipitates is much higher compared to the Cr-oxides and Cr-carbide precipitates. Although, precipitation strengthening could be the main cause,^[19] other factors such as the starting microstructure, the annihilation of induced-dislocations, residual stresses, and the grain coarsening could also contribute to this difference in part properties between the SA+AH and D-AH treatments.

4.4. Influence of Post-Heat-Treatments on Mechanical and Electrical Properties and their Interrelationship with Microstructural Evolution

4.4.1. Mechanical Properties

Overall, higher YS and UTS were obtained at the expense of a slight loss of ductility for the carbon-mixed-CuCr0.3 alloy parts compared to the corresponding virgin CuCr0.3 alloy. This could be due to the in situ de-oxidation effect, making that more metallic chromium as well as chromium carbides, which especially form upon solution annealing heat-treatment, pin the grain boundaries; hence, inhibit the grain growth.

For the virgin CuCr0.3 alloy, the highest tensile strength was obtained after the D-AH heat treatment. This could be due to the fine chromium precipitates and/or chromium oxide particles pinning the dislocations and grain boundaries. On the other hand, the highest ductility obtained for the SA+AH heat treatment could be attributed to the annihilation of the residual stresses and grain coarsening during the high-temperature solution annealing treatment ^[19].

The carbon-mixed-CuCr0.3 alloy in the as-built condition showed reduced ductility and slightly higher standard deviation in the yield (YS) and ultimate tensile strength (UTS) values. Two phenomena are responsible for this observation. First, segregation of elemental carbon forms weak points at the grain boundaries oriented parallel to the building direction (BD), as shown in **Figure 8(b)**. (It is important to note that the carbon segregation defects are much smaller and less dense compared to the one reported for pure copper parts,^[7] this indicates that chromium was helpful in reducing the intensity of carbon segregation defects). Second, the tensile testing itself that was carried out transverse to the BD i.e., perpendicular to the carbon segregated grain boundaries, leading to premature failure. Furthermore, the post-heat-treatment of the carbon-mixed-CuCr0.3 alloy offered a better combination of mechanical properties (tensile strength and ductility) compared to the corresponding as-built part. The improved mechanical strength of the post-heat-treated parts could be attributed to the precipitation strengthening effect caused by the uniform distribution of the nanometer-sized chromium and chromium carbide precipitates. The improved ductility could be attributed to a reduction in the number of carbon segregation defects caused by the conversion of the elemental carbon into chromium carbide (**Figure 8(d-f)**), which was present along the grain boundaries in the as-built condition, as shown in **Figure 8(c)**.

4.4.2. Electrical Conductivity

Generally, lower electrical conductivity values were obtained for the parts in the as-built condition due to the solid solubility of chromium in copper.^[21] Upon post-heat-treatments (SA+AH and D-AH), chromium precipitates out, offering purer copper matrix. This increases the electron's mean free path length, enhancing the electrical and thermal conductivity after heat treatments. Additionally, other phenomena such as grain coarsening, the annihilation of dislocations, and the reduction in defects due to carbon segregation, contribute to the higher conductivity of the heat-treated parts, but they are not as effective as the chromium precipitation.^[19] The highest electrical conductivity ($88 \pm 1\%$ IACS) was recorded for the virgin CuCr0.3 alloy when the optimum SA+AH heat-treatment condition was applied. The highest conductivity could be attributed to the precipitation of chromium-oxides, depleting the copper

matrix with chromium, the annihilation of induced-dislocations, residual stresses and, and the grain coarsening during the solution annealing and age hardening treatment.

5. Conclusions

Mixing 0.05 wt.% carbon nanoparticles with pre-alloyed CuCr0.3 powder improved the optical absorption of the virgin CuCr0.3 powder from 44% to 56% at room temperature, while the only minimal influence on the powder flowability and packing density was observed. Despite the improved optical absorption offered by the carbon nanoparticle addition, the SLM processability remained unaffected, as similar laser scan settings were required to achieve dense parts. The similar SLM processability of the virgin and carbon-mixed-CuCr0.3 alloy powders could be attributed to the poor wettability of liquid copper on solid carbon and the low solid-solubility of carbon in copper. This causes the segregation of carbon along the melt pool boundary, and thus, it does not considerably contribute to the optical absorption during the actual laser melting process. A significant influence of the laser scan speed on the part density was observed for both alloys, as dense parts could only be produced at low laser scan speeds (300-400 mm/s), regardless of the employed laser power (350, 450, or 600W). This was attributed to the formation of deep melt pool geometry at low scan speeds, which could increase the effective laser absorptivity (by multiple internal reflections of laser light) and complete the fusion of the powder particles.

Overall, the carbon-mixed-CuCr0.3 alloy showed better mechanical properties (the best combination of the tensile strength of 298 MPa, yield strength of 208 MPa and electrical conductivity of 87% IACS) compared to the virgin CuCr0.3 alloy (the tensile strength of 242 MPa, yield strength of 129 MPa and electrical conductivity of 88% IACS). The better mechanical properties of the carbon-mixed-CuCr0.3 alloy were attributed to the formation of nanometer-sized, uniform, and coherent metallic chromium (and chromium-carbide) precipitates during the post-heat-treatments. On the other hand, the ineffective mechanical strengthening of the virgin CuCr0.3 alloy was attributed to the formation of chromium-oxide precipitates upon post-heat-treatment. This result established the positive influence of the carbon addition that offered metallic chromium in the as-built SLM part by in situ de-oxidation of the CuCr0.3 alloy.

Since undesired oxide film formation is inevitable during the storage of copper alloy powders, the recyclability and storage life of copper powders can be rather limited. This can, however, be improved by the addition of carbon nanoparticles, which reduce the exposure of the copper powder surface to air and promote the in situ de-oxidation during the process.

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Data availability

The raw/processed data required to reproduce these findings can be made available upon request.

Declaration of Interest

None

7. References

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Selective laser melting of virgin and carbon-mixed-CuCr0.3 alloy powder is performed to produce dense parts. The combination of electrical and mechanical properties of the parts

fabricated using carbon-mixed-CuCr0.3 powder are superior to the virgin CuCr0.3 alloy powder-based parts. This enhancement is attributed to the precipitation of metallic chromium in the copper matrix.