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# Selective laser remelting of an additively manufactured Cu-Al-Ni-Mn shape-memory alloy

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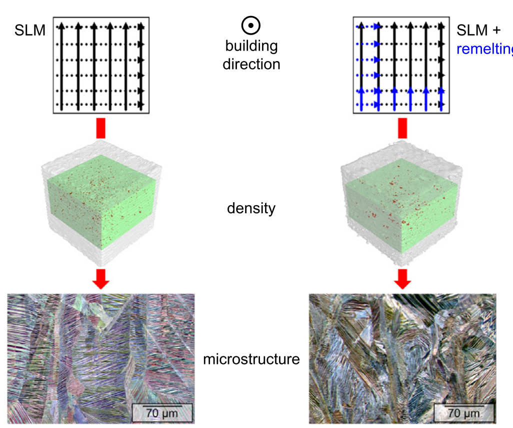
## HIGHLIGHTS

- Selective laser remelting enhances the relative density of Cu-11.85Al-3.2Ni-3Mn shape-memory parts.
- The grain size can be modified during remelting.
- The transformation temperatures can be adjusted in a broad range via remelting.
- Selective laser remelting is a promising approach for optimizing the properties of shape-memory alloys.

## GRAPHICAL ABSTRACT

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## ABSTRACT

Selective laser melting (SLM) was used to manufacture fully martensitic ($\beta'_1$) samples of the shape-memory alloy 81.95Cu-11.85Al-3.2Ni-3Mn (wt%). Crack-free specimens with a high relative density of about 98.9 ± 0.1% were produced. Immediate remelting of already processed layers during SLM enhances the relative density (99.5 ± 0.3%). Primarily by varying the scanning speed in the remelting step, the thickness of the remelted zone can be adjusted. Moreover, remelting alters the microstructure as well as the transformation temperatures, which tend to rise with the volumetric energy input. In this way, the shape-memory properties can be modified without compromising the relative density and the considerable plasticity of the samples. Thus, the remelting procedure proves to be an interesting tool for 81.95Cu-11.85Al-3.2Ni-3Mn and related alloys in order to optimize and tailor their performance already during SLM processing and without applying additional post-processing steps.

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## 1. Introduction

Selective laser melting (SLM) is a laser-based additive manufacturing technique in which a bulk part is created layer-by-layer through melting defined areas of a powder bed [1–4]. In general, SLM is used for the fabrication of metallic components with a high degree of geometrical freedom and often yields materials properties comparable or

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even superior to their conventionally processed counterparts [3,5–8]. Another considerable benefit is that the processing of a thin powder layer (generally between 20 and 100 μm) on massive substrate plates in combination with small laser spot diameters results in high intrinsic cooling rates [2]. These unique processing conditions have a strong impact on the microstructure (grain refinement) and, in turn, this can enhance the performance of additively manufactured materials [6].

One key aspect in SLM is the resulting porosity, which can be relatively high compared to conventionally processed material. Because of its versatility a large number of process parameters can be varied during selective melting most of which influence the final density and the properties of a part [3,9]. Despite the difficulties in optimizing the SLM processing parameters for obtaining high-quality parts, the general applicability and potential of SLM has been already demonstrated for a variety of alloys [2–4,6,7,10,11]. Among these materials, shape-memory alloys (SMAs) are one group of materials that have drawn attention due to new and innovative actuator designs that are now feasible to be produced in a cost effective way via SLM [12–14].

The most widely used and studied SMA for applications, equiatomic NiTi (Nitinol), has been already processed via SLM by numerous groups and the capability of this approach in terms of manufacturing medical devices has been shown [13–20]. However, industrial use of Nitinol is limited due to transformation temperatures below 100 °C [21,22]. Extensive research has been carried out to increase the transformation temperatures by developing new alloy compositions that can be applied at temperatures above 100 °C, the so-called high-temperature shape-memory alloys (HTSMAs) [23]. But especially promising HTSMAs based on NiTi, viz. Ti-Ni-Pd, Ti-Ni-Pt or Ni-Ti-Hf, are relatively expensive and difficult to manufacture [22,24,25]. For example, the HTSMA Ni-Ti-Hf was recently produced by SLM [21]. These samples suffered from porosity, cracking and a significant oxygen uptake, which are inherent challenges when NiTi-based SMAs are being processed by SLM.

With regard to HTSMAs, there is a major interest in materials derived from Cu-based SMAs (e.g. Cu-Al-Ni-Mn [26]) due to their relatively low manufacturing costs, good processability as well as promising shape-memory properties [24,26–28]. The improved thermodynamic stability and a broad regime of transformations temperatures above 100 °C is characteristic for the family of Cu-Al-Ni-based alloys and has drawn interest in terms of applications [28]. Yet, the applicability of these alloys is limited by their brittleness in the coarse-grained polycrystalline state, which is caused by a strong elastic anisotropy and the precipitation of certain intermetallic phases [29,30]. Thus, intensive research was conducted in the processing of Cu-based SMAs with small grain sizes to overcome the poor deformability resulting from intergranular cracking [26,27,31–33]. Besides refining the microstructure through the addition of grain refiners such as Ti [26,28,33] or Zr [26], another approach constitutes in rapid solidification techniques like melt spinning [33], spray-forming [27] and SLM [34].

As shown previously [34,35], a Cu-Al-Ni-Mn shape-memory alloy can be successfully processed by SLM. An optimized parameter setup has been developed to obtain crack-free and fully martensitic ($\beta'_1$) specimens with a high relative density of up to 99%, low oxygen content, small grain sizes between 10 and 80 μm and transformation temperatures above 100 °C [1,36]. Although the relative density has been found to be very high in Cu-Al-Ni-Mn parts for a range of energy inputs [1], optimization to 100% relative density has not been feasible. The mechanical properties (i.e. deformability, yield strength) strongly depend on the remaining porosity, especially when it comes to fatigue [5]. One method to reduce or eliminate remaining defects (pores, cracks) in SLM parts without drastically changing the chemical composition or the phase formation of the material [37,38], is to employ a second laser process such as re-scanning or remelting of previously solidified layers [5,6,39–41]. This step can be easily implemented in the SLM process and is also called selective laser remelting (SLRM) [1].

The transformation temperatures of Cu-based SMAs strongly depend on the (i) chemical composition [26,31,42], (ii) the phases present [28] and (iii) the grain size [36,43,44]. For example, it has been reported that SLM processing, in contrast to manufacturing via spray forming, reduces the transformation temperatures due to a smaller grain size [36]. Furthermore, the transformation temperatures of Cu-based SMAs can be adjusted to higher values by changing the process parameters, viz. increasing the energy input during the manufacturing of SLM bulk samples [43]. Similar investigations have been reported for equiatomic NiTi alloys [12,15–17,20]. Interestingly, the possibility of an additional remelting step to further optimize and adjust the properties of SMAs has not been investigated so far. This makes employing an additional remelting step in the SLM process a very interesting approach, because the main parameters for remelting like laser power, scanning speed and track overlap result in additional energy dissipated in previously solidified layers and should affect the grain size and with it the transformation temperatures.

In this study, the SLM process has been combined with an additional remelting step to process the shape-memory alloy 81.95Cu-11.85Al-3.2Ni-3Mn (wt%). The remelting step is implemented in order to further increase the relative density. The main focus of this work, yet, is to study the effect of the additional remelting step on the microstructure and the transformation characteristics. Therefore, we correlate the influence of the additional energy input with the microstructural details (e.g. phase formation, grain size), the corresponding transformation temperatures and the mechanical properties.

## 2. Experimental details

### 2.1. Materials and processing

The Cu-based shape-memory alloy 81.95Cu-11.85Al-3.2Ni-3Mn (wt%) was processed by selective laser melting in a SLM 250$^{HL}$ (SLM Solutions Group AG) equipped with a 400 W Nd:YAG-based fibre laser. Based on our previous work [1], an optimized parameter setup with a laser power ($P_L$) of 330 W (wavelength $\lambda = 1.064$ μm, laser spot diameter = 80 μm), a scanning speed ($v_s$) of 740 mm/s and a track overlap of 30% (hatching distance ($h_c$) = 0.13 mm) was selected by using a stripe-hatching strategy (rotation per layer: 90°) and a layer thickness of 90 μm for the fabrication of SLM bulk samples with various geometries (cubes, rods and beams). The powder used had a particle size between 30 and 90 μm ($d_{50}$ = 48 μm) and is suitable for the SLM process as described in [1].

During manufacturing of particular SLM samples, every layer was additionally illuminated by the laser [1] in order to remelt the material with various scanning speeds and track overlaps. First, the processing parameters for the additional remelting procedure were optimized in single-track experiments on a 81.95Cu-11.85Al-3.2Ni-3Mn baseplate. The laser power was varied between 300 and 340 W (laser spot diameter = 130 μm) and the scanning speed was altered between 500 and 2500 mm/s. The width and morphology of the individual tracks was measured using a digital microscope (Keyence VHX-2000). Subsequently, bulk material with a surface area of 4 × 4 mm$^2$ was remelted at a constant laser power ($P_L$ = 330 W) with scanning speeds and track overlaps between 500 and 1500 mm/s as well as 30 and 90%, respectively. The remelting depth of the remelted surfaces was measured along the cross-sections using an optical microscope (Nikon Epiphot 300). In the next step, the same parameter combinations were applied to SLM bulk samples in order to remelt previously processed layers. These samples will be abbreviated as SLRM.

### 2.2. Sample characterization

The chemical composition of the powder and the bulk material was analyzed using ICP-OES (Inductively-Coupled Plasma-Optical Emission Spectroscopy, IRIS Intrepid II XUV, Thermo Fischer Scientific).

The density of all specimens was measured via the Archimedean method using a Sartorius MC210P balance. Relative densities were

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