Laser-Based Additive Manufacturing of WC–Co with High-Temperature Powder Bed Preheating†

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The field of additive manufacturing (AM), and especially laser powder-bed fusion (LPBF), is constantly growing. Process windows for a large variety of materials are already developed. Nevertheless, some materials are still difficult to manufacture with LPBF. One of these materials is the tungsten carbide/cobalt-based hard metal (WC–Co), which is conventionally produced by powder metallurgy including liquid-phase sintering. Most approaches to manufacture WC–Co with LPBF show a high porosity, undesirable phases in the microstructure, and inhomogeneous carbide distribution. However, the production of WC–Co cutting tools by LPBF will offer some benefits such as production of geometrically optimized inner cooling channels or optimized geometry of the cutting edges. Herein, WC with 17 wt% Co is processed by LPBF with a powder-bed heating of 900 °C. Afterwards microstructure, density, and hardness are determined. In addition, X-ray diffraction (XRD) analysis is performed to determine the phase composition. To investigate the edge-holding properties of LPBF-manufactured WC–Co cutting tools, stock removal tests are conducted on three different workpiece materials.

1. Introduction

Laser powder-bed fusion (LPBF) is a technology spanning application from prototyping to the production of small batch sizes. Process parameters for various steels and nonferrous materials, e.g., nickel-based alloys, are already fully developed.[1] Composites of tungsten carbide (WC) and cobalt (Co), known as cemented carbides (WC–Co), are essential for the tool-manufacturing industry for diverse applications. A wide range of hardness, fracture toughness, and wear resistance is achieved by the variation of WC grain size and cobalt content.[2]

However, up to now, cemented carbides cannot be successfully processed by LPBF. First approaches to build crack-free dense parts have been conducted in recent years.[3] Process parameters, such as laser power and scanning speed, have been found to be crucial to obtain high material density.[4] Several problems have been identified that prevent a reliable production of LPBF parts made of WC–Co. Cobalt evaporation during the process occurs, which leads to locally lowered binder contents. Pronounced cracks and delamination have been found due to internal stresses.[5] Preheating of the build platform is a promising approach to avoid cracks.[6] However, detrimental brittle phases, e.g., W₂C and η-phase (W₃Co₂C), are still found in the binder phase.[6–8] For this study, WC–Co with 17 wt% Co (WC–17Co) has been processed by LPBF with a powder-bed heating of 900 °C. The microstructure, density, and hardness of the produced specimens have been determined. In addition, the phase composition has been characterized by X-ray diffraction (XRD) analysis. To investigate the edge-holding properties of LPBF-manufactured WC–Co cutting tools, stock removal tests have been conducted on brass (CuZn39Pb4), titanium (TiAl₆V₄), and nickel alloys (Inconel 718).

2. Experimental Section

2.1. Materials

The raw material used in this investigation was the tungsten carbide–cobalt powder AMPERIT 526 by Starck.[9] The WC–17Co...
powder had a composition of 83 wt% tungsten carbide and 17 wt% cobalt, as shown in Table 1.

For typical applications in thermal spraying, WC–17Co was agglomerated and presintered. WC and cobalt powder were mixed with water, solvent, and organic binder. After spray-drying and debinding, porous WC–Co agglomerates were obtained. The particle size distribution of the agglomerate was measured by a laser diffraction method (Horiba LA 950). The median of the particle size, \(d_{50}\), and the particle diameter corresponding to 90% of the cumulative undersize distribution, \(d_{90}\), were determined. The median of the agglomerate diameter was 35 \(\mu m\). In the agglomerates, WC particles with \(d_{50}\) of 3 \(\mu m\) and \(d_{90}\) of 9 \(\mu m\) were found.

WC–17Co agglomerates with a rough surface are shown in Figure 1a; carbides (light grey) embedded in Co-binder phase (dark grey) and internal porosity are shown in Figure 1b. In addition, an inhomogeneous binder distribution was observed along different powder particles. A build platform with a diameter of 55 mm made of stainless steel (AISI 316 L) in a sand-blasted surface condition was used.

2.2. Laser Powder-Bed Fusion

The samples for this study were produced using a laboratory scale LPBF machine “Aconity MIDI” (Aconity3D GmbH, Herzogenrath). The LPBF machine was equipped with a fiber laser system YLR-1000-WC (IPG Photonics GmbH, Burbach), providing a maximum output of 1 kW with a specific wavelength of 1070 nm. The used dynamic focusing unit “varioSCAN60” and galvanometric scanning unit “intelliSCAN30” (SCANLAB GmbH, Puchheim) allowed the precise adjustment of a Gaussian-shaped beam diameter. An inductive heating device coupled with a generator “TruHeat HF 3005” (Trumpf Hüttinger GmbH & Co.KG, Freiburg) was integrated in the LPBF machine allowing substrate preheating of up to 1200°C.

To establish a suitable process window for manufacturing of bulk material, various specimens of \(10 \times 10 \times 8 \text{ mm}^3\) were manufactured by varying the laser power \((P_{L,H})\) and scan speed \((v_{s,H})\), as shown in Figure 2a. Other process parameters were kept constant throughout all experiments (see Table 2). The samples were built without support structures using a chess-like exposure pattern in the hatch area, with the xy-plane divided into 25 equal squares of \(2 \times 2 \text{ mm}^2\). Between every layer, the hatch was rotated consecutively (33°). For the contour area, the same process parameters were used for all specimens. Higher scan speed \((v_{c,C})\) of 400 mm s\(^{-1}\) compared with the hatch area (180–220 mm s\(^{-1}\)) was chosen to increase surface quality. The build chamber was filled with pure argon atmosphere with oxygen content below 100 ppm and \(80 \text{ mbar}\) overpressure compared with ambient atmosphere. To investigate the capability of this material for tooling applications, a tool geometry (Figure 2b,c) was manufactured. The tools T1 and T2 were built with best-of parameters (Table 2) from the previously performed parameter study.

2.3. Analysis of Microstructure and Mechanical Properties

The density of the WC–17Co powder was determined as average of ten measurements using helium pycnometry (Micromeretics, Unterschleissheim). The Archimedes density was determined for tools T1 and T2. The density of the WC–17Co powder was determined as average of ten measurements using helium pycnometry (Micromeretics, Unterschleissheim). The Archimedes density was determined for tools T1 and T2. For the microstructural analysis, the samples

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**Table 1.** Chemical composition of WC–17Co according to powder manufacturer.

| Element | Co | C total | W | Fe | Other |
|---------|----|---------|---|----|-------|
| wt%     | 16.9 | 5.07 | 78.0 | 0.0 | < 0.5 |

**Figure 1.** Agglomerated and sintered WC–17Co powder: a) powder morphology; b) cross-section of a powder agglomerate.

**Figure 2.** a) WC–17Co cubic samples (as build), the selected samples “1,” “2,” and “3” are C1, C2, and C3, respectively. The arrow indicates the gas flow in the build chamber. b) Tool geometry. c) Tool geometry samples (as build).
were cut along the building direction (BD), ground, and polished. The porosity of cubic samples was determined by image analysis. The powder and LPBF samples were analyzed by a scanning electron microscope (Zeiss, Oberkochen). The WC grain size was determined on SEM micrographs using a linear interception method. Since cobalt evaporation during LPBF can be expected, analysis of the cobalt content was made by energy-dispersive X-ray spectroscopy (EDS) mapping (Oxford Instruments, Abingdon, UK). Furthermore, the chemical composition of selected samples was determined by X-ray fluorescence. Phase analysis was performed using XRD technique (30° ≤ 2θ ≤ 150°, Cu Kα, XRD Eigenmann GmbH, Ahrensburg). The Vickers hardness profile (HV 10) was measured along the building direction.

2.4. Fundamental Cutting Tests

To evaluate the performance of the additively built cemented carbide grade, fundamental cutting tests of different workpiece materials were conducted. Orthogonal cutting tests were conducted on a test bench based on the external broaching machine tool Forst RASX 8 × 2200 × 600 M/CNC. (Forst Technologie GmbH, Solingen). The machine tool has three broaching tracks (C1, C2, C3) and a high-speed camera— as shown in Figure 3. The test workpieces with dimensions of 45 mm × 30 mm × 6 mm (length × width × depth) were clamped into the tool holder, whereby the machining was conducted along the long side of the workpiece. The machining tests were conducted on the brass alloy CuZn39Pb3, the titanium alloy TiAl6V4, and the nickel-based alloy Inconel 718. The machining of steels was aborted due to the high wear tendency and therefore the necessity of coatings or the use of other carbide grades containing cubic carbides.

In the as-built condition, the additively manufactured inserts showed an insufficient surface quality. Therefore, a grinding of the cutting edge was necessary. The clearance angle was set to α = 3° and the rake angle to γ = 12°. A further reduction of the wedge angle β was not carried out to guarantee sufficient cutting-edge stability, although larger clearance angles α are recommended for cutting titanium alloys, due to the low Young’s modulus of ≈114 GPa. Before use, a defined cutting-edge rounding of rβ = 20 μm was applied by a brushing process. To avoid a mechanical overload of the cutting edge, the chip thickness was set to f = 0.05 mm for the titanium- and nickel-based alloys and to f = 0.2 mm for the brass in a first step. The cutting speed in all investigations was set to v = 10 m min⁻¹, which does not correspond to the state of the art for machining with cemented carbide tools. The reason for this was to reduce the thermomechanical tool load as much as possible in the first investigations with the additively manufactured tools.

3. Results and Discussion

Volume energy density $E_V$, which is described by the laser power $P_L$, scan speed $v_s$, layer thickness $d_l$, and hatch distance $d_h$, is estimated using Equation (1)

$$E_V = \frac{P_L}{v_s \cdot d_l \cdot d_h}$$

The volume energy density within the specimens has been varied between 463 and 617 J mm⁻³. For the contour scan, a higher scan speed and therefore a lower energy density of 278 J mm⁻³ were used. The evaluation of the material properties has been conducted for samples exposed to a wide range of volume energy density, as shown in Table 3.

Two cutting tools with the geometry shown in Figure 2a,b were also examined.
3.1. Microstructure

The density of the WC–17Co powder was $13.56 \pm 0.03 \text{ g cm}^{-3}$. Since the theoretical density of WC–17Co ($\rho(\text{WC}) = 15.7 \text{ g cm}^{-3}$; $\rho(\text{Co}) = 8.85 \text{ g cm}^{-3}$) is $13.87 \text{ g cm}^{-3}$, closed porosity of the agglomerates is 2.2%. Archimedes density, Table 4, was determined for tools T1 and T2. Since the extent of cobalt evaporation is uncertain, the theoretical density was not estimated. Therefore, relative density was not calculated. Large scatter in density indicates low reproducibility of the LPBF process. Porosity of cubic samples is between 2–5% with a trend with increasing volume energy density $E_V$. Nevertheless, the sample C3 with the highest $E_V$ shows the highest porosity.

The microstructure of the cuboids is shown in Figure 4a. Since no melt pools are visible, the densification is assumed due to partial melting, which is equivalent to consolidation via liquid-phase sintering in the conventional manufacturing route.

In higher magnification, an unequal distribution of the binder and of the WC grain size can be observed.

In the heat-affected zones, an increase in WC grain size due to diffusion is expected in underlying layers during LPBF process. In the hatch area of cubic samples, the observed pores have an equivalent diameter between 11 and 17 $\mu$m. The maximum pore size identified is between 105 and 158 $\mu$m, as shown in Table 4. Furthermore, accumulations of cobalt binder are found, which are described later.

The microstructure of a sample built with a shell–core structure due to the specified scanning strategy is shown in Figure 4a. In the hatch area, WC grains with $d_{50}$ values of 3–4 $\mu$m were observed, with individual large grains up to 50 $\mu$m, as shown in Figure 4b,d. A distinction as to whether these are single large grains or agglomerates was not possible by means of optical microscopy. Therefore, electron backscatter diffraction (EBSD) analysis was performed, showing large WC grains with one single orientation, as shown in Figure 4e. The pronounced grain growth from 2.5 $\mu$m to up to 50 $\mu$m is probably caused by the interaction between laser energy and the preheated powder bed with $T_P = 900^\circ\text{C}$. Thermal cycles are present during the solidification of upper layers. The peak temperature of the material during the process was not measured. However, computed thermal cycles predict an increase in peak temperature. Nevertheless, the temperature exceeds 900 $^\circ\text{C}$, where surface and volume diffusion are expected in cobalt.

Inside the WC grain, Figure 4d, both, round- and irregular-shaped pores, are found. The origin of this porosity could not be clarified yet. By changing LPBF parameters, given microstructure can be triggered. With regard to future applications of LPBF-manufactured WC–Co, controlled local microstructure can be advantageous. Up to a depth of $\approx 200 \mu$m, in the contour area, considerably smaller WC grains with $d_{50}$ of 3 $\mu$m are found, as shown in Figure 4c.

Cobalt is inhomogeneously distributed in the microstructure, visible by an EDS mapping in Figure 5. A cobalt depletion is observed between the individual laser tracks. In these regions,
the overall energy input is higher due to partial remelting of the solidified material during processing of the following layer. This supports local cobalt evaporation. For lower energy input, the deviation in cobalt content along the BD appears to be less significant. Therefore, lower $E_V$ values are preferable for a homogeneous cobalt distribution. X-ray fluorescence measurements indicated a global cobalt depletion compared with the cobalt content in the powder agglomerate (16.9 wt%). In the as build samples, cobalt was determined as 16.6 wt% for lower energy input (556 J mm$^{-3}$) and 16.3% for higher energy input (611 J mm$^{-3}$). This method allows to measure the total cobalt content in binder as well as in $\eta$-phases. It can be supposed that the depletion is caused by evaporation.

Furthermore, segregation of cobalt is visible leading to relatively large Co clusters, as shown in Figure 6b. This unwanted effect is probably caused by inhomogeneous cobalt distribution in the powder agglomerates. The influence of process parameters on the homogeneity of the Co distribution needs to be investigated in future studies. A powder with a more uniform cobalt distribution and less porosity would be beneficial for the LPBF process.

XRD diffractograms of WC–17Co powder and three LPBF samples with different energy inputs are shown in Figure 6a. Diffractograms are plotted as a square root of the observed intensity to emphasize smaller peaks compared with the dominant WC peaks. Furthermore, the intensity of all samples is normalized to the main peak of WC (41.6°)\cite{18}. Peak identification showed a peak shift toward lower diffraction angles and thus higher distances between lattice planes. Compared with powder, a signal at $2\theta = 34°$ is recorded in all LPBF specimens. This phase is probably $\alpha$-$W_2C$ (hex)\cite{19} which can be found dispersed in the cobalt binder, as shown in Figure 6b. This assumption is supported by EDS element mappings, as shown in Figure 6c,d.

Figure 5. Element mapping for Co (green), including a line scan for Co (white), for a) lower (556 J mm$^{-3}$) and b) higher (611 J mm$^{-3}$) energy input.

Figure 6. a) XRD diffractograms of the powder and LPBF samples. b) SEM image with the fine-dispersed $W_2C$ phase in the cobalt binder (arrows). c) EDS element mapping for cobalt. d) EDS element mapping for tungsten.
where the dispersed phase contains tungsten and is deficient in cobalt. The phase $\alpha$-W$_2$C was also found by Gläser\cite{8} using electron backscatter diffraction measurements and by Schubert et al.\cite{6} using EDS. The signal of cubic $\beta$-Co (51.8°, 60.6°)\cite{20} disappears in LPBF samples. In the same time, a higher amount of hexagonal $\alpha$-Co (57°) is observed.\cite{21} The proportion $\alpha$-Co/$\beta$-Co determined by X-ray diffraction is uncertain. Cobalt grains up to 200 $\mu$m are expected\cite{22} and the area exposed to X-ray diffraction is 5 mm$^2$. Therefore, the relations between intensities of different Bragg reflections change drastically depending on the position of the measurement.

3.2. Hardness

Hardness measurements were conducted on cuboids along the BD beginning at the top of the cube (0 mm), as shown in Figure 7. There is no trend visible along the BD. Considerable scatter in hardness values indicates inhomogeneities in microstructure. The mean values for samples C2 and C3 are equal, whereas the sample C1 shows higher hardness values. Similar hardness values are obtained for microstructure with equal WC grain size and different porosities.

The hardness of the cutting edge is measured after the stock removal tests. In the region around the cutting edge (contour region with finer WC grains), Vickers hardness of 899 ± 156 HV0.1 (T1) and 909 ± 20 HV0.1 (T2) has been measured. The enhanced hardness is in accordance with the knowledge about hard metals, which finer WC grains result in higher hardness.\cite{23} The large scatter of the hardness values of T1 could be explained by an inhomogeneous microstructure. However, this indicates that in these first attempts, process reproducibility is rather low and that samples with different microstructures were produced using nominally the same LPBF process parameters. Therefore, the root causes for the low reproducibility need to be understood and the reliability of the LPBF process should be optimized in future studies. The hardness level of LPBF-produced hard metal is lower compared with conventionally produced hard metal. According to a previous study,\cite{23} the Vickers hardness for a cobalt content of 17 wt% and coarse tungsten carbides lies in the range of 900–1000 HV10.

3.3. Cutting Performance

For machining the brass alloy, the additively manufactured tool showed a sufficient cutting-edge stability and almost no tool wear after 20 broaching strokes. The chip formation and the cutting forces were indistinguishable to standard cemented carbide tools. In machining of TiAl6V4 and Inconel 718, a high width of flank wear land (VB) already occurred after the first cut, which did not allow a continuation of the machining tests, as shown in Figure 8. The tool wear can be described as a combination of abrasive wear and cutting-edge chipping. In case of TiAl6V4, the chipping tendency of the cutting edge can be attributed to the strong elastic formability of the workpiece material. To reduce the tool wear, an increase in hardness of the cemented carbide in combination with an optimization of the geometric cutting-edge design as well as proper coating is necessary.

4. Conclusion and Outlook

With preheating of the build platform up to 900 °C, it was possible to build crack-free tools of WC–17Co by the additive manufacturing process LPBF. The achieved density, the microstructural, and mechanical properties are lower compared with the properties of conventional WC–Co hard metal. Laser energy input effects the Vickers hardness, the overall cobalt content, and phase composition. The observed finer microstructure in the

![Figure 7. Vickers hardness along the BD for samples with different volume energy densities.](image)

![Figure 8. Chip formation and flank wear when machining different work piece materials.](image)
contour area leads to the hypothesis, which a lower laser energy input is favorable for a homogeneous, fine, and dense microstructure. In further research, the influence of higher powder-bed preheating temperature and simultaneously reduced energy input should be investigated. Moreover, further investigation will consider the influence of porosity of the agglomerated powder, which is probably an influencing factor for the porosity of the resulting additively built parts. For machining the brass alloy, the additively manufactured tool showed a sufficient cutting-edge stability and almost no tool wear. For the titanium alloy TiAl6V4 the additively built parts. For machining the brass alloy, which is probably an influence should be investigated. Moreover, further investigation will consider the influence of porosity of the agglomerated powder, which is probably an influencing factor for the porosity of the resulting additively built parts. For machining the brass alloy, the additively manufactured tool showed a sufficient cutting-edge stability and almost no tool wear. For the titanium alloy TiAl6V4 the additively built parts. For machining the brass alloy, which is probably an influence should be investigated. Moreover, further investigation will consider the influence of porosity of the agglomerated powder, which is probably an influencing factor for the porosity of the resulting additively built parts. For machining the brass alloy, the additively manufactured tool showed a sufficient cutting-edge stability and almost no tool wear. For the titanium alloy TiAl6V4 the additively built parts. For machining the brass alloy, which is probably an influence should be investigated. Moreover, further investigation will consider the influence of porosity of the agglomerated powder, which is probably an influencing factor for the porosity of the resulting additively built parts. For machining the brass alloy, the additively manufactured tool showed a sufficient cutting-edge stability and almost no tool wear. For the titanium alloy TiAl6V4 the ad

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Conflict of Interest
The authors declare no conflict of interest.

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