Comparative investigations into microstructural and mechanical properties of as-cast and laser powder bed fusion (LPBF) fabricated duplex steel (1.4517)

Vergleichende Untersuchungen zu mikrostrukturrellen und mechanischen Eigenschaften von Gussstahl und durch Laser-Pulver-Bett-Schmelzen hergestelltem Duplexstahl (1.4517)

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A methodology is presented that compares the microstructural and mechanical properties of as-cast and additive-made ferritic-austenitic duplex steel 1.4517. Microstructure of approximately equal amounts of ferrite and austenite measured in as-cast material could not be replicated in post heat-treated laser powder bed fusion samples after 30 min and 60 min of post heat treatment. This is attributed to nitrogen loss during powder atomization which left fewer austenite formers. Post-heat treated laser powder bed fusion samples of duplex structure had its austenite content repeatedly adjusted between 38 % and 40 %. As-built laser powder bed fusion tensile specimens which had a ferritic microstructure recorded high tensile and yield strength but had very poor elongation. Post heat-treated duplex laser powder bed fusion tensile specimen built in both horizontal and vertical orientations had good tensile and yield strength comparable to conventional casting processes; Tensile strength – 739 MPa (horizontal), 759 MPa (vertical); Yield strength (\(R_{p0.2}\)) – 489 MPa (horizontal), 525 MPa (vertical). The horizontally built duplex specimen had a very high elongation of 32 % than the vertical (11 %) or conventionally reported (22 %). This work establishes the 1.4517 duplex steel as a good candidate with good mechanical properties when processed by additive manufacturing.

Keywords: Duplex-steel / additive manufacturing / microstructure / tensile test / Calphad method

Schlüsselwörter: Duplex-Stahl / additive Fertigung / Mikrostruktur / Zugversuch / Calphad-Methode

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1 Introduction

With increasing use of laser powder bed fusion manufactured components, it is obvious that there is the need to expand the material range for this generative manufacturing method especially with duplex steels whose good strength and toughness properties are usually obtained by targeted adjustment of the microstructure by conventional production routes. The use of duplex steels can be attractive if they lead to cost and raw material reduction by replacing nickel-base alloys. Duplex steels, such as UNS31803 and SAF 2507 have already been investigated with regards to their microstructure [1–5].

In this study, the cast steel GX2CrNiMoCuN25-6-3-3 (1.4517) is considered for the first time as another variant of duplex steels. Compared to the steel SAF 2507, this steel is made from a less expensive alloy design approach. While it also contains copper and the same chromium content, it has a significantly less nickel, molybdenum and manganese.

Duplex steels are typically set to equal proportions of ferrite and austenite. While the austenite ensures toughness and corrosion resistance, the ferrite is mainly responsible for the increase in strength [2].

Based on chemical composition, microstructure components can be predicted through the use of constitutional charts such as the Schaeffler or the DeLong chart. Alloying elements are usually grouped into ferrite and austenite formers and their effect on phase transformation is determined. Although these charts are known to have originated from welding high alloy stainless steels, they have also shown to be applicable in other processes as well. Constitutional charts are not readily applicable in laser powder bed fusion production because unlike the casting process, its temperature profile is quite different and the steels are sensitive to changes in chemical composition. For this reason, predictions are usually done by performing simulations. Theoretical calculations are however performed initially to determine the formation and expected equilibrium phases. The focus of this work is therefore to practically investigate whether the 1.4517 duplex steel has suitable mechanical properties using laser powder bed fusion production route through a tailoring of the microstructure.

2 Experiments/experimental procedure

The starting material (cast ingot) used for the powder production was the cast stainless steel 1.4517 (GX2CrNiMoCuN25-6-3-3) obtained from Otto Junker GmbH. The nominal composition of the alloy (wt.%) has been given, Table 1. The cast ingot was gas atomized at Nanoval GmbH & Co. KG under argon atmosphere through a crucible process to powder for the laser powder bed fusion process. The particle diameters were between 15 μm and 51 μm. With a monomodal particle distribution, the average particle diameter was 29 μm. While the higher silicon/manganese ratio in this alloy gives an indication of a better castability compared to duplex steel SAF 2507, copper was alloyed to improve specific corrosion properties [1, 6]. Based on this chemical composition, simulations both at equilibrium solidification and non-equilibrium solidification using the Calphad method, ThermoCalc (TCF8. Steels/Fe-Alloys v.8.0 database) were performed.

With light optical microscope and scanning electron microscope, the initial as-cast material and
laser powder bed fusion feedstock powder particles were characterized. Among several etchants, Bera-ha II was selected in order to reveal the ferrite and austenite components [7]. The nitrogen content of the powder and the laser powder bed fusion built sample was measured by carrier gas heat extraction (ON-Mat 8500, Ströhlein Laboratory, Measurement and Environmental Technology).

From the powder material, test cubes of 10 mm × 10 mm × 10 mm dimensions were produced at the Fraunhofer Institute (ILT Aachen) using the laser powder bed fusion process, and their microstructure was likewise characterized by both light and scanning electron microscope coupled with energy dispersive spectroscopic detector, INCA X-Sight 7426, Oxford Instruments. Each of the 16 fabricated test cubes were printed with a different parameter set. From the 16 different parameter sets, the optimal parameters which yielded the densest sample were determined by light optical pore measurement. The printing which took place in an inert argon atmosphere was performed using a bidirectional X-Y scanning strategy at 900 mm/s scanning speed and 200 W laser power. With a constant layer thickness of 30 μm and a hatch distance of 100 μm, an average density of 99.95 % was recorded. The density was determined with the software “analySIS pro” on unetched samples under 25x magnification.

Using the optimized parameters, 8 mm × 8 mm × 8 mm dimensional test cubes and mechanical test specimen were manufactured using an Aconity-Mini Laser Powder Bed Fusion equipment. Blanks for tensile specimen were fabricated in both XY (horizontal) orientation, perpendicular to the building direction and Z (vertical) orientation which was parallel to the building direction. Six tensile specimens were fabricated in total, three for each orientation. This was done to probe the mechanical anisotropic properties. The tensile blank specimen were separated from the powder bed by wire erosion. Tensile tests were carried out on the tractor INSTRON 8033 with a crosshead speed of 0.3 mm/min according to the B4 DIN 50125 2009-07 standard. The transmission electron microscopic characterization of the powder particles and of the laser powder bed fusion sample microstructure was carried out using the FEI Tecnai F20 TEM microscope.

Microstructural phases and defects were also determined by electron backscatter diffraction technique and light-optical method (ImageJ software from Fiji) respectively. Phases that could not be quantified due to their low level concentrations (<5 %), such as chromium nitride, small local austenite phases, production-related pores and cracks were identified as so-called zero-solutions and were not taken into account in determining the ferrite and austenite phase compositions.

Post-heat treatments at 1120 °C were performed on two laser powder bed fusion sample sets (sample size: 10 mm × 10 mm × 3 mm) for 30 min and 60 min, respectively, followed by water quenching. The temperature fluctuations in the furnace used was 25 °C.

3 Simulation results

3.1 Thermodynamic equilibrium calculation

In order to gain a better insight into the precipitation and phase formation of the 1.4517 steel, based on its as-cast chemical composition and equilibrium phase diagram, the post-heat treatment temperature was chosen, Figure 1a. The heat treatment targeted adjustment of the austenite and ferrite proportions in the microstructure and the dissolution of material embrittling intermetallic phases present in the cast material. The formation of embrittling phases such as σ-phase which is promoted without heat treatment is to be avoided as it reduces the corrosion resistance due to chromium and molybdenum depletion in the matrix.

From the phase diagram, the phase formation temperatures were read, Figure 1a. As it can be read on the phase diagram, the possibility of attaining a microstructure of equal ferrite and austenite proportion only lies between 900 °C and 1200 °C phase transformation temperatures, but not without the simultaneous formation of the undesirable embrittling chromium nitride. In principle, the achievement of the desired microstructure with its proportionate phases: austenite, ferrite and the other various precipitates, depend on its rate of formation. This is in turn determined by the local chemical composition, the initial structure, the temperature, the holding time and the quenching rate.
Another phase diagram was simulated for the laser powder bed fusion built samples whose chemical composition when measured, was found to be slightly different from that of the starting as-cast material, Figure 1b. This phase diagram was simulated to probe the effect of the chemical composition variation on phase formation. From the phase diagram, after post-heat treatment of the laser powder bed fusion sample at 1120 °C, the expected austenite proportion is 45 %. Since the cooling rates for laser powder bed fusion production are between $10^5$ K/s and $10^7$ K/s and therefore do not represent equilibrium cooling, a Scheil-Gulliver diagram was simulated to show the temperature sequence of the phase formation with rapid cooling [2], Figure 1c. From the non-equilibrium simulation, it can be seen, that austenite formation begins at 1360 °C and that the precipitation of hexagonal chromium nitrides is to be expected from a temperature of about 1300 °C, Figure 1c. Phases with negligible proportions such as manganese sulfide and metal carbides were also calculated, but was not the subject of this investigation.

### 3.2 Theoretical calculations

In practice, several methods for determining phase components are available, such as the De-Long and various versions of the Schaeffler diagram. According to the known formulas for both nickel and chromium equivalents, and with the given as-cast composition, chromium and nickel equivalents of 28.04 and 12.32 were respectively calculated. For the heat treated cast material, the general Schaeffler diagram showed, despite the inclusion of nitrogen, a ferrite content of 40 %, which was too low compared to the experimental measurements of 52 %, Figure 2c. Using the Schoefer diagram according to ASTM A800 (variant 1 according to AVS D63 A), the ferrite content of the heat-treated cast material was calculated to be 51 % using a ferrite index of 1.737, which was fairly consistent with the experimentally measured ferrite content (52 %) [8], Figure 2b–c.

The ferrite content of the heat-treated laser powder bed fusion sample was estimated with the help of Schaeffler diagram and the Schoefer diagram. While the first estimated a ferrite content of 56 % (chromium equivalent of 27.54, nickel equivalent of 10.32), the second calculated a ferrite content of 68 % (ferrite index of 1.951). The actual experimental measurements of the ferrite content lay between 60 % and 62 %. The influence of the nitrogen content on the nickel equivalent and thus on the prediction of the austenite content was taken very seriously into account.

The variation in the values when using these calculation aids lie on the exactness of the chemical composition as well as the different multiplicators of the alloying elements.
The as-cast material had a ferritic-austenitic microstructure with finely distributed unwanted intermetallic phases (black spots), Figure 2a. These were dissolved after annealing at 1120 °C, Figure 2b–c. After the heat treatment, the ferrite content was 52 % and the austenite, 48 % (Norwegian standard of petrochemical NORSOK Standard M-650 requirement), Figure 2c. This standard allows a ferrite content of between 35 % and 65 % since the determination of the phase fractions is challenging due to the strong dependence on the measurement position and sample section [9]. These results were fairly consistent with the predictions of the Schaeffer diagram (ferrite content 51 %) and the Thermo-Calc® simulation; ferrite content 52 %, Figure 1a. Energy dispersive spectroscopic analysis of the laser powder bed fusion powder feedstock showed only a single matrix phase with even distribution of elements (in contrast to metal powders such as Inconel or austenitic chromium-nickel steel powders) with a measured nitrogen content of 0.149 wt.%, Figure 2e. Higher magnifications using transmission electron microscope also detected aluminum and silicon-containing mixed oxides in the matrix.

### 4.1 Laser powder bed fusion test cube structure

#### 4.1.1 As-built laser powder bed fusion test cube microstructure

The rapid cooling (between $10^4$ K/s and $10^6$ K/s) associated with laser powder bed fusion process suppressed austenite formation to form a nitrogen supersaturated ferrite with nitride precipitates on the melt lines and grain boundaries. With electron backscatter diffraction measurements, various etching methods and scanning electron microscopy, an almost ferritic matrix was determined, Figure 3. In the ferritic microstructure, the meltlines were clearly visible, but no preferential growth orientation was observed.

The light and scanning electron micrographs show a ferrite matrix with chromium nitrides on the meltlines and grain boundaries, Figure 3b–c. These chro-
mium nitrides formed due to the nitrogen super-
saturation in the ferrite as a result of the rapid cooling
and the high diffusion rate of the nitrogen [2–3, 10–
11]. The plate-like formation of chromium nitrides on
the grain boundaries is due to its hexagonal structure
and its relatively low coherence with the ferrite [12].
Other authors also reported primary austenite on the
grain boundaries and showed with selected area dif-
fraction pattern method that austenitic phases could
also exist in nitrogen-enriched areas [1–2]. From the
Scheil-Gulliver simulation, both phases are to be ex-
pected, Figure 1c.

Transmission electron micrographs of the as-
built ferritic test cube show grain boundary cover-
ages and nanometer-sized round silicon and mixed
oxides formation due to local supersaturation as
well as Chromium nitrides visible on the grain
boundaries, Figure 4a–b.

4.2 Microstructure of post-heat-treated laser
powder bed fusion test cube

After heat treatment (1120 °C 1 h/water), the elon-
gated chromium nitrides dissolved and were no lon-
ger detectable on the grain boundaries, Figure 4b.
Other precipitates were retained. As in the powder
and the ferritic as-built sample, this heat-treated
sample also contained mixed oxides with silicon,
aluminum, chromium and manganese. Since the
chromium nitrides dissolved above a temperature of
1000 °C, they served as nucleating sites for the aus-
tenite, because, it has a higher solubility for nitrogen
than the ferrite [13]. Austenite, also called iso-
thermal austenite, formed heterogeneously at the
ferrite and chromium nitride interfaces, Figure 5a.

The orientation relationship between the three
phases: ferrite, austenite and the chromium nitride
on the grain boundaries behaves in a way that the
densest crystal planes of all three phases are par-
allel [11]:

\[(110) \alpha || (0001) Cr_2N || (111) \gamma\]

The interfacial energy is therefore low and facil-
itates the austenite formation. Under these orien-
tation relationships, austenite could also form
within the ferrite grains when the ferrite is depleted
of chromium by chromium nitrides [1]. As a result
of the heat treatment, new grain structure was
formed from the movement and the arrangement of
the dislocations, Figure 5b. Depending on the sam-
ple section, electron backscatter diffraction meas-
urements showed a ferrite content between 60 %
and 62 %. The measured austenite content ranged
between 38 % and 40 %. After annealing, the melt-
lines were still clearly visible due to austenite for-
mation (light phase), Figure 6a–b.

From the light optical micrographs, in contrast
to the as-cast, the laser powder bed fusion built
sample is seen to have a much finer grain structure,
Figure 2c. The samples were annealed for 30 min
and 60 min. After 30 min of annealing, not more
than 40 % austenite content was measured, Fig-
ure 6a. 60 min annealing led to a coarsening of the
austenitic phases although up to 40 % austenite was
equally measured, Figure 6b. The laser powder bed
fusion samples consistently recorded lower austen-
ite content than the cast samples. While the meas-
ured austenite content of the cast samples was con-
sistent with the simulated phase diagram, this was
not the case for the laser powder bed fusion sam-
ples. Since the laser powder bed fusion samples
showed a very fine-grained microstructure, there obviously must have been sufficient nucleation sites for the austenite formation due to the high dislocation density and nitride precipitates, the phase formation of duplex steels is primarily dependent on the chemical composition.

The measured chemical composition of the laser powder bed fusion sample shows a drop in nitrogen concentration compared to the as-cast starting material (as-cast: 0.203 %, powder and laser powder bed fusion component 0.148 %), Table 1. The drop in nitrogen concentration can be attributed to nitrogen losses during the powder production stage. From the phase diagrams, it is clearly seen that the application of the initially simulated phase diagram of the as-cast starting material to the building of the laser powder bed fusion samples will be erroneous due to the occurrence of a phase shift as a result of the loss of alloying elements, Figure 1a. The desired phase fractions can only be attained in a narrow range of temperature, owing to the drop in nitrogen concentration, Figure 1b. This necessarily led to a steeper rise and fall of the austenite or ferrite curves at an annealing temperature of 1120 °C,
so that temperature fluctuations had a more extreme effect on the proportion of the phases than on the as-cast chemical composition. The temperature range necessary for adjusting equal proportions of ferrite and austenite was therefore very limited and ranged between 970 °C and 1015 °C, Figure 1b. It was however between 900 °C and 1200 °C (optimally 940 °C and 1100 °C) in the as-cast, Figure 1a. From the equilibrium phase diagram simulation of the Laser Powder Bed Fusion samples at 1120 °C, the expected phases are 42 % austenite and 58 % ferrite, Figure 1b. Electron backscatter diffraction and light microscopic measurements of the austenite proportions were however between 38 % and a maximum of 40 %. The variation in the exact phase fraction measurement is not only attributed to laser powder bed fusion fabrication process but also from local alloying differences which led to uneven distribution of the phases, Figure 6a–b. Although a small magnification in the measurement covers a larger measurement area, yet it does not detect phases with small proportions. Samples with higher magnification can be measured more accurately, but smaller image sections lead to poorer statistical evaluation. This applies to both the electron backscatter diffraction and light microscopy measurements. Higher content of austenite formers offer a greater degree of freedom from temperature and also lead to higher strength such as through interstitial incorporation of nitrogen and substitutional mixed crystal formation in the ferrite. With knowledge of the phase diagram, temperatures could therefore be selected specifically for microstructure optimization if the design of the alloy composition takes into account the nitrogen losses due to atomization.
5 Tensile test

Mechanical properties are primarily material-dependent. Among the microscopic factors are the lattice structure, crystal orientation, the orientation between phases, particle sizes, particle types, quantity and distribution of precipitates, segregation, substitutional and interstitial atoms, dislocations and their sliding systems. Manufacturing-related macroscopic factors are lack of fusion, cracks, porosity, laser scanning parameters, the microstructure formed by the scanning process and the properties of the meltlines, which generally differ from the matrix properties.

The unannealed ferritic specimen recorded an ultimate tensile strength of about 1100 MPa and a yield strength at R$_{p0.2}$ of 1035 MPa and an elongation at break value of 8 %. The good strength values were achieved despite the relatively low manganese content of 0.6 % due to the fine-grained microstructure of the additive production. The poor toughness is due to the brittle chromium nitrides on the grain boundaries and melt lines between the individual melt layers, Figure 4a.

The building direction was parallel to the tensile load direction for the ferritic tensile test specimen fabricated in vertically orientation. The break occurred between the melt lines of the individual layers.

In the case of the specimen fabricated in horizontal orientation, the tensile load direction was perpendicular to the building direction, a better fusion, parallel to the tensile loading direction was visible. The tensile specimen fabricated in horizontal orientation had a higher ductility than the specimen fabricated in vertical orientation.

The material-dependent anisotropic properties of laser powder bed fusion fabricated components depend on the scanning strategy and the building direction, which was particularly evident in the heat-treated samples with duplex structure. Here, the greater toughness of horizontally built tensile test specimen is significant. Apparently, there was a better bonding and bonding of the melting traces, Figure 7a. This bonding is significantly influenced by the microstructure of the melt lines, since this structure differs from the matrix microstructure. Basically, segregation phases and coarser crystals with different crystal orientations show up in the melt lines [14].

The duplex structure was formed from the ferritic microstructure after heat treatment at 1120 °C. The embrittling chromium nitride precipitations on grain boundaries and melt lines dissolved [13]. The austenite formed from nitrogen-rich phases and chromium nitrides. Austenite, with its face-centered cubic crystal structure and the associated higher number of slip systems, influences the ductility of the duplex steel, while the strength of a duplex steel is due to the ferrite.

Since the only difference in the duplex specimen was the building direction, the anisotropy of the ductility can be explained by the growth orientation of the austenite and the crystal structure in the melt lines. The austenite phase can form active slip systems in the direction of pull on the melt lines. The fused lines are embedded in a ductile matrix, as shown by the fracture pattern. Because of the scanning strategy, the austenitic phases were not interrupted by ferritic phases in the tensile load direction, Figure 7a.

After the heat treatment at 1120 °C, 1 h, specimen fabricated in horizontal orientation recorded an ultimate tensile strength of 739 MPa and a yield strength at R$_{p0.2}$ of 489 MPa. The elongation at break was 32 %. The tensile specimen fabricated in vertical orientation recorded an ultimate tensile strength of 759 MPa and a yield strength at R$_{p0.2}$ of 525 MPa. The elongation at break was 11 %, Figure 8. The ultimate tensile strength of conventionally produced duplex steels is between 600 MPa and 800 MPa and yield strength between 400 MPa and 550 MPa, with hardness given as 200 HV 1 to 250 HV 1 [15].

The samples after 30 min of annealing exhibited a finer phase distribution and higher hardness values (284 HV 0.2) than the longer annealed samples (269 HV 0.2). Since the strength can be adjusted via the grainsize of the microstructure, it can be assumed that a shorter heat treatment time (30 min instead of 60 min) at the same temperature leads to higher strength values. In order to guarantee a fast heating of the workpiece and a fast dissolution of the chromium nitrides, which become unstable from approximately 1000 °C, a temperature of 1120 °C is suitable. The annealing time must be optimized.

With the knowledge of the phase diagram, it is possible to optimize the mechanical properties of the steel through its alloy content, annealing temperature and duration. A higher content of austenite formers offers greater freedom for choosing a temperature to
achieve a microstructure of equal proportion of ferrite and austenite, which also leads to better mechanical properties due to the increased solid solution hardening. It can therefore be understood that the local chemical composition of the laser powder bed fusion workpiece affects the local austenite content and results in an uneven austenite distribution.

The choice of the laser scanning strategy not only affects the ductility, but also influences the strength. Due to different scanning strategies, the temperature load of the individual tracks is very different and different materials react differently due to their individual solidification behavior in terms of structure and precipitation. Therefore, further studies will deal with the complex precipitation behavior of the duplex steel, especially at the melt lines: Transmission electron microscope analyses shall characterize the microstructure and compare it with simulation results that predicted small amounts of the Laves, $\sigma_1$ and $\sigma_2$ phases, metal carbides, titanium nitride and others in addition to nitride precipitation. In addition, controlling the formation and the contribution of the mixed oxides to the strength will to be investigated.

6 Results

The cast steel 1.4517 is basically suitable for laser powder bed fusion production after post-heat treatment. The atomization of the starting cast material into powder and the additive production step may alter the alloy composition, which must be taken into account and compensated for during the alloy design. In this alloy particularly, the nitrogen content has to be taken into consideration during the alloy design.

The mechanical properties are strongly dependent on the building orientation of the sample and not only on the constituent phases. Duplex tensile specimen fabricated in horizontal orientation showed after the heat treatment much better tough-

Figure 7. Duplex laser powder bed fusion tensile specimen: (a) Horizontally made, tensile load direction perpendicular to the building direction. (b) Vertically made, tensile load direction parallel to the building direction. Fractured duplex tensile specimen laser powder bed fusion structure: (c) Tensile load direction perpendicular to the building direction, austenite white (d) Tensile load direction parallel to the building direction, austenite light blue.
ness than the vertical built specimens. The elongation at break was 32%, the ultimate tensile strength was 739 MPa and the yield strength at $R_{p0.2}$ was 489 MPa, Figure 8. The non-heat treated tensile specimen with ferritic microstructure recorded ultimate tensile strength of 1100 MPa and yield strength at $R_{p0.2}$ of 1035 MPa with very poor elongation of 8%. These good strength values were achieved despite the relatively low manganese content of 0.6%. The mechanical properties of laser powder bed fusion duplex steel 1.4517 are convincing compared to cast material (Data provided by supplier: $R_{p0.2} \geq 480$ MPa, $R_m$ between 650 MPa–850 MPa, $A_5 \geq 22\%$).

Electron backscatter diffraction analysis showed that the as-built laser powder bed fusion microstructure was almost completely ferritic. Light optical microscope investigations showed isolated chromium nitrides on the grain boundaries. After 30 min annealing at 1120 °C and subsequently water quenching, a two-phase ferritic-austenitic structure was formed. The extension of the annealing time led to coarsening of the austenitic phases and an eventual loss of hardness.

To reveal structural constituents such as ferrite, austenite, chromium nitrides, grain boundaries and melting traces, the optimal method was selected from various etching methods, color etching and erosive etching. Etching with Beraha II was suitable for optical determination of the ferritic and austenitic phases.

The average measured austenite content of the laser powder bed fusion samples was between 38% and 40%, locally maxima of 45% could be measured, which was still lower than that of the cast material (48%). This is due in particular to the loss of alloying elements and, in particular, the loss of nitrogen due to powder atomization.

Different phase diagrams for the casting and the laser powder bed fusion sample chemical compositions had to be simulated to show the influence of the change in the chemical composition on the microstructure. With the loss of nitrogen, a narrowing of the temperature range and an increased sensitivity of the material in relation to the phase portion compared to temperature changes were observed. Under equilibrium conditions, with the given composition and the annealing temperature of 1120 °C,
in the laser powder bed fusion sample an austenite amount of only 40% can be achieved. In the as-cast chemical composition however, about 50% austenite could be attained within a temperature range of 920 °C and 1120 °C as against 950 °C and 1020 °C for the laser powder bed fusion material. The smaller the temperature range, the stronger the influence of the temperature on the equilibrium portion of the ferrite and austenite phases. It can therefore be explained that the local chemical composition of the laser powder bed fusion workpiece influences the local austenite content and leads to an uneven austenite distribution.

A higher content of austenite formers offer greater freedom from temperature and leads to higher strengths. Only with knowledge of the exact chemical composition of the as-built laser powder bed fusion workpiece, that with the help of equilibrium diagram, the appropriate heat treatment temperature can be chosen.

In order to achieve a microstructure of equal proportions of austenite and ferrite structure at a certain temperature, the alloy composition must be adapted. Using simulated phase diagrams, the alloy composition, annealing temperature and annealing time can be optimized. An increase in strength can be achieved by optimizing the annealing temperature and duration.

The Schoefer diagram gives a useful approximation for the cast material, but calculated too low austenite content for the laser powder bed fusion-fabricated material. The Schaeffler diagram cannot be used for the cast material.

Round mixed oxides with a maximum size of 100 nm were found in all three process stages (powder, ferritic and duplex laser powder bed fusion samples) both in the grain and on the grain boundaries. A targeted investigation of the influence of these particles on the material properties could not be done within the framework of these investigations.

7 Conclusion and outlook

Using the duplex steel 1.4517 as an example, it has been shown that the additive process route offers a high potential in terms of setting suitable material properties, but requires careful alloy and process development. The atomization of the starting material into powder and the subsequent generative additive step can alter the alloy composition. The alloy design for additive manufacturing must therefore take these effects into account and must be tailored precisely to the generative methods and the process parameters.

The reasons for the anisotropic properties due to the scanning strategy and building orientation of the sample in the laser powder bed fusion process must be specifically investigated and appropriate strategies for the entire process chain should be developed from this.

Furthermore, the intrinsic formation of fine oxide particles in the laser powder bed fusion process provides a potential source of increase in strength. Their influence on the material properties and the specific adjustability through process control will be determined in further investigations.

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