

Performance Evaluation of Directed Energy Deposited (DED) Built-up of Titanium (Gr-2) Alloy using Metallurgical and Mechanical Properties

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Abstract

Wire arc additive manufacturing (WAAM) enables near-net-shape fabrication of titanium components, but different metal-transfer modes can produce distinct thermal histories and build-height-dependent properties. This study comparatively evaluates commercially pure Titanium Grade 2 deposited using Cold Metal Transfer (CMT) and Pulse Multi Control (PMC) under their respective selected processing conditions. Two 20-layer deposits were fabricated using 1.2 mm titanium wire and characterized for microstructure, hardness, tensile behavior, and Charpy impact response. The calculated heat inputs were 248.4 and 473.6 J/mm for CMT and PMC, respectively. The characteristic microstructural feature size increased from the top toward the bottom of both deposits, ranging from 2.35–3.46 μm for CMT and 1.97–3.20 μm for PMC. PMC exhibited higher average hardness (324 vs. 298 HV) and ultimate tensile strength (527.4 vs. 516.2 MPa), although the higher average UTS of PMC was governed by a single bottom-region specimen, whereas CMT showed slightly higher average yield strength (500.2 vs. 493.8 MPa) and Charpy energy (0.80 vs. 0.76 J). The results reveal pronounced spatial variation in microstructure and mechanical response and indicate the importance of process-dependent thermal history in WAAM-fabricated commercially pure titanium, while the elevated hardness and limited ductility of both deposits indicate that possible interstitial pick-up requires further investigation.

Keywords: Directed energy deposition, Wire arc additive manufacturing, Cold metal transfer, Pulse multi control



The Technology Readiness Level (TRL) of this research is Level 3 (Proof of Concept). The concept was demonstrated through experimental fabrication and characterization of Titanium Grade 2 using WAAM with CMT and PMC modes. Metallographic, microhardness, tensile, and Charpy impact tests revealed variations in microstructure and mechanical properties along the build height. These findings provide initial experimental evidence of the feasibility of both deposition modes. Further independent builds, process optimization, and thermal characterization are required to validate reproducibility and support progression toward higher TRL stages.

1. Introduction

Titanium is employed in various high-performance applications owing to its excellent strength-to-weight ratio, corrosion resistance, and thermal stability compared with many conventional structural metals [1]. The aerospace sector is one of the principal consumers of titanium mill products. In aerospace structures, titanium is also favoured because of its galvanic compatibility with polymer matrix components and ability to be made into compact designs needing high strength and constrained dimensions [1], [2], [3]. However, the high cost of titanium and the challenges associated with its conventional processing can limit material utilization [4], [5]. Additive Manufacturing (AM) provides an alternative manufacturing route because material can be deposited close to the required geometry with reduced material wastage [6]. AM produces components through layer-by-layer addition of material and can reduce material wastage compared with conventional subtractive manufacturing approaches [7], [8].

Directed Energy Deposition (DED) is an AM process in which metallic feedstock in powder or wire form is delivered to a localized energy source and deposited onto a substrate [9]. Wire Arc Additive Manufacturing (WAAM) is a wire-fed form of DED in which an electric arc provides the energy required for melting and deposition [10], [11]. The process offers high deposition efficiency and material utilization compared with several other metal AM routes [5], [12]. In WAAM, the wire feedstock is continuously supplied to the arc, while the previously

deposited material can undergo partial remelting during subsequent layer deposition [10], [13]. WAAM is therefore particularly relevant for producing large metallic structures where deposition rate and material utilization are important considerations [14].

Despite its advantages, WAAM can exhibit limitations associated with surface quality, dimensional accuracy, residual stress, porosity, and other process-dependent defects [5], [7]. More importantly for titanium deposition, the repeated thermal cycles inherent to layer-by-layer fabrication produce a non-uniform thermal history through the build height, which can influence subsequent microstructural evolution and mechanical response. Among arc-based WAAM approaches, Cold Metal Transfer (CMT) uses a controlled short-circuit metal-transfer mechanism and is generally associated with comparatively low thermal input [12], [15], [16], [17]. These characteristics can influence melt-pool behavior, deposition conditions and the thermal exposure experienced by previously deposited material. Pulse Multi-Control (PMC) employs a pulsed spray-transfer mode and can operate at comparatively higher thermal input than CMT under the selected deposition conditions [17].

Commercially pure Titanium Grade 2 is predominantly an α -titanium material and is used where a combination of corrosion resistance, ductility and specific strength is required [18]. During layer-by-layer deposition, however, material deposited in earlier layers is repeatedly exposed to subsequent thermal cycles. The resulting reheating, cooling and heat dissipation conditions vary with build height and can therefore contribute to spatial variations in microstructure and mechanical properties.

Previous studies have established the feasibility of fabricating commercially pure titanium by wire-arc additive manufacturing and have examined its microstructure, defects and mechanical properties [19], [20], [21]. Studies on Ti Grade 2 have also demonstrated that the deposited material is strongly influenced by the thermal conditions associated with layer-by-layer deposition and subsequent processing [20], [21]. However, the available literature identified in the present study provides limited direct comparison of CMT and PMC deposition conditions for commercially pure Titanium Grade 2 within a common experimental framework. In particular, the relationship between the selected deposition condition, position-dependent thermal history, microstructural evolution and mechanical response through the build height requires further examination.

The present study therefore investigates the process–microstructure–property relationship of commercially pure Titanium Grade 2 deposited using CMT and PMC modes. The comparison is made between the respective selected operating conditions rather than treating transfer mode as an isolated variable, because voltage, current and wire-feed rate also differ between the two deposition conditions. Particular attention is given to the variation in microstructure and mechanical properties with build height, where each region experiences a different cumulative thermal history arising from repeated reheating, cooling, heat dissipation and thermal cycling during subsequent layer deposition. Five tensile specimens were extracted from different heights of each build to characterize this through-height variation; these specimens represent spatial subsamples of the corresponding build rather than independent experimental replicates.

2. Materials and Methods

2.1. Experimental procedure

For this study, commercially pure titanium (CP-Ti) designated as ASTM Grade 2 was used in the form of 1.2 mm diameter wire. The chemical composition of the Grade 2 titanium wire used in the investigation is presented in Table 1, while its reported physical, thermal, and mechanical properties are listed in Table 2. For WAAM deposition, 100% argon was used as the shielding gas at a flow rate of 15 L/min, with a pre-flow time of 0.4 s and a post-flow time of 0.6 s. A titanium plate measuring 200 × 50 × 5 mm was used as the substrate. Two separate deposits were fabricated using Cold Metal Transfer (CMT) and Pulse Multi Control (PMC) metal-transfer modes. The selected process parameters used for the CMT and PMC deposits are presented in Table 3. For both deposits, the deposition system consisted of a KUKA KR10 R140 industrial robot (KR Cybertech nano series) integrated with a Fronius CMT 400 TPSi power source for CMT deposition and a PMC deposition using CMT and pulsed module. The welding torch and CTWD were maintained at the preset values specified in the welding-system software. The deposition was performed using a unidirectional travel pattern. The interlayer waiting time was maintained until the temperature of the previous bead decreased to 50 °C.

Table 1. Composition of titanium Gr-2

Element	Titanium (Ti)	Iron (Fe)	Oxygen (O)	Carbon (C)	Nitrogen (N)	Hydrogen (H)
% by weight	99.2	max 0.3	max 0.25	max 0.08	max 0.03	max 0.015

Table 2. Physical, thermal, and mechanical properties of titanium Gr-2

Property Type	Property	Value
Physical	Density	4.51 g/cc
Thermal Properties	Heat of fusion	325 J/g
	Specific Heat Capacity	0.523 J/g·°C
	Thermal Conductivity	16.4 W/m·K
	Melting Point (Max)	1665
	Liquidus	1665
Mechanical Properties	Beta Transus	913
	Knoop Hardness	170 (HK)
	Rockwell B Hardness	80 (HRB)
	Vickers Hardness	145 (HV)
	Ultimate Tensile Strength	344 MPa
	Yield Strength	270 MPa
	Elongation at break	20%
	Reduction of Area	35%
	Modulus of Elasticity	105 GPa
	Poisson's ratio	0.37

Table 3. Selected process parameters for CMT and PMC deposition techniques on a Grade 2 titanium substrate plate

Technique	Voltage(V)	Current(A)	Welding Speed (m/min)	Wire feed (m/min)
CMT	11.5	120	0.3	6.9
PMC	20	148	0.3	8.8

2.2. Specimen extraction and testing procedure

Both CMT and PMC deposits consisted of 20 deposited layers and had a deposition length of 155 mm (Table 4). Specimens for mechanical and metallurgical characterization were extracted from the deposited walls using wire electrical discharge machining (WEDM). For tensile testing, five specimens were extracted from the central region of each deposit along the longitudinal deposition direction. These five tensile specimens were spatially distributed along the build height and were treated as subsamples representing different through-height locations rather than as five independent experimental replicates. Their purpose was to evaluate the variation in mechanical properties associated with differences in local thermal history, including the number of subsequent deposition cycles, reheating, cooling, and heat accumulation experienced at different build heights. Specimens were extracted from the deposits for Charpy impact testing and metallographic characterization. The portions at both ends of the deposits were discarded to minimize the influence of non-uniform deposition conditions near the start and end regions.

For Vickers hardness testing, cross-sections of the CMT and PMC deposits were sectioned and polished to remove surface irregularities. Hardness measurements were performed using a Vickers indenter under a load of 0.5 kgf with a dwell time of 15 s. The indentation matrix was positioned 1.7 mm away from both the bottom and top edges of the specimen, with individual indentations spaced 1.75 mm apart. Tensile testing was performed in accordance with ASTM A 370-07a using a Tinius Olsen 100ST universal testing machine with a 100 kN capacity and 25 kN grip capacity at ambient temperature. Five sub-size tensile specimens were extracted from different locations along the build height of each deposit, with the longitudinal specimen axis parallel to the deposition direction. The specimens were designated T1–T5 from bottom to top. The reported test rate of 2 mm/min represents the crosshead/displacement rate rather than a calculated strain rate. Yield strength was determined using the 0.2% offset method from the engineering stress–strain response,

Table 4. Geometry and bead deposition characteristics of CMT and PMC

Sample	CMT	PMC
Length*Height*Width	155*65*5.5 (mm)	155*67*6 (mm)
Average bead height (reinforcement)/layer	3.09 mm	3.35 mm
Total number of layers deposited	20	20

and elongation was calculated using a gauge length of 25 mm. The average cross-sectional areas of the CMT and PMC tensile specimens were 20.616 and 20.741 mm², respectively. Sub-size Charpy V-notch specimens with dimensions of 55 × 5 × 5 mm were extracted and machined in accordance with ASTM A 370-07a. Measurement on five specimens has been employed while, the max and min value has been discarded and the average value from rest three specimens is used in the study to reduce the scattering effect. The specimens were oriented such that the notch was positioned in the specified deposit region and aligned with the welding/deposition direction. Fracture surfaces of selected tensile and Charpy specimens were subsequently examined by scanning electron microscopy (SEM) using JEOL-JSM-6390 at accelerating voltage 5kV.

2.3. Metallographic procedure

The deposits were sectioned into two halves to facilitate grinding, polishing, and metallographic examination, as shown in **Figure 1**. One half contained the top and middle-top regions, while the other contained the middle-bottom and bottom regions. The sectioning arrangement was used to examine microstructural variation along the build-height direction. The sectioned surfaces were ground and polished using progressively finer abrasive papers followed by diamond paste. For color metallography, Beraha's reagent was prepared by dissolving 2.5 g of ammonium bifluoride in 50 mL of distilled water. The polished specimens were immersed in the reagent for an appropriate duration to obtain sufficient color contrast. The etched microstructures were examined and imaged at magnifications of 50×, 100×, 200×, 500×, and 1000×.

Quantitative microstructural measurements were performed using ImageJ image-analysis software. The micrographs corresponding to the top, middle, and bottom regions of the CMT and PMC deposits were calibrated using the scale bar available in each micrograph. The characteristic dendritic/microstructural feature size was then measured from the calibrated images using the measurement tools in ImageJ. Multiple representative measurements

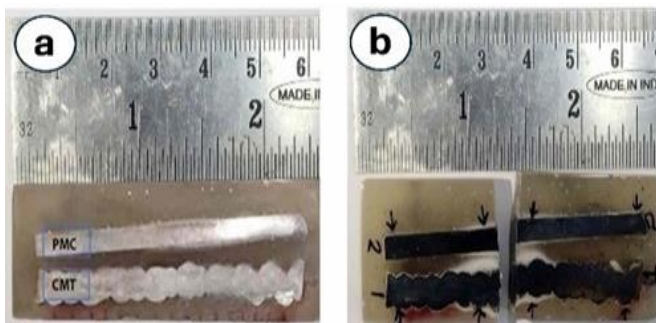


Figure 1. Photographic view of sample (a) before being sectioned, and (b) after being cut through the middle

were taken from each selected region and the corresponding average value accumulated from minimum of 30 measurements was used for the quantitative comparison. The measured values were 2.35, 3.16, and 3.46 μm for the top, middle, and bottom regions of the CMT deposit, respectively, and 1.97, 2.70, and 3.20 μm for the corresponding PMC regions. These measurements represent the characteristic microstructural/dendritic feature size obtained from the metallographic images and are not interpreted as crystallographic grain diameter.

3. Results and Discussion

3.1. Weld thermal cycle and heat input calculation

Since microstructural evolution and mechanical response are influenced by the thermal history of layer-wise deposition, consideration of the heat input provides a useful basis for comparing the two deposition conditions. The thermal history of WAAM is governed by several interacting factors, including electrical energy input, material deposition rate, component geometry, heat dissipation, and repeated reheating by subsequent layers. Therefore, the heat-input values calculated below are used as comparative process indicators rather than direct measurements of local cooling rate or temperature history. The nominal heat input per unit length was estimated from the electrical power and travel speed using Eq. (1):

$$Q = k \frac{V \times I \times 0.06}{S} \text{ (J/mm)} \quad (1)$$

where Q is the nominal heat input per unit length (J/mm), k is the assumed thermal efficiency, V is the voltage (V), I is the current (A), and S is the travel speed (m/min).

Using the process parameters listed in **Table 3**, the corresponding nominal heat-input values were calculated as Eq. (2):

$$Q(CMT) = 248.4 \text{ J/mm}, Q(PMC) = 473.6 \text{ J/mm} \quad (2)$$

The calculated nominal heat input for PMC was approximately 90.67% higher than that for CMT [22]. This difference indicates substantially different thermal-energy conditions between the two selected deposition regimes and provides a basis for interpreting the observed differences in microstructural morphology and mechanical response. In the present calculation, a thermal efficiency of 90% was assumed for CMT based on reported CMT-WAAM calculations [22], while an efficiency of 80% was adopted for PMC based on reported arc-based CMT/PMC welding conditions [17]. These values were used as calculation assumptions rather than experimentally measured efficiencies for the present Ti Grade 2 deposition system. It should also be noted that the cited efficiency values were reported for aluminium-alloy deposition and may differ for titanium. However, the calculated heat input does not by itself establish the local cooling rate or complete thermal history of the deposits.

3.2. Microstructure evolution and analysis

3.2.1. Top region

The microstructure of the DED deposits exhibited variation along the build-height direction, which can be associated with the different thermal conditions experienced by successive layers during deposition. Each deposited layer is subjected not only to its own solidification cycle but also to subsequent reheating by the deposition of overlying layers. The top region exhibited a relatively finer dendritic morphology than the middle and bottom regions [23]. The finer morphology may be associated with the thermal conditions prevailing near the final deposited layers, where subsequent reheating cycles are limited compared with material located below the deposition surface. The top region represents material deposited during the final stages of the build and therefore experiences fewer subsequent reheating cycles than regions located below it. The surrounding environment also provides a pathway for heat dissipation from the exposed deposition surface. These factors can contribute to differences in the thermal history of the top region; however, the present study did not directly measure the local temperature or cooling rate.

In both CMT and PMC deposits, fine, plate-like or acicular morphological features (α_f) were observed in the top region. In the present work, the labels α_f , α_m and α_a are used only as morphological descriptors for fine plate-like, coarser plate-like and needle-like features, respectively, and do not denote definitive phase identification. The PMC deposit exhibited a finer morphological scale in the top region than the CMT deposit. This difference is consistent with the different process conditions and nominal heat inputs calculated in Section 3.1. This finer morphology under the higher calculated heat input of PMC is counter-intuitive and may reflect differences in arc pressure, droplet transfer and melt-pool dynamics between the two selected conditions; however, a direct relationship between heat input and local cooling rate cannot be established from the present measurements. It should be noted that the comparison concerns the selected process conditions rather than the transfer mode in isolation. Additional plate-like and needle-like morphological features (α_m) were observed in the top regions of both deposits [24]. Needle-like features (α_a) were more readily distinguishable in the CMT top region.

Figure 2 to Figure 5 illustrate the microstructural morphology of the top regions of the PMC and CMT deposits at different magnifications. In the CMT deposit, fine plate-like features were observed along the boundaries of the larger lamellar morphological regions [25]. Similar build-dependent microstructural characteristics have also been reported for commercially pure titanium fabricated by wire-arc directed energy deposition [26].

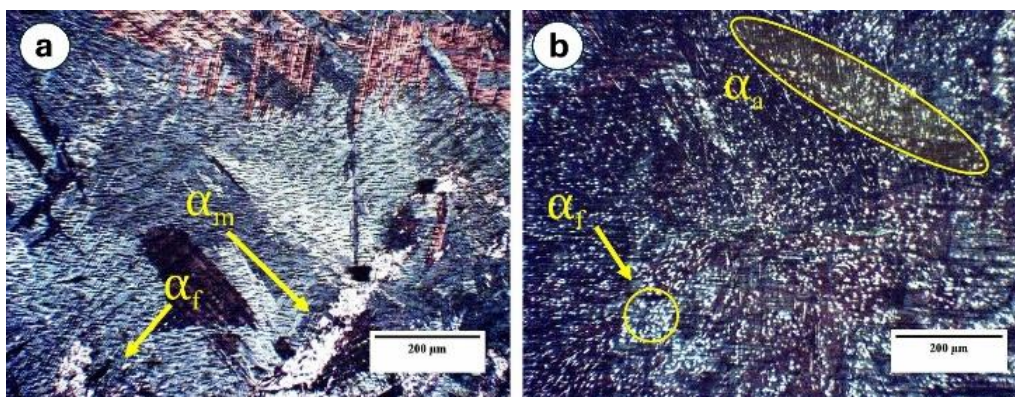


Figure 2. Interior microstructure of top region belongs to (a) PMC, and (b) CMT technique

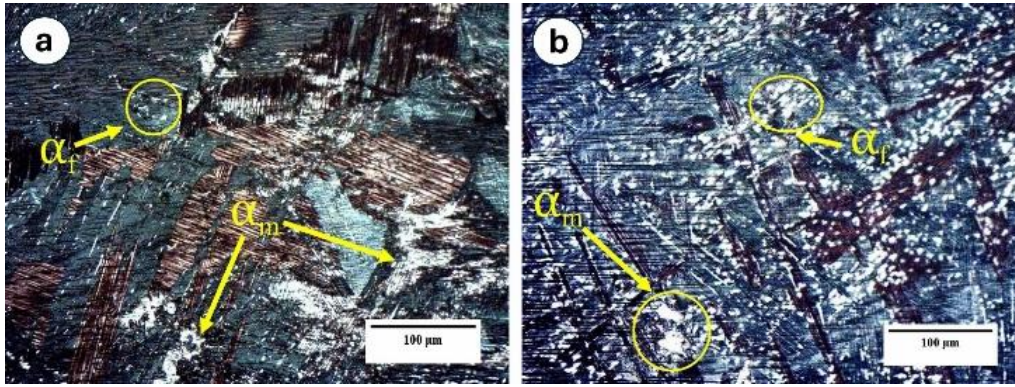


Figure 3.
Interior microstructure of top region belongs to (a) PMC, and (b) CMT technique with improved magnification

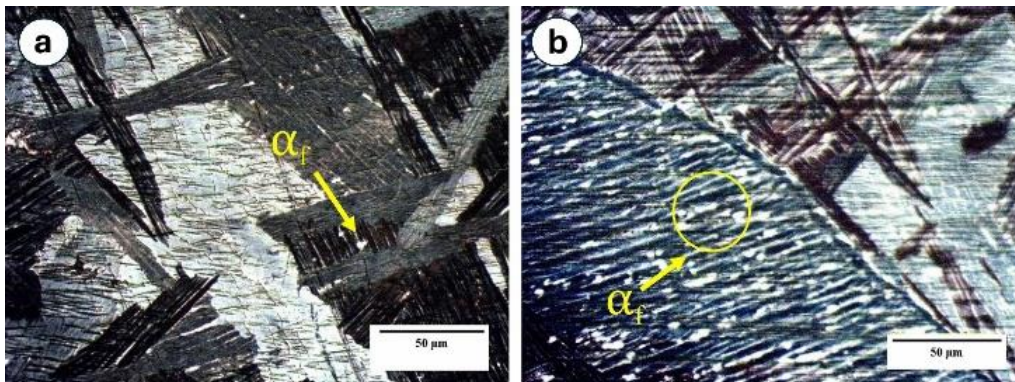


Figure 4.
Interior microstructure of top region belongs to (a) PMC, and (b) CMT technique with higher magnification

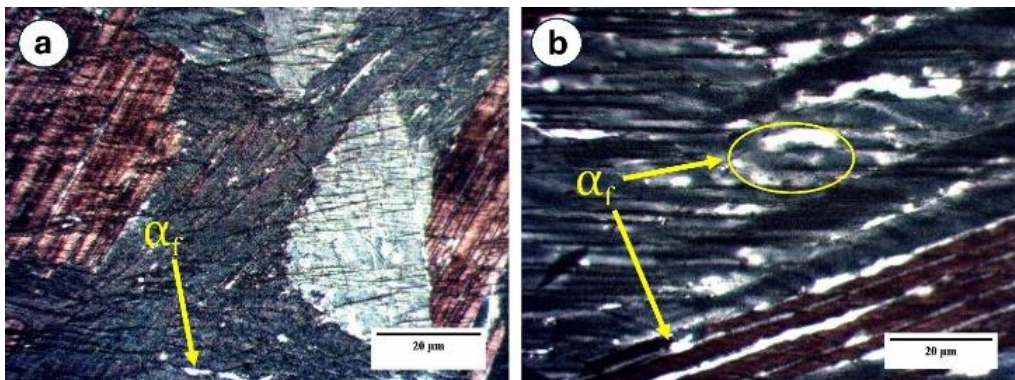


Figure 5.
Close interior microstructure of top region belongs to (a) PMC, and (b) CMT technique

3.2.2. Middle region

a. Middle-top region

The middle-top region exhibited a relatively coarser morphological scale than the top region. This region had experienced a greater number of subsequent deposition cycles than the final layers and therefore underwent repeated reheating during the continued build process. Such repeated thermal exposure can contribute to differences in morphology between the middle-top and top regions. However, the local cooling rate was not directly measured in the present study. The micrographs showed a combination of relatively equiaxed and elongated/columnar morphological features in the middle-top region. In the PMC deposit, relatively fine plate-like features (α_f) were sparsely distributed, together with coarser plate-like (α_m) and needle-like morphologies (α_a). The morphology observed in this region was consistent with the previously reported features associated with thermally cycled titanium deposits [19], [27]. In the CMT deposit, the microstructural features appeared more heterogeneous and overlapped within the observed field. Needle-like morphological features (α_a) were also observed in this region. **Figure 6** to **Figure 9** illustrate the microstructural morphology of the PMC and CMT deposits in the middle-top region at different magnifications.

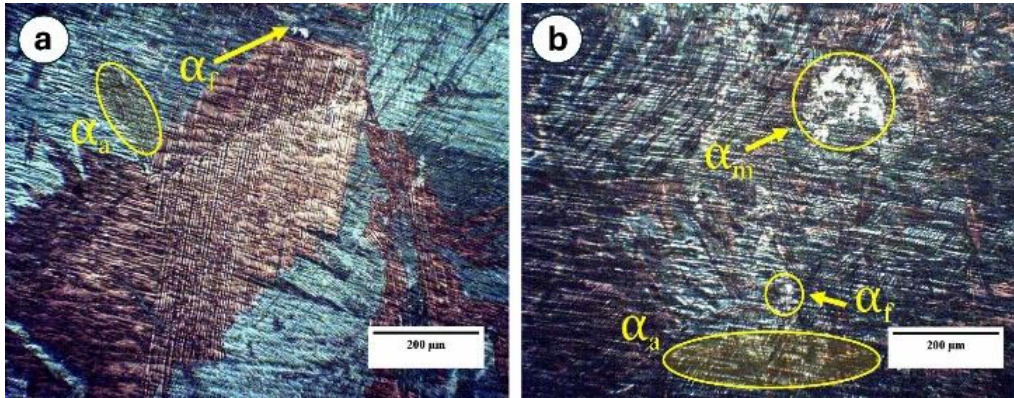


Figure 6.
Interior
microstructure of
middle-top region
belongs to (a) PMC,
and (b) CMT
technique

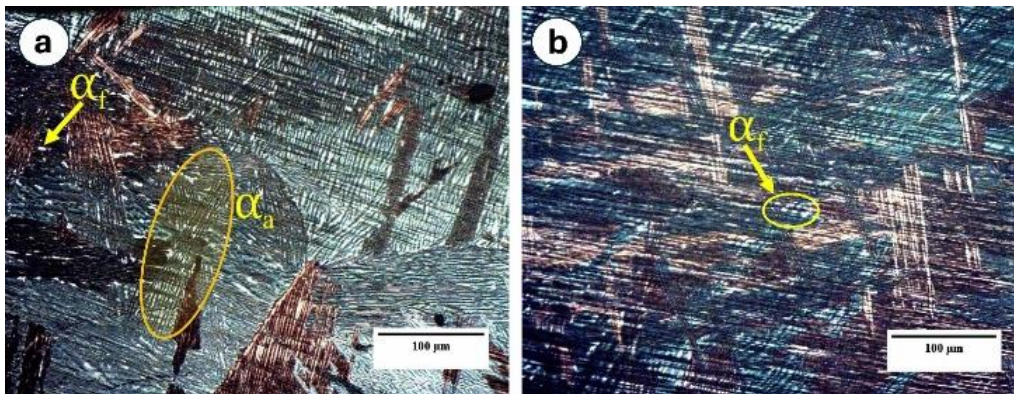


Figure 7.
Interior
microstructure of
middle-top region
belongs to (a) PMC,
and (b) CMT
technique with
improved
magnification

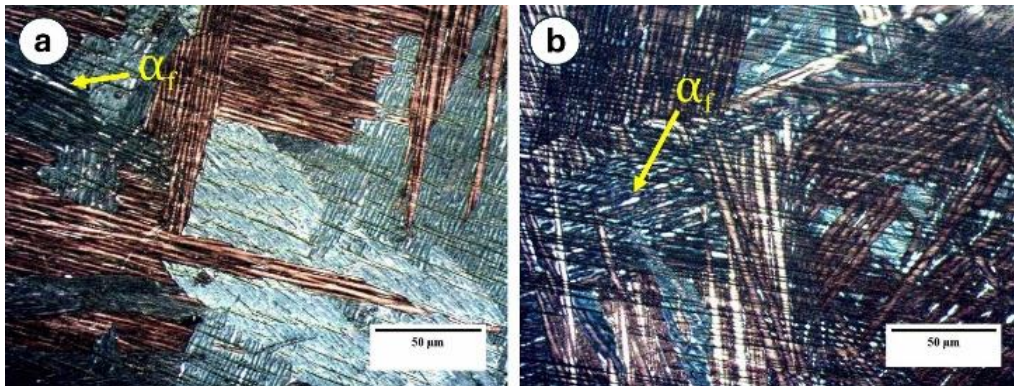


Figure 8.
Interior
microstructure of
middle-top region
belongs to (a) PMC,
and (b) CMT
technique with higher
magnification

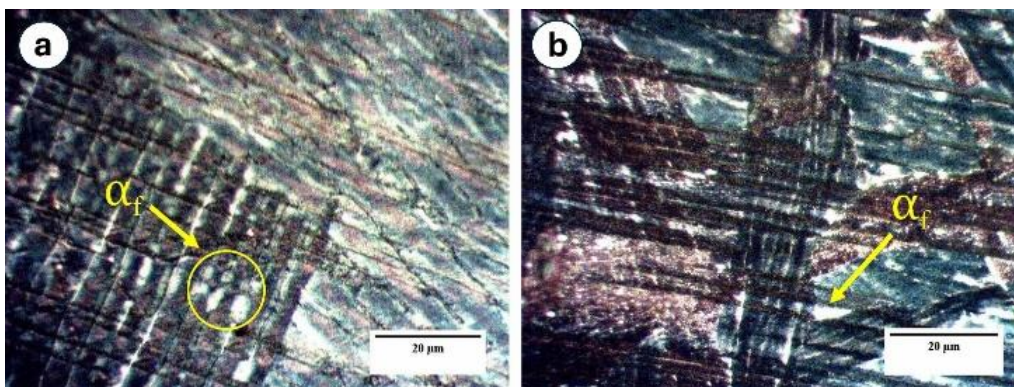


Figure 9.
Close interior
microstructure of
middle-top region
belongs to (a) PMC,
and (b) CMT
technique

b. Middle-bottom region

Figure 10 to Figure 13 show the microstructural morphology of the PMC and CMT deposits in the middle-bottom region at different magnifications. Compared with the top and middle-top regions, this region exhibited a coarser morphological scale. Because the middle-bottom region was exposed to a greater number of subsequent deposition cycles, it underwent repeated reheating as the build continued, which may account for the coarsening relative to the upper regions. However, since the local cooling rate and thermal history were not measured in the present study, this explanation remains qualitative. The micrographs revealed a mixture of near-equiaxed and

elongated (columnar) features. In the PMC deposit, coarse plate-like (α_m) and needle-like (α_a) morphologies dominated, while finer plate-like features (α_f) were only sparsely distributed. This morphology is consistent with features previously reported for thermally cycled titanium deposits [19], [27]. In contrast, the CMT deposit displayed a more heterogeneous microstructure than the PMC deposit, with the different morphological features overlapping within the observed field; needle-like features (α_a) were also present.

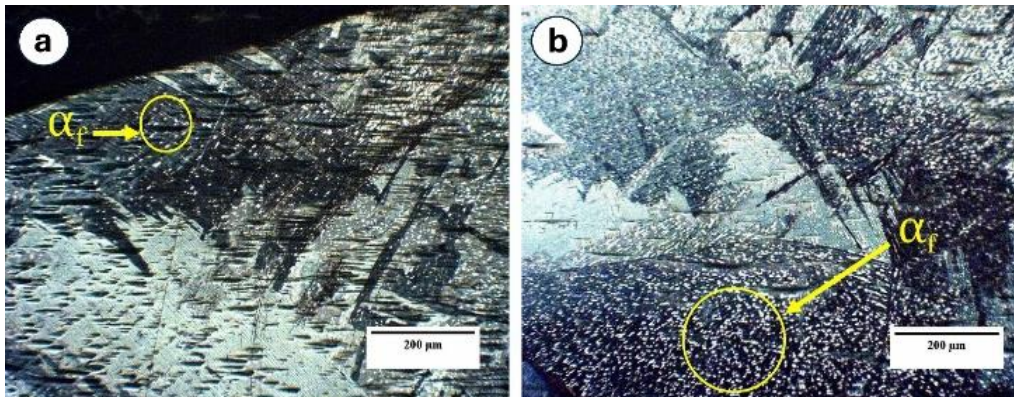


Figure 10. Interior microstructure of middle-bottom region belongs to (a) PMC, and (b) CMT technique

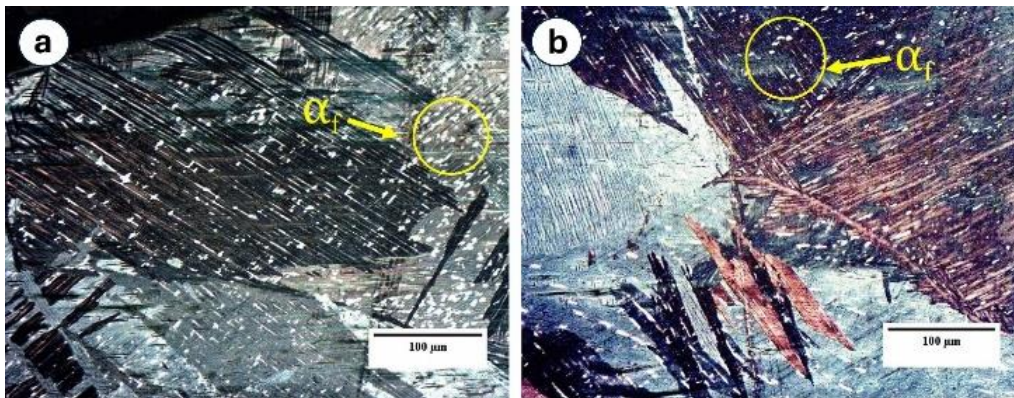


Figure 11. Interior microstructure of middle-bottom region belongs to (a) PMC, and (b) CMT technique with improved magnification

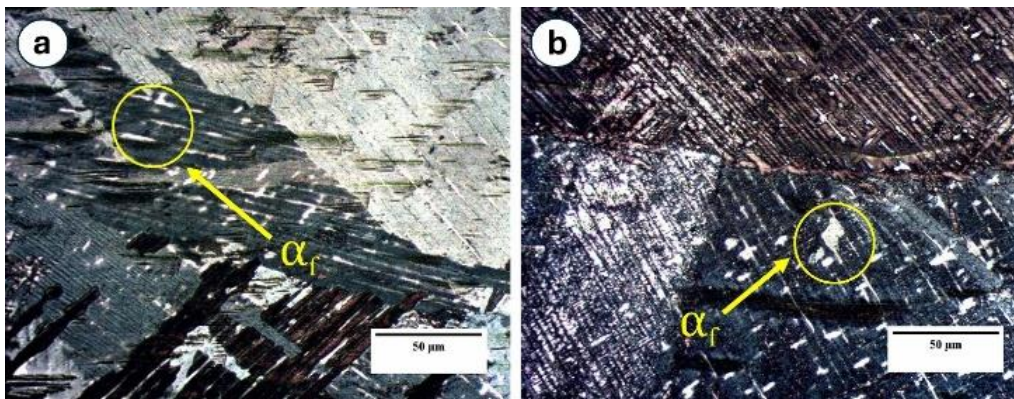


Figure 12. Interior microstructure of middle-bottom region belongs to (a) PMC, and (b) CMT technique with higher magnification

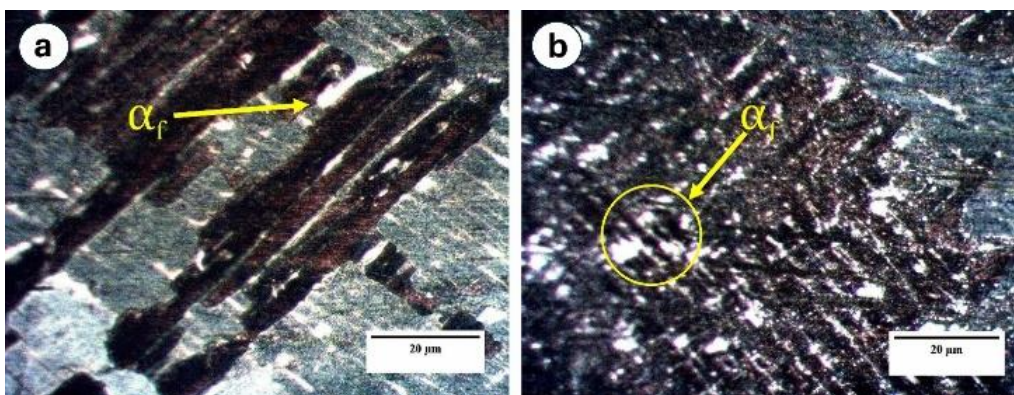


Figure 13. Close interior microstructure of middle-bottom region belongs to (a) PMC, and (b) CMT technique

3.2.3. Bottom region

The bottom region experienced a distinct thermal history because it was deposited directly above the titanium substrate and subsequently subjected to repeated thermal cycles as additional layers were deposited. During the initial stages of deposition, heat extraction through the substrate can influence the thermal conditions of the lower layers. As the build progressed, these same layers were repeatedly reheated by subsequent deposition passes. Consequently, the bottom region should be considered as a region subjected to both substrate interaction and repeated reheating rather than being assigned a specific cooling-rate value from the present experimental data. The relatively coarser morphology observed in the bottom region is consistent with the cumulative influence of repeated thermal exposure and microstructural evolution during the layer-wise build [20]. Because local temperature histories were not measured, the specific contribution of cooling rate to this coarsening cannot be isolated from the effects of reheating and heat accumulation. The bottom region of the CMT deposit exhibited a relatively coarser morphological scale than the corresponding PMC region. Fine plate-like features were observed near the lower layers of both deposits. In the CMT deposit, needle-like features were also visible near the upper part of the bottom region (Figure 17), while similar plate-like and needle-like morphologies were observed in the PMC deposit. These observations indicate spatial variation in morphology along the build height; however, the present optical characterization does not by itself establish the exact phase identity or confirm remelting and recasting of specific regions. Figure 14 to Figure 17 show the microstructural morphology of the bottom regions of the CMT and PMC deposits at different magnifications.

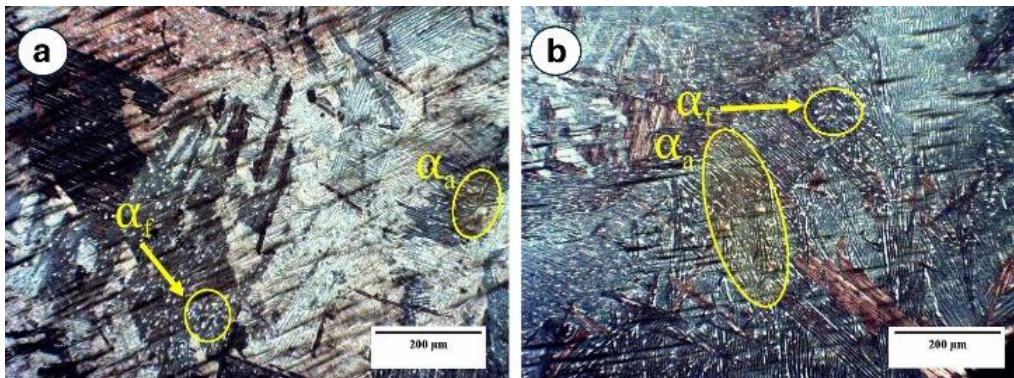


Figure 14.
Interior
microstructure of
bottom region belongs
to (a) PMC, and (b)
CMT technique

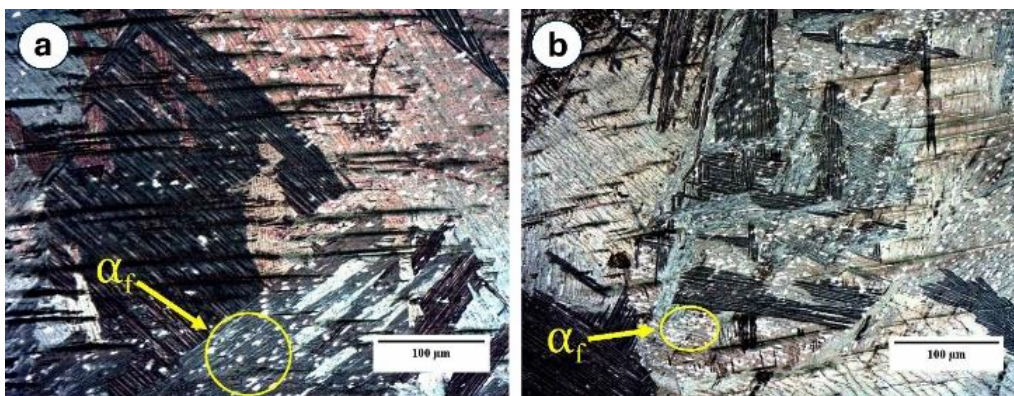


Figure 15.
Interior
microstructure of
bottom region belongs
to (a) PMC, and (b)
CMT technique with
improved
magnification

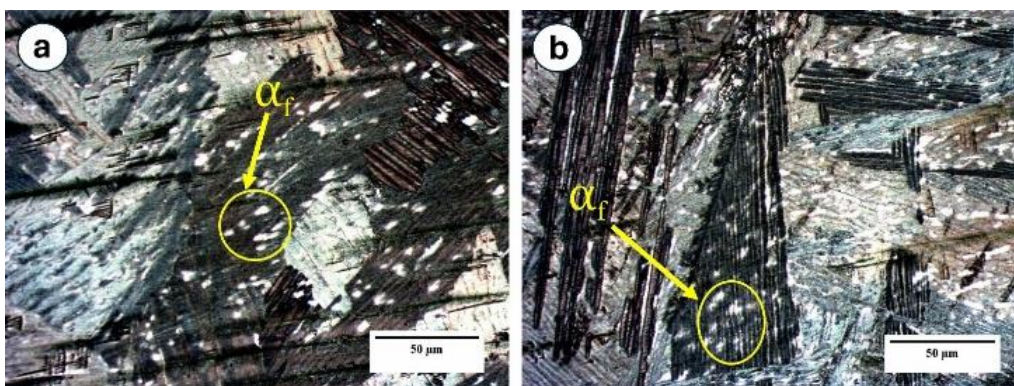


Figure 16.
Interior
microstructure of
bottom region belongs
to (a) PMC, and (b)
CMT technique with
higher magnification

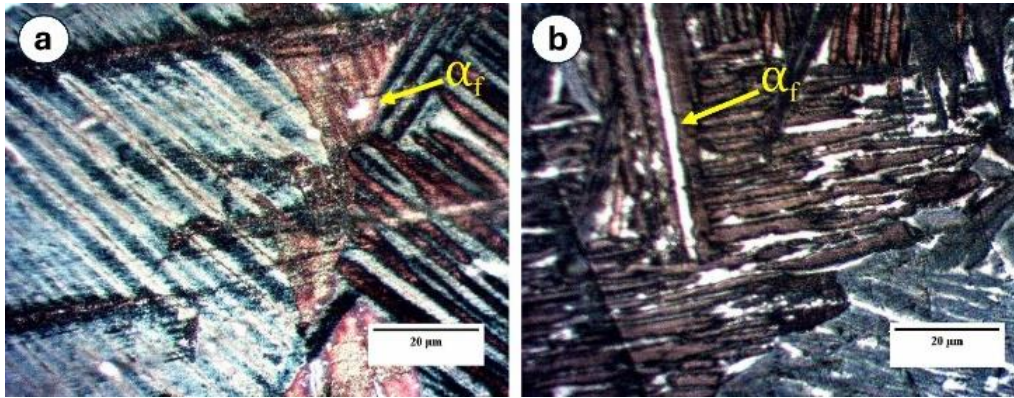


Figure 17. Close interior microstructure of bottom region belongs to (a) PMC, and (b) CMT technique

3.2.4. Microstructural feature and morphological-size variation

Quantitative analysis of the microstructural features was performed using ImageJ, as described in Section 2.3. The metallographic images from the top, middle, and bottom regions of the CMT and PMC deposits were calibrated using their respective scale bars, and the characteristic dendritic/microstructural feature size was measured from the calibrated images. The average values obtained from the image analysis are presented in **Table 5**. The ImageJ measurements show a consistent increase in the characteristic microstructural feature size from the top toward the bottom region in both deposition conditions. For the CMT deposit, the measured values were 2.35 μm , 3.16 μm , and 3.46 μm for the top, middle, and bottom regions, respectively. The corresponding values for the PMC deposit were 1.97 μm , 2.70 μm , and 3.20 μm . Thus, both deposits exhibited a similar top-to-bottom trend, although the measured feature size was lower for PMC at all three investigated locations.

The observed variation indicates that the microstructural scale was not uniform throughout the build height. Such spatial variation can be associated with the different thermal histories experienced by material deposited at different heights, including repeated deposition cycles, reheating, and heat accumulation. However, because the present study did not include direct thermal measurements or numerical thermal analysis, the specific contribution of cooling rate or local thermal gradients to the measured feature-size variation cannot be established experimentally. The smaller ImageJ-measured feature size in the PMC deposit is therefore reported as an experimental observation under the selected PMC and CMT processing conditions rather than being attributed solely to the higher calculated heat input of PMC. Since the two processes also employed different voltage, current, and wire-feed conditions, the measured difference should be interpreted as a response to their respective selected processing conditions rather than as an isolated effect of transfer mode [21].

Table 5. ImageJ-measured characteristic dendritic/microstructural feature size in CMT and PMC deposits

Process	Top (μm)	Middle (μm)	Bottom (μm)
CMT	2.35	3.16	3.46
PMC	1.97	2.70	3.20

3.3. Micro-hardness test

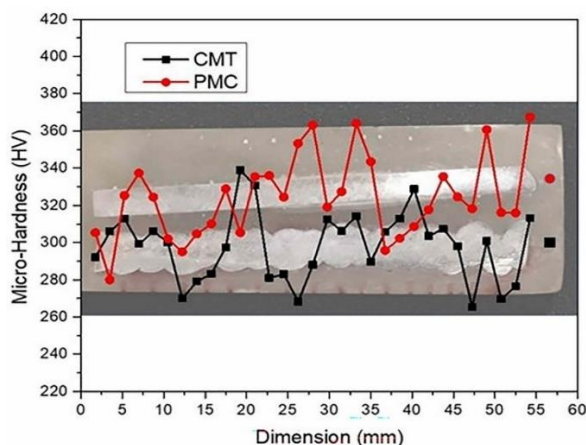


Figure 18. Micro-hardness variation of CMT and PMC deposits along the build height

The micro-hardness of the CMT and PMC deposits was measured on polished cross-sections along the build-height direction (**Figure 18**). The Vickers indents were positioned from the bottom toward the top of the deposited region, with an approximate spacing of 1.75 mm between successive measurements. The CMT deposit exhibited relatively limited variation in hardness across the measured height, with an average hardness of approximately 298 HV. The PMC deposit exhibited a higher average hardness of approximately 324 HV.

The higher average hardness measured in the PMC deposit is consistent with the finer measured microstructural features reported in Section 3.2; however, the present

experiments do not provide direct thermal measurements or cooling-rate data. Therefore, the hardness difference should not be attributed solely to the higher calculated heat input or to a specific cooling-rate mechanism. The observed variation may instead reflect the combined influence of the selected process parameters and the resulting thermal history, microstructural morphology, and local deposition conditions [28], [29], [30]. It is also noted that the average hardness of both deposits (approximately 300 HV) is about twice the reference value for wrought Grade 2 titanium (145 HV, Table 2). Because the hardness and strength of commercially pure titanium are highly sensitive to interstitial oxygen and nitrogen, possible interstitial pick-up during deposition cannot be excluded, and chemical analysis of the deposits is required to clarify this contribution.

3.4. Tensile properties

Five sub-size tensile specimens were extracted from each deposit at different locations along the build height and tested according to ASTM A370. The specimens were extracted with their longitudinal axis parallel to the deposition direction. For the CMT deposit, the specimen extracted from the bottom region was designated CMT T1, followed sequentially by CMT T2, CMT T3, CMT T4, and CMT T5 toward the top of the deposit. Similarly, PMC T1 was extracted from the bottom region of the PMC deposit, followed by PMC T2, PMC T3, PMC T4, and PMC T5 along the build height toward the top region. Thus, the specimen designations represent their spatial positions within the respective deposits.

These five specimens from each process were spatial subsamples from a single build rather than independent experimental replicates. Their purpose was to investigate the variation in tensile properties with build height and the corresponding differences in local processing and thermal history associated with successive deposition layers, reheating, heat accumulation, and heat dissipation. Therefore, the results are used to characterize the through-height variation within the investigated CMT and PMC builds rather than to establish statistical build-to-build reproducibility. Such location-dependent variation in mechanical response is consistent with the microstructural heterogeneity commonly reported in additively manufactured titanium deposits [31], [32]. After tensile testing, the specimens exhibited fracture at different locations along the gauge length, as shown in **Figure 19**. The variation in fracture location indicates spatial differences in the response of the deposited material; however, the present observations alone do not establish a specific defect or microstructural feature as the cause of the fracture location [31]. The tensile stress-strain data were processed using Origin software, and the resulting tensile properties are presented in **Table 6**. The combined stress-strain curves for the CMT and PMC specimens are shown in **Figure 20**.

The average ultimate tensile strength (UTS) of the PMC deposit was higher than that of the CMT deposit. The average UTS values were 527.4 MPa for PMC and 516.2 MPa for CMT. Within the PMC deposit, the UTS varied substantially with specimen position, with PMC T1 from the bottom region exhibiting the highest UTS of 736 MPa, whereas the UTS values of PMC T2–T5 were lower. When PMC T1 is excluded, the average UTS of PMC T2–T5 is 475.3 MPa, which is lower than the CMT average; the higher overall PMC average is therefore governed by a single bottom-region specimen. Because PMC T1 represents the bottom region of the deposited build, its substantially higher tensile strength should be considered as a location-dependent result rather than attributed to the substrate without additional evidence.

For the CMT deposit, the UTS also varied with specimen position. CMT T3, extracted from the middle region, exhibited the lowest UTS of 446 MPa, whereas the specimens extracted from the other investigated heights showed

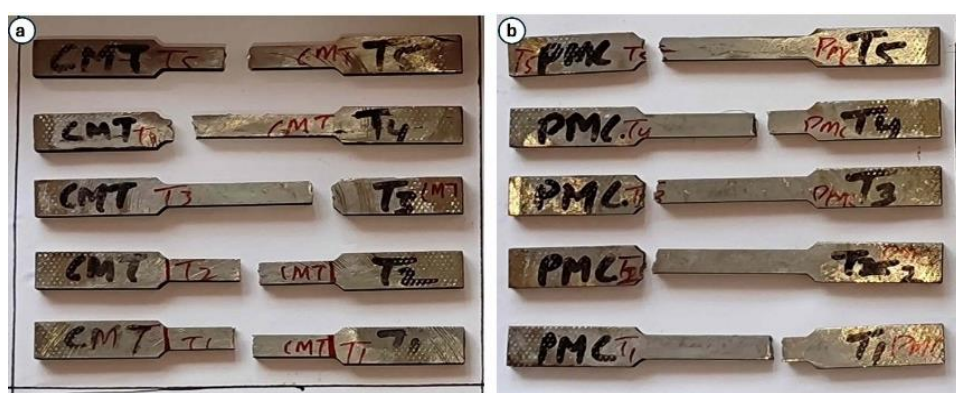


Figure 19.
After tensile test specimens
of (a) CMT, and (b) PMC

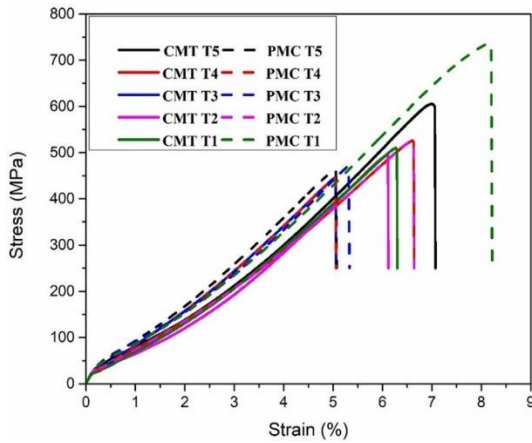


Figure 20. Combined stress-strain diagram of CMT and PMC samples

particularly because of the high UTS measured for PMC T1. Therefore, the tensile results indicate a pronounced location-dependent response within both deposits [31], [32]. Notably, the tensile trend in the PMC deposit did not follow the feature-size trend: although the top region exhibited the finest features, PMC T5 showed one of the lowest UTS values, whereas the bottom specimen PMC T1 showed the highest. Feature size alone therefore does not explain the tensile response. In addition, the yield-to-tensile strength ratios of both deposits were high (approximately 0.88–0.99), and the elongation values (5.06–8.21%) were considerably lower than the reference value for wrought Grade 2 titanium (20%, Table 2), indicating limited work-hardening capacity and ductility.

Table 6. Tensile test data for CMT and PMC tensile sample

Sample Name	Yield Strength (MPa) by 0.2% offset method	Ultimate Tensile Strength (MPa)	Elongation (%)
CMT T1	492	510	6.29
CMT T2	486	493	6.11
CMT T3	428	446	5.46
CMT T4	510	526	6.64
CMT T5	585	606	7.07
Avg. Value	500.2	516.2	6.314
PMC T1	708	736	8.21
PMC T2	465	526	6.64
PMC T3	445	470	5.32
PMC T4	416	446	5.06
PMC T5	435	459	5.06
Avg. Value	493.8	527.4	6.058

3.5. Charpy impact properties

All Charpy specimens fractured during the impact test, with most specimens fracturing near the mid-section. CMT C2 exhibited fracture closer to the specimen end (Figure 21). The reason for this different fracture location cannot be established from the present observations alone and is therefore not attributed to a pre-existing crack without additional fracture-surface or defect characterization. The location and progression of fracture in titanium can be influenced by local microstructural characteristics and grain-boundary effects; however, the present Charpy observations are insufficient to establish such a mechanism for the individual specimens [33], [34]. The measured absorbed-energy values are presented in Table 7.

Both the CMT and PMC specimens exhibited absorbed energies between 0.7 and 0.9 J. The corresponding average absorbed energies were 0.80 J for CMT and 0.76 J for PMC. Because the specimens were sub-size Charpy specimens, the measured absorbed energy is strongly dependent on specimen dimensions and test configuration. Therefore, these values should be interpreted as comparative measurements for the present specimens rather than as directly equivalent to standard-size Charpy impact energy reported for commercially pure Grade-2 titanium. Nevertheless, the low absorbed energies, together with the reduced elongation, are consistent with limited ductility of the deposited material; the underlying cause could not be established without chemical and fractographic analysis. The relatively small difference between the average absorbed energies of the CMT and PMC specimens indicates comparable impact-energy absorption within the investigated deposits. However, because only one build was produced for each process condition and the specimens were extracted from different locations within those builds, the observed difference should be regarded as a descriptive comparison rather than a statistically validated process-level difference. A summary of average values of all the mechanical properties studied is outlined in Table 8. The Charpy absorbed-energy values for the present CMT and PMC specimens were obtained using sub-size specimens and therefore should not be directly compared with reference values obtained using different specimen dimensions or test configurations.

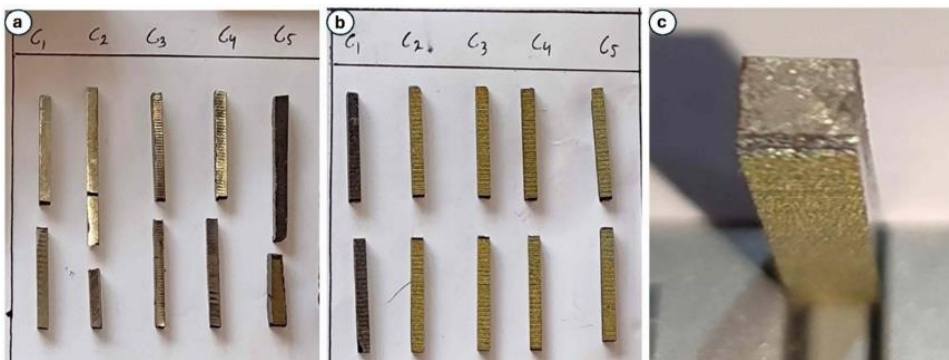


Figure 21. Charpy samples after impact toughness test of (a) CMT, (b) PMC, and (c) fractured surface of a Charpy sample

Table 7. Tabulated results from impact toughness test on Charpy specimens

	Sample Name	C1	C2	C3	C4	C5	Average
Energy Absorbed (J)	CMT	0.8	0.8	0.7	0.8	0.9	0.78 (J)
	PMC	0.8	0.8	0.9	0.7	0.7	0.76 (J)

Table 8. Average mechanical properties of Ti Grade 2 samples prepared through CMT and PMC techniques

Sample Name	Yield Strength (MPa)	Ultimate Stress (MPa)	Hardness (HV)	Charpy V-notch Impact (J)	Elongation (%)
CMT	500.2	516.2	298	0.8	6.314
PMC	493.8	527.4	324	0.76	6.058

3.6. Field emission scanning electron microscopy

Field emission scanning electron microscopy (FESEM) was used to examine the fracture surfaces of the tensile and Charpy specimens and to obtain higher-magnification information on the morphology of the fractured regions.

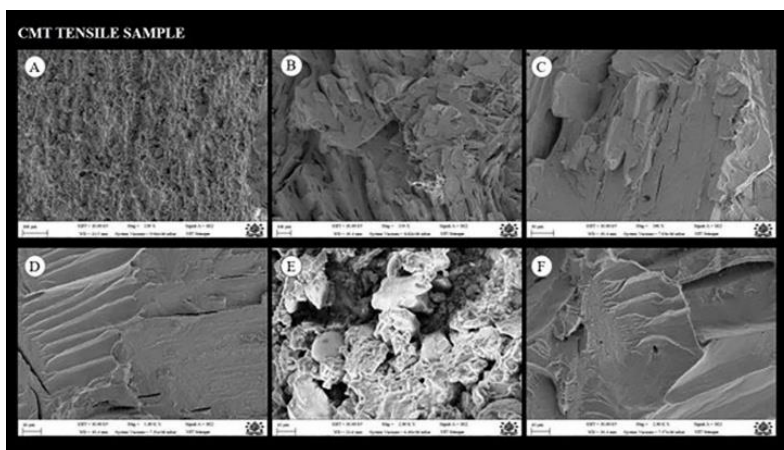


Figure 22. Fractured surface morphology of CMT tensile sample

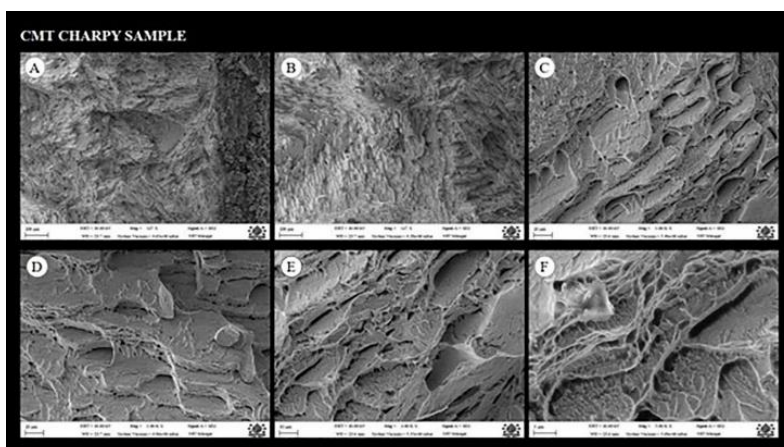


Figure 23. Fractured surface morphology of CMT Charpy sample

The FESEM observations were used as complementary evidence to the tensile and Charpy test results discussed in Sections 3.4 and 3.5 (**Figure 22** and **Figure 23**). The fracture surfaces exhibited variations in surface morphology, including relatively flat facets and irregular localized features. These observations indicate spatial variation in the fracture morphology of the tested specimens; however, the present FESEM images alone do not provide sufficient evidence to classify the entire fracture surface as exclusively cleavage, quasi-cleavage, or ductile fracture. Therefore, the observed features are described morphologically rather than being assigned to a specific fracture mechanism without additional quantitative fractographic analysis. In the PMC tensile specimen, localized particulate or contrast features can be observed in the regions indicated in **Figure 24**. These features are reported only as localized morphological features, since their chemical composition was not determined by

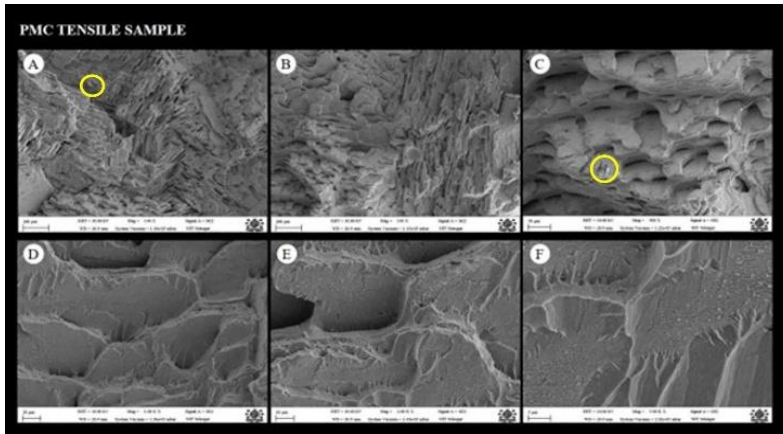


Figure 24. FESEM fracture-surface morphology of the PMC tensile specimen

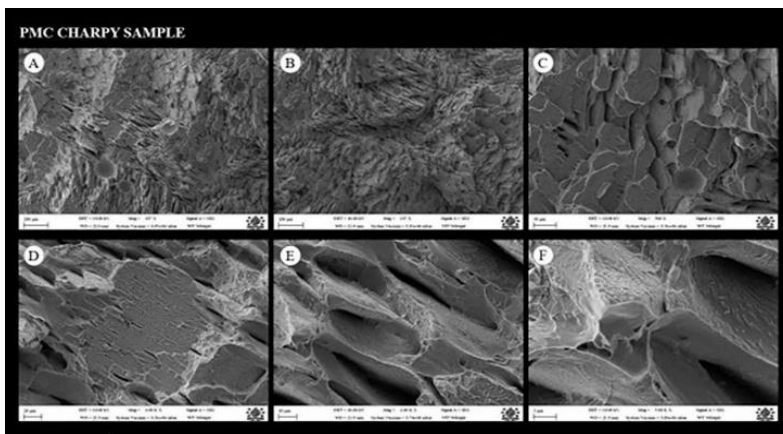


Figure 25. FESEM fracture-surface morphology of the PMC Charpy specimen

EDS point analysis or elemental mapping. Consequently, they cannot be conclusively identified as secondary-phase particles or inclusions from the present FESEM observations alone.

The fracture-surface observations provide additional information about the failure morphology of the tested specimens but do not establish a direct causal relationship between the observed localized features and the higher strength or fracture behavior of the PMC specimens (Figure 25). In particular, the substantially higher UTS of PMC T1 compared with the other PMC tensile specimens should be interpreted together with its specific location at the bottom region of the deposit and the spatial variation in properties discussed in Section 3.4. Further compositional analysis, such as EDS, would be required to determine the chemical nature of the localized features and to establish their possible contribution to fracture behavior.

4. Conclusion

The present study compared the microstructural and mechanical characteristics of commercially pure Titanium Grade 2 deposited using Cold Metal Transfer (CMT) and Pulse Multi Control (PMC) under their respective selected processing conditions. The results showed that the characteristic dendritic or microstructural feature size varied with build height in both deposits, with ImageJ measurements indicating smaller features in the top regions and larger features toward the bottom regions. The measured feature sizes ranged from 2.35 to 3.46 μm for CMT and from 1.97 to 3.20 μm for PMC. Although the PMC deposit exhibited smaller measured features at the investigated locations, the present results do not establish cooling rate as the sole factor responsible for this difference. The mechanical properties also exhibited clear spatial variations along the build height. The average microhardness values were approximately 298 HV for CMT and 324 HV for PMC, while the average ultimate tensile strengths (UTS) were 516.2 MPa and 527.4 MPa, respectively, although the higher PMC average was governed by the single bottom specimen PMC T1. In contrast, the average yield strength was slightly higher for CMT (500.2 MPa) than for PMC (493.8 MPa). Notably, the PMC T1 specimen, representing the bottom region of the deposit, exhibited an unusually high UTS of 736 MPa, highlighting the importance of specimen location within the build when evaluating mechanical performance. Under the applied sub-size Charpy configuration, the impact responses of the two deposits were comparable, with average absorbed energies of 0.80 J for CMT and 0.76 J for PMC. Overall, the selected CMT and PMC processing conditions produced distinct build-height-dependent microstructural and mechanical responses in Ti Grade 2 deposits. However, because only one build was produced for each process condition and the tensile specimens represented spatial subsamples rather than independent experimental replicates, the observed differences should be interpreted as comparative experimental observations rather than statistically validated process-level differences. Further investigations involving independent builds and additional thermal and compositional characterization, including interstitial (O/N) analysis, are therefore required to establish the reproducibility of these findings and clarify the mechanisms underlying the observed variations.

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References

- [1] M. Peters, J. Kumpfert, C. H. Ward, and C. Leyens, "Titanium alloys for aerospace applications," *Advanced Engineering Materials*, vol. 5, no. 6, pp. 419–427, 2003, doi: 10.1002/adem.200310095.
- [2] R. R. Boyer, "Titanium for aerospace: Rationale and applications," *Advanced Performance Materials*, vol. 2, pp. 349–368, 1995, doi: 10.1007/BF00705316.
- [3] R. R. Boyer, "An overview on the use of titanium in the aerospace industry," *Materials Science and Engineering: A*, vol. 213, no. 1–2, pp. 103–114, 1996, doi: 10.1016/0921-5093(96)10233-1.
- [4] J. C. Williams and R. R. Boyer, "Opportunities and issues in the application of titanium alloys for aerospace components," *Metals*, vol. 10, no. 6, p. 705, 2020, doi: 10.3390/met10060705.
- [5] Z. Lin, K. Song, and X. Yu, "A review on wire and arc additive manufacturing of titanium alloy," *Journal of Manufacturing Processes*, vol. 70, pp. 24–45, 2021, doi: 10.1016/j.jmapro.2021.08.018.
- [6] T. S. Tshephe, S. O. Akinwamide, E. Olevsky, and P. A. Olubambi, "Additive manufacturing of titanium-based alloys: A review of methods, properties, challenges, and prospects," *Heliyon*, vol. 8, no. 3, p. e09041, 2022, doi: 10.1016/j.heliyon.2022.e09041.
- [7] D. Chakraborty, T. Tirumala, S. Chitral, B. N. Sahoo, D. V. Kiran, and A. Kumar, "The state of the art for wire arc additive manufacturing process of titanium alloys for aerospace applications," *Journal of Materials Engineering and Performance*, vol. 31, pp. 6149–6182, 2022, doi: 10.1007/s11665-022-07128-1.
- [8] D.-G. Ahn, "Directed energy deposition (DED) process: State of the art," *International Journal of Precision Engineering and Manufacturing-Green Technology*, vol. 8, pp. 703–742, 2021, doi: 10.1007/s40684-020-00302-7.
- [9] D. Svetlizky *et al.*, "Directed energy deposition (DED) additive manufacturing: Physical characteristics, defects, challenges and applications," *Materials Today*, vol. 49, pp. 271–295, 2021, doi: 10.1016/j.mattod.2021.03.020.
- [10] N. Khanna, H. Salvi, B. Karaş, I. Fairoz, and A. Shokrani, "Cost modelling for powder bed fusion and directed energy deposition additive manufacturing," *Journal of Manufacturing and Materials Processing*, vol. 8, no. 4, p. 142, 2024, doi: 10.3390/jmmp8040142.
- [11] U. Alonso, F. Veiga, A. Suárez, and T. Artaza, "Experimental investigation of the influence of wire arc additive manufacturing on the machinability of titanium parts," *Metals*, vol. 10, no. 1, p. 24, 2020, doi: 10.3390/met10010024.
- [12] R. P. Meena, N. Yuvaraj, and Vipin, "A review on wire arc additive manufacturing based on cold metal transfer," *Materials and Manufacturing Processes*, vol. 39, pp. 1315–1341, 2024, doi: 10.1080/10426914.2024.2323441.
- [13] L. P. Raut and R. V. Taiwade, "Wire Arc Additive Manufacturing: A Comprehensive Review and Research Directions," *Journal of Materials Engineering and Performance*, vol. 30, no. 7, pp. 4768–4791, Jul. 2021, doi: 10.1007/s11665-021-05871-5.
- [14] S. Singh, S. K. Sharma, and D. W. Rathod, "A review on process planning strategies and challenges of WAAM," *Materials Today: Proceedings*, vol. 47, pp. 6564–6575, 2021, doi: 10.1016/j.matpr.2021.02.632.
- [15] S. Selvi, A. Vishvaksean, and E. Rajasekar, "Cold metal transfer (CMT) technology: An overview," *Defence Technology*, vol. 14, no. 1, pp. 28–44, 2018, doi: 10.1016/j.dt.2017.08.002.
- [16] B. Tomar and S. Shiva, "Cold metal transfer-based wire arc additive manufacturing," *Journal of the Brazilian Society of Mechanical Sciences and Engineering*, vol. 45, p. 400, 2023, doi: 10.1007/s40430-023-04084-2.
- [17] A. Işitan, S. Aytakin, and V. Onar, "Enhancing the microstructure and mechanical properties of cast aluminum matrix composites through nano-Al₂O₃ reinforcement via CMT and PMC welding," *International Journal of Metalcasting*, 2024, doi: 10.1007/s40962-023-01234-z.
- [18] F. A. Shah, M. Trobos, P. Thomsen, and A. Palmquist, "Commercially pure titanium (cp-Ti) versus titanium alloy (Ti6Al4V) materials as bone anchored implants: Is one truly better than the other?," *Materials Science and Engineering: C*, vol. 62, pp. 960–966, 2016, doi: 10.1016/j.msec.2016.01.032.

- [19] A. Dash, V. K. Champagne, and A. Bandyopadhyay, "Commercially pure titanium via wire arc directed energy deposition using localised shielding," *Philosophical Magazine Letters*, vol. 105, p. 2533788, 2025, doi: 10.1080/09500839.2025.2533788.
- [20] B. Palacios *et al.*, "Role of structural hierarchy on mechanics and electrochemistry of wire arc additive manufactured (WAAM) single phase titanium," *Journal of Manufacturing Processes*, vol. 93, pp. 239–249, 2023, doi: 10.1016/j.jmapro.2023.03.025.
- [21] P. S. Deshmukh, A. Katiyar, A. Sahu, D. Sathiaraj, I. A. Palani, and A. Sonawane, "Wire arc additive manufacturing of commercially pure titanium bio-medical alloy," *Materials Today: Proceedings*, vol. 90, pp. 6–10, 2023, doi: 10.1016/j.matpr.2023.03.346.
- [22] B. Cong, Z. Qi, B. Qi, H. Sun, G. Zhao, and J. Ding, "A comparative study of additively manufactured thin wall and block structure with Al-6.3%Cu alloy using cold metal transfer process," *Applied Sciences*, vol. 7, no. 3, p. 275, 2017, doi: 10.3390/app7030275.
- [23] H.-J. Yi, J.-W. Kim, Y.-L. Kim, and S. Shin, "Effects of cooling rate on the microstructure and tensile properties of wire-arc additive manufactured Ti-6Al-4V alloy," *Metals and Materials International*, vol. 26, pp. 1235–1246, 2020, doi: 10.1007/s12540-019-00563-1.
- [24] J. Zhang *et al.*, "Effect of interlayer-compressed argon gas active cooling on microstructure and properties of Ti-6Al-4V fabricated by WAAM," *Advanced Engineering Materials*, 2023, doi: 10.1002/adem.202300618.
- [25] D. Nursyifaulkhair, N. Park, E. R. Baek, and S. Kim, "Influence of cooling rate on volume fraction of α massive phase in a Ti-6Al-4V alloy fabricated using directed energy deposition," *Materials Letters*, vol. 257, p. 126671, 2019, doi: 10.1016/j.matlet.2019.126671.
- [26] X. Gao, W. Zeng, S. Zhang, and Q. Wang, "A study of epitaxial growth behaviors of equiaxed alpha phase at different cooling rates in near alpha titanium alloy," *Acta Materialia*, vol. 122, pp. 298–309, 2017, doi: 10.1016/j.actamat.2016.10.012.
- [27] N. W. Hrabe, P. Heintl, B. Flinn, C. Körner, and R. K. Bordia, "Compression-compression fatigue of selective electron beam melted cellular titanium (Ti-6Al-4V)," *Journal of Biomedical Materials Research Part B: Applied Biomaterials*, vol. 99B, no. 2, pp. 313–320, 2011, doi: 10.1002/jbm.b.31901.
- [28] T. Mutuk and M. Gürbüz, "Effect of pure titanium particle size on density, hardness, wear resistance and microstructure properties," *Journal of Metals, Materials and Minerals*, vol. 29, no. 2, pp. 54–59, 2019, doi: 10.14456/jmmm.2019.34.
- [29] Q. Wang, Z. Liu, B. Wang, Q. Song, and Y. Wan, "Evolutions of grain size and micro-hardness during chip formation and machined surface generation for Ti-6Al-4V in high-speed machining," *The International Journal of Advanced Manufacturing Technology*, vol. 82, pp. 1725–1736, 2016, doi: 10.1007/s00170-015-7508-1.
- [30] V. Klimenov *et al.*, "Influence of the coarse grain structure of a titanium alloy Ti-4Al-3V formed by wire-feed electron beam additive manufacturing on strain inhomogeneities and fracture," *Materials*, vol. 16, no. 11, p. 3901, 2023, doi: 10.3390/ma16113901.
- [31] H. Shao, D. Shan, K. Wang, G. Zhang, and Y. Zhao, "Massive α precipitation selectivity and tensile fracture behavior of TC18 alloy," *Journal of Alloys and Compounds*, vol. 797, pp. 10–17, 2019, doi: 10.1016/j.jallcom.2019.04.315.
- [32] H. Attar, M. Calin, L.-C. Zhang, S. Scudino, and J. Eckert, "Manufacture by selective laser melting and mechanical behavior of commercially pure titanium," *Materials Science and Engineering: A*, vol. 593, pp. 170–177, 2014, doi: 10.1016/j.msea.2013.11.038.
- [33] K. Tokaji, T. Ogawa, and K. Ohya, "The effect of grain size on small fatigue crack growth in pure titanium," *International Journal of Fatigue*, vol. 16, no. 8, pp. 571–578, 1994, doi: 10.1016/0142-1123(94)90483-9.
- [34] C. Liu, X. Xu, T. Sun, R. Thomas, J. Q. da Fonseca, and M. Preuss, "Microstructural effects on fatigue crack initiation mechanisms in a near-alpha titanium alloy," *Acta Materialia*, vol. 253, p. 118957, 2023, doi: 10.1016/j.actamat.2023.118957.