Ti6Al4V/SiC Metal Matrix Composites Additively Manufactured by Direct Laser Deposition

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Abstract

Nowadays, research on additive manufacturing of Ti6Al4V alloy is growing exponentially but there are just a few studies about additive manufacturing of metal matrix composite components. In this work, highly reinforced Ti6Al4V matrix composites with SiC particles have been additively manufactured by direct laser deposition (DLD). Ti6Al4V powder and SiC particles have been deposited layer by layer to form an additive thin wall structure. The geometry, microstructure, and microhardness of the samples are strongly influenced by the laser scanning speed used during the fabrication process. In addition, the effect of the SiC increment in reinforcement concentrations and the influence of SiC particle sizes in the microstructure have been evaluated, and the reaction mechanisms have been established. The percentage of reinforcement measured is lower than expected due to the reinforcement-matrix reactivity that results in partially dissolved SiC particles and the formation of a TiC and Si5Ti3 ring around them. The size and number of particles and reaction products depend on the initial size and percentage of reinforcement and the DLD scanning speed. The higher the size and percentage of SiC particles and reaction products in the matrix, the higher the hardening effect of the composite matrix.

Keywords Direct laser deposition · Ti6Al4V · SiC · Metal matrix composite · Interfacial reactivity

1 Introduction

Additive manufacturing (AM) or 3D printing is a fabrication method with which it is possible to obtain 3D parts with complex and custom geometries in a short period by depositing material on a layer-by-layer basis [1]. Therefore, AM enables the creation of designs that cannot be manufactured via conventional methods and it offers the advantage of producing the least amount of waste material [2]. Raw ceramic and metallic materials apply to the 3D printing technology to give rise to metal matrix composites (MMCs), which are currently highly investigated [3]. One of the most commonly used AM techniques is direct laser deposition (DLD), where a high-energy laser acts as a source to melt materials in powder or wire form following a layer-by-layer model that is built from a CAD file with the desired part design [4]. During this process, the creation of a molten pool takes place, and the selected fabrication parameters influence its size and shape, modifying and improving the resulting microstructure of the material [5]. In addition, DLD originates less heat damage and distortion to the part, and the high level of control over the laser leads to exact energy values supplied in specific substrate areas.

The variation in the DLD processing parameters, such as laser power energy and laser scanning speed, has an influence on the shape and dimensions of the molten pool and in the heat-affected zone (HAZ), thus these parameters affect the morphology, microstructure, and mechanical properties of the final composite material [6, 7, 8]. Both the laser power and the scanning speed can result in layer re-melting, different cooling rates, and modifications in the grain size [8]. Therefore, the manufacturing parameters can be tuned to achieve specific material properties.

There are several alloys available in the market to be utilized in additive manufacturing, and Ti6Al4V (an α–β type alloy) is the most used material in AM [9, 10]. It represents 70% of titanium alloys, due to its excellent properties like light weight, good structural stability, low thermal conductivity, high elastic limit, great fatigue and corrosion resistance and high biocompatibility and osseointegration. Hence, Ti6Al4V is widely used not only in components fabrication
in the aerospace and transport industry [11], but also, in biomedicine, where it is the majority component for the manufacture of hip prostheses. Even though the titanium alloy used in the additive manufacturing process costs more than the forging of the traditional process, the cost of each component can be reduced by 50% without compromising the mechanical properties [12].

Nevertheless, the main disadvantage that this alloy presents is the low resistance to wear due to its high coefficient of friction. This leads to the material failure in applications like hip implants, where the formation of abrasive particles that cause toxicity, inflammation and pain in the patient occur. Consequently, few studies have been performed to improve Ti6Al4V characteristics, which involves surface treatments and coating depositions [13, 14, 15]. All these aspects must be studied for the specific case of titanium matrices.

Titanium matrix composites (TMCs) based on Ti6Al4V as the metal matrix reinforced with ceramic materials, which have a great potential to improve mechanical properties like wear resistance, are currently being investigated [16, 17]. Several reinforcement materials can be introduced in the titanium matrix, such as boron nitride, alumina, or silicon carbide. In particular, in other metal matrices, SiC has great chemical and mechanical properties, like high thermal shock, wear and corrosion resistance and high hardness [18].

Previous research has shown that SiC particles improved the composite strength as a consequence of an increase in hardness [19]. Kang et al. [20] reinforced porous titanium materials for medical applications with SiC by conventional powder metallurgy methods and demonstrated that just a 3 wt% of SiC increased the hardness by 170%. Aigbodion and Hassan [21] studied the microstructure and properties of alloys of Al–Si–Fe reinforced with 5–25 wt% of SiC particles, and have determined that the higher the SiCp content, the greater the hardness, the elastic limit, and the maximum tensile strength of the material. The interaction between the SiC reinforcement and the metal matrix is also of high importance. Riquelme et al. [22] have used laser cladding to reinforce an aluminum matrix (Al–Si) with SiC particles, and have evaluated the matrix–reinforcement reactivity, which can be reduced with the addition of alloying elements such as Si or Ti, and the effects on tribological properties in fabricated composite materials.

Ti6Al4V metallic matrix composites (reinforced with TiC, TiB, TiN between others) used for aircraft components have even higher hardness, better strength-to-weight ratio and wear resistance than the unreinforced material, especially at high temperatures [23]. The manufacture of these composites by additive manufacturing has been successfully carried out in several research studies [24], although their widespread use in aerospace applications has some way to go.

One of the main problems that occur in composite materials reinforced with ceramic particles is the poor matrix–reinforcement interaction, which depends on the wettability (surface chemistry) of the ceramic particle when in contact with the metal in the molten state. The poorer the particle–matrix interfacial bonding, the worse the properties of the composite; this is because the ceramic particles will tend to detach from the matrix, resulting in a low wear resistance [25]. Sivakumar et al. [26] used SiC nanoparticles in concentrations of 0–15 wt% to reinforce Ti6Al4V metal matrices, increasing both the hardness and resistance and also the wettability and the adhesion forces when adding 5 wt% content of SiC. Kloosterman et al. [27] examined the interface between Ti6Al4V matrix and SiC injected particles after laser embedding, and determined that the degree of dissolution of the particles in the matrix influences the mechanical properties of the composite material.

The influence of the reinforcement content in the microstructure and mechanical properties of TMCs has been analyzed by a few authors. Li et al. [28] fabricated gradient Ti–SiC coatings on Ti64 substrate using pure Ti and increasing percentages of SiC powder. They started from pure Ti without SiC and increased the SiC percentage by 10 vol% on each new layer deposited until a final proportion of 90 vol% SiC + 10 vol% Ti. They concluded that different reaction products were formed at each layer as a result of the in-situ reactions between Ti and SiC and that the Vickers microhardness increased gradually from the Ti64 substrate with the increase of SiC volume fraction. Li et al. [29] also manufactured two cladding layers of 90%Ti + 10%SiC and 80%Ti + 20%SiC on Ti6Al4V substrates and have reported that TiC and Ti3Si3 phases were formed, leading to a higher Vickers microhardness of the composite layer compared to that of the substrate. Li et al. [30] in whose work TiCp is selected in concentrations ranging from 5 to 50 wt% for Ti6Al4V matrices, where an increase in microhardness of up to 94% compared to the material unreinforced has been observed. Zhang et al. [31] have characterized the interfacial reaction that takes place between matrix and reinforcement in composite materials of Ti43Al9V reinforced with SiC fibers by SEM and TEM, identifying three types of reaction products formed between the C of SiC and the titanium matrix, such as TiC.

Other fabrication techniques such as selective laser melting (SLM) have been used to create Ti6Al4V with similar reinforcement contents. Krakhmalev et al. [32] fabricated pure Ti composites reinforced with 20, 30, and 40 wt% SiCp coatings on Ti6Al4V substrates with SLM and have studied the microstructure phase formation. The composites had higher hardness and higher abrasive wear resistance.

However, the influence of the DLD process parameters like the scanning speed in the microstructure, tribological and mechanical properties for Ti6Al4V/SiCp composites
has not been deeply studied yet. Yang et al. [33] have used a hybrid additive manufacturing method of laser cladding and layer deposition manufacturing (LCLD) to manufacture Ti6Al4V/\(\text{SiC}\) fiber composites and studied and optimized the processing parameters. Shojaei et al. [34] analyzed the effect of applying a Ti/\(\text{SiC}\) MMC coating on a Ti64 substrate for enhancing the hypervelocity impact resistance. Nevertheless, these authors did not fully characterize the fabricated composites nor made a microstructural evaluation. Also, some mechanical properties like microhardness and nanohardness were not made. Das et al. [35, 36] modified the surface properties of pure Ti by adding SiC particles at different laser power values and scanning speeds, creating two single layers of Ti–SiC composite coatings that exhibited higher hardness than the substrate. The researchers concluded that the composite fabricated with 400 W and 10 mm/s demonstrated the highest hardness, which is attributed to the formation of reaction products.

Moreover, just a few bibliographic studies show evidence of the reactivity that occurs between the SiC particles and the titanium from Ti6Al4V, and the work has been done just for low reinforcement percentages. This matrix–reinforcement reactivity leads to the formation of reaction products that can change the material properties and its stability. Furthermore, the effect of the variation of the percentage and size of SiC particles in the final material of Ti6Al4V/\(\text{SiCp}\) and its properties should be further analyzed.

This paper aims to determine the effect of the scanning speed of DLD additive manufacturing of Ti6Al4V/\(\text{SiC}\) metal matrix composites, in the morphology, microstructure, and micro and nanohardness properties of the fabricated samples. The role of the different SiC particle sizes and content in the matrix has been researched. The possible reaction mechanisms between the SiC and the titanium matrix have been proposed together with their effect on the microstructure. In addition, the relation of the process parameters with these mechanisms and their contribution to the sample microhardness have been analyzed.

2 Experimental procedure

Thin wall samples (length = 60 mm and height = 15 mm) of Ti6Al4V/\(\text{SiC}\) (Fig. 1a) were manufactured by DLD using 316 L sheets (100 × 100 × 20 in mm) as the 150 °C preheated support base. Ti6Al4V commercial powder has been used, and the effect of the SiC size has been evaluated. As received titanium and SiC particles are shown in Fig. 2a–c, and the particle size distribution for each material is shown in Fig. 2c–f. In all cases, the particle...
The size distribution was small (most Ti6Al4V particles were between 15 and 39 μm, the SiC F240 particles were between 43 and 83 μm, and the SiC F360 were between 20 and 38 μm.

The laser used was a high-power diode laser (1300 W) (ROFIN DL013S) with a wavelength between 808 and 940 nm. The carrying gas was argon (4.5 atm pressure and 0.05 L s\(^{-1}\) flow rate). The powder and the laser focus were
12.5 mm below the nozzle tip. The laser spot had a rectangular shape with 1.25 mm × 0.6 mm of size. The Ti6Al4V and SiC particles were in two different hoppers that worked simultaneously and were sprayed coaxially with the laser beam through a Fraunhofer IWS COAX 8 coaxial nozzle. The powders were sprayed at the Ti6Al4V:SiCp ratios shown in Table 1. The powder feed ratio was 5 g/min and the laser focus overlapped with the powder focus. In addition, this system is placed in an ABB IRB2400 robot. A hot plate connected to a temperature control system (150 °C) was used for substrate support.

The effect of using laser scanning speeds between 5 and 25 mm/s has been analyzed. Table 2 shows the set of samples fabricated with their corresponding fabrication conditions. To obtain the height increase per layer for each laser parameter used, a single layer with a length of 60 mm was deposited for each condition before the full thin wall fabrication (Fig. 1b). In this way, single layer heights were measured using a caliper with a precision of ± 0.01 mm, and these values determined the height increase per layer for the manufacturing process. Once this height increment per layer was established, thin-wall samples with a length of 60 mm made of Ti6Al4V/SiC were built layer by layer using the parameters shown in Table 2, as is shown in Fig. 1c. The samples height chosen was 15 mm. The number of layers depended on the height increment per layer required to reach the established height of the sample. In all cases, the laser power gradually decreases during the deposition process to avoid heat build-up. Some tests were performed at different laser powers to optimize the manufacturing parameters. Laser power values below 300 W could not melt the titanium particles. Increasing these values provided AM structures that were not homogeneously built. At the laser power value of 650 W, a high-quality first layer was deposited but maintaining this value in the following layers led to the melting of the manufactured samples. To avoid an excessive heat input, the laser power was gradually reduced from the initial 650 W to a final value of 100 W used for the last layer. Therefore, the power reduction per layer depended on the height of the samples and the number of deposited layers. Three samples were manufactured for each condition for reproducibility.

X-ray diffraction (XRD) was performed using a Panalytical X’Pert PRO diffractometer and analyzed the patterns by X’Pert HighScore Plus software. The radiation source was the Cu Kα. (λ = 1.5406 Å) operating at 45 kV and 300 mA. The 2θ scanning range used was from 20° to 95° with a step of 0.02° and an acquisition time of 20 s.

To evaluate the actual proportion of SiC particles incorporated into each sample, an X-ray fluorescence (XRF) analysis was performed using a Panalytical MagiX XRF equipped with a 4 kW X-ray generator and a rhodium anode. Samples were metallography prepared on their cross-section to measure their microhardness and study their micro-structure. For it, the samples were mounted in an electrically conductive resin, wet grounded using a sequence of abrasive

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### Table 1 Powder combinations used

| Ratio (Ti6Al4V:SiC) | SiC size | Percentage | Abbreviated name |
|---------------------|----------|------------|------------------|
| 4:1                 | F360     | Ti6Al4V–20 wt% SiC | Ti6Al4V/20SiC F360 |
| 4:1                 | F240     | Ti6Al4V–20 wt% SiC | Ti6Al4V/20SiC F240 |
| 3.5:1.5             | F240     | Ti6Al4V–30 wt% SiC | Ti6Al4V/30SiC F240 |

### Table 2 Set of DLD fabrication process parameters evaluated

| Condition | Reinforcement (% and size) | Scanning speed (mm/s) | Height increment per layer (mm) | Material abbreviated name |
|-----------|-----------------------------|-----------------------|-----------------------------|--------------------------|
| C1        | 20%, F360                   | 5                     | 1.25                         | Ti/20SiC F360-5          |
| C2        | 10                          | 0.77                  | Ti/20SiC F360-10             |
| C3        | 15                          | 0.57                  | Ti/20SiC F360-15             |
| C4        | 20                          | 0.38                  | Ti/20SiC F360-20             |
| C5        | 25                          | 0.15                  | Ti/20SiC F360-25             |
| C6        | 20%, F240                   | 5                     | 1.25                         | Ti/20SiC F240-5          |
| C7        | 10                          | 0.77                  | Ti/20SiC F240-10             |
| C8        | 15                          | 0.57                  | Ti/20SiC F240-15             |
| C9        | 20                          | 0.38                  | Ti/20SiC F240-20             |
| C10       | 25                          | 0.15                  | Ti/20SiC F240-25             |
| C11       | 30%, F240                   | 5                     | 1.25                         | Ti/30SiC F240-5          |
| C12       | 10                          | 0.77                  | Ti/30SiC F240-10             |
| C13       | 15                          | 0.57                  | Ti/30SiC F240-15             |
| C14       | 20                          | 0.38                  | Ti/30SiC F240-20             |
| C15       | 25                          | 0.15                  | Ti/30SiC F240-25             |
silicon carbide (120–4000 grit), and finally polished using a 1 μm diamond with ethylene glycol as a lubricant.

The percentage of SiC particles was measured with an image analysis software (Leica Application Suite) on the captured images obtained with a light optical microscope (OM; Leica DMR). Microstructures were evaluated by scanning electron microscope (SEM; Hitachi S-3400N) equipped with an energy dispersive X-ray spectrometer (EDX).

Vickers microhardness (HV$_{0.1}$ and 10 s) values were measured by using a Shimadzu microhardness tester on the cross-section of the specimens. The microhardness measurements were made from the substrate to the top of the sample with distances of 1500 μm between each indentation.

The nanohardness of the different phases of the composite DLD samples was measured using a Berkovich indenter (MTS nanoindenter). Nanoindentation matrices of 15 × 15 were made by applying 5 gf for 10 s, to analyze the nanohardness distribution. In addition, 10 × 10 nanoindentation matrices were made to evaluate the hardness of the individual phases in the material by applying 1 gf for 10 s.

3 Results

3.1 Dimensional characterization

A macro image of a Ti/20 SiC F360-15 part fabricated by DLD additive manufacturing is shown in Fig. 3a. The thin wall-like morphology of the sample can be observed. In addition, a well-defined layer-by-layer configuration can be appreciated on the image. An area of greater height is observed at the edges of the longitudinal section, where the height is 2 mm higher than in the rest of the part. This variation is caused by the higher time that the laser remains in those zones when it ends a line, changes its height, and starts the following layer in the opposite direction.

The thickness of the samples varied as a function of the process parameters. Figure 3b shows the thickness and speed used for each sample. The influence of the SiC particle size and the amount of reinforcement has been analyzed.

An increment in the scanning speed resulted in a decrease in the thickness of the samples independently of the rest of the factors. The scanning speed was related to the laser net energy input deposited on each layer and also to the amount of deposited material per layer and per unit of time. In addition, at lower scanning speed (up to 5 mm/s), the thickness increased with the increment in the SiC particle size, possibly because the SiC particles can absorb a greater amount of energy from the laser since they have higher absorbance than the Ti matrix. Nevertheless, at higher scanning speeds (above 5 mm/s), the effect of the SiC particle size was less significant and the thicknesses of the samples were rather similar. In the same way, the effect of the reinforcement rate followed a similar behavior: up to 10 mm/s resulted in a higher thickness (due to the effect of the higher energy absorption by the SiCp), and the thickness was similar for all the samples above 10 mm/s. This suggests that the effect of the scanning speed is the processing parameter that is the most relevant for the thickness, more than the SiC particle size or the amount of reinforcement deposited.

3.2 XRD characterization

Figure 4 shows the X-ray diffraction tests performed on the surface of different samples to analyze the effect of the scanning speed during the fabrication process, and the amount and size of reinforcement particles on the different compounds formed. In all cases, α and β phase peaks of the titanium matrix were observed.

The XRD spectrums of the unreinforced titanium fabricated by DLD using different laser scanning speeds Fig. 5a, b. The increment in the scanning speed from 5 (Fig. 5a) to 25 mm/s (Fig. 5b) resulted in an important decrease of the β phase peak.

The β phase peak appeared with similar intensity in all the reinforced samples independently of the scanning speed used. Matrix–reinforced interfaces act as new nucleation
zones for the phases. However, in all cases, an increase in the scanning speed leads to a decrease in the β phase. Furthermore, the addition of SiC particles in the titanium matrix produces a relative increase of the (100) α-titanium peak compared to the (101) α-titanium peak, which can be attributed to the preferential orientation of the alpha grains.

SiC peaks have been observed in all the metal matrix composites manufactured. The increment of the SiC peak (200) when the SiC size becomes larger (Fig. 4e–h) was detected. Additionally, titanium carbide (TiC) and titanium silicide (Si5Ti3) peaks appeared in all the metal matrix composite spectra. The partial dissolution of the SiC particles and the reactivity between Si and C and the Ti matrix leads to the formation of these compounds. Previous research has shown that the initial Ti and SiC powders resulted in the in-situ formation of multiphase intermetallic composites that contained TiC, Ti3SiC2, Ti5Si3, and SiC phases, while an equilibrium Ti3SiC2 was not detected [32]. Other authors have manufactured single-layer pure Ti/SiC composite coatings and have determined that laser parameters influence the dissolution of SiC, leading to the formation of Ti5Si3, Ti3Si1, and TiC intermetallic compounds [36]. For metastable states, Ti3Si1 is preferentially formed instead of Ti3SiC2 and TiSi2 [37].

### 3.3 Evaluation of SiC percentage on the manufactured composite materials

Table 3 shows the percentage of SiC particles measured by XRF in the DLD samples. The XRF identifies the presence of Si and, as this element does not appear in Ti6Al4V, from the proportion of Si, the amount of SiC that has been...
incorporated into the DLD samples can be determined. In all cases, the percentage of SiC in the samples was lower than the amount of SiC used during the fabrication. The values measured were independent of the scanning speed and only depended on the percentage of SiC in the mixture and the size of the particles. For the smallest SiCp used, the effective reinforcement rate was about 50% of the SiCp used in the fabrication. For the biggest SiC particles, the reinforcement rate was 20% lower for the Ti6Al4V/20SiC F240, and less than 5% lower in the case of the Ti6Al4V/30SiC F240 samples. Figure 5a shows the reinforcement percentage, for each material as a function of the laser scanning speed. This reinforcement rate was measured by image analysis and represents the percentage of SiCp that were still in the composite as SiCp. The values were smaller in all cases than those shown in Table 3, indicating that part of the SiCp dissolved in the matrix. The percentage of reinforcement increased with the amount of SiCp fed and with the size of the SiCp. The increase in the SiC particle size (from F360 to F240) produced a higher laser power density attenuation and, for the same scanning speed, the higher size of SiC particles need more time to complete their dissolution. Equally, the higher initial amount of SiC particles (from 20 to 30 wt%) increases the percentage of reinforcement after the AM fabrication process.

The undissolved reinforcement content in the materials after the AM fabrication process is directly related to the laser scanning speed. The higher the laser scanning speed, the higher the percentage of SiC particles. The cause of these differences in the percentage is the reactivity between the SiC particles and the Ti matrix. The matrix–reinforcement reactivity decreases when the laser scanning speed increases because it takes place in a liquid state during the formation of the molten pool. The cooling rates are higher with the scanning speeds increments, thus the reactivity decreases since particles get a shorter reaction time.

From the difference in the percentage of the SiCp in the composites and the incorporation of SiCp in the samples (calculated from XRF results), the proportion of dissolved SiCp was determined. Figure 5b shows the percentage of dissolved SiC particles in the composite matrix for the different manufacturing conditions. The smallest SiC particles were more prone to dissolve in the matrix than the biggest ones, as they have a greater specific surface. The percentage of dissolution decreased when the scanning speed increased due to the lower net energy input and the shorter contact time between the SiC particles and the molten Ti. An increase of 10 wt% in the SiCp (20–30 wt%) produces the double % of SiC dissolved in the matrix, which may be related to the higher temperatures achieved due to the higher absorption of the SiC particles.

### 3.4 Effect of the laser scanning speed on the microstructure

Figure 6 shows the Ti6Al4V matrix microstructure of samples manufactured with different scanning speeds (Fig. 6a, b for 5 mm/s and Fig. 6c, d for 25 mm/s). Figure 6a exhibits the microstructure of the composite layers

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**Table 3** Final SiC percentage incorporated in the composite material during the additive manufacturing process

| Material       | SiC percentage (wt%) |
|---------------|----------------------|
| Ti6Al4V/20SiC F360 | 10.4 ± 1.1          |
| Ti6Al4V/20SiC F240 | 16.3 ± 3.9          |
| Ti6Al4V/30SiC F240 | 28.5 ± 0.9          |

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**Fig. 5** a Percentage of reinforcement particles in the composites measured by image analysis and b percentage of dissolved SiC particles in the composite matrix
that are closer to the substrate, which consists of lamellas of α-Ti in a β-Ti matrix. The growth of primary α-phase with a fine platelet morphology and the decomposition of the β-matrix formed the characteristic microstructure called basket-weave. Similar structures have been found by other authors [38]. Figure 6b shows a micrograph at the top of the sample (last deposited layers) that reveals a mixture of equiaxed and lamellar α-Ti grains.

Figure 6c, d displays the Ti6Al4V microstructures at the bottom and top zones of the 25 mm/s manufactured sample. In both cases, the α–β Widmanstätten microstructure was identified, and the grains in the top layer were longer than in the bottom ones.

Figure 7a shows the cross-section of the Ti/20SiC F360-5 samples. Figure 7b depicts the microstructure of the different samples at different heights. A higher amount and bigger SiCp were observed for the faster laser scanning speeds. During the deposition, the particles can dissolve while the matrix is in the molten state, and the faster the laser scanning speed, the lower the heat input, and, therefore, the time in which the particles can dissolve, giving rise to a lower number of dissolved particles. This can be seen in Fig. 5, where a lower percentage of SiC was found compared to the initial Ti6Al4V:SiC ratio, due to that particle dissolution.

In addition, the amount and size of reinforcement are different in function of the zone of the cross-section. At the top of the sample, SiC particles are smaller than in the reception state. At the bottom, SiC particles are larger than those at the top but still smaller than the reception ones. As mentioned above, the SiC particles size in the composite depends on the heat input of the process.

In this case, the cooling rate at the bottom of the sample is higher due to the heat transferred by conduction through the substrate. However, the heat loss at the top of the sample is lower because it takes place principally by convection and conduction through the previous layers that remain at high temperatures.

Additionally, in Fig. 7, the products formed from the reaction between SiC particles and the titanium matrix can be observed around the partially dissolved SiCp. The reaction products are more dispersed in the matrix and are further from the surface of SiC particles when laser scanning speed is lower since the material is in a molten stage for longer and thus, the particles are transferred due to Marangoni forces.

Figure 8 shows details of SiC particles in the matrix of the Ti/20SiC F360-5 sample. Figure 8a exhibits the microstructure at the top of the sample and Fig. 8b at the bottom. In both, a Widmanstätten α–β titanium matrix can be observed, where the α phase corresponds to elongated grains...
(dark grey thick plaques) and the β phase is located at the grain boundaries (light grey fine plaques). DLD process was performed with high energy input conditions so high temperatures took place in the alloy that provided an in-process heat treatment of α′ martensite and decomposes α′ to the equilibrium α + β microstructure [39].

A partially dissolved SiC particle located at the top of the sample and its EDX element map is shown in Fig. 8a. The SiC is smaller than the reception size one because the particle is partially dissolved. No reaction products around the particle were observed, indicating that C and Si were dissolved in the matrix.

Figure 8b shows a SiCp partially dissolved at the bottom of the sample, together with its corresponding EDX element map. The SiCp was slightly smaller than the reception ones because the particle was partially dissolved. No reaction products around the particle were observed, indicating that C and Si were dissolved in the matrix.

Figure 8b shows a SiCp partially dissolved at the bottom of the sample, together with its corresponding EDX element map. The SiCp was slightly smaller than the reception ones because the particle was partially dissolved. No reaction products around the particle were observed, indicating that C and Si were dissolved in the matrix.

Figure 7 a Cross-section of Ti/20SiC F360-5; b Microstructures of the Ti/20 SiC F360 samples at the top, center, and bottom (rows) for different scanning speeds (columns)

conditions were less critical, as was analyzed before. In addition, smaller particles were observed around the main particle, forming a reaction products ring. EDX element analysis indicates that there is an Al-poor zone around the SiC particle. This area is rich in silicon, carbon, titanium, and vanadium because the silicon from the SiCp is a β eutectoid stabilizer element. Some Si-rich particles can be found distributed in the titanium matrix. C-rich dendrites and eutectic compounds are located around the SiC particle. During the fusion-solidification process, the partially dissolved SiC particles increased the Si and C content in the titanium matrix. Si and C react with Ti and form different compounds.

Figure 9 shows the microstructure of the Ti/20SiC F360 sample fabricated using 20 mm/s. As in the previous case, a Widmanstätten α + β titanium matrix was observed, and partially dissolved SiC particles were present. However, in this case, the size of these particles was higher because of the
faster laser scanning speed, so the heat input was lower than in the previous case. Particles were preferentially observed in the central zones of each layer, while there were no particles at the overlapping zones between layers (arrowed zones in Fig. 9a), and no significant differences between the SiC particles at the top or the bottom of the sample. This suggests that the particles dissolved in the overlapping zones, where the highest temperatures were achieved. Also, lower adhesion between the reinforcement and the matrix was observed in many particles (Fig. 9b).

Around the particles, there was a reaction zone (Fig. 9c). C and Si from the particles also react with the liquid titanium matrix and form different compounds. C-rich dendrites zone near the partially dissolved SiC particles is smaller than in the previous case due to the lower heat input (Fig. 9c, d). This reduced the temperatures reached by the molten pool and the tendency of the alloy to dissolve the particles. Therefore, the amount of C and Si diluted in the matrix and the migration distance were smaller. This caused the local formation of TiC and the appearance of the eutectoid. In addition, eutectic compounds are observed in zones far away from the SiC particles. Finally, strip-like eutectoid Si-rich compound is identified in the α-Ti phase limits (Fig. 9c). According to XRD (Fig. 4), the C-rich compound that is originated around partially dissolved SiCp is TiC and the Si-rich compound corresponds to Ti₅Si₃.
3.5 Effect of the SiC particle size on the microstructure

The effect of the increase in the SiC particle size (from F360 to F240) was analyzed. Figure 10 shows the effect of the laser scanning speed on the microstructure of the Ti/20SiC F240 samples. As in Fig. 7, three cross-section images of each sample are shown to illustrate the microstructure at the top, middle, and bottom of the different samples.

SiC particles partially dissolved independently of the scanning speed and location. However, due to their higher size, the dissolution was not complete in any case. Samples...
presented a homogeneous distribution in the part with no significant differences in the microstructure between the top and bottom zones.

Nevertheless, these differences are noticeable depending on the laser scanning speed. As in the previous case, reaction products originated from the reaction between Ti and SiC were identified, and its distribution depended fundamentally on the scanning speed and, therefore, on the heat input.

A ring of reaction products around the SiC particles is formed for all manufacturing conditions. The ring thickness was lower when the reaction time between the Ti matrix and the SiC particle is shorter, i.e., when samples were manufactured at lower scanning speeds.

From 20 mm/s laser scanning speeds, thinner rings around the SiCp with reaction products around them and dispersed in the matrix are observed. Moreover, cracks between the SiC particles and the titanium matrix appear since they are interfacially detached due to the higher cooling rates. However, the rings formed in the samples fabricated at 5 mm/s are thicker and the reaction products are homogeneously distributed in the matrix.

Figure 11a–c shows the microstructure of the Ti/20 SiC 240-5 sample. Figure 11a (taken at low magnification), exhibits partially dissolved SiCp in a matrix in which two different microstructures were present. These are formed because of the interfacial reaction between matrix and reinforcement during the manufacturing process, which implies a C and Si enrichment in the zones closer to particles. In zone 1 (around particles), the formation of a TiC compact ring has taken place, but this TiC gets more dispersed with a dendritic morphology as the distance to the SiCp increases. Equally, Si combines with Ti forming the intermetallic Ti₅Si₃ phase (brighter on the images) that appears following the grain boundaries in the matrix and that belongs to the Ti₅Si₃-α eutectoid microconstituent in the zones closer to the SiCp (zone 2).

Figure 11d–f exhibits the Ti/20SiC 240-25 sample microstructure. Here, a TiC ring is created around SiCp and it is much thinner and less compact (Fig. 11e), which indicates that particles dissolve to a less extent due to a lower heat input for greater scanning speeds. For this reason, the α-Ti₅Si₃ eutectoid microconstituent zone is less extensive and matrix areas with Ti₅Si₃ that follow grain boundaries of the α-β titanium matrix increase (Fig. 11f).

### 3.6 Effect of the percentage of reinforcement on the microstructure

Figure 12 shows the effect of the laser scanning speed on the microstructure of the Ti/30SiC F240 samples. Three images of the cross-section of each sample are shown to illustrate the microstructure at the top, middle, and bottom of the different samples.

The increase in the amount of SiC particles in the deposited material results in the presence of partially dissolved SiC particles in all the samples. As in the previous cases, a ring of reaction products was observed around the particles. An increment in the laser scanning speed
Fig. 11  a general microstructure of Ti/20 SiC F240-5; b detail of SiCp of Ti/20 SiC F240-5; c detail of zone 2 in the matrix of Ti/20 SiC F240-5; d general microstructure of Ti/20 SiC F240-25; e detail of SiCp of Ti/20 SiC F240-25; f detail of zone 2 in the matrix of Ti/20 SiC F240-25.
resulted in a variation in the thickness of the reaction products layer, which was greater for lower scanning speeds, i.e., higher input energy. The reaction products were dispersed in the Ti–30SiC F240 matrixes and, in all cases, regardless of the scanning speed, equiaxial dendrites of reaction products were identified. The higher the amount of SiCp in the raw material, the higher the absorption, because these particles have lower reflectivity than the titanium powder, thus, the molten pool temperature increased along with subcooling (higher thermal conductivity) and this resulted in the formation of equiaxial dendrites [28].

Figure 13 shows a detail of the microstructure of two different samples fabricated with the same laser scanning speed (10 mm/s) but different amounts of reinforcement in the raw material. The microstructures of a sample with 20 wt% of SiC reinforcement (Ti/20SiC F240) and a sample with 30 wt% SiC (Ti/30 SiC F240) are shown in Fig. 13a, b, respectively.

The matrix–reinforcement interface presented cracks in the case of lower amounts of SiC particles (Fig. 13a, SE image), and higher dissolution evidence was identified at the SiC surface (black arrow) for samples with higher amounts of SiC particles (Fig. 13b, SE image).

Furthermore, in Fig. 13a (BSE image) the reaction products were less equiaxial than in Fig. 13b (BSE image) due to the lower temperatures reached during the molten pool formation as a result of the lower SiCp percentage.

Figure 14a, b shows the microstructure of Ti/30SiC F240 samples fabricated at 5 and 25 mm/s of scanning speed, respectively. As before, there were reaction products rings around the particles. In both cases, this ring was mainly composed of TiC dendrites. A detail of the matrix-reinforcement interface for each scanning speed can be seen in Fig. 14c, d. These TiC dendrites were smaller and rounder for higher scanning speeds (Fig. 14d) because an increase in the scanning speed during the fabrication process results in a shorter reaction time. In addition, the TiC dendrites have less time to distribute in the molten titanium matrix before solidification takes place.

There was greater silicon enrichment at lower scanning speeds compared to higher speeds (demonstrated in Fig. 5). In this way, primary Ti5Si3 with polygonal morphology and TiC precipitated in zones close to the SiCp at low speeds were observed (Fig. 14a, c). However, at higher scanning speeds, a greater presence of the α-Ti5Si3 eutectoid microconstituent in the titanium matrix was detected (Fig. 14b, d).

Figure 14e shows a detail of the matrix in a zone that is far away from the SiC particle in the sample manufactured using 25 mm/s of laser scanning speed. A high amount of α-Ti5Si3 eutectoid microconstituent was identified in the α–β matrix.
3.7 Microhardness and nanohardness

Figure 15a, d shows the evolution of the microhardness across the section of the manufactured samples. The microhardness of the non-reinforced Ti6Al4V sample can be seen in Fig. 15a. For scanning speeds of up to 15 mm/s, there is a slight increase in the microhardness from the bottom to the 4 mm height. From here on, the microhardness values remained constant. For scanning speeds above 15 mm/s, the microhardness values were similar across the entire sample and were higher for the 25 mm/s scanning speed. These variations are related to the microstructural modifications observed in Fig. 6. As discussed above, for the lower scanning speeds, the microstructure considerably changed from basket-weave type to a mixture of equiaxed and lamellar α-titanium. For high scanning speeds, the microstructure was similar in the sample. It consisted of an α–β Widmanstätten structure with grains that became longer in the last deposited layers.

Figure 15b shows the microhardness of the Ti/20SiC F360 sample. There is an increment in the microhardness values in the first 4 mm; then the values stabilize. The values obtained for the two fastest scanning speeds were higher than for the other conditions. At the top of the samples, all the particles were diluted, but particles remained partially undiluted in the bottom. Microhardness measures the hardness of the matrix, and the matrix was hardened by the reaction products formed from the dissolution of the SiCp in the titanium matrix, which had a very small size and were homogeneously distributed.

The microhardness values for Ti/20SiC F240 and Ti/30SiC F240 samples are shown in Fig. 15c, d, respectively. Once again, no clear variation in the microhardness was detected in the cross-section of the samples or between layers. There was a partial dissolution of particles in the matrix that would contribute to the increase in the hardness values. The hardness values were less homogeneous than those with smaller particles, presumably because of the effect of having a smaller number of bigger particles, which induce more local differences between zones in the samples.

The microhardness evolution from the bottom to the top of the sample usually follows similar trends, but it is also affected by the speed. To identify the effect of the scanning speed for the different materials manufactured, a resume of the average values in the central zone (between 4 and 10 mm) has been plotted in Fig. 15f. The addition of SiC particles as reinforcement of the titanium matrix resulted in an increase in the microhardness in most cases. The samples fabricated with the bigger SiC (F240) present higher microhardness values than the unreinforced samples, independently of the percentage of reinforcement and the laser
scanning speed. However, in the case of the samples fabricated with SiC F360 size, and for scanning speeds under 15 mm/s, the microhardness of reinforced and unreinforced samples is very similar and slightly lower for the reinforced samples. This occurred as a result of the dissolution extent of the SiC particles during the fabrication process. When the scanning speeds were 15 mm/s and 20 mm/s, the reactivity decreases, and the microhardness of the reinforced samples was higher than that of the unreinforced. At 25 mm/s, the microhardness of the unreinforced sample is similar to that of the reinforced one probably because higher cooling rates resulted in a finer microstructure.

A correlation between the microhardness and the size and amount of reinforcement has been established. For a given material (SiC size and percentage), there is a limit value of the laser scanning speed from which the microhardness remains constant, independently of the scanning speed used. For the material reinforced with SiC F360, the scanning speed limit value was 20 mm/s. At this speed, the percentage of reinforcement was maximum for this SiC size (Fig. 4a) and an increase in the scanning speed did not increase the microhardness. However, below 20 mm/s, the higher the scanning speed, the higher the microhardness. These differences in the microhardness are due to two different factors: (i) the higher the scanning speed, the higher the cooling rates, and this affects the microstructure of the samples, which results in a finer microstructure that leads to a higher microhardness; (ii) the SiCp-Ti matrix interactions and reactivity result in the decrease of the SiC particles percentage and the consequent lower microhardness.

Furthermore, the microhardness increases with the increment in the size of the SiC particles. In all cases, the samples fabricated with a larger SiC (F240) present a higher microhardness than the samples fabricated with SiC F360, regardless of the laser scanning speed (similar microhardness is measured no matter what values of the laser scanning speed are used during their fabrication process). The reason is that the percentage of reinforcement in the middle of the sample is similar in all cases (Fig. 4a).

In addition, the microhardness increases with the increment of reinforcement percentage. Samples fabricated with 30 wt% SiC particles have presented a higher microhardness. However, they have similar microhardness values when the scanning speed is 10 mm/s or higher, even though the higher real percentage of reinforcement was found at 20 mm/s (Fig. 4a) and was similar in all cases for 10 mm/s of laser scanning speed.

In previous research, it has been demonstrated that pure Ti/SiC coatings present an increasing Vickers microhardness with the increase of SiC volume fraction, from 340 HV of the Ti64 matrix to 1600 HV of 10%Ti/90%SiC [28]. Li et al. [29] also manufactured pure Ti/SiC materials on Ti6Al4V substrates. They concluded that Vickers microhardness was increased from 340 HV for the Ti64 substrate to 930 HV for the 20%Ti/80%SiC layer. Krakhmalev et al. [32] proved that the increase in the SiC content in pure Ti/SiC composites increased the hardness of the coatings, due to the obtained high values of 26–36 GPa for SiC compared to those for TiC, Ti5Si3, and TiSi2 constituents. Das et al. [35] created Ti/SiC composite coatings with a microhardness of 900 HV
Fig. 15  Evolution of the microhardness across the section of the samples: a Ti6Al4V; b Ti/20SiC F360; c Ti/20SiC F240; d Ti/30SiC F240; f Effect of the scanning speed in the microhardness for each AM material. Microhardness measured on the middle cross-section zone.
for materials manufactured at 300 W and 20 mm/s, and 1100 HV for 400 W and 10 mm/s, compared to the Ti substrate that exhibited values of 232 HV. Other authors reinforced porous Ti getting microhardness values of 625 HV for 3 wt% SiC content, but by using conventional powder metallurgy methods and not AM [20].

However, no data has been found in the literature for additively manufactured Ti6Al4V/SiC bulk composites fabricated with a one-step method. Therefore, evidence of improved mechanical properties (Vickers microhardness) has been shown, with values of ~500 HV for Ti64 up to ~950 HV for Ti/30%SiC F240 manufactured with the right processing parameters (15 mm/s of scanning speed and 650–100 W of laser power).

Nevertheless, the effect of the reaction products originated from the reaction between the SiC particles and the titanium matrix and their distribution across the titanium matrix from the surface of the SiC particles do not seem to affect the microhardness of the sample. Previous microstructures showed a relation between the reaction products distribution and the laser scanning speed. For this reason, this effect of the reaction products distribution in the titanium matrix has been analyzed by the nanoindentation technique.

Figure 16 shows the distribution of nanohardness on the analyzed surfaces. All the measurements were carried out in the central zone of the cross-section of the samples. Figure 16a shows the hardness for the Ti/20SiC F360-5 part. It can be observed that the measured hardness was between 5 and 12 GPa, with an average hardness of 10 GPa. There were no areas of high hardness, and hardness remained homogeneous throughout the entire surface.

Figure 16b shows the nanohardness for the Ti/20SiC F360-25 part. Increasing the speed increased the hardness of the matrix by 33% (from an average value of 10 GPa when samples were manufactured at 5 mm/s to an average value of 15 GPa (between 12 and 17 GPa) when they were manufactured at 25 mm/s). In addition, when increasing the speed, areas with hardness above 30 GPa were observed, while this increase did not take place when the speed was 5 mm/s. These areas correspond to zones in which there was an agglomeration of SiC particles. Likewise, there was an increase in the matrix hardness around these SiC agglomeration zones with an average hardness of 20 GPa. This hardness can be associated with the presence of dispersed reaction products in the matrix from the partial dissolution of the SiCp and subsequent reaction between silicon and carbon (SiC) with the titanium from the matrix.

In Fig. 16c the nanohardness for the Ti20SiC F240-25 part is represented. As the particle size increases, an increase in the matrix hardness of 25% relative to the part manufactured with the same scanning speed and smaller particle size takes place (average value of 20 GPa). As the particles did not dissolve, they created a more accused gradient of hardness because of the presence of the particles and the reaction products formed around them, and because they form defects and residual stress during cooling that hardens the material. Therefore, the hardness pattern is directly correlated with the particle distribution in the samples.

Figure 17 shows a detail of the nano-hardness distribution in the reinforcing matrix for two parts manufactured at different speeds for the Ti/20SiC F240. The nanohardness of the sample manufactured at 10 mm/s can be seen in Fig. 17a and at 25 mm/s in Fig. 17b.

The particles shown in Fig. 17a are partially dissolved, which reduces the hardness of the particle when compared to that of Fig. 17b, where just the surface is partially dissolved. A ring of reaction products is observed around SiC particles in Fig. 17a, which is visibly wider than that of the particle in Fig. 17b. Although the microstructures showed that the reaction products were more homogeneously distributed in the samples fabricated at low scanning speeds, their presence does not have a relevant influence on the nanohardness distribution in the area close to the particle. The decrease in hardness from the particle to the matrix is equally gradual in both cases, regardless of the ring thickness. Another research has shown results for the nanohardness of Ti/SiC composite coatings manufactured with similar scanning speeds, and obtained 19.6 ± 2.6 GPa for 20 mm/s and 300 W, and 15.0 ± 2.2 GPa for 10 mm/s and 400 W [36]. However, in the present study, values of up to 90 GPa were measured in the matrix–reinforcement interface, followed by a zone with an average nanohardness of 60 GPa, which corresponds to the ring of reaction products. The matrix in the sample manufactured at 10 mm/s had a nanohardness between 5 and 20 GPa and values higher than 20 GPa were measured for the part manufactured at 25 mm/s. This increase in the matrix hardness was associated with the higher content of non-dissolved SiC particles in the lower layers of the bulk material. The presence of non-dissolved SiC particles had a higher contribution to the matrix hardening than the reactivity SiC–matrix products.

4 Discussion

The hardening of the titanium matrix in Ti6Al4V/SiC composites is associated with (i) the scanning speed used during the sample fabrication process; (ii) the amount and size of SiC particles; and (iii) the amount and distribution of reaction products in the composite matrix. Regardless of these factors, the increase in the scanning speed, which reduces the heat input and increases the cooling rates, resulted in finer microstructures with higher microhardness.

On the other hand, the microstructural analysis revealed that microstructures were similar across the samples in most
Fig. 16  Nanohardness distribution map of a Ti/20SiC F360-5; b Ti/20SiC F360-25 and c Ti/20SiC F240-25
Fig. 17 Detail of the particle–reinforced interface and nanohardness and elastic modulus distribution map for a Ti–20SiC F240-10 and b Ti–20SiC F240-25
manufacturing conditions. The greatest differences occurred for unreinforced matrices and the smallest reinforcement size at low scanning speeds. This indicates that the cooling rate was homogeneous throughout the manufactured sample, what has been achieved by the reduction of the laser power during the manufacture, and the increase of the temperature of the platform. The variation in the microstructures of the composite materials is related to the difference in the heat input as a function of the laser scanning speed. The microstructural transformations will be influenced by the scanning speed.

When the molten pool is formed during the manufacturing process, the titanium matrix in liquid state dissolves the SiC particles, resulting in an enrichment in C and of the molten matrix. This mechanism is more pronounced when the temperature is higher and the contact time between the molten matrix and the particles is longer. Reducing the scanning speed increases the heat input and the interaction time, so lower scanning speed promote the dissolution of the SiCp.

According to Das et al. [36], the formation of reaction layers takes place in SiC/Ti materials as a result of the reactivity between Ti and C (preferably rather than between Ti and Si). They reported that the solidification begins at the SiC surface after the partial dissolution of the particles and that Si combines with Ti in the molten pool to form Ti5Si3. Near the SiC particle, the carbon content is high, and the alloy is in the hypereutectic zone of the Ti–C phases diagram. When the TiC rich layer on the surface particle is smaller, primary TiC dendrites near the SiCp are formed. The reactions that occur during the solidification and the cooling process are the following:

\[ L \rightarrow L + \text{primary TiC} \]  
\[ L \rightarrow \beta + Ti + TiC \]  
\[ \beta + TiC \rightarrow \alpha + Ti + TiC \]  

On the other hand, the silicon content in the titanium after the reaction layer is very high. As discussed above, Si is a \( \beta \) eutectoid element that leads to \( Ti_3Si_3 \) when combined with Ti at high temperatures. Primary \( Ti_3Si_3 \) is formed at the grain boundaries in the titanium matrix following the reaction [29]:

\[ L_1 \rightarrow L + Ti_3Si_3 \]  

During the cooling times of the liquid phase, the \( \beta \)-Ti\( _3 \)Si\( _3 \) eutectoid microconstituent forms, which will turn into a during the solid cooling. As the cooling of the parts occurs in metastable conditions, the transformation of the Ti\( _3 \)Si\( _3 \) compound will not happen, according to Fiore et al. [37].

\[ L_2 \rightarrow \beta Ti + Ti_3Si_3(\text{eutectic}) \]  
\[ \beta Ti + Ti_3Si_3(\text{eutectic}) \rightarrow \alpha Ti + Ti_3Si_3(\text{eutectoid}) \]  

The reactivity between SiC particles and the Ti matrix results in a decrease in the SiC particles percentage and lower microhardness; however, the lower the heat input, the lower the reactivity and the higher the microhardness.

This approach implies that the TiC dendrites zone near the partially dissolved SiC particles in the Ti/20SiC F360-20 sample is smaller than in the sample fabricated using 5 mm/s, because the C rich zone is smaller due to the lower heat input. In the same way, the silicon content is lower due to the lower dissolution of the SiC particles, so, strip-like eutectic Ti\( _3 \)Si\( _3 \) can be observed in the \( \alpha \)-Ti phase limits.

The increment in the SiC size in the Ti/20 SiC F240 samples increases the microhardness, mainly at low scanning speeds because, although there is a reactivity between the particles and the titanium matrix, the size of the particles after the partial dissolution is higher. The increment in the scanning speed resulted in less reactivity since the net energy input is lower, so, for this reason, the difference in the microhardness in samples fabricated at high scanning speeds of 25 mm/s and with a variation in the particle size is lower. The increase in the SiC size has another effect on the microstructure: higher contents of Si and C are present in the matrix due to the higher reactivity at the surface of the SiC particles. For this content of Si and C, the formation of titanium carbides and silicides takes place, independently of the laser scanning speed and the zone of the 3D sample. A possible reaction mechanism may follow reactions (1) and (4) to form TiC and primary Ti\( _3 \)Si\( _3 \), followed by reactions (2), (3) and (4).

Finally, the increment in the percentage of large SiC particles (F240) results in a higher microhardness independently of the chosen process parameters. This is due to the partial dissolution of SiC particles, they are well distributed in the matrix and keep their hardening effect. TiC also provides a hardening effect, thus, the higher the TiC proportion, the higher the microhardness. In addition, the molten pool temperature and the subcooling increase given the higher absorptivity of the SiC particles which increases the heat absorption. Therefore, the thermal conductivity increases, and this results in the formation of equiaxial TiC dendrites.

5 Conclusions

In the present work, the influence of DLD additive manufacturing process parameters, such as laser scanning speed, in the morphology, microstructure, and microhardness of the resulting Ti6Al4V/SiCp composite materials has been
analyzed. Similarly, the effect of both size and concentration of SiC reinforcement introduced into the titanium matrix has been studied. It can be concluded that:

- A relation between the laser scanning speed and the geometry, microstructure, and microhardness of the manufactured parts has been established. The amount and size of reinforcement particles also influence these properties.
- The laser scanning speed is the most relevant factor in the thickness, more than the SiC particle size or the amount of reinforcement in the matrix. The faster the laser scanning speed, the lower the measured thickness of the fabricated composite. Furthermore, the thickness increases with the increment in the SiC particle size and percentage for slower scanning speeds and hardly varies for higher speeds.
- The final percentage of incorporated particles measured in the manufactured materials is lower than expected due to the matrix-reinforcement reactivity. The higher the laser scanning speed, the higher the final percentage and size of SiC particles.
- TiC, primary Ti₅Si₁₃, and eutectoid αTi + Ti₅Si₁₃ were formed as a result of the reaction between the Si and C from the reinforcement and the titanium of the matrix.
- The Ti matrix is an α–β Widmanstätten type reinforced with partially dissolved SiC and the reaction products form a ring in which both microstructure and thickness vary with the used scanning speed during the additive manufacturing.
- The matrix–reinforcement reactivity decreases with increasing laser scanning speeds since this reactivity takes place in the molten pool. The heat energy input decrease with higher scanning speeds and therefore, reaction times are shorter.
- Micro and nano hardness increase with the percentage and size of SiC reinforcement and with increasing laser scanning speed. The higher the scanning speed, the lower the heat energy input rates which result in finer grain size and higher hardness values.

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**Declarations**

**Conflict of interest** All authors declare that they have no conflict of interest.

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