

Article

Synchronized Multi-Laser Powder Bed Fusion (M-LPBF) Additive Manufacturing: A Technique for Controlling the Microstructure of Ti-6Al-4V

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Abstract: One of the technological hurdles in the widespread application of additive manufacturing is the formation of undesired microstructure and defects, e.g., the formation of columnar grains in Ti-6Al-4V—the columnar microstructure results in anisotropic mechanical properties, a reduction in ductility, and a decrease in the endurance limit. Here, we present the potential implementation of a hexagonal array of synchronized lasers to alter the microstructure of Ti-6Al-4V toward the formation of preferable equiaxed grains. An anisotropic heat transfer model is employed to obtain the temporal/spatial temperature distributions and construct the solidification map for various process parameters, i.e., laser power, scanning speed, and the internal distance among lasers in the array. Approximately 55% of the volume fraction of equiaxed grains is obtained using a laser power of $P = 500$ W and a scanning speed of $v = 100$ mm/s. The volume fraction of the equiaxed grains decreases with increasing scanning velocity; it drops to 38% for $v = 1000$ mm/s. This reduction is attributed to the decrease in absorbed heat and thermal crosstalk among lasers, i.e., the absorbed heat is higher at low scanning speeds, promoting thermal crosstalk between melt pools and subsequently forming a large volume fraction of equiaxed grains. Additionally, a degree of overlap between lasers in the array is required for high scanning speeds ($v = 1000$ mm/s) to form a coherent melt pool, although this is unnecessary for low scanning speeds ($v = 100$ mm/s).

Keywords: laser powder bed fusion; equiaxed microstructure; multi-beam laser; melt pool; process variables; Ti-6Al-4V



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1. Introduction

Laser powder bed fusion (LPBF) additive manufacturing techniques can fabricate complex 3D parts on demand using localized networks [1,2], drastically changing the landscape of future manufacturing. However, its current application is limited by defect formation (e.g., pores, spatter, and cracks), dimensional inaccuracy, and undesired microstructures that result in the final part's poor material properties (e.g., lower yield stress, anisotropic mechanical properties, and fatigue endurance limit [3]). Among all the issues above, the formation of undesirable microstructures and the consequent degradation of mechanical properties pose a significant hurdle to the full adoption of this technology. For instance, the laser powder bed fusion (LPBF) of Ti-6Al-4V is prone to forming a columnar microstructure aligned with the built direction [4], which reduces the endurance limit. Note that these microstructural issues are more process-dependent (due to a fast heating and cooling rate in the LPBF process) than material-dependent and have been observed in the LPBF manufacturing of a wide range of materials, such as high-entropy alloys [5], high-strength aluminum alloys [6], 316L steel [7], Invar [8], and Ni-based superalloys [9]. Therefore, any proposed method to control the microstructure has a significant impact on the field.

Previous studies [10,11] show that the thermal gradient and solidification rate are two parameters that can control the microstructure (equiaxed vs. columnar). One novel

method to tune the microstructure by controlling the thermal gradient and cooling rate is beam shaping [12–15], which utilizes a non-traditional beam shape instead of the Gaussian beam profile, e.g., ring, top-hat, and elliptical beams. The temperature profile is more uniform than that of Gaussian beams, which leads to a higher percentage of equiaxed grains. This objective can also be achieved by employing the synchronized multi-beam additive manufacturing technique, which can combine numerous Gaussian beams to form a wide range of beam shapes [16,17]. The synchronized multi-beam can be used in various configurations, such as hexagonal, circular, square, and linear, making it a versatile beam-shaping approach [17]. However, few studies have investigated the relationship between synchronized multi-laser-based additive manufacturing and microstructure. For instance, a multi-beam Rosenthal solution and a semi-analytical approach were used to model the thermal gradients of two coordinated laser beams for Ti-6Al-4V [18]. The results showed that this technique could not alter the microstructure, and the material still comprised columnar grains.

Additionally, an analytical solution was used in [19] to optimize the multi-beam laser configurations (five laser beams) to investigate the lasers' configuration on the microstructure of Inconel. The computational results showed that one could increase the percentage of equiaxed grains by precise control over laser positioning, increasing the probability of equiaxed grain formation to 99.8%. Notably, the analytical solutions are overly simplified and only provide a steady-state solution, cannot predict complex beam paths, and ignore the effects of latent heat, radiation, and convection [20]. A dual laser configuration was implemented in [21], in which the secondary laser acts as a post-treatment process. This dual configuration created a more uniform microstructure; however, the grains were still columnar. A similar configuration was implemented for Inconel, which showed that having a secondary laser beam near the melt pool significantly alters the microstructure, density, and surface roughness [22]. Finally, a comprehensive computational model, including temperature dependency of material properties, Marangoni effects, radiation, and convection, was developed for two synchronized lasers with various configurations, i.e., layer-wise and spatially synchronized [23]. However, the impact of laser configuration on the microstructure was not investigated. Other studies on multi-beam lasers focused on reducing residual stress by using a laser to preheat the powder bed [24–27], increasing productivity [28], and variation in mechanical properties (microhardness and tensile strength) of the overlapped region in a dual laser configuration [29–31]. In short, there is no study on the effect of a synchronized hexagonal laser configuration on the microstructure, and a basic understanding of the underlying physics is still lacking. Note that the hexagonal laser configuration is selected here because of its potential to change the thermal distribution around the melt pool and its complexity compared to linear configurations.

It is worth noting that the main focus of this work is on grain structure, and other microstructural issues, such as the martensitic phase transition, were ignored. The martensitic phases reduce the ductility of Ti-6Al-4V; however, it can be limited through alloying [32], post-treatments [33], changes in border design [34], and optimization of process parameters [35]. It can also be refined through high-energy pulsed laser surface re-melting, which increases the corrosion resistivity of the material compared to as-cast components [36]. Finally, note that the thermal gradient and solidification rates are not the only available methods to control the grain structure, and a wealth of other approaches, such as grain boundary modification, optimization of process control, and alloy compositions, have been actively pursued to attain the desired microstructure in as-built fabricated parts [37]. A seminal review of the correlation between melt pool characteristics and microstructure can be found in [38].

2. Governing Equations and Model Verification

The MOOSE (Multi-Physics Object-Oriented Simulation Environment) software is used to solve the heat conduction equation and predict melt pool dimensions. The powder bed dimensions are $1 \times 1 \times 0.5$ mm for a single laser and $5 \times 1.5 \times 1.5$ mm for the multi-

laser configuration. Eight-node hexahedral elements with dimensions of $10 \times 10 \times 7.5 \mu\text{m}$ are employed to mesh the simulation domain; note that the powder diameter is typically around $30\text{--}50 \mu\text{m}$. Additionally, a finer mesh, $5 \times 5 \times 5 \mu\text{m}$, was utilized to check the sensitivity of the results to the mesh size, and it was found that the melt pool dimension only changed by 0.1%. Therefore, the $10 \times 10 \times 7.5 \mu\text{m}$ mesh is used to reduce the computational time and cost for all simulations. The size of the simulation domain is adequate for transient thermal analysis since it is at least ten times larger than the thermal diffusion length based on the laser velocity range, $l_{th} = \sqrt{Dr_0/v} = 0.036\text{--}0.025 \text{ mm}$ [39], where v is the scanning velocity, D is thermal diffusivity, and r_0 is laser beam radius. The laser power and velocity range (see Table 1) are selected to ensure that melting occurs predominantly in the conduction regime [40]; see Figure S1 in the Supplementary Material. All lasers in the hexagonal configuration have the same power, scanning speed, and beam radius. The distance between the lasers in the hexagonal configuration is constant during scanning, and the powder bed is scanned simultaneously by all lasers in the array.

Table 1. Model and laser parameters for LPB-AM of Ti-6Al-4V [41].

Parameter	Notation	Value
Solidus temperature (K)	T_S	1878
Liquidus temperature (K)	T_L	1993
Melting temperature (K)	T_m	1923
Absorption coefficient	A	0.25
Laser beam radius (μm)	r_o	75
Laser power (W)	P	50, 300, 500
Laser speed (mm/s)	v	100–1000
Melting enthalpy (kJ/kg)	H_f	370
Thermal conductivity enhancement factors	$\lambda_x, \lambda_y, \lambda_z$	10, 10, 15
Heat transfer convection ($\text{W}/\text{m}^2\text{K}$)	h_c	50
Heat transfer radiation ($\text{W}/\text{m}^2\text{K}$)	h_r	1

Governing Equations: The transient temperature distribution is obtained using the anisotropic thermal conductivity model.

$$c_p \rho \frac{\partial T}{\partial t} = \frac{\partial}{\partial x} \left(\lambda_x k \frac{\partial T}{\partial x} \right) + \frac{\partial}{\partial y} \left(\lambda_y k \frac{\partial T}{\partial y} \right) + \frac{\partial}{\partial z} \left(\lambda_z k \frac{\partial T}{\partial z} \right) - \rho H_f \frac{\partial \alpha}{\partial t}. \quad (1)$$

where c_p is the specific heat capacity, ρ is density, λ_n is the thermal conductivity enhancement factor, k_n is the thermal conductivity, H_f is the melting enthalpy, α is the melt fraction ($\alpha = 1$ indicates the liquid (melt) and $\alpha = 0$ represents the solid phase), T is the powder bed temperature, and t represents time. The melt fraction parameter can be defined as follows:

$$\alpha = \begin{cases} 0, & T < T_S \\ (T - T_S)/(T_L - T_S), & T_S < T < T_L \\ 1, & T > T_L \end{cases} \quad (2)$$

where T_S is the solidus temperature, and T_L is the liquidus temperature. This simplified model mimics the effect of Marangoni convection in the melt pool and considers heat

absorption during melting. Additionally, the Gaussian heat flux was used to model a moving laser heating the top surface.

$$q(x, y) = \frac{2PA}{\pi r_o^2} \exp\left(-\frac{2((x - vt)^2 + y^2)}{r_o^2}\right). \quad (3)$$

where A is the laser absorption coefficient, and v , P , and r_o are the laser beam velocity (scanning speed), power, and spot radius, respectively. Additionally, convection and radiation boundary conditions were added to all surfaces except the bottom of the powder bed.

$$\dot{q} = h_c(T - T_0) + h_r(T^4 - T_0^4) \quad (4)$$

$$\dot{q} = q(x, y) + h_c(T - T_0) + h_r(T^4 - T_0^4) \quad (5)$$

where h_c and h_r are the convective and radiation heat transfer coefficients, respectively, and T_0 is room temperature. The symmetric boundary condition was applied to reduce computational cost. All model parameters are summarized in Table 1, and the Ti-6Al-4V alloy's thermo-physical properties are included in Table S1 in the Supplementary Materials.

Model Verification for Single LPBF: Due to the lack of experimental results on the melt pool dimensions of multi-laser synchronized LPBF, the single laser scenario is modeled first, and its melt pool width/height is compared to its experimental counterparts of Ti-6Al-4V [10] (Figure 1c). The rationale behind this verification approach is based on the validity of the superposition principle for the heat equation, i.e., the thermal distribution of multiple heat sources can be written as the sum of the solutions of each heat source in conduction mode. Therefore, if a computational model correctly predicts the thermal distribution of a single heat source (common LPBF with one laser), it should also be able to predict the solution of multiple heat sources correctly based on the superposition principle.

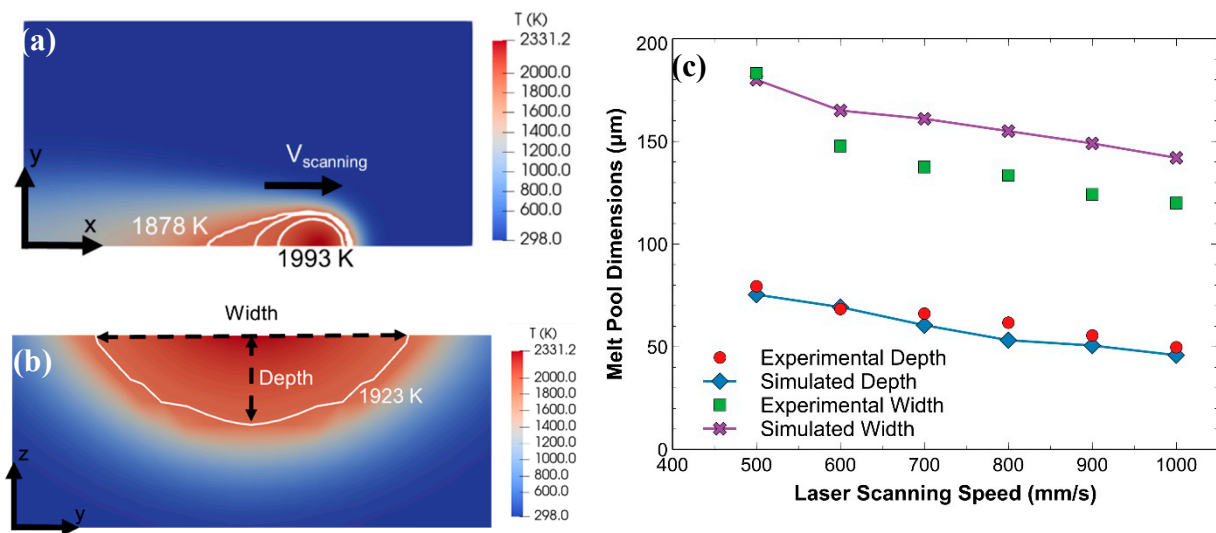


Figure 1. The melt pool dimensions for a single laser beam. (a) Scanning direction, the overall shape of the melt pool, and solidus/liquidus contours. (b) Temperature distribution and definition of depth and width of the melt pool based on melting temperature. (c) Comparison between experimental and numerical simulations for melt pool dimensions (Reproduced with permission from Ref. [16]. Copyright 2022, Springer Nature).

The model agrees well with experiments, considering all uncertainties in modeling and physical parameters, e.g., the temperature dependence of the laser absorption coefficient [42], various mathematical models for laser-beam heat sources [43,44], and the temperature dependence of density and thermal conductivity with powder porosity [45].

Additionally, note that the dependency of the anisotropic thermal conductivity on laser power and speed [44] was ignored to reduce computational modeling expenses. Finally, the overall shape of the melt pool agrees with other numerical studies, i.e., it constitutes a long tail in the form of a comet [46].

Solidification Map: The thermal gradient ($\left|\vec{G}\right|$) and solidification rate (R) are calculated to construct the solidification map and consequently analyze the microstructure:

$$\left|\vec{G}\right| = \sqrt{\left(\frac{\partial T}{\partial x}\right)^2 + \left(\frac{\partial T}{\partial y}\right)^2 + \left(\frac{\partial T}{\partial z}\right)^2}, \quad R = v \frac{G_x}{\left|\vec{G}\right|}. \quad (6)$$

The calculated solidification map for a single laser with various scanning speeds is depicted in Figure 2b. The R values are calculated for the cooling zone of the melt pool, $R > 0$; Figure 2a shows the contour of R and $\left|\vec{G}\right|$ in the cooling zone. The local values are used to provide a better picture of solidification due to the significant variation of R and $\left|\vec{G}\right|$ along the liquidus surface, i.e., the temperature gradient at the bottom is two times larger than the values at the top [20,47]. Generally, the solidification rate increases as it moves toward the tail of the melt pool (Figure 2a). Therefore, the columnar grain appears at the bottom of the melt (point B in Figure 2b), and the mixed microstructure appears at the tail (point S in Figure 2b). Increasing scanning speed leads to an increase in the $R \times G$ value (constant grain size), representing a reduction in grain size [11,48]. In general, an increase in R accompanied by a reduction in G causes larger undercooling and, consequently, promotes the equiaxed microstructure [49]. These trends agree well with the observed experimental microstructures in the literature [48].

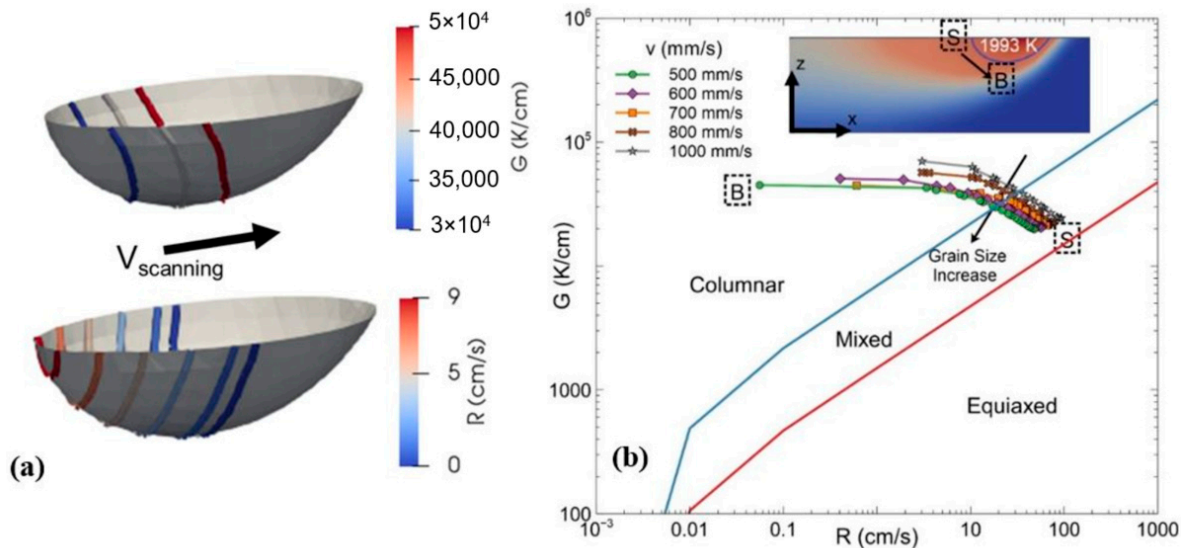


Figure 2. The melt pool shape and solidification map of a laser beam. (a) The G and R contours on the liquidus surface for a 300-W laser beam with a scanning velocity of 500 mm/s. (b) Solidification map for a 300-W laser beam with various scanning velocities.

3. Results and Discussion

The laser power in the hexagonal configuration varies between 50 and 500 W (low to medium level), and the velocity range is 100 to 1000 mm/s. The distance between laser beams, r , in the hexagonal array varies from 50 to 250 μm for all processing parameters (Figure 3a). The solidification map for the 50-W laser is shown in Figure 3b, which predicts the columnar and mixed microstructures for all process parameters, i.e., internal laser distance and speed. However, an increase in scanning speed results in more refined grains

(larger values of $R \times G$). It also increases the fraction of the mixed region by increasing R and, consequently, increasing the undercooling. It is worth noting that the equiaxed grain morphology cannot be attained with this set of processing parameters. Figure 4c shows the melt pool shape in the x - y and y - z planes at $v = 100$ mm/s as a representative case. The melt pool shape becomes irregular at $r \geq 150$ μm and loses its coherency for $r \geq 200$ μm with a scanning speed of 100 mm/s. The melt pool shape incoherency, i.e., multiple melt pools, occurs for $r > 75$ μm for 500 mm/s and 1000 mm/s scanning speeds. Therefore, the thermal gradients and solidification rates for incoherent melt pools were omitted from Figures 3b and 4a,b.

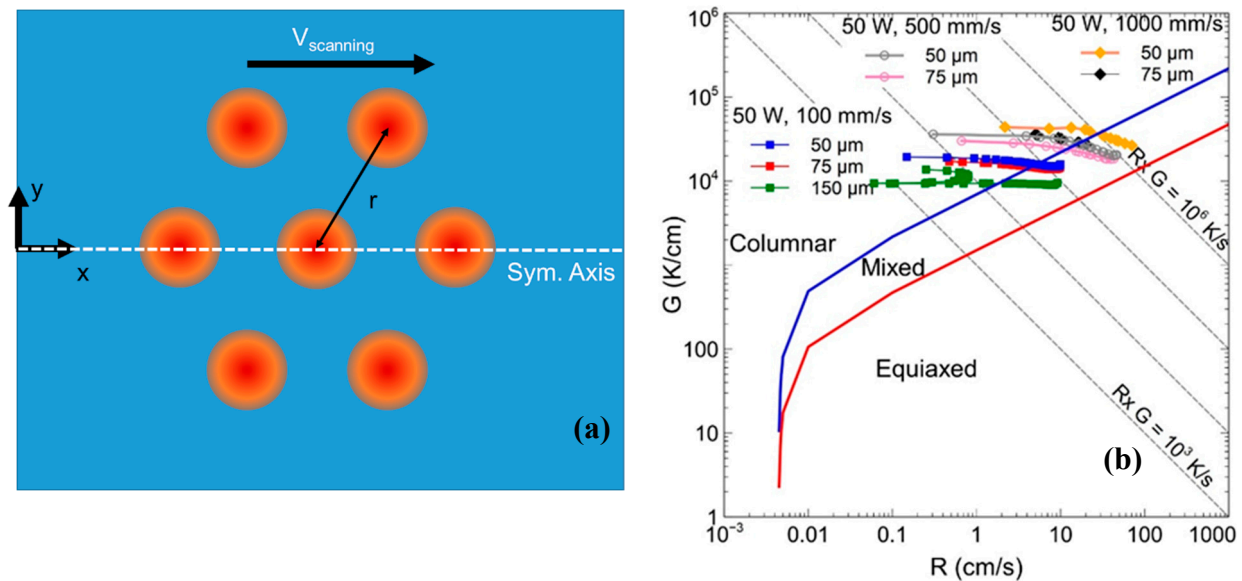


Figure 3. The hexagonal laser configuration and solidification map for the low-power laser beam. (a) The schematic of the hexagonal laser array. The distance between laser spots inside the hexagonal configuration is designated by r , and it is constant during scanning. (b) Solidification map for various scanning velocities and laser configurations, i.e., various distances between laser spots.

The melt pool's variation in width and depth for various internal laser spacings (i.e., the distance between two lasers in the hexagonal configuration as depicted in Figure 3a) and scanning speeds is shown in Figure 4a,b. The width increases by increasing the distance between lasers due to distributing heat flux over a larger area. The melt pool depth shows a reverse trend stemming from overlapping heat sources at small r values and, consequently, an increasing penetration depth; the depth decreases by increasing r . The shape of the melt pools for $P = 50$ W and $v = 100$ mm/s is shown in Figure 4c,d. Figure 4d depicts the shape of the melt pool at the top of the powder bed, i.e., the x - y plane. The laser beam profiles overlap at $r = 50$ and 75 μm , forming a large and coherent melt pool. The melt pool at 200 μm laser spacing consists of two separate melt pools; the front laser forms its melt pool, while the overlap between temperature fields of the other beams forms an irregular melt pool shape (Figure 4d). It is worth noting that the width of the back laser melt pool is larger than the others for $r = 250$ μm since the length of the pool is typically larger than the width and depth. Therefore, there is a considerable thermal crosstalk between the temperature fields of the rest of the beams and the back laser. Any further increase in inter-laser distance, r , results in discrete melt pools due to an insufficient temperature field overlap, i.e., thermal crosstalk. Generally, the discrete/incoherent melt pools can form defects such as pores and are not desirable.

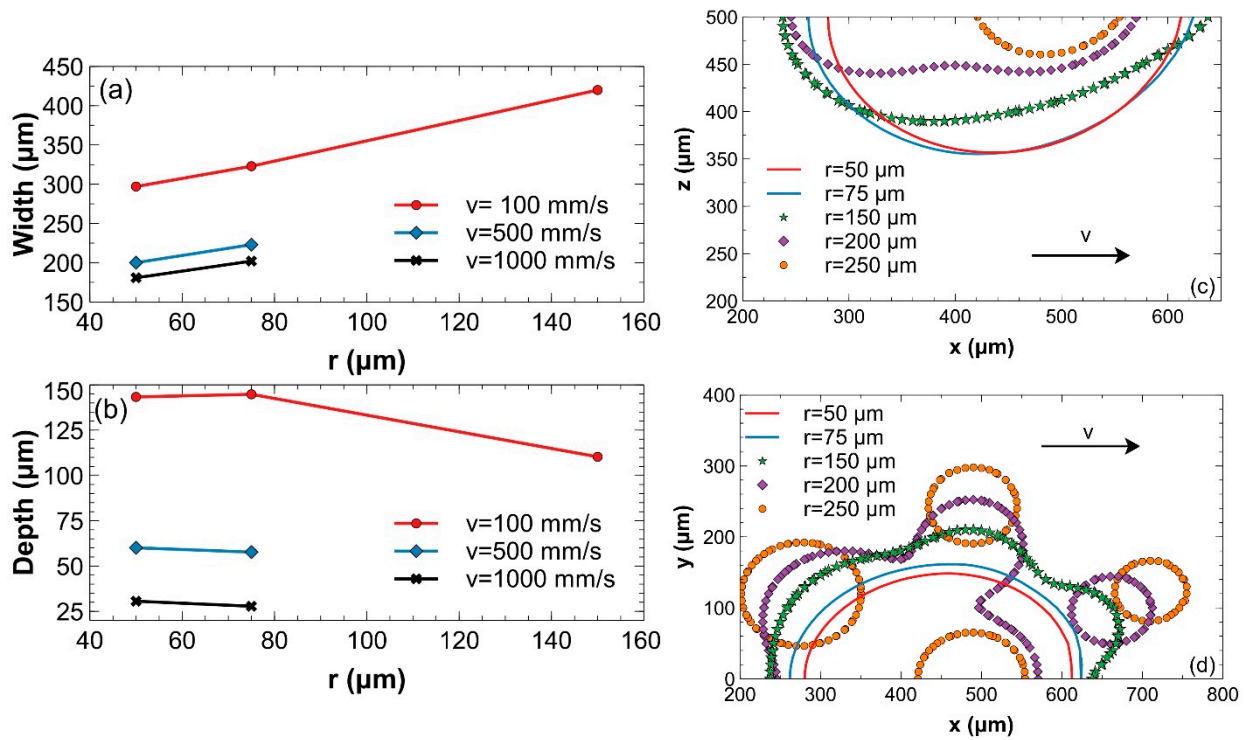


Figure 4. Melt pool dimension and shape for low-power hexagonal laser configuration and low speed ($P = 50$ W, $v = 100$ mm/s). (a) Variation of the width of the melt pool for various velocities and laser spacings. (b) Variation of the depth of the melt pool for various velocities and laser spacings. (c) The melt pool shape on the symmetric plane (x - z). (d) The melt pool shape at the surface (x - y).

Figure 5 depicts the temperature distribution at two scanning speeds ($v = 100$ and 1000 mm/s) for 50 W lasers at the top plane to better illustrate the thermal crosstalk among lasers in the array. The formation of three melt pools at the back of the configuration for $v = 100$ mm/s and $r = 250$ μm shows that one 50 W laser is not enough to melt the powder at this speed; however, the thermal crosstalk among all lasers increases the temperature at the tail and forms three distinct melt pools. These separated melt pools merge by reducing the distance between lasers in the hexagonal array and eventually form a coherent melt pool when laser profiles overlap with each other, i.e., $r \leq 75$ μm .

In contrast to the low laser power scenario ($P = 50$ W), the laser power of 300 W leads to the appearance of an equiaxed microstructure (see Figure 6). The equiaxed microstructure is expected to appear at the surface and extend to the depth of the melt pool, and the columnar microstructure will form at the deepest point of the melt pool. Figure 6b shows the solidification map for a processing parameter set, $P = 300$ W and $v = 100$ mm/s, as a representative case to investigate the effect of laser spacing. Increasing the scanning velocity makes the microstructure finer and reduces the equiaxed grain volume fraction. Figure 7 shows the melt pool dimensions and shape variations with velocity and laser spacing. The results follow the trend of low laser power ($P = 50$ W) when overlapping lasers cause a larger melt pool depth and a smaller width. Figure 7c,d shows the melt pool shape on the x - y and x - z planes for $v = 1000$ mm/s; the melt pool shapes for $v = 100$ – 500 mm/s are depicted in Figure S2 in the Supplementary Materials.

The melt pool is continuous at the low-scanning speed regime; however, the shape shows irregularity when $r > 150$ μm for $v = 500$ – 1000 mm/s (Figure 7d and Figure S2d), i.e., the lasers do not overlap. Note that the irregularity of the melt pool shape can affect the surface quality and printing resolution. Figure 8 shows the temperature distribution for lasers with $P = 300$ W, along with the appearance of the irregular melt pool shapes at high scanning velocity (1000 mm/s) and large laser spacing ($r > 150$ μm). Additionally, the effect

of thermal crosstalk becomes more pronounced in $r > 150 \mu\text{m}$ cases when the continuous back melt pool appears.

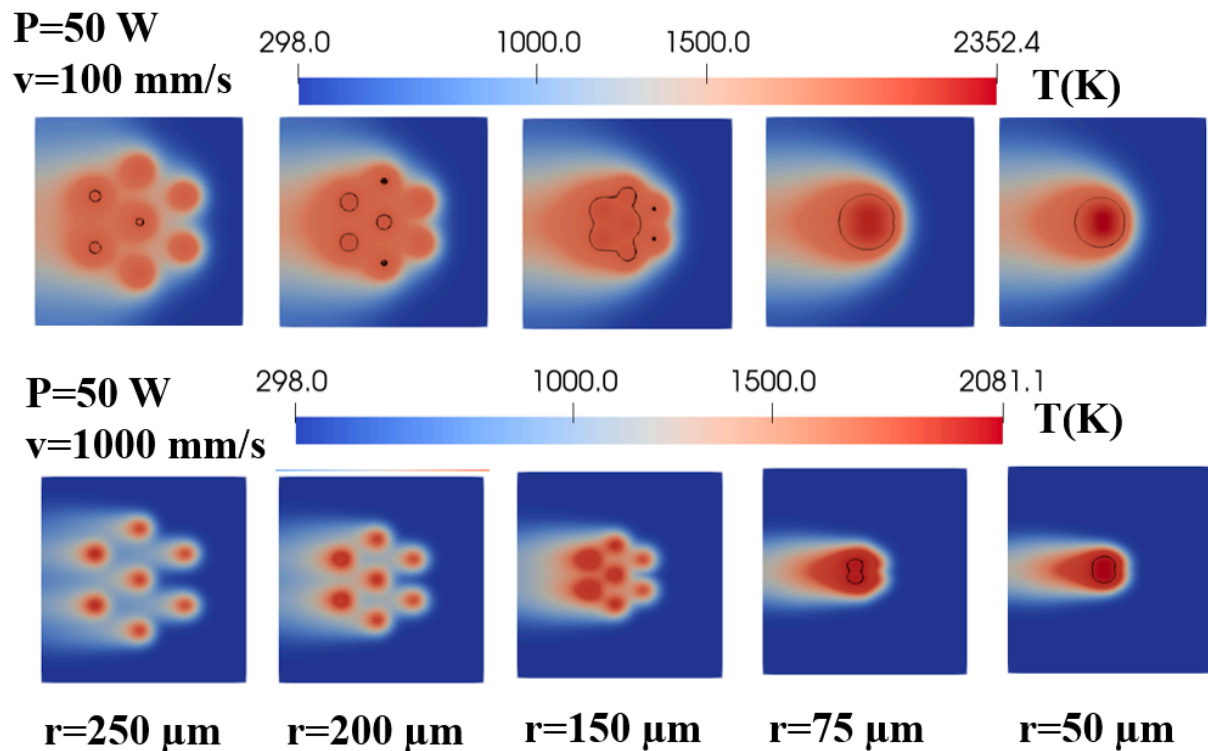


Figure 5. The temperature distribution of the hexagonal laser array with $P = 50 \text{ W}$ at high (1000 mm/s) and low-velocity (100 mm/s) scanning speed regimes for various internal laser spacings on the x - y plane. The black contour represents the melting temperature, i.e., melt pool shape.

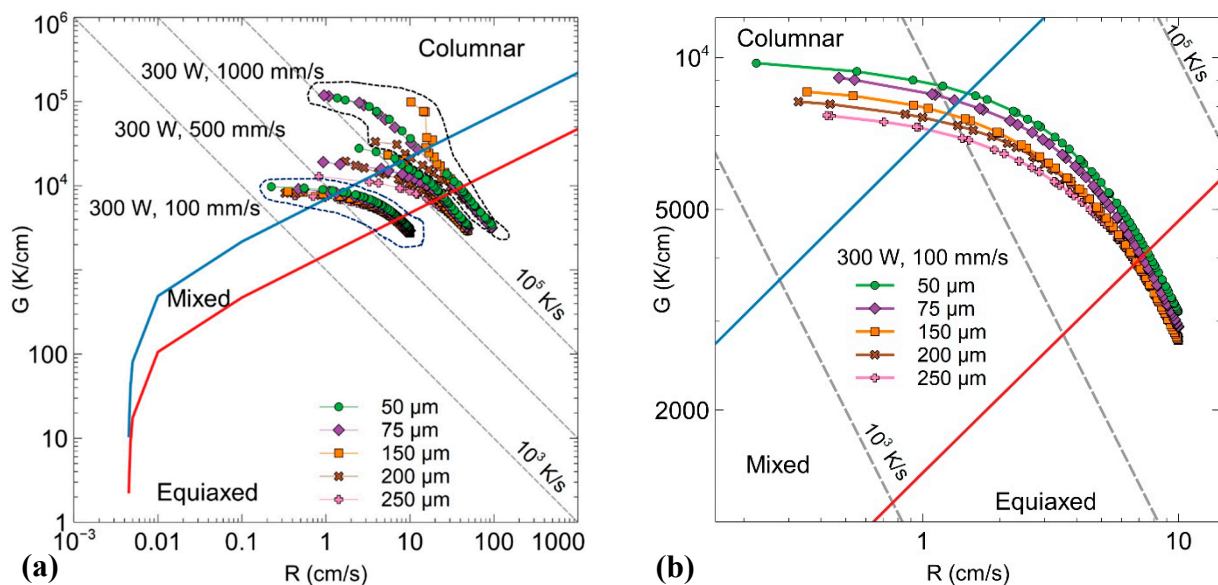


Figure 6. The solidification map for $P = 300 \text{ W}$. (a) Variation of microstructure and grain size with scanning velocity and laser spacing. A dashed curve encompasses each velocity set for clustering. (b) Solidification map for $P = 300$ and $v = 100 \text{ mm/s}$ as a representative set.

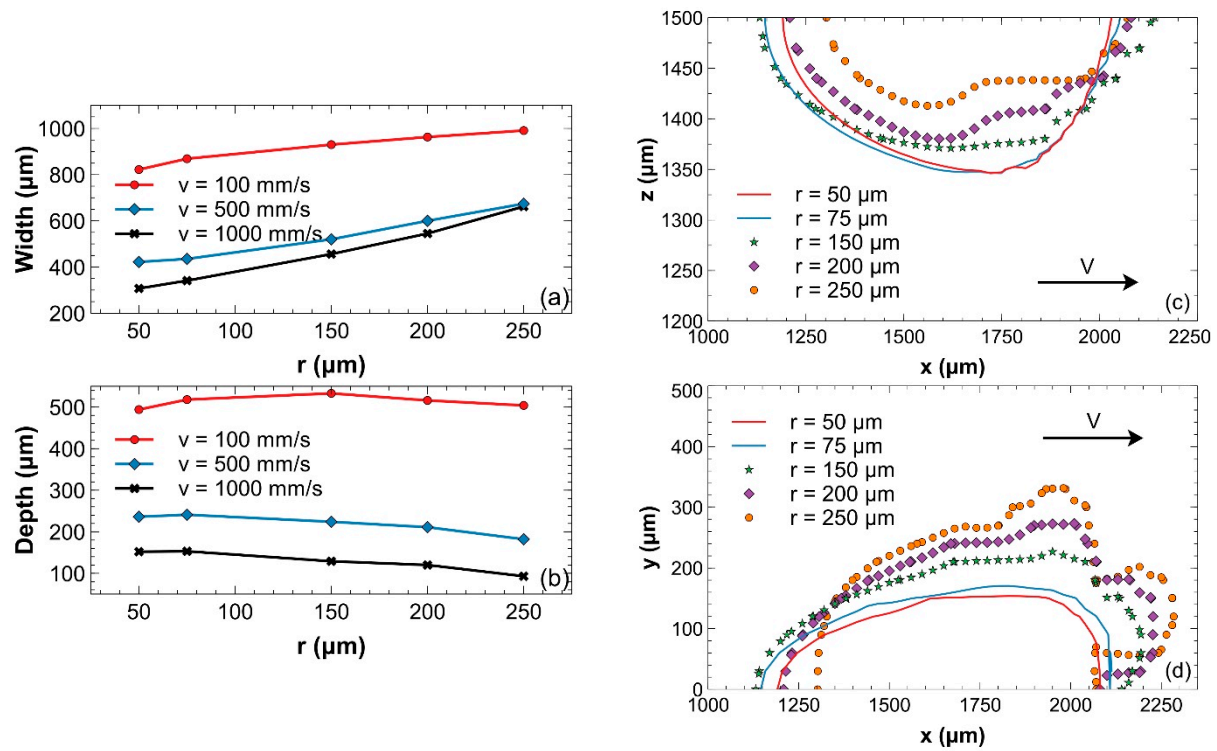


Figure 7. Melt pool dimension and shape for intermediate-power hexagonal configuration ($P = 300$ W, $v = 1000$ mm/s). (a) Variation of the width of the melt pool for various velocities and laser spacings. (b) Variation of the depth of the melt pool for various velocities and laser spacings. (c) The melt pool shape on the symmetric plane (x - z). (d) The melt pool shape at the surface (x - y).

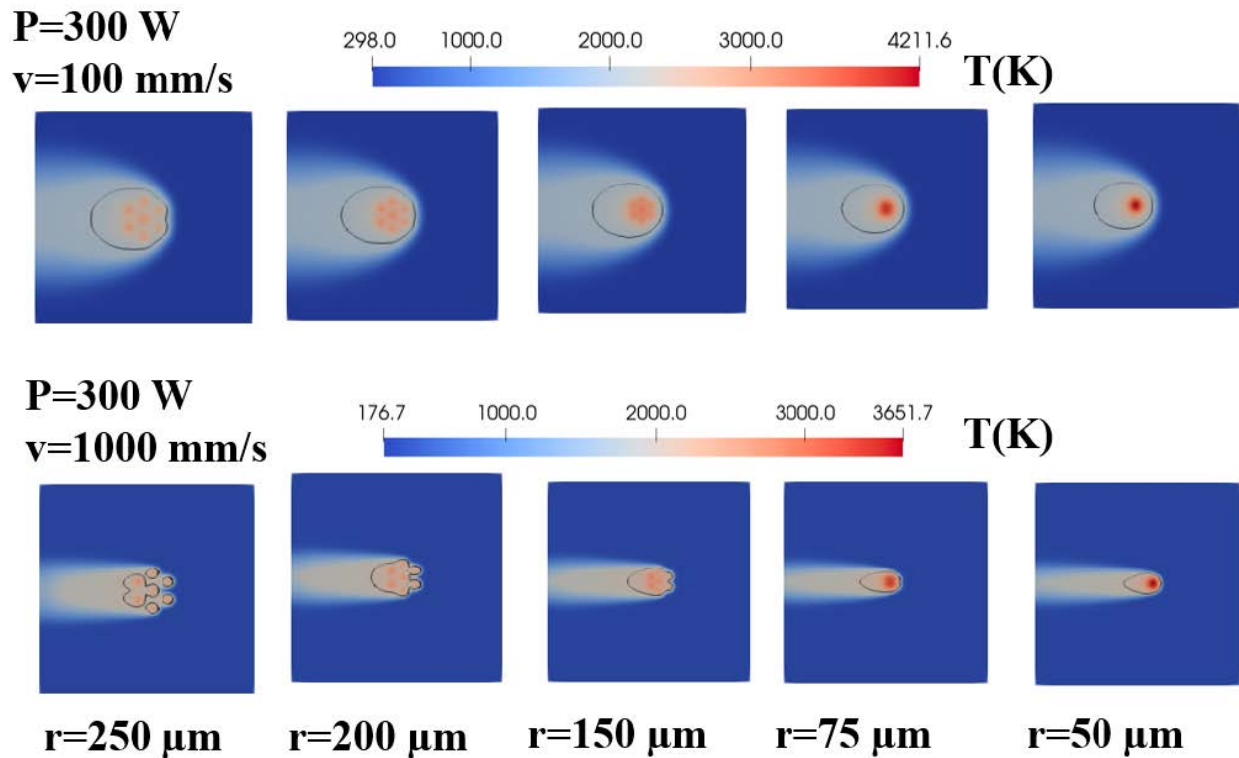


Figure 8. The temperature distribution of the hexagonal laser array with $P = 300$ W at high (1000 mm/s) and low-velocity (100 mm/s) scanning speed regimes for various internal laser spacings on the x - y plane. The black contour represents the melting temperature, i.e., melt pool shape.

Finally, Figure 9 illustrates the solidification map for a high-power laser, i.e., 500 W. All velocities are plotted in the same graph for comparison. Increasing the velocity at the same power level increases the $R \times G$ values and, consequently, it represents a more refined microstructure that agrees with $P = 300$ W. Figure 9b shows the solidification map for a processing parameter set, $P = 500$ W and $v = 100$ mm/s, as a representative case to focus on the effect of laser spacing at the same scanning speed. The arc length of the liquidus line can be used to roughly estimate the volume fraction of each microstructure zone (equiaxed, mixed, and columnar); the Hunt criterion [49] was used to separate these zones on the liquidus line, i.e., columnar: $G^{1.91}/R > 1.92 \times 10^6 \text{ K}^{1.91}/\text{cm}^{2.91} \cdot \text{s}$ and equiaxed: $G^{1.91}/R < 1.04 \times 10^6 \text{ K}^{1.91}/\text{cm}^{2.91} \cdot \text{s}$. Figure 10a,b shows the volume fraction of equiaxed and columnar grains for $P = 500$ W and $v = 100$ –1000 mm/s. In this scenario, a significant portion of the microstructure will be equiaxed (about 55% for $v = 100$ mm/s). The volume fraction of equiaxed grains will decrease as scanning velocity increases, i.e., the average volume fraction of equiaxed grain will decrease from ~55% for $v = 100$ mm/s to ~37% for $v = 1000$ mm/s. Generally, the laser-matter interaction time is higher at lower scanning speeds, which results in a larger melt pool and higher temperature that enhances the thermal crosstalk as the hexagonal laser array moves. Any increase in velocity at the same power level reduces absorbed heat and decreases the thermal crosstalk. Therefore, the thermal spatial distribution becomes closer to the condition of a single laser beam, and a reduction in the volume fraction of equiaxed grains is observed (see Figure 10a). Additionally, the variation of equiaxed volume fraction with the internal laser distancing in the array is slight at $v = 100$ –500 mm/s. However, it drastically changes at $v = 1000$ mm/s, i.e., it drops to 26% at $r = 250 \mu\text{m}$ (see Figure 9a). The decrease in absorbed heat and thermal cross-talk can be linked to the drastic reduction in the equiaxed volume fraction at high scanning velocities. A similar trend is observed for $P = 300$ W in Figure 10c,d. However, the average volume fraction of equiaxed grains is lower, i.e., it decreases from ~42% for $v = 100$ mm/s to ~34% for $v = 1000$ mm/s.

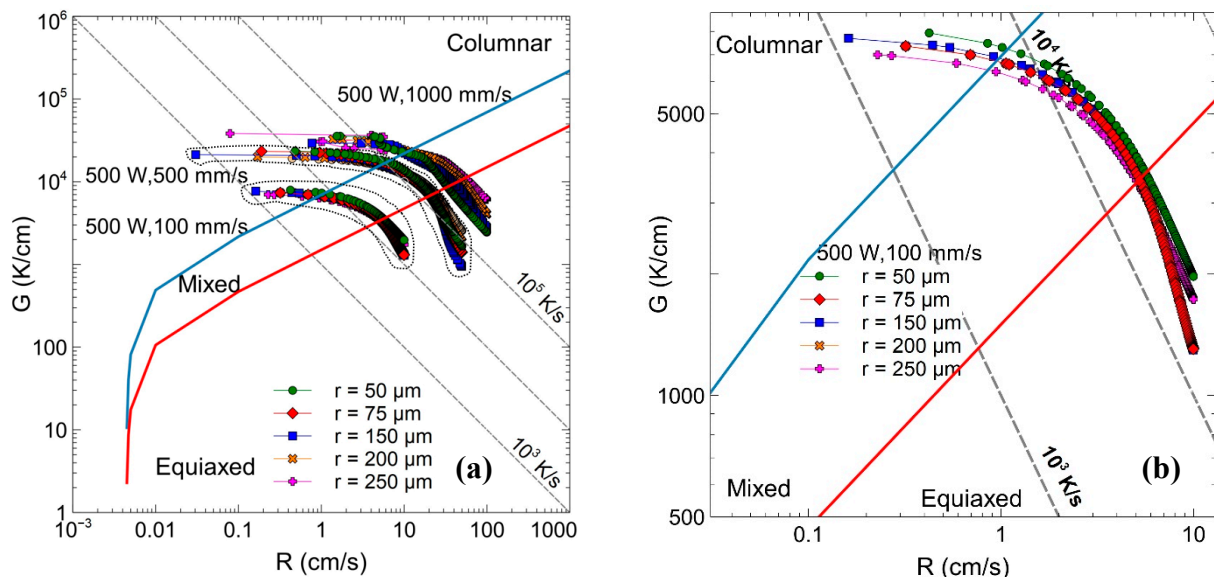


Figure 9. The solidification map for $P = 500$ W. (a) Variation of microstructure and grain size with scanning velocity and laser spacing. A dashed curve encompasses each velocity set for clustering. (b) Solidification map for $P = 500$ W and $v = 100$ mm/s as a representative set.

Figure 11 shows the melt pool geometry and shape variation with laser spacing and speed for $P = 500$ W and $v = 1000$ mm/s. A similar irregularity in shape was observed in the non-overlapping regime, consistent with low and intermediate laser powers. Additionally, the melt pool dimensions follow the same trend, and the only difference is an increase in overall width and depth compared to low (50 W) and intermediate powers (300 W). The

increase in overall melt pool dimensions is due to higher laser powers and increased heat flux. In addition, the melt pool shapes for $P = 500$ W and $v = 100$ – 500 mm/s are shown in Figure S3 of the Supplementary Material.

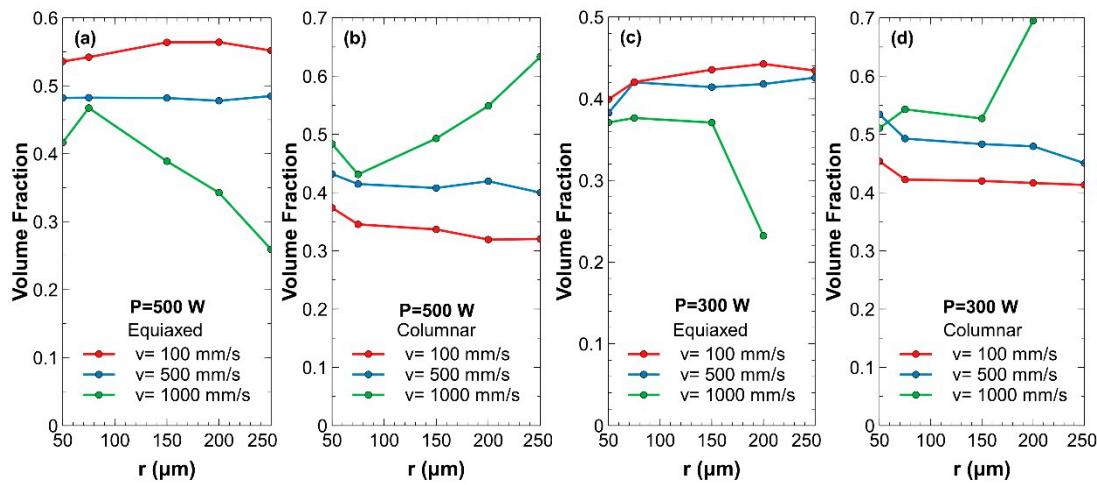


Figure 10. The variation of volume fraction of equiaxed and columnar microstructure for (a,b) $P = 500$ and (c,d) 300 W, with different scanning velocity ($v = 100$ – 1000 mm/s), and laser spacing ($r = 50$ – 250 μm).

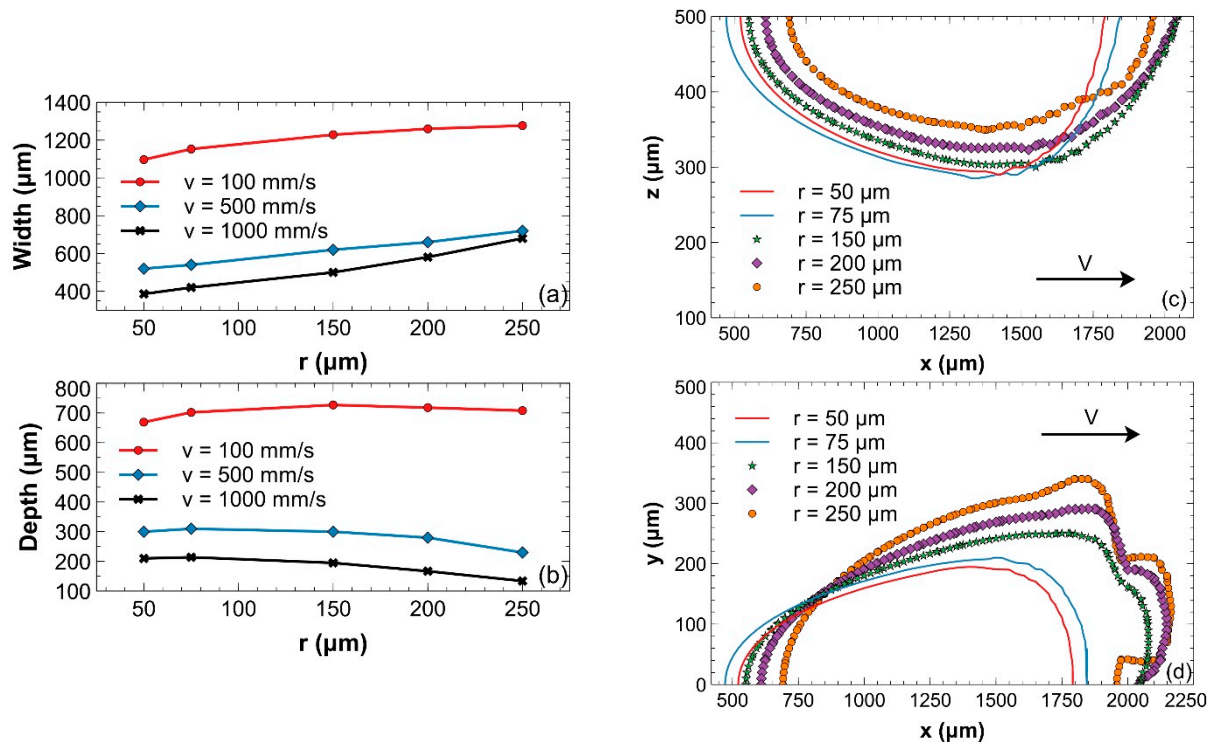


Figure 11. Melt pool dimensions and shape for high-power hexagonal configuration ($P = 500$ W, $v = 1000$ mm/s). (a) Variation of the width of the melt pool for various velocities and laser spacings. (b) Variation of the depth of the melt pool for various velocities and laser spacings. (c). The melt pool shape on the symmetric plane (x - z). (d) The melt pool shape at the surface (x - y).

In general, using a hexagonal laser configuration increases the depth and width of the melt pool compared to that of a single laser beam. Increasing depth can result in better coherence between consecutive layers. Additionally, a power larger than 50 W is required to form the equiaxed region in the current scanning speed range. A slow scanning speed, e.g., $v = 100$ mm/s, can potentially form a small equiaxed region based on the data trend in

Figure 3b. The results recommend using a high laser power with a low scanning speed to maximize the equiaxed grain volume fraction. Finally, one should consider the potential spatter in a high-power/low-velocity/small-laser-interspacing scenario (large degree of laser overlap), which deteriorates part quality. However, one can avoid spatter by increasing the laser interspacing while keeping a high percentage of equiaxed microstructures.

4. Conclusions

The solidification map of a synchronized hexagonal array of lasers was constructed to predict the microstructure of Ti-6Al-4V with various processing parameters, such as scanning speed, laser power, and laser spacing. The laser spacing changed the melt pool shape and thermal gradients. It was shown that about 55% of the printed track has the equiaxed structure in a high power and low scanning velocity regime ($P = 500$ W and $v = 100$ mm/s). However, any increase in velocity reduces the equiaxed portion of the track. Additionally, a degree of laser overlap ($0 \leq r \leq 150$ μm) is recommended to form a continuous melt pool and coherent printed track. The proposed laser array can open a new path in LPBF technology due to its control over microstructure, increased productivity, and printing speed.

Supplementary Materials: The following supporting information can be downloaded at: <https://www.mdpi.com/article/10.3390/jmmp8060242/s1>, Figure S1: Processing diagram of Ti-6Al-4V for LPBF using dimensionless heat input (E^*) and dimensionless laser velocity (v^*) [50]; Figure S2: The melt pool shape on the x - z and y - z planes for $P = 300$ W and $v = 100$ – 500 mm/s. (a,b) $v = 100$ mm/s, (c,d) $v = 500$ mm/s; Figure S3: The melt pool shape on the x - z and y - z planes for $P = 500$ W and $v = 100$ – 500 mm/s. (a,b) $v = 100$ mm/s, (c,d) $v = 500$ mm/s; Table S1: Density, thermal conductivity, and specific heat of the Ti-6Al-4V [40].

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