

Article

Process Optimization of Metal FDM AA6061: Thermal Debinding and Sintering Toward Enhanced Component Quality

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Abstract

Fused Deposition Modeling (FDM) of AA6061 aluminum alloys faces several critical challenges, including limited parameter control, inefficient thermal debinding, and sintering difficulties associated with oxide formation. These factors hinder the broader adoption of FDM compared to other additive manufacturing (AM) techniques, often resulting in high porosity and suboptimal mechanical performance. In this study, a commercially available AA6061–PLA filament feedstock was processed via fused deposition modeling (FDM), followed by thermal debinding and sintering. The printing parameters, debinding cycle, and sintering profile were systematically optimized to produce dense AA6061 components. The effects of these processing conditions on relative density, microstructure, and mechanical properties were investigated. Optimized printing parameters, combined with a staged thermal treatment, enabled the fabrication of high-density parts with reduced defects. Furthermore, the introduction of a pre-sintering pressing step significantly enhanced the final properties, achieving a relative density of 88.34% (11.66% porosity), a hardness of 83.67 ± 0.67 HV ($n = 3$), and a compressive strength of 115 ± 3 MPa ($n = 3$). Scanning electron microscopy (SEM) revealed a relatively uniform microstructure with a refined pore distribution. These results demonstrate the potential of FDM as a viable and promising approach for producing aluminum components with improved structural and mechanical characteristics, and provide initial forming–microstructure–property relationships for sintered AA6061 fabricated via metal FDM.

Keywords: aluminum alloys; FDM; AA6061; printing; thermal debinding; sintering



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1. Introduction

According to the ISO/ASTM 52900-15 [1] standard, additive manufacturing (AM)—commonly referred to as 3D printing—is a process that fabricates parts by adding material layer-by-layer based on a digital 3D model [1,2]. Current commercial AM technologies are classified into seven primary categories: material extrusion (MEX), binder jetting (BJ), material jetting (MJ), sheet lamination (SL), vat photopolymerization (VP), powder bed fusion (PBF), and directed energy deposition (DED) [3,4]. Among these, material extrusion (MEX) has emerged as the most prevalent and accessible category due to its cost-effectiveness, high availability, and capacity for producing complex geometries [5,6]. MEX processes significantly reduce raw material waste and are considered highly efficient for producing components with uniform microstructures [7].

However, fabricating metallic components with satisfactory mechanical and microstructural attributes via MEX remains a formidable challenge for researchers. Despite these difficulties, Fused Deposition Modeling (FDM), a subset of MEX, is a widely adopted method because of its operational simplicity and lower capital investment compared to other metal AM technologies [8]. Consequently, this technology is predominantly utilized in the production of patterns, molds, prostheses, prototypes, and structural components for defense applications [5,9].

Current research has primarily focused on a narrow range of metal AM technologies. For instance, Roshchupkin et al. [10] developed a specialized extruder for metal–polymer filaments, enabling direct winding at the extruder outlet even with high powder loadings. Building on this, Suwanpreecha [11] concluded that such processes are indeed capable of producing functional, end-use metallic components. Nevertheless, significant technical gaps persist across all stages of the MEX process; factors such as material homogeneity, mechanical integrity, geometric precision, and production throughput still require substantial optimization. Achieving a filament with ideal extrusion properties and refined processing parameters remains a priority for further technological advancement.

Specific studies have addressed these challenges in various alloys. Malone [12] investigated the influence of 3D printing parameters on the surface roughness and porosity of green samples using in-house developed AA6061-binder filaments, highlighting similarities to FDM of other aluminum alloys. However, the resulting sintered specimens exhibited relatively low densities (47% to 70%) and poor hardness. Tang et al. [13] proposed a hybrid sequential AM approach, using ultrasonic welding to join polymer components (produced via FDM) with metallic components (fabricated through PBF), utilizing partially melted powder to create micro-mechanical interlocking. Similarly, Falck et al. [14] introduced the so-called “AddJoining” method for fabricating layered metal–polymer hybrid structures by depositing melted polymer onto metallic substrates. In the context of high-strength alloys, Ding et al. [15] established specific FDM protocols for Al7075, detailing the transition from green to brown bodies and final sintering. Their study contrasted the microstructures of Al7075 with FDM-printed 17-4 PH steel, emphasizing the use of debinding and sintering to mitigate voids. Furthermore, Yakout et al. [16] explored the impact of critical printing parameters on the microstructure and mechanical properties of titanium, aluminum, and nickel-based alloys.

Despite the growing body of literature, there is still a noticeable scarcity of viable feedstocks for aluminum alloys in AM. This is a critical oversight, as aluminum alloys offer an exceptional strength-to-density ratio, making them indispensable lightweight materials for the automotive and aerospace sectors.

Over the last decade, Al-based AM has gained traction for its ability to produce intricate geometries with reduced lead times. However, AM of aluminum is often plagued by the development of coarse columnar grains, especially in techniques involving materials melting (such as PBF). Most research has focused on eutectic or near-eutectic Al–Si alloys due to their narrow solidification ranges, which minimize hot-cracking susceptibility [17,18]. While these Al–Si alloys exhibit commendable ultimate tensile strength (UTS) exceeding 300 MPa, they often suffer from limited ductility (elongation < 8%) and significant property anisotropy caused by the presence of columnar grains [19,20]. Currently, only a small subset of alloys—such as AlSi10Mg, Ti6Al4V, CoCr, and Inconel 718—can be reliably printed [21]. The vast majority of the approximately 5500 existing alloys remain unsuitable for AM due to unfavorable melting and solidification dynamics, which lead to detrimental microstructures and periodic cracking [21,22].

Although prior studies have explored aluminum–binder composites for material extrusion, including Malone’s work on general MF3 processing of AA6061 filaments and

Aguilar-García's focus on sustainable aluminum-based feedstock formulation, these efforts primarily address feedstock design and early-stage printing. They consistently report persistent sintering challenges and relatively low relative densities in the final aluminum parts, highlighting an unresolved bottleneck in achieving fully dense, functional components via standard thermal processing routes. In contrast, the present work targets this specific gap by introducing three key advancements: (i) a mechanical 1250 MPa pre-sintering pressing step designed to promote a denser microstructure and enhance interparticle bonding during sintering; (ii) a refined thermal strategy combining an ultra-slow heating rate (0.1 °C/min) with a Ti–Mg oxygen-trap getter under vacuum; and (iii) a comprehensive microstructural evaluation of successfully sintered AA6061 parts, supported by quantitative mechanical validation through compression testing (compressive strength of 114.79 MPa). Together, these contributions move beyond feedstock formulation to address the critical post-processing limitations that have so far restricted the performance of aluminum parts produced by metal FDM.

This study addresses a pivotal challenge in the field: the lack of affordable and sustainable feedstock materials, specifically for widely used aluminum alloys, and the associated difficulties in obtaining dense, mechanically robust parts via metal FDM. While AA6061 filaments show promise for FDM, their thermo-physical behavior during debinding and sintering is not yet fully understood, and standard thermal routes have so far yielded parts with limited density and mechanical performance. To bridge this gap, this paper investigates the feasibility of processing AA6061 filaments via FDM and optimizing their post-processing stages through pre-sintering pressing and a controlled sintering cycle. The findings provide critical baseline data for processing and performance, evaluating the potential of AA6061 as a practical material for low-cost metal additive manufacturing and demonstrating a pathway toward denser, functional aluminum components.

2. Materials and Methods

2.1. Feedstock Material and Filament Characterization

The AA6061 filament utilized in this study was supplied by Filament2D Print (Nigrán, Spain), which sources this filament from The Virtual Foundry (TVF) (Stoughton, WI, USA), a company specializing in metal-infused filaments for FDM applications. As illustrated in Figure 1a,b, the feedstock is composed of 50 vol.% metallic powder and 50 vol.% polylactic acid (PLA) binder. To verify the true metallic content, the chemical composition of the AA6061 powder was quantified after completely burning out the polymer binder via thermal debinding.

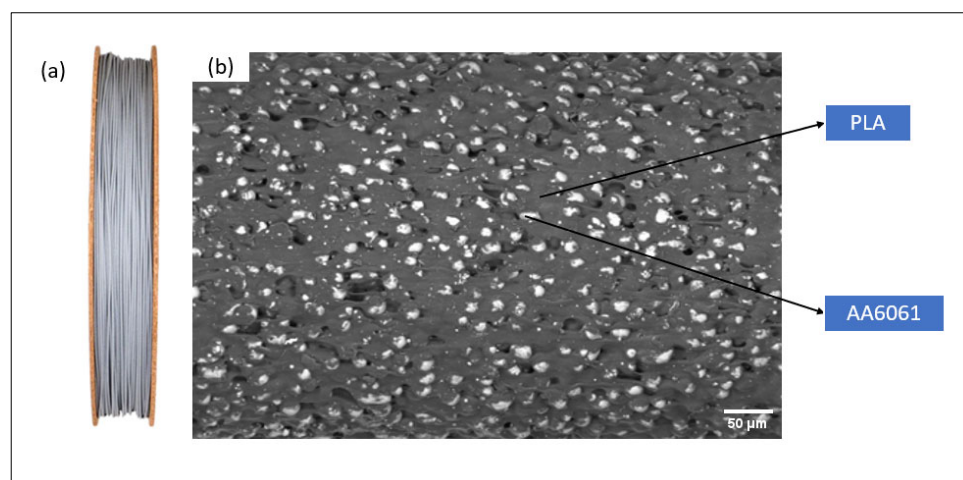


Figure 1. (a) AA6061 filament and (b) microstructural observation under SEM.

As shown in Table 1, the Mg and Si contents of the metallic fraction of the commercial AA6061-designated feedstock were 1.64 wt.% and 1.4 wt.%, respectively. These values exceed the conventional composition ranges for wrought AA6061. This variation reflects the composition of the commercial feedstock batch analyzed in the present study. The increased Mg and Si levels may influence sintering by modifying the oxide-controlled particle-bonding behavior and by facilitating local liquid-phase-assisted transport at the sintering temperature. Nevertheless, because a liquid phase was not directly characterized in this study, this mechanism is proposed only as a possible contribution to the observed densification).

Table 1. Chemical composition of AA6061 (in wt.%).

Al	Mg	Si	Mn	Cu	Fe
Bal.	1.64	1.4	0.007	0.30	0.18

2.2. Additive Manufacturing Process and Parameter Optimization

Components were fabricated using a Creality Ender 3 Pro 3D printer (Creality Ender 3 Pro 3D printer, Shenzhen, China) (Figure 2) at the Fundacio Centre CIM facilities. The system is equipped with four NEMA 17 stepper motors and an Einsy Rambo control board, which regulates the movement of the print head and build platform across the X, Y, and Z axes. To enhance the handling and reliability of the metal-filled filament, a filament warmer device was integrated into the feeding system. This device employs a temperature-controlled heating tube to reduce the intrinsic curvature of the filament, thereby increasing its flexibility and mitigating the risk of fracture during the extrusion process. The FilaWarmer is compatible with both direct-drive and Bowden-extrusion systems using 1.75 mm filaments.

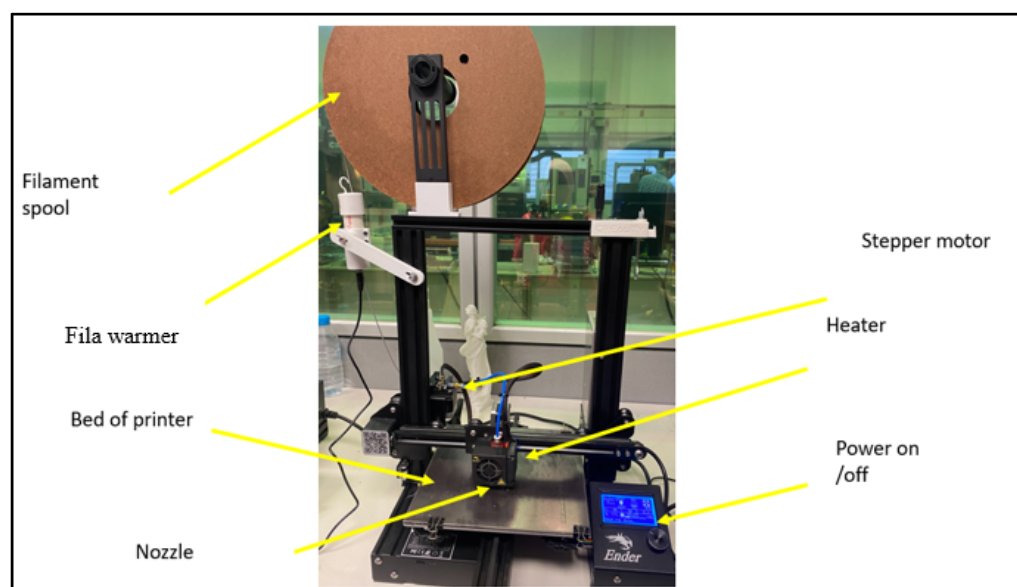


Figure 2. FDM 3D printer used with filament warmer.

The quality of the FDM-printed parts is highly sensitive to various processing parameters, including layer height, line width, infill density and pattern, infill overlap, thermal settings (nozzle and bed), flow rate, and printing speed. A systematic evaluation was conducted across five distinct printing configurations, as detailed in Table 2. Initial trials using Setting 1 encountered significant technical challenges, primarily insufficient material flow

and nozzle clogging. These issues were attributed to the small nozzle diameter (0.4 mm), which resulted in filament breakage within the extrusion assembly (Figure 3a,b).

Table 2. AA6061 trial printing parameters.

Printing Setting Parameter	#1	#2	#3	#4	#5
layer height (mm)	0.2	0.2	0.2	0.2	0.2
Initial layer height (mm)	0.3	0.3	0.3	0.3	0.3
Linewidth (mm)	2.4	2.2	1	2	0.5
Infill density (%)	100	100	100	100	100
Infill pattern	line	line	line	line	line
Infill overlap percentage (%)	30	50	90	100	100
Printing temperature (°C)	220	215	210	205	205
Build plate temperature (°C)	50	50	50	50	50
Flow (%)	130	120	120	125	125
Printing speed (mm/s)	40	45	30	20	20
Nozzle diameter (mm)	0.4	0.8	0.8	0.8	0.8

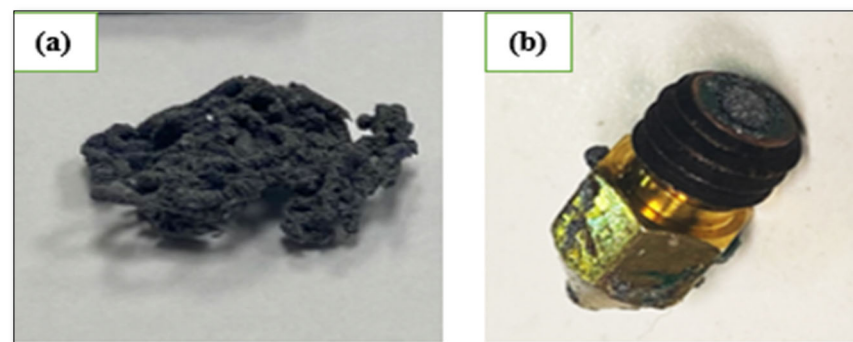


Figure 3. (a) Printed part according to experimental setting #1, (b) clogged and broken filament inside of a 0.4 mm nozzle.

Subsequent iterations (Settings 2 through 4) also revealed defects related to suboptimal nozzle temperatures, inadequate infill overlap, and excessive printing speeds. High speeds, in particular, led to filament “grinding” within the Bowden drive system, where the gear teeth would strip the filament, leading to extrusion failure (Figure 4a–c). Optimization efforts culminated in Setting 5, which utilized a nozzle temperature of 205 °C, 100% infill overlap, a 0.8 mm nozzle diameter, and a reduced line width of 0.5 mm. Reducing the print speed while narrowing the line width led to substantial improvements in dimensional accuracy, interlayer adhesion, and overall structural integrity. All specimens were printed with nominal dimensions of 10 mm (diameter) by 6 mm (height), as shown in (Figure 5).

2.3. Thermal Debinding and Sintering

The removal of the polymer binder was achieved through a multi-stage thermal debinding process. To establish the degradation profile of the AA6061-PLA feedstock, thermogravimetric analysis (TGA) was performed. Scans were conducted in a nitrogen (N₂) atmosphere from room temperature to 600 °C at a heating rate of 3.5 °C/min. The results indicated that the organic binder begins to degrade at 350 °C, with complete decomposition occurring by 550 °C.

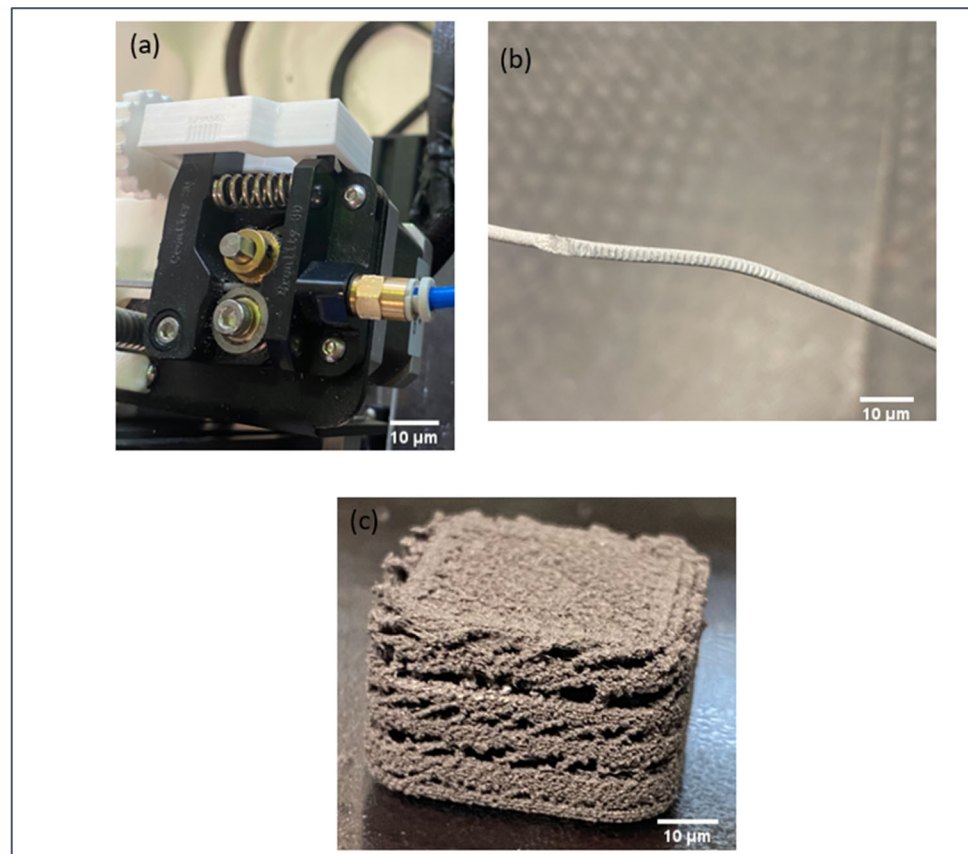


Figure 4. Printing experiments according to printing settings #1 to #4 (a) extruder gear of the filament, (b) filament bite by extruder gear, (c) printed part after stopping extruding the filament (under extrusion).

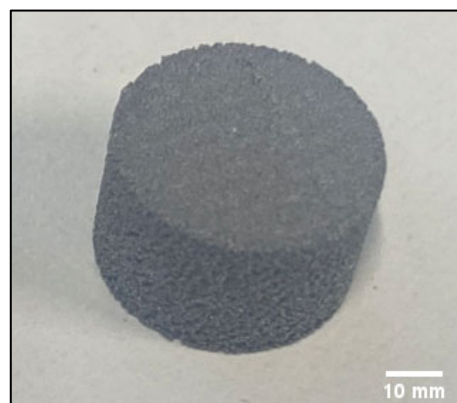


Figure 5. Shape of printed part after optimal setting condition #5.

Based on these thermal profiles, green parts were subjected to debinding in a nitrogen-purged furnace (Hobersal, Caldes de Montbui, Spain) with a heating rate of 3.5 °C/min. Staged isothermal holds were evaluated at 350, 450, 500, and 550 °C for durations of 1, 2, and 3 h. Comprehensive microstructural analysis identified the optimal debinding cycle as 550 °C for 3 h. Following debinding, to eliminate residual voids and enhance particle contact, a pre-sintering compaction step was introduced to promote higher final densification. As illustrated in the experimental setup, compaction was performed immediately following thermal debinding using a dedicated custom steel die designed for cylindrical specimens (10 mm) in diameter and (6 mm in thickness). A uniaxial load of 1250 MPa was applied vertically at room temperature without external heating, maintaining a dwell time of 10 s.

From a structural integrity standpoint, this uniaxial configuration successfully improved particle packing along the loading axis. Although uniaxial pressing inherently creates differential stress fields, the rigid radial confinement provided by the matching 10 mm steel mold effectively constrained unconstrained lateral expansion, preventing transverse micro-cracking, interlayer separation, or macro-distortion. Furthermore, cross-sectional analyses oriented both parallel and perpendicular to the pressing direction confirm homogeneous densification without introducing deleterious anisotropic structural defects.

Sintering experiments were subsequently conducted on both pressed and unpressed specimens under identical conditions. The furnace was evacuated using a vacuum pump and subsequently purged with nitrogen. The specimens were then sintered in a nitrogen atmosphere at a controlled heating rate of 1 °C/min. An oxygen trap consisting of a Ti–Mg powder mixture (50:50 wt.%) was used to minimize oxidation.

Sintering temperatures of 625, 635, 645, and 655 °C were investigated for holding times of 1, 2, and 3 h. Additionally, differential scanning calorimetry (DSC) was performed using a TA Instruments DSC2920 (TA Instruments, New Castle, DE, USA) in a nitrogen atmosphere to characterize the thermal transitions of the filament.

2.4. Microstructural and Mechanical Characterization

Post-sintered samples were sectioned, mounted in Bakelite resin, and progressively ground using SiC abrasive paper (240 to 2500 grit). Final polishing was achieved using diamond suspensions (9, 6, 3, and 1 µm). Microstructural features were examined using a ZEISS GeminiSEM scanning electron microscope (ZEISS, Oberkochen, Germany) (SEM). The density of the processed components was determined via the Archimedes principle using an OHAUS precision balance (OHAUS, Parsippany, NJ, USA). Mechanical performance was evaluated through Vickers hardness (HV) testing with a 100 gf load and a 15 s dwell time, and three duplicate samples. To ensure statistical reliability, four measurements were recorded at different locations across three duplicate samples. Following ASTM E9 [23] (adapted to the lab's specimen size and equipment), compression tests were performed on cylindrical AA6061 specimens ($\varnothing 10 \text{ mm} \times 6 \pm 0.5 \text{ mm}$) using an Instron 4507 (Instron GmbH, Darmstadt, Germany) with a 100 kN load cell and a strain rate of 0.001 s^{-1} . Specimen ends and fixture blocks were cleaned with acetone (or equivalent), and a TFE fluorocarbon lubricant was applied to minimize clamp–sample friction. Stress–strain data were recorded and processed with the instrument software.

3. Results and Discussions

3.1. Feedstock Characterization and Morphology

The morphological integrity and particle-size distribution of the AA6061 feedstock are vital for ensuring consistent extrusion and proper densification during sintering. To analyze the powder characteristics following thermal debinding, particles were isolated directly from the debinded matrix. Representative scanning electron microscopy (SEM) images were collected and processed using ImageJ software (v1.54k) to measure individual particle diameters, utilizing a robust statistical approach across multiple randomized micrographs and hundreds of particles to guarantee reliability. The resulting particle-size distribution and cumulative frequency curves (Figure 6a,b) were analyzed using Origin software (9.9, 2022). From these curves, the key percentile diameters D10, D50, and D90 were determined. Specifically, D10 indicates the size where 10% of the distribution is smaller (39), D50 represents the median particle size where 50% of the particles are smaller (49), and D90 marks the point where 90% of the particles are finer (68). This clear statistical breakdown confirms that the powder possesses a well-controlled particle size suitable for subsequent compaction and sintering as shown in (Figure 6). These dimensions are

considered optimal for the FDM process, as they allow for high metal powder loading while maintaining sufficient flowability through a 0.8 mm nozzle. In metal-infused composite filaments, a homogeneous distribution of particles within the polymer binder matrix is essential to prevent localized stress concentrations and ensure isotropic shrinkage during post-processing. Furthermore, SEM elemental mapping (Figure 7) confirmed the uniform distribution of primary alloying elements—Magnesium (Mg), Silicon (Si), Copper (Cu), and minor trace Silver (Ag)—dispersed homogeneously within the aluminum matrix. According to the manufacturer’s technical specifications (MSDS for Aluminum Filament), silver is incorporated as a minor trace element within a range of 0.0 to 0.7 wt.%. Meanwhile, the detection of Carbon (C) and Oxygen (O) stems from residual PLA binder remnants and the native oxide layer inherent to aluminum powder particles. This high chemical homogeneity provides an optimal metallurgical foundation for the subsequent formation of transient eutectic phases during sintering, which relies heavily on uniform localized diffusion and melting.

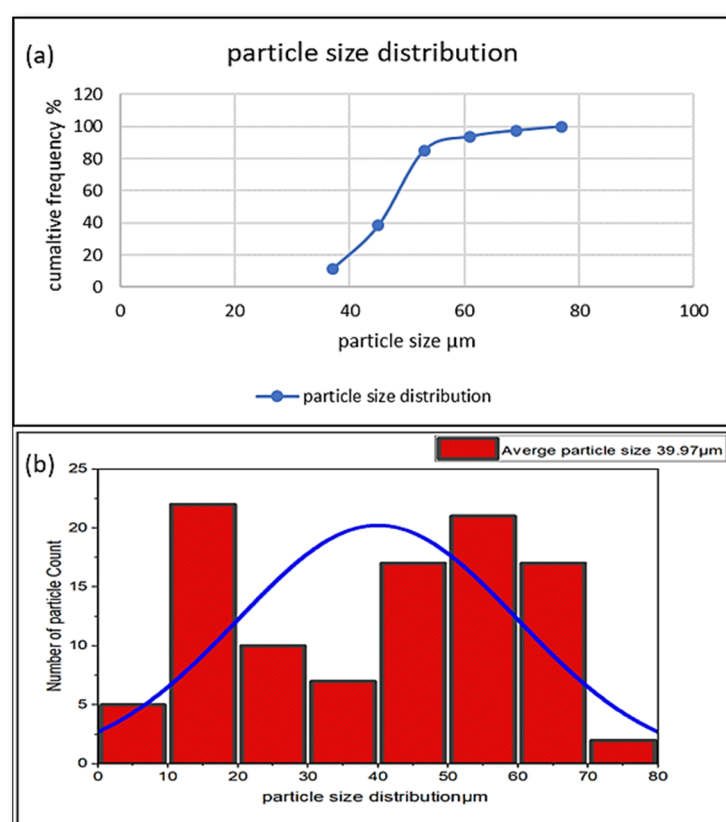


Figure 6. (a) Accumulated particle size distribution, (b) frequency distribution of particle size.

3.2. Thermal Properties and Binder Degradation Kinetics

The thermal behavior of the commercial AA6061–PLA feedstock was characterized by differential scanning calorimetry (DSC) and thermogravimetric analysis (TGA) to establish suitable temperature windows for printing and thermal debinding. As shown in (Figure 8) and summarized in Table 3, the polymeric binder exhibited a glass-transition temperature (T_g) of 80 °C, a crystallization temperature (T_c) of 110 °C, and a melting temperature (T_m) of 190 °C. Based on these thermal transitions, a nozzle temperature of 205 °C was selected to provide sufficient melt flow during extrusion while remaining below the onset of significant thermal degradation.

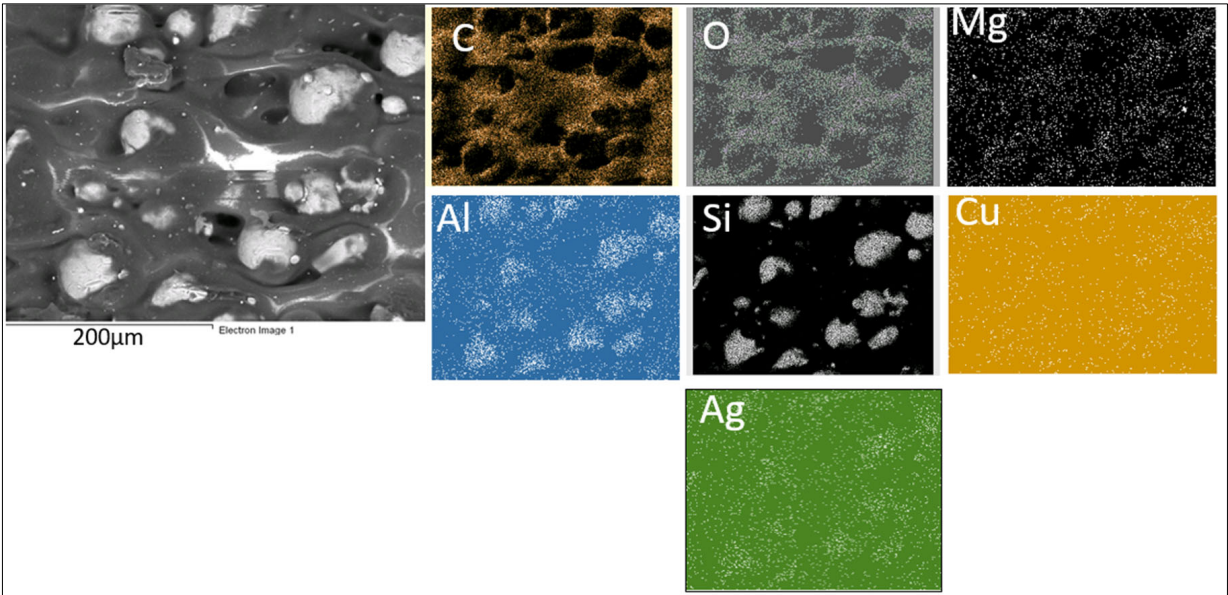


Figure 7. Backscattered electron-SEM image and EDS mapping of the chemical elements present in the current AA6061.

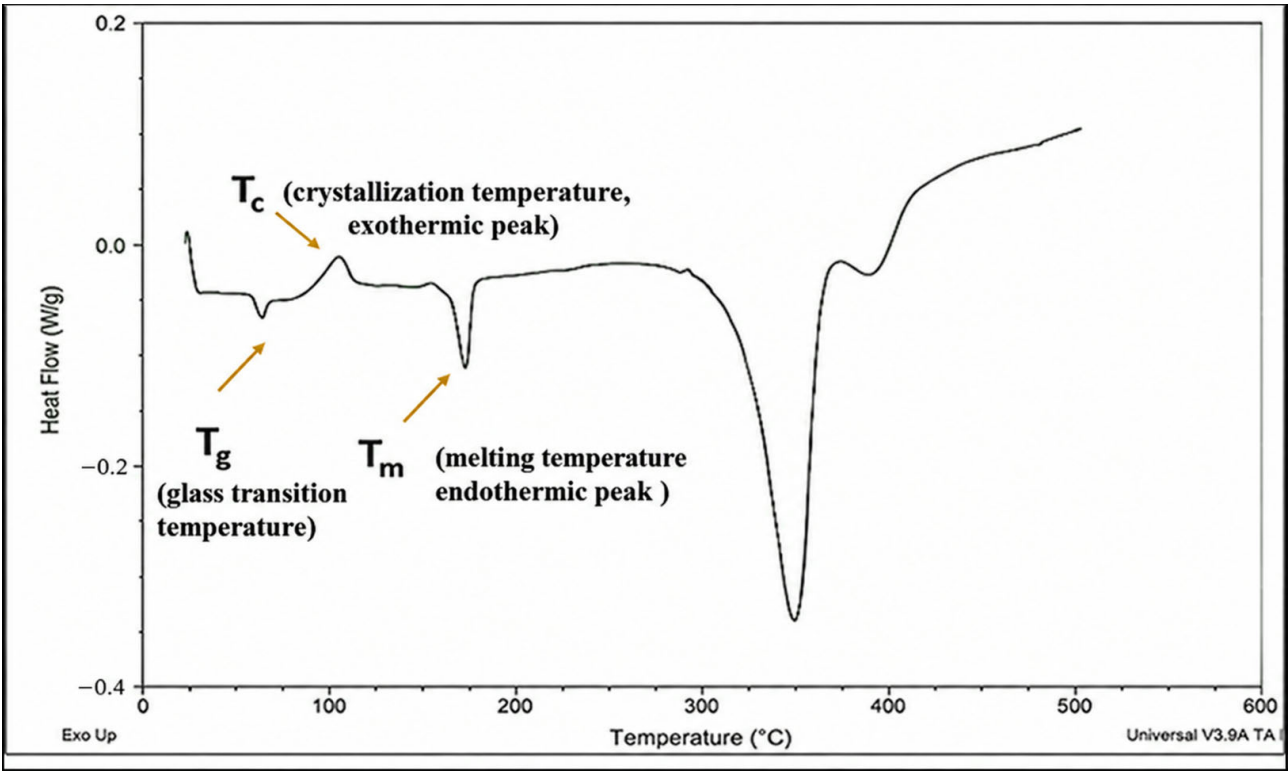


Figure 8. DSC of the filament AA6061.

Table 3. Temperature used in the TGA.

Temperature	T_{start}	T_{max}	T_{end}
	350 °C	450 °C	550 °C

In addition to the main thermal transitions, the DSC curve shown in Figure 8 exhibits a broad endothermic event between approximately 300 and 400 °C, followed by an exothermic event between approximately 400 and 500 °C. The endothermic feature is consistent

with heat absorption during the principal thermal degradation and volatilization stages of the polymeric binder system. The subsequent exothermic feature may be associated with secondary reactions involving degradation products and/or residual organic species during continued heating, such as structural rearrangement, cross-linking, oxidation, or carbonaceous-residue-related reactions, depending on the binder composition and test atmosphere [24]. However, because the evolved gases and residual products were not characterized in the present work, the specific mechanisms responsible for these DSC events cannot be assigned unambiguously.

Therefore, the thermal-debinding window was defined primarily from the TGA and derivative thermogravimetry (DTG) results as shown in (Figure 9), the TGA/DTG curves indicate a multistep binder-degradation process. The main mass-loss event begins at approximately 350 °C and reaches its maximum rate at 450 °C, corresponding to the DTG peak temperature, T_{max} . The mass loss then continues and is essentially completed near 550 °C, where the TGA curve approaches a stable plateau. The total mass loss was approximately 31%. Consequently, the thermal-debinding cycle was designed to apply controlled heating through the 350–550 °C interval to enable gradual removal of volatile products. Rapid heating within this range may generate internal gas pressure, causing cracking, blistering, or distortion, whereas incomplete debinding can leave residues that hinder particle contact and subsequent sintering [24].

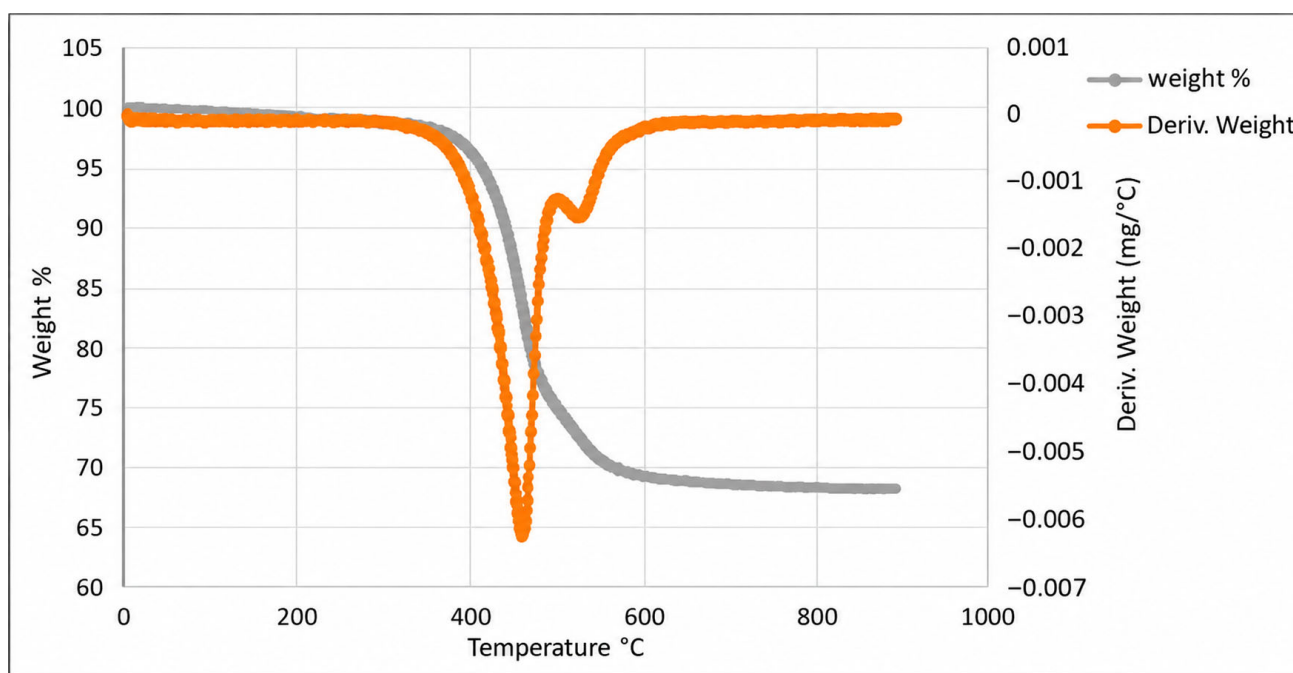


Figure 9. TGA graph of present AA6061.

3.3. Analysis of the Thermal Debinding Process

The transition from a “green” body to a “brown” body is the most critical phase of the MEX-Metal process. Polymer removal must be complete enough to allow for sintering but controlled enough to prevent structural collapse due to gravity (slumping). Experiments conducted between 350 °C and 550 °C (Figure 10) demonstrated that weight loss is strongly temperature dependent. At 350 °C, weight loss remained below 29%, indicating the persistence of residual binder. Conversely, at 550 °C, the threshold of 31% was reached, confirming that the thermal energy was sufficient to crack the PLA chains and volatilize its components. The effect of the isothermal holding time was equally decisive (Figure 11). It was observed that at lower temperatures (350–400 °C), increasing the hold time from 1 to

3 h produced only modest increases in weight loss, which was insufficient for total removal. However, at 550 °C, a 3 h hold guaranteed near-complete degradation, allowing even the core of the 10 mm diameter samples to be effectively processed. SEM observations of the brown parts (Figure 12) reveal an intricate network of interconnected micropores across the powder particles. These channels play a critical role by serving as pathways for the evacuation of residual gases and facilitating the capillary action of the liquid phase during sintering. (Figure 12a–d) demonstrates that polymer degradation is incomplete at holding temperatures between 350 °C and 500 °C, leaving residual binder within the structure. In contrast, (Figure 12e) shows that raising the temperature to 550 °C successfully achieves complete polymer breakdown and binder removal within a 3 h holding time. Additionally, a minor weight increase was observed under certain conditions, which is attributed to localized oxidation caused by aluminum's high affinity for residual oxygen, even within a controlled nitrogen atmosphere.

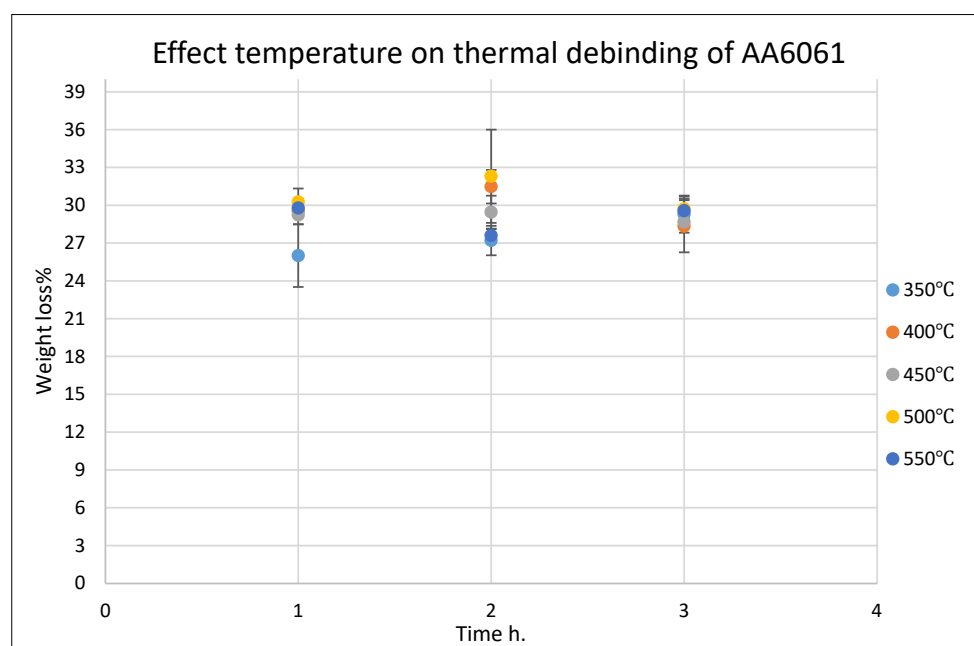


Figure 10. Effect of temperature on the thermal debinding process of printed AA6061 parts. Weight loss % after debinding at various temperatures: 350 °C, 400 °C, 450 °C, 500 °C, and 550 °C.

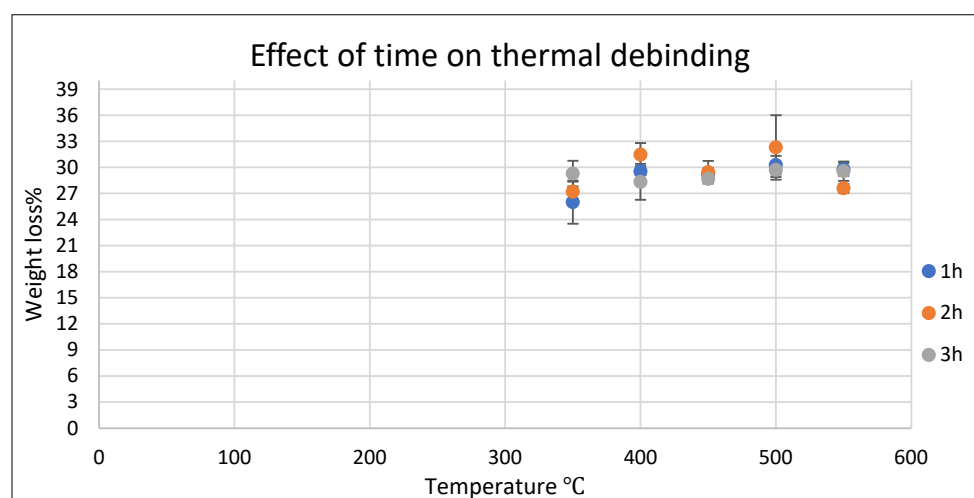


Figure 11. Effect of time on the thermal debinding process at different times at 1 h, 2 h, and 3 h.

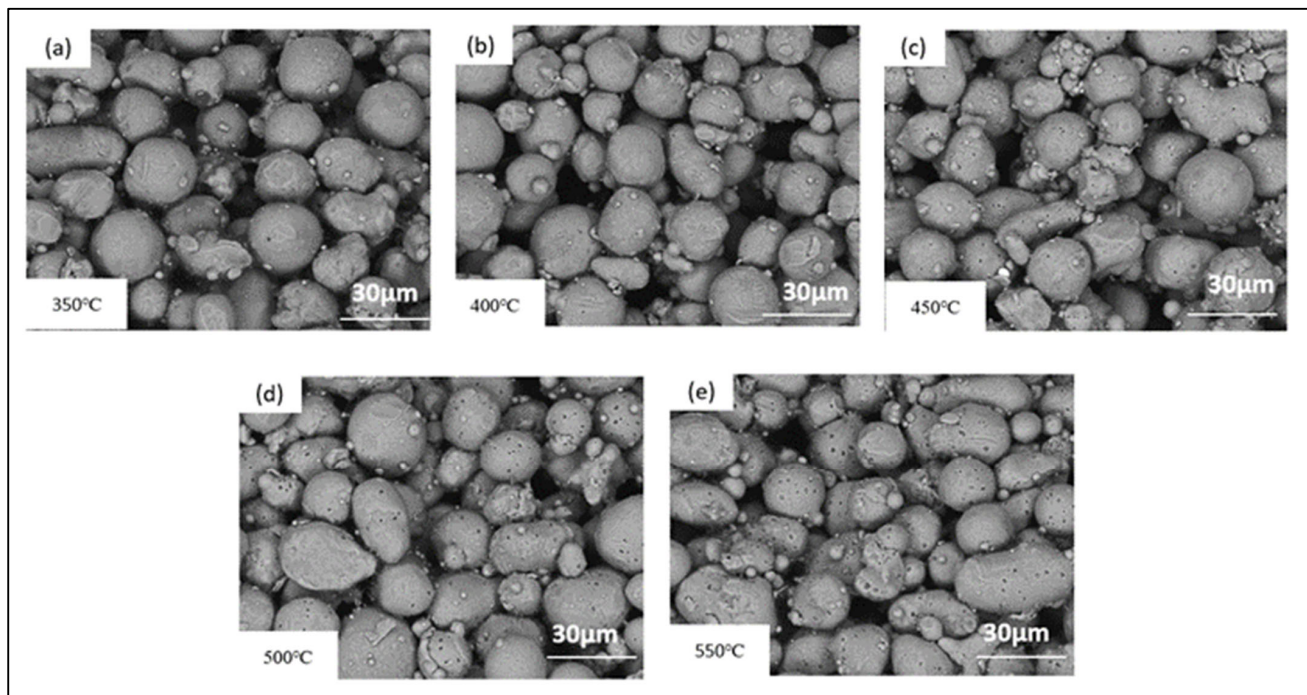


Figure 12. Microstructure appearance by SEM after thermal debinding at different temperatures (a) 350 °C, (b) 400 °C, (c) 450 °C, (d) 500 °C, and (e) 550 °C.

3.4. Sintering Optimization and the Role of Pre-Pressing

The sintering of aluminum alloy feedstock produced by FDM is challenging because the metallic particles are covered by a stable aluminum-oxide layer, which can restrict atomic diffusion and limit metallurgical bonding. In the present study, a 1250 MPa pre-sintering compaction step, a Ti–Mg oxygen-getter system, and a vacuum-assisted atmosphere were used to improve particle contact, reduce oxygen exposure, and promote oxide-layer disruption. These measures were intended to facilitate densification during subsequent thermal treatment. Twelve sintering conditions were evaluated using temperatures between 625 and 655 °C and holding times of 1–3 h, as summarized in Tables 4 and 5.

Table 4. Sintering process values under different temperatures and times in Cycle 1 unpressed.

Run/No	Temperature (°C)	Holding Time (h)	Heating Rate (°C/min)	Weight Loss (%)	Dimension Change (%)	Porosity (%)	Hardness HV
1	625	1	1	27	9.2	12.59	63
2	625	2	1	29	10	11.53	55.3
3	625	3	1	26.6	10	10.83	73.3
4	635	1	1	28.18	10	7.9	68
5	635	2	1	29.5	10	6.87	68.7
6	635	3	1	28.55	10.5	5.36	76
7	645	1	1	29.75	3.8	16.91	62
8	645	2	1	30.02	7.2	15.52	75
9	645	3	1	28.73	2.2	11.69	51.7
10	655	1	1	31.06	3.04	18.51	65.7
11	655	2	1	29.42	4.3	7.85	66.3
12	655	3	1	30.02	0.46	33.18	74

Table 5. Temperatures selected in the modified cycle of the sintering process in Cycle 2 unpressed.

Run/No	Temperature (°C)	Holding Time (h)	Heating Rate (°C/min)	Weight Loss (%)	Dimension Change (%)	Porosity (%)	Hardness HV
1	625	3	1	29.06	23.29	19.09	65.3
2	635			28.85	0.37	1.6	77
3	645			29.13	3.7	1.9	59.7
4	655			29.12	1.14	5.3	67.7

The microstructural and dimensional responses were strongly dependent on the sintering temperature. Specimens treated at 625 °C exhibited limited structural integrity, insufficient particle bonding, and a powdery or friable appearance, as shown in Figure 13a–c. The dimensional change of 23.29% measured for the 625 °C, 3 h condition may have resulted from particle rearrangement, binder removal, gas release, and localized deformation rather than uniform shrinkage. Therefore, this condition is described as having insufficient densification rather than being attributed exclusively to a specific phase-transformation mechanism.

Increasing the temperature to 635 °C improved densification and mechanical performance. Under the Cycle 1 conditions, the 635 °C, 3 h treatment produced the lowest measured porosity, approximately 5.36%, and a hardness of 76 HV. These results indicate that the combination of temperature and holding time promoted improved particle bonding and pore reduction. At temperatures of 645 and 655 °C, the specimens exhibited increased dimensional instability, pore opening, and mechanical degradation, as shown in Figure 13g–l and Tables 4 and 5, with porosity reaching approximately 33.18% for the 655 °C, 3 h condition. This behavior is likely associated with over-sintering, localized particle rearrangement, gas entrapment, or pore coalescence. Although this abrupt transition is consistent with potential liquid-phase-assisted phenomena in the Al–Mg–Si system, the present study did not perform differential scanning calorimetry (DSC) or direct phase analysis; consequently, the active participation of a transient liquid phase remains a proposed mechanism rather than a confirmed observation, warranting future investigation via phase-equilibrium calculations and post-sintering microstructural characterization.

Two heating cycles were also compared to evaluate the effect of the thermal profile on densification. Cycle 1 used a linear heating rate of 1 °C/min, whereas Cycle 2 employed a slower heating rate of 0.1 °C/min during the critical debinding-to-sintering interval between 550 and 580 °C. The slower heating rate was designed to allow more gradual binder decomposition and gas release, thereby reducing the risk of internal gas entrapment and localized swelling. For unpressed specimens, the modified Cycle 2 reduced the local pore-area fraction measured by SEM/ImageJ from 5.36% in Cycle 1 to 1.6% at 635 °C in Cycle 2, indicating lower local microstructural porosity in the analyzed polished cross-sections (Tables 4 and 5).

As shown in Table 5, Cycle 2 produced improved densification and lower porosity than the corresponding Cycle 1 conditions. The hardness results in (Figure 14) and the microstructural observations in (Figure 15) indicate that the modified heating cycle promoted a more cohesive metallic structure with fewer localized defects.

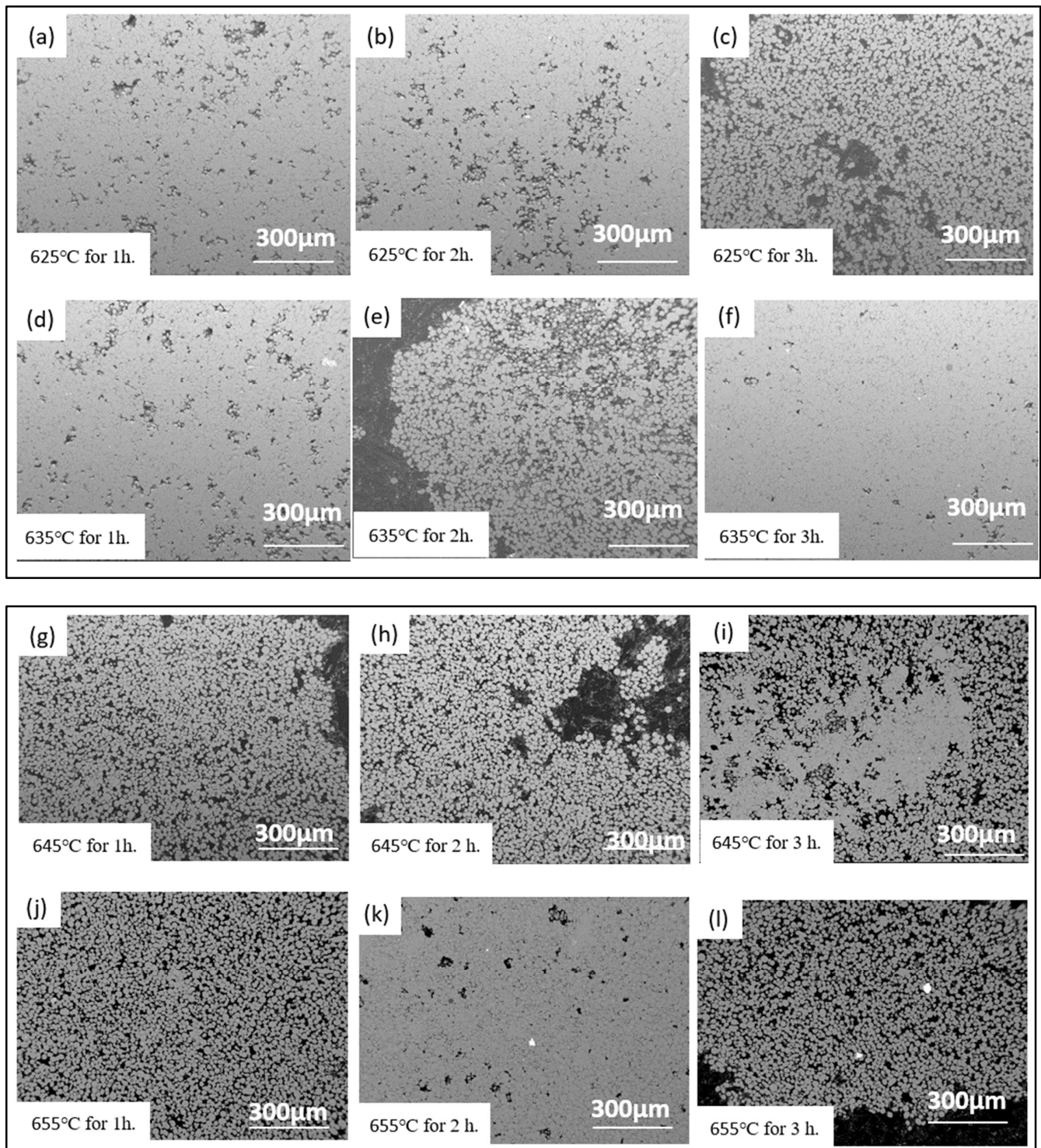


Figure 13. Microstructural evolution by SEM after the sintering process of Cycle 1 (a) 625 °C for 1 h, (b) 625 °C for 2 h, (c) 625 °C for 3 h, (d) 635 °C for 1 h, (e) 635 °C for 2 h, (f) 635 °C for 3 h, (g) 645 °C for 1 h, (h) 645 °C for 2 h, (i) 645 °C for 3 h, (j) 655 °C for 1 h, (k) 655 for 2 h, (l) 655 °C for 3 h.

The aluminum-oxide layer remains an important factor affecting particle bonding during sintering. The 1250 MPa pre-sintering load increased the packing density and may have mechanically disrupted portions of the oxide layer, thereby increasing the area of metallic contact between adjacent particles. In addition, the Ti–Mg oxygen-getter system and vacuum-assisted atmosphere may have reduced the oxygen available for further oxidation and contributed to improved interparticle bonding. Because oxide-

layer disruption and oxygen removal were not directly quantified in the present work, these effects are considered possible contributing mechanisms rather than experimentally confirmed processes [25–27].

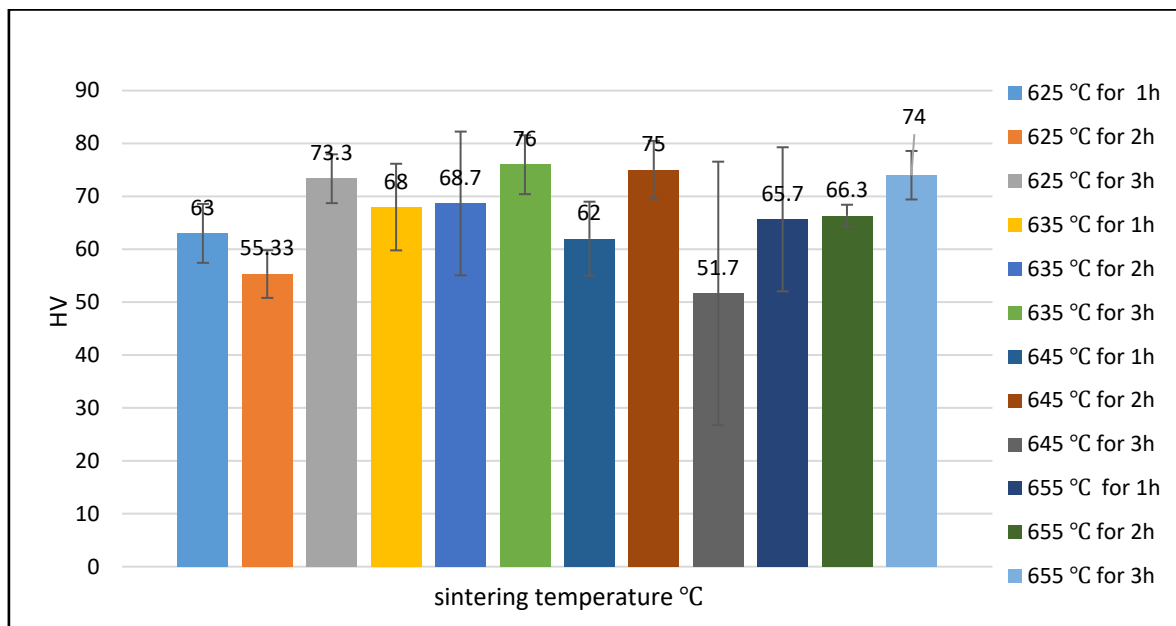


Figure 14. Vickers hardness (HV) of sintered AA6061 at different temperatures in Cycle 1.

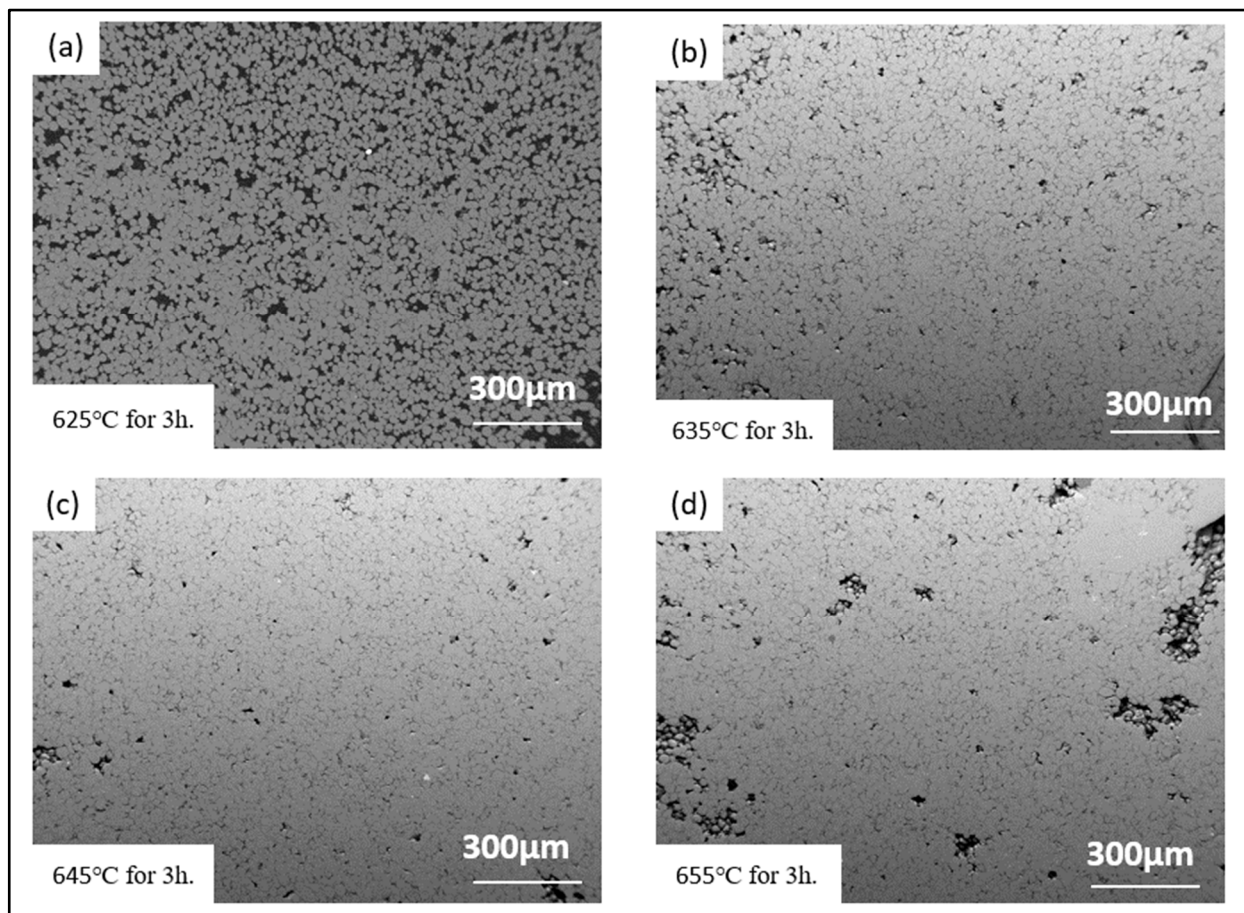


Figure 15. Microstructural evolution by SEM after sintering by the modified Cycle 2 at a holding time of 3 h. (a) 625 °C, (b) 635 °C. (c) 645 °C, (d) 655 °C.

Overall, the results show a transition from insufficient bonding and high residual porosity at lower thermal inputs to improved densification and mechanical performance under the optimized condition of 635 °C, 3 h, and Cycle 2. While these improvements are consistent with the proposed role of oxide-layer disruption and potential transient liquid-phase contributions during optimized heating, confirming these mechanisms will require targeted surface analytics in future work.

The optimized specimens exhibited improved interparticle bonding, reduced visible defects, and a hardness approaching 77 HV, as reported in Table 5. The observed trends demonstrate that pre-sintering compaction and thermal-profile control substantially affect the densification behavior of the FDM-produced aluminum feedstock as shown in (Figure 16). Although the results are consistent with possible liquid-phase-assisted sintering, direct confirmation of a transient eutectic liquid phase requires further thermal and phase characterization.

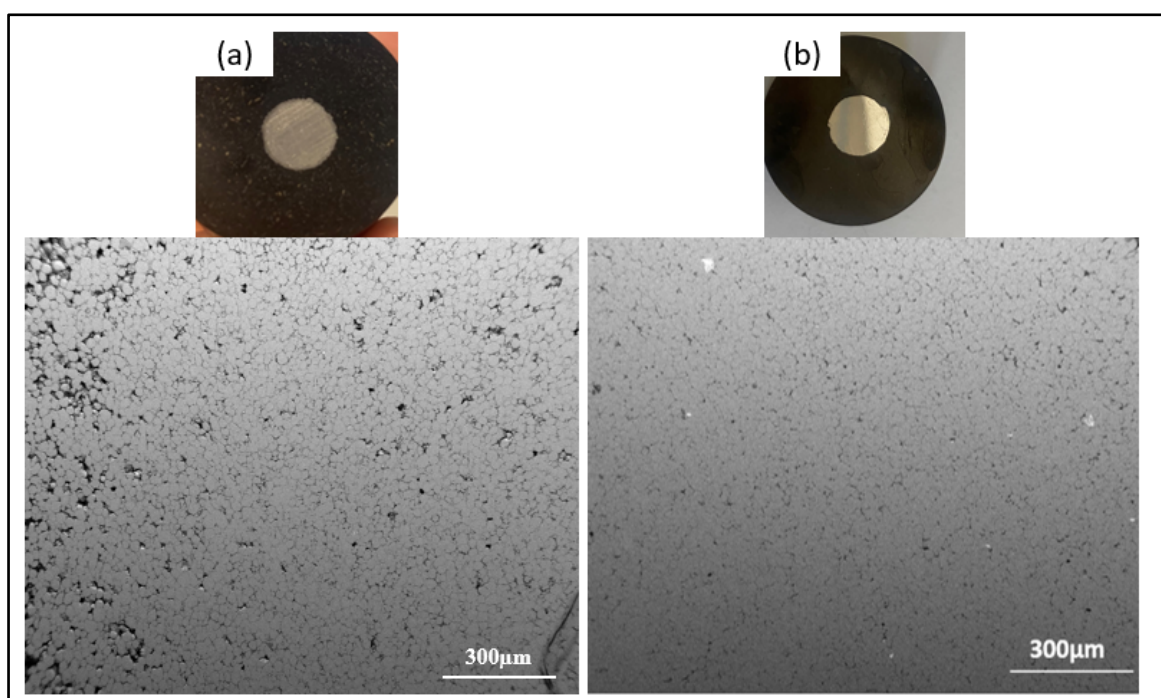


Figure 16. SEM images of samples sintered at Cycle 2. (a) 635 °C without pressing, and (b) 635 °C with pressing, vacuum pump, and oxygen tramp. The optimal debinding cycle (550 °C, 3 h, 3.5 °C/min).

3.5. Density, Porosity, and Mechanical Performance Analysis

The effectiveness of the optimized FDM-sintering route was quantitatively evaluated through Archimedes density measurements and hardness testing. The theoretical density of AA6061 was taken as approximately 2.70 g/cm³. As summarized in Table 6, the unpressed specimens sintered at 635 °C for 3 h reached a density of 2.00 g/cm³, corresponding to a relative density of 74.07%. The corresponding bulk porosity, calculated from the Archimedes-derived relative density, was 25.93%.

Table 6. Archimedes-derived relative density of unpressed and pre-sintering-pressed AA6061 specimens sintered at 635 °C for 3 h.

Relative density	Without pressing	With pressing
	74.07%	88.34%

The relatively high bulk porosity of the unpressed specimens indicates incomplete overall densification under this condition. After pre-sintering pressing, the relative density increased to 88.34%, corresponding to a bulk porosity of 11.66%, as presented in Table 7. These results demonstrate that the pre-sintering pressing step improved bulk densification and reduced the overall pore volume of the sintered AA6061 specimens [28].

Table 7. Corresponding bulk porosity of unpressed and pre-sintering-pressed AA6061 specimens sintered at 635 °C for 3 h, calculated from Archimedes-derived relative density.

Porosity (%)	Without pressing	With pressing
	25.93%	11.66%

In marked contrast, applying a 1250 MPa pre-sintering pressing load, combined with the optimized Cycle 2 and a vacuum-assisted atmosphere, successfully elevated the relative density to 88.34% (2.38 g/cm³). Mechanical performance followed this densification trend, as seen in the hardness values reported in Table 8. While the unpressed samples averaged a Vickers hardness of 76 HV, the optimized “pressed and vacuum-sintered” samples achieved 83.67 ± 0.67 HV. Although this value remains below the ~107 HV typical of the fully dense, precipitation-hardened AA6061-T6 wrought condition [29], it represents a clear enhancement over standard material extrusion aluminum parts, which routinely experience high residual porosity and limited inter-particle bonding [11,30]. Within the framework of powder-based extrusion manufacturing, achieving 83.67 ± 0.67 HV directly correlates with suppressed porosity (11.66%) and the mechanical disruption of surface oxide layers facilitated by the combined 1250 MPa re-sintering load and transient liquid phase mechanisms [30,31].

Table 8. Vickers hardness of unpressed and pre-sintering-pressed AA6061 specimens sintered at 635 °C for 3 h.

HV	Without pressing	With pressing
	76	83.67 ± 0.67

These results confirm that the proposed methodology provides a viable pathway for manufacturing functional aluminum components. Furthermore, the ultimate compressive strength of the filament samples reached 115 ± 3 MPa ($n = 3$), governed strongly by processing parameters such as sintering temperatures, pre-sintering pressure, and hold times. Higher relative densities, lower porosities, and greater Vickers hardness consistently correlate with improved compressive performance. As demonstrated by the true stress–strain curves in Figure 17a, it should be noted that the reported value of 115 ± 3 MPa represents the statistical average across multiple tested specimens, whereas the curve displayed in Figure 17a corresponds to a representative sample exhibiting a peak stress of ~120 MPa, reflecting normal experimental scatter. As further highlighted by the SEM fractographs in Figure 17b, these denser, more uniform microstructures promote more homogeneous plastic deformation.

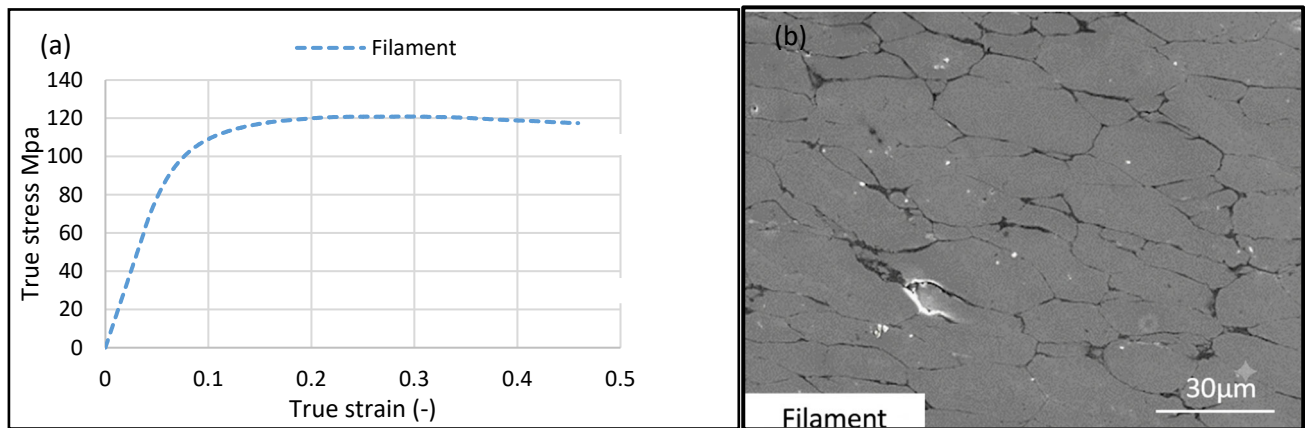


Figure 17. (a) True stress–strain curve from the filament compression test, and (b) corresponding SEM fractograph of the AA6061 filament.

4. Conclusions

This research successfully established a comprehensive methodology for the fabrication of AA6061 aluminum alloy components using FDM-based additive manufacturing. By optimizing the transition from green to sintered bodies, the following conclusions were drawn:

1. **Optimization of Printing and Feedstock Parameters:** The study identified that a nozzle diameter of 0.8 mm, a temperature of 205 °C, and a high infill overlap of 100% are critical to achieving structural integrity in green parts. The integration of a FilaWarmer device proved essential in mitigating filament brittleness, ensuring a continuous and defect-free extrusion process for the metal-infused PLA feedstock;
2. **Effective Thermal Debinding Strategy:** Thermogravimetric analysis (TGA) and differential scanning calorimetry (DSC) provided the foundation for a staged debinding protocol. It was determined that a peak temperature of 550 °C with an isothermal hold of 3 h is necessary to ensure the near-complete removal of the organic binder (>31% weight loss). This stage is pivotal in creating the interconnected capillary networks required for successful liquid phase sintering;
3. **The Critical Role of Pre-Sintering Pressing:** To overcome the inhibitory effect of the native oxide layer, introducing a 1250 MPa pre-sintering pressing step was fundamental. This mechanical intervention, combined with a Ti-Mg oxygen getter and a vacuum-assisted atmosphere, effectively disrupted the oxide shells and promoted metallic necking, a feat often unattainable through conventional pressureless sintering of aluminum;
4. **Sintering Kinetics and Thermal Cycles:** The optimal sintering window was identified at 635 °C for 3 h. The implementation of a refined heating cycle (Cycle 2), characterized by an extremely slow ramp rate (0.1 °C/min) during the final stages of degasification, significantly improved the cohesion of the metallic matrix and reduced internal micro-porosity compared to standard linear heating profiles;
5. **Final Performance Metrics:** The AA6061 filament sample exhibited a compressive strength of 115 ± 3 MPa. The optimized processing route produced parts with 88.34% relative density (11.66% porosity) and a Vickers hardness of 83.67 ± 0.67 HV. Although these values are lower than T6 heat-treated AA6061, they represent a notable improvement for low-cost MEX-metal processes and show that functional AA6061 components with good structural reliability can be produced;
6. **Industrial Feasibility and Cost Implications:** While the proposed methodology successfully utilizes low-cost MEX-metal feedstock to circumvent traditional high-end

metal AM equipment expenses, the introduction of a 1250 MPa pre-sintering pressing step introduces important industrial considerations. Although uniaxial pressing increases tooling requirements and shifts the process from a purely continuous additive workflow to a semi-automated hybrid route, it remains significantly more cost-effective than HIP or laser- or electron-beam powder bed fusion systems, making it a viable alternative for batch-scale component manufacturing;

7. **Limitations and Future Outlook:** While this work demonstrates the feasibility of processing AA6061 components via FDM and pressure-assisted sintering, several limitations should be noted. First, the experimental findings are based on a single feedstock batch; subtle batch-to-batch variations or minor deviations from the nominal AA6061 composition may influence debinding and densification behavior. Second, although an optimized relative density of 88.34%, corresponding to 11.66% porosity, was achieved, the proposed contribution of a transient liquid phase during sintering remains an unverified hypothesis, as direct high-temperature XRD or DSC analyses were not performed. Third, the reported properties are based on a limited sample set: $n = 3$ specimens for compressive testing (115 ± 3 MPa) and Archimedes density measurements, and three representative cross-sectional regions for microstructural porosity quantification (1.6%). Therefore, the mechanical and physical results should be regarded as a preliminary statistical assessment. Larger sample lots, together with dynamic and fatigue testing, will be required for broader industrial qualification. Future work will investigate secondary alloy modifications, post-sintering hot isostatic pressing (HIP), and advanced off-gas and phase characterization to further reduce residual porosity and confirm the underlying sintering mechanisms.

In summary, this work provides a scalable and cost-effective framework for aluminum additive manufacturing, bridging the gap between polymer-based extrusion and high-end metal sintering through strategic mechanical and thermal optimizations.

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Abbreviations

The following abbreviations are used in this manuscript:

FDM	Fused Deposition Modeling
AM	Additive manufacturing
MEX	Material extrusion
SEM	Scanning electron microscopy

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