

Process–Microstructure–Property Relationships in Binder-jetted AISI D2 Tool Steel: Effects of Sintering Temperature and Heat Treatment

Jong Wan Ko

jwko@kitech.re.kr

Korea Institute of Industrial Technology

Heejun Cho

Korea Institute of Industrial Technology

AWAL AFLIZAL ZUBIR

Korea Institute of Industrial Technology

Research Article

Keywords: AISI D2 tool steel, binder jetting, sintering, heat treatment, mechanical properties, microstructure

Posted Date: September 4th, 2026

DOI: <https://doi.org/10.21203/rs.3.rs-10757310/v1>

License:  This work is licensed under a Creative Commons Attribution 4.0 International License.

[Read Full License](#)

Additional Declarations: No competing interests reported.

Process–Microstructure–Property Relationships in Binder-jetted AISI D2 Tool Steel: Effects of Sintering Temperature and Heat Treatment

Abstract

AISI D2 tool steel is a high-carbon, high-chromium cold-work tool steel widely used in wear-resistant applications due to its excellent hardness and compressive strength. However, its poor machinability and high carbide content limit the fabrication of complex geometries using conventional manufacturing techniques. Binder jetting (BJT), a sintering-based additive manufacturing technique that avoids localized melting, has therefore attracted increasing attention as a promising approach for fabricating sophisticated D2 components. Nevertheless, the limited sintering process window and intricate carbide dissolution behavior of AISI D2 impose significant challenges in achieving complete densification and stable mechanical properties. Furthermore, a systematic understanding of the combined effects of sintering temperature and subsequent heat treatment remains insufficient.

In this study, the effects of sintering temperature and subsequent heat treatment on the microstructural evolution and mechanical properties of BJT-fabricated AISI D2 tool steel were systematically investigated. Sintering was conducted at temperatures ranging from 1,200 to 1,260 °C, followed by austenitizing at 1,010 °C, air cooling, and tempering at 200 °C. An optimal sintering temperature of 1,250 °C was identified, resulting in near-complete densification (~99%). This densified microstructure enabled a significant improvement in mechanical performance after heat treatment, with a tensile strength of approximately 1,300 MPa and an elongation of 6.81%. EBSD analysis indicated that these enhancements were attributable to martensitic transformation and the homogenization of internal lattice strain. These results establish a narrow processing window for the near-full densification of binder-jetted AISI D2 and demonstrate that subsequent hardening and tempering can substantially improve its tensile performance, providing a process–microstructure–property basis for further development of binder-jetted tool-steel components.

Keywords

AISI D2 tool steel, binder jetting, sintering, heat treatment, mechanical properties, microstructure

Highlights

- Near-full densification (~99%) of AISI D2 tool steel was achieved at 1,250 °C during binder jetting based sintering.
- Sintering temperature governed carbide evolution, pore development, and mechanical performance.
- Post-sintering heat treatment increased tensile strength from 860 MPa to 1,300 MPa and hardness from 692 HV to 720 HV.
- EBSD analysis indicated austenite-to-martensite transformation and homogenization of internal lattice strain after heat treatment.

1. Introduction

AISI D2 tool steel, a high-carbon, high-chromium alloy, contains a carbon content of 1.4-1.6 wt% and a chromium content of 11-13 wt% [1-3]. It is renowned for its exceptional hardness, wear resistance, and compressive strength [1-4]. These characteristics have contributed to its extensive application in cold-working tools, including cutting dies, blanking dies, shear blades, and forming tools across diverse industrial sectors [1,3,4]. However, AISI D2 tool steel comprises a significant amount of Cr-rich hard carbides (M_7C_3), which can accelerate tool wear during cutting operations, leading to cracks and dimensional inaccuracies [5-7]. Tang et al. indicate that the friction and wear behavior of hardened AISI D2 tool steel are substantially affected by the materials' hardness levels, with rapid wear occurring under high-load conditions due to local stress concentrations [8]. The inherent difficulties in machinability constrain the fabrication of complex geometries, and the manufacturing process is often hindered by high costs and extended lead times associated with mold production and processing [1,7].

To address these low machinability and restricted design issues, near-net-shape processes such as powder metallurgy and metal injection molding have been employed; however, these approaches still have limitations, including high initial costs associated with mold fabrication and challenges in manufacturing internal channels and intricate geometries [9]. In recent years, driven by a substantial increase in industry demand for high-performance and complex-shaped cold molds, conventional manufacturing technologies have struggled to concurrently attain the dual objectives of geometrical constraints and cost reduction.

Additive manufacturing (AM) is increasingly recognized as a promising technology capable of addressing limitations in design flexibility. Laser-based AM, including powder bed fusion (PBF) and directed energy deposition (DED), manufactures 3-dimensional objects using computer-aided design. They utilize a laser as the energy source to fuse laminated metal powder layers or wire as the initial feedstock material. They provide high precision and facilitate the fabrication of complex geometries, thereby enabling the production of a diverse range of alloys. However, AISI D2 tool steel, a high-carbon alloy, exhibits a high cooling rate during welding, leading to poor weldability due to low temperature cracking caused by martensitic transformation [10,11]. Lim et al. reported that in laser-based AM of AISI D2 tool steel, high-temperature cracks are primarily caused and propagated along M_7C_3 carbide and eutectic austenite interfaces, along with tensile residual stress during repeated reheating [12]. Yadav et al. analyzed the heat treatment and microstructural changes following the laser-DED manufacturing of AISI D2 tool steel, noting that the high carbon and high chromium content of AISI D2 steel constitute a significant constraint on process stability [13]. Moreover, segregation of carbon and chromium, as well as carbide reprecipitation, are observed during the melting process [13,14]. Consequently, laser-based AM, such as PBF and DED, possess significant potential for the fabrication of AISI D2 tool steel. Nevertheless, this potential is contingent upon resolving the limitations associated with the heat cycle through ongoing process optimization and heat treatment advancements and the cost reduction still remains a challenge.

Binder jetting (BJT), a sintering-based AM, is a promising alternative for avoiding residual stresses having a lower probability of inducing localized melting unlike laser-based techniques [15-17]. It is particularly advantageous for the industrial sector, as it reduces thermal cracking issues associated with high-strength, high-carbon materials such as cold-work tool steel. BJT selectively applies binders to a repeatedly laminated powder layer to form a specific 3-dimensional shape. After that, the 3-dimensional shape gains strength through a curing process known as crosslinking, which enables the binder to establish physical bonds with thermal energy. The crosslinked components were subsequently debound and sintered to obtain dense final parts. In addition, the BJT offers strong potential for high-throughput mass production with cost reduction, as it enables layer-wise fabrication without localized melting, allowing faster build rates, efficient powder reuse, and scalable batch processing [18]. Although the BJT effectively mitigates issues such as hot cracking and is well-suited for industrial-scale production, only a limited number of studies have investigated its application to tool steels, including AISI D2. Nandwana et al. revealed that in H13 tool steel manufactured through BJT, there is a narrow process window of about 40 °C during supersolidus liquid phase sintering (SLPS), and that carbide precipitation, grain growth, and residual stress changes caused by the sintering temperature are the dominant factors affecting hardness and mechanical properties [19]. Choudhari et al. examined various printing and sintering parameters to assess their influence on the quality of AISI M2 tool steel, demonstrating that its mechanical properties can be tailored for specific applications [20]. A preliminary study presented by Rodríguez et al. examined the influence of sintering parameters on densification, shrinkage rate, and microstructural development in AISI D2 fabricated via BJT [21]. Though Rodríguez et al. demonstrated the feasibility of densification, the relative density behavior as a function of sintering temperature continued to exhibit abrupt transitions (from 90% to 99%), suggesting a highly sensitive densification behavior within the investigated temperature range. Furthermore, the sintering

temperature, within the SLPS regime, not only influences densification but also dictates the fundamental microstructure state, such as carbide morphology and the chemical composition of the matrix [20]. Since the initial microstructure state directly governs the phase transformation kinetics and hardenability during subsequent heat treatments, any inconsistency in the sintering process can lead to unpredictable mechanical performance [22]. Recent studies have demonstrated the densification of binder-jetted D2 tool steel and highlighted the high temperature sensitivity of its SLPS response. However, the coupled effects of a narrowly controlled sintering-temperature window and subsequent hardening and tempering on tensile behavior, phase evolution, and local misorientation remain insufficiently quantified. In particular, a processing condition that simultaneously achieves near-full density while avoiding grain coarsening and grain-boundary carbide-network formation has not been clearly established for binder-jetted D2.

Accordingly, this study (i) maps densification and anisotropic shrinkage over a narrow sintering range of 1,200–1,260 °C, (ii) correlates sintering temperature with microstructure, hardness, and tensile strength, and (iii) evaluates the effect of subsequent austenitizing and tempering on tensile behavior and EBSD-derived phase and local-misorientation characteristics.

2. Materials and methods

2.1 Material and specimen fabrication

In this study, gas atomized AISI D2 tool steel powder supplied by INDO-MIM was employed. Fig. 1 and Fig. 2 show the morphology and particle size distribution of the AISI D2 tool steel powder used. The powder exhibits a spherical shape with a characteristic particle size of $D_{10} = 2.87 \mu\text{m}$, $D_{50} = 10.6 \mu\text{m}$, and $D_{90} = 21.6 \mu\text{m}$. The chemical composition of the AISI D2 tool steel powder, determined by inductively coupled plasma (ICP) analysis, is summarized in Table 1.

In this experiment, printing was performed using the Shop system (Desktop Metal, USA) with a water-based binder to create three-dimensional shapes. The powder layer was applied to a thickness of $50 \mu\text{m}$, and the binder saturation was set to 55%.

Crosslinking was carried out using a Desktop Metal Drying Oven. The specified crosslinking conditions were maintained at 100°C for a duration of 6 hours with a heating rate of $5^\circ\text{C}/\text{min}$, and at 180°C for 10 hours with heating rate of $5^\circ\text{C}/\text{min}$. Following the process, the oven was permitted to cool to ambient temperature. The crosslinked specimens, hereafter referred to as green parts, were then depowdered utilizing a Desktop Metal Depowdering Station, where loose powder was removed from the surface using a brush.

Debinding was conducted in a tube furnace (SH-FU-100TS, Samheung, Korea) under an argon atmosphere (99.999% purity), with the furnace pressure maintained at 17 Torr. The specified conditions for the debinding process involved heating to 500°C for a duration of 1 hour with a heating rate of $4^\circ\text{C}/\text{min}$, followed by an elevation to 600°C for 2 hours with a heating rate of $2^\circ\text{C}/\text{min}$.

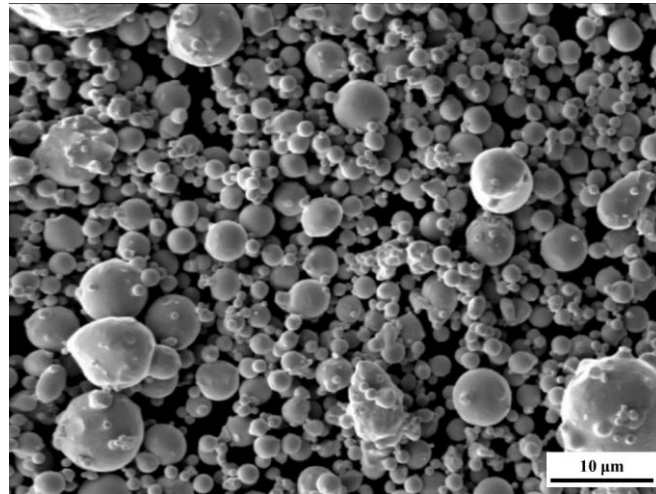


Fig. 1 Scanning electron microscope (SEM) of the AISI D2 tool steel powders

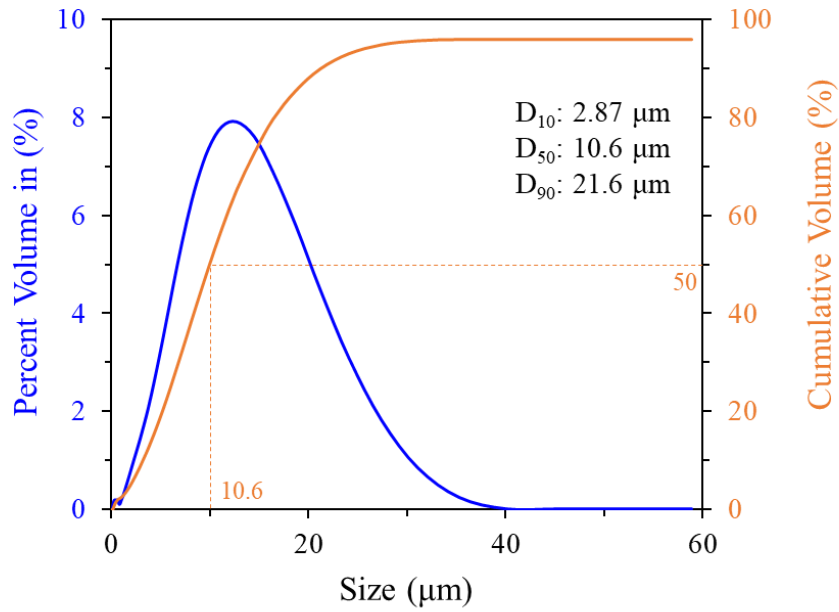


Fig. 2 Particle size distribution analysis of AISI D2 tool steel powder

Table 1. Chemical composition (wt%) of AISI D2 tool steel

	C	Cr	Mo	Mn	Si	V	Cu	Ni	Fe
AISI D2 tool steel	1.59	12.16	0.93	0.37	0.43	0.51	0.46	0.04	Bal.

2.2 Sintering process

The sintering process was performed continuously subsequent to the debinding step within the same tube furnace. The target sintering temperatures were determined based on the Fe-Cr-C phase diagram (shown in Fig. 3 a) and the existing literature concerning the SLPS regime of D2 tool steel [23-25]. The selected temperatures were 1,200 °C, 1,220 °C, 1,240 °C, 1,250 °C, and 1,260 °C. This narrow temperature range (10–20 °C) encompasses the critical transition from solid-state sintering, characterized by M_7C_3 dissolution, to the SLPS regime, and further into the higher liquid phase formation regime. Table 2 and Fig. 4 illustrate the selected temperatures and the sintering temperature profile. Prior to reaching the final sintering temperature, the specimens were held at 1,100 °C for 1 hour with a heating rate of 5 °C/min to ensure structural stabilization. The final sintering phase involved heating from 1,100 °C to the designated target temperatures (illustrated in Table 2) at a rate of 2 °C/min and held for 2 hours, followed by furnace cooling. Furnace temperature was monitored and controlled using process temperature control rings.

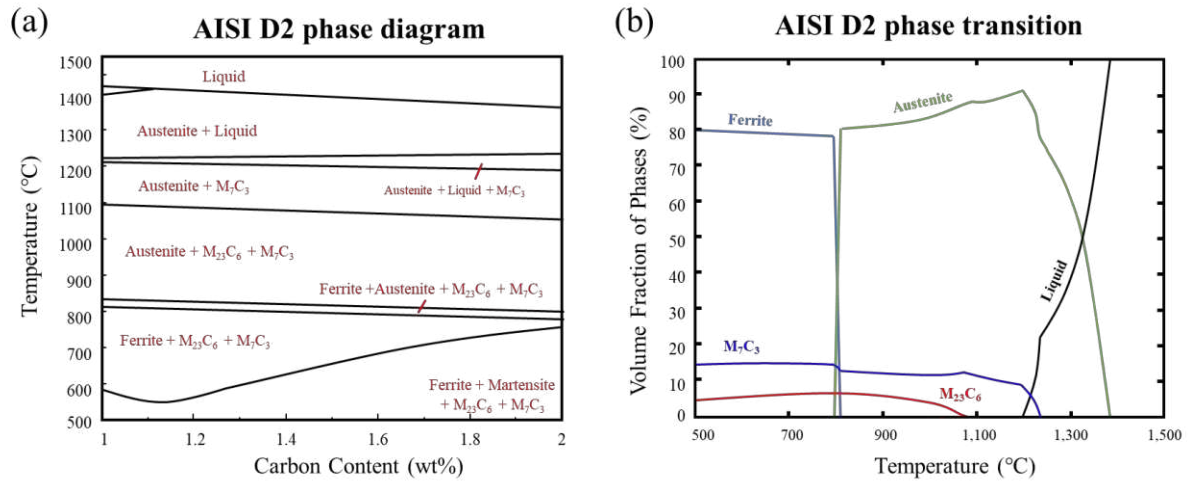


Fig. 3 Thermodynamic graphs related to AISI D2 tool steel: (a)equilibrium phase diagram and (b)phase transition behavior with temperature

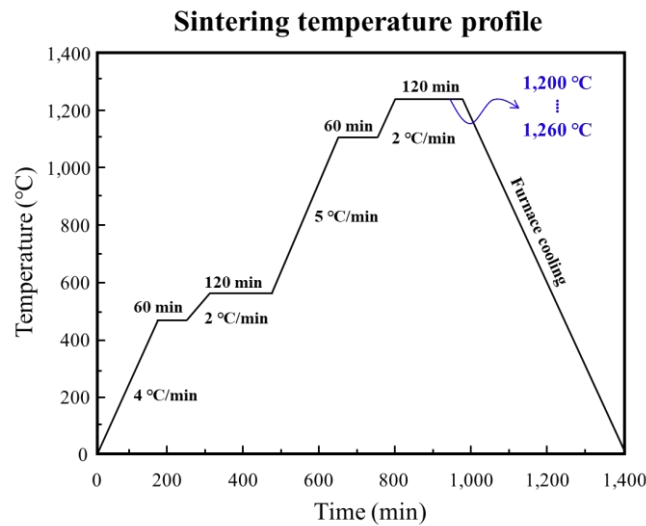


Fig. 4 Sintering temperature profile: the maximum sintering temperature was varied from 1,200 °C to 1,260 °C

Table 2. The sintering temperature conditions and corresponding sintering regime for AISI D2 tool steel

Sintering temperature (°C)	Experimental rationale
1,200	The condition of solid-state diffusion dominant
1,220	The condition of the onset of partial M_7C_3 dissolution
1,240	The condition of the active dissolution of M_7C_3
1,250	The condition of the SLPS commencing
1,260	The condition of the liquid-phase fraction within the SLPS regime

2.3 Heat treatment process

Heat treatment was performed with reference to the standard AISI D2 tool steel heat-treatment procedure [3]. The sintered samples were austenitized at 1,010 °C for 30 minutes with a heating rate of 5 °C/min, followed by air cooling at room temperature. Subsequent tempering was carried out at 200 °C for 2 hours with a heating rate of 10 °C/min. Fig. 5 shows the heat treatment temperature profile for AISI D2 tool steel fabricated by BJT.

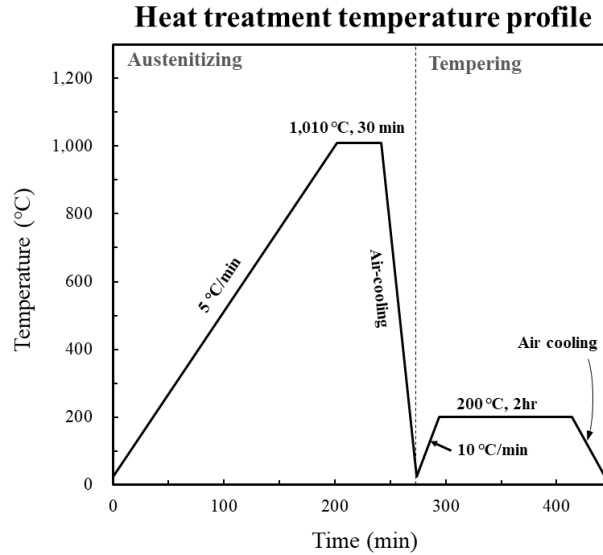


Fig. 5 The heat treatment temperature profile: 1,010 °C austenitizing – air cooling – 200 °C tempering

2.4 Microstructural and mechanical characterization

As shown in Fig. 6, cube specimens measuring 10 × 10 × 10 mm were designed for density measurement and surface observation. For mechanical properties such as tensile test and hardness test, rectangular bar specimens (60 × 12 × 15 mm) were designed to ensure dimensional stability and to avoid deformation during sintering. The final tensile specimens were then machined from the sintered bars using wire-cut electrical discharge machining (EDM). The tensile specimens had an effective gauge length of 8 mm and a cross-sectional area of 2 mm x 4 mm, resulting in a non-standard geometry.

Various characterizations were performed on the sintered and heat-treated samples to analyze their physical properties and microstructure. The density of the sintered samples was measured using the Archimedes method according to ASTM B962, with each sample measured three times and the average value calculated. The relative density was then calculated using equation (1):

$$\text{Relative density (\%)} = \frac{\text{measured density}}{\text{theoretical density}} \times 100 \quad (1)$$

where the theoretical density of AISI D2 tool steel is 7.7 g/cm³.

The shrinkage was calculated by comparing the pre-sintered green part dimensions measured with a digital vernier caliper with the as-sintered dimensions using equation (2):

$$\text{Shrinkage rate (\%)} = \frac{\text{green part dimension} - \text{sintered part dimension}}{\text{green part dimension}} \times 100 \quad (2)$$

For microstructural analysis, the specimens were surface-prepared by grinding (180-1,200 grit SiC paper) and polishing (1 μm diamond suspension and 0.04 μm colloidal silica). The metallographic structure was observed through an optical microscope (Hirox RX-100, Japan). SEM analysis was performed to match the surface quality to the electron backscatter diffraction (EBSD) results presented in Chapter 3. Etching was performed for 25 seconds using a 5% Nital solution (95 mL of C₂H₅OH + 5 mL of HNO₃).

Mechanical property evaluation consisted of tensile and hardness tests. Tensile tests were performed using a Galdabini Quasar 50 instrument, with a crosshead displacement rate of 1 mm/min, taking into account the non-standard sample characteristics due to the short 8 mm gauge length. This corresponds to the range of typical tensile tests ($2.1 \times 10^{-3} \text{ s}^{-1}$).

Hardness measurements were performed on surface-prepared cubic specimens using a micro-Vickers hardness tester (402MVD, Wolpert Willson Instrument, USA) under conditions of a 9.8 N load and a 10-second dwell time. Five measurements were repeated, and the average value was calculated.

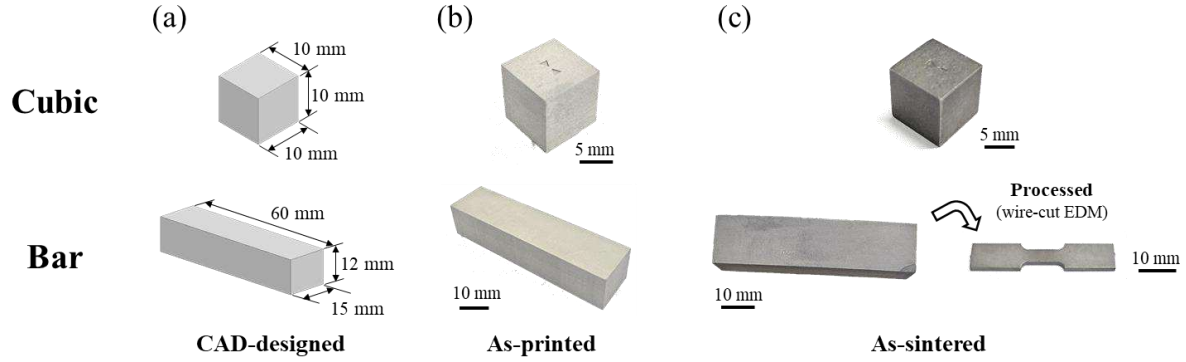


Fig. 6 Specimen geometry and appearance fabricated by BJT of AISI D2 tool steel at various processing stages: (a) CAD-designed models of cubic and bar specimens, (b) green parts after printing and depowdering, (c) sintered specimens following furnace cooling, and the final tensile test specimen machined by wire-cut EDM

3. Results and discussion

3.1 Sintering behavior

The sintering temperature is the most critical parameter that directly influences the densification and shrinkage behavior during the sintering process. In this study, a comprehensive analysis was conducted on the relative density, directional shrinkage, and porosity, with the sintering temperature varied from 1,200 °C to 1,260 °C. The macroscopic effect of the temperature variation is clearly visible in the sintered specimens shown in Fig. 7. As the sintering temperature increases from 1,200 °C to 1,260 °C, the specimens exhibit a visually clear increase in densification and overall volume shrinkage. Fig. 8 (a) and (b) illustrate the relative density and shrinkage across different sintering temperatures, respectively. As the temperature increased from 1,200 °C to 1,240 °C, the relative density showed a gradual increase from 81.35% to 93.92%. At 1,250 °C, a high relative density of 99.13% was attained. At 1,260 °C, the relative density further increased; however, the relative-density difference between 1,250 and 1,260 °C was smaller than the measurement scatter. This suggests that a sufficiently high relative density had been achieved, and densification was complete at 1,250°C. This trend aligns with the densification temperature range documented in studies of AISI D2 tool steel and reduced tool steel [21]. Furthermore, shrinkage was more pronounced along the z axis than along the x or y axes. This anisotropy results from the lower powder packing density caused by the layered structure formed along the z axis and the uneven distribution of the powder binder [15,26,27]. The increase in shrinkage was most evident between 1,200 °C and 1,240 °C, which correlates with the temperature range in which a substantial change in relative density was observed. As the sintering temperature increased, the shrinkage in all directions showed a tendency to rise. Similar to the changes in relative density, the shrinkage along the x and y axes increased from 1,250 °C to approximately 15%, while along the z axis it increased to approximately 21%, yet no significant difference was observed between 1,250 °C and 1,260 °C. This tendency in shrinkage followed by convergence has also been documented in SLPS studies [21,28].

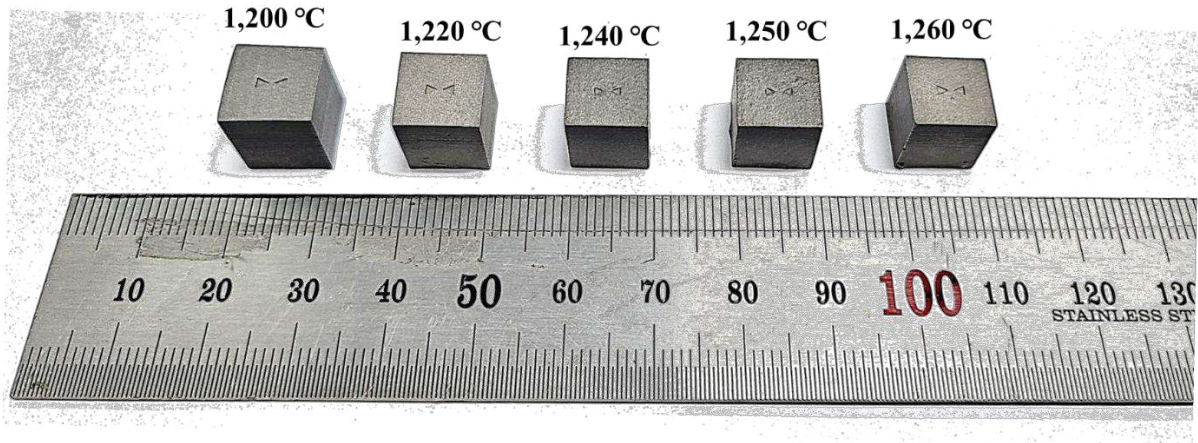


Fig. 7 Images of AISI D2 tool steel cubic specimens after sintering at various temperatures (1,200–1,260 °C)

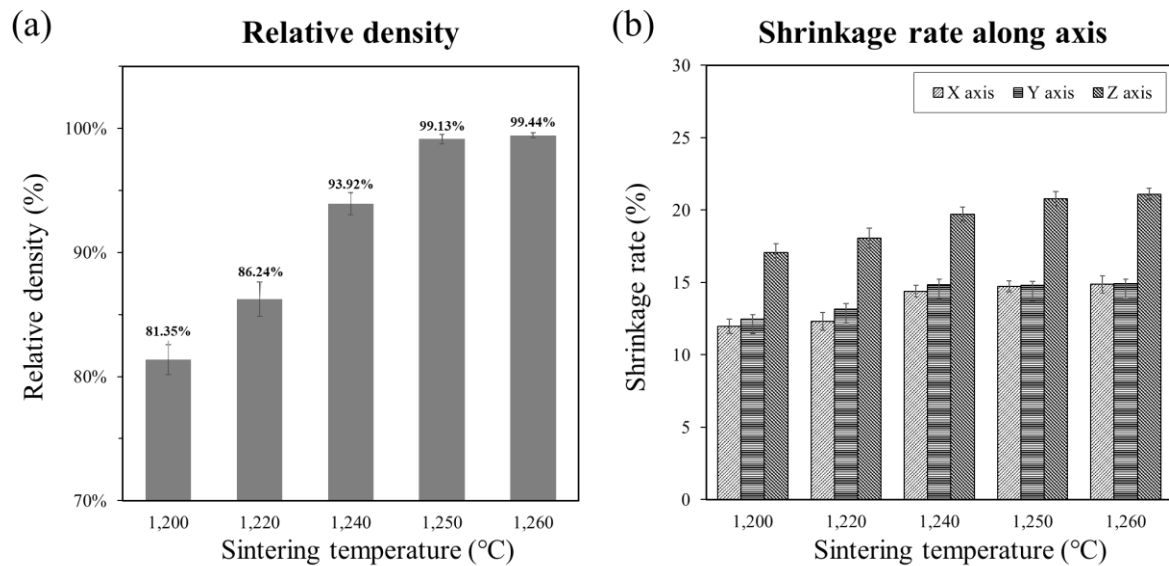


Fig. 8 Effects of sintering temperature on densification and shrinkage behavior of AISI D2 tool steel: (a) relative density as a function of sintering temperature, (b) directional shrinkage along the x, y, and z axes at different sintering temperatures

The evolution of densification and shrinkage is characterized by the surface morphologies observed at varying temperatures. Fig. 9 illustrates the changes in surface structure at various sintering temperatures, magnified at 200x and 700x. At 1,200°C, large pores were observed to be continuously connected, indicating the initiation of neck growth. At 1,220°C, pore connectivity diminished, and pore spheroidization commenced. At 1,240°C, the number and size of pores decreased, resulting in a more uniform structure, in accordance with the temperature range in which the relative density increased sharply. At 1,250°C, pores nearly vanished, yielding a uniform, dense structure that signifies high structural integrity. This densification behavior aligns with prior research reporting the gradual dissolution and diffusion of M_7C_3 carbides [24,29,30]. At 1,260°C, although the total pore volume was minimal, the grain size apparently increased, and over-sintering phenomena were again observed along the grain boundaries with newly emerging pores. This observation suggests that residual pores may have formed during the solidification of the localized liquid phase [28,29,31].

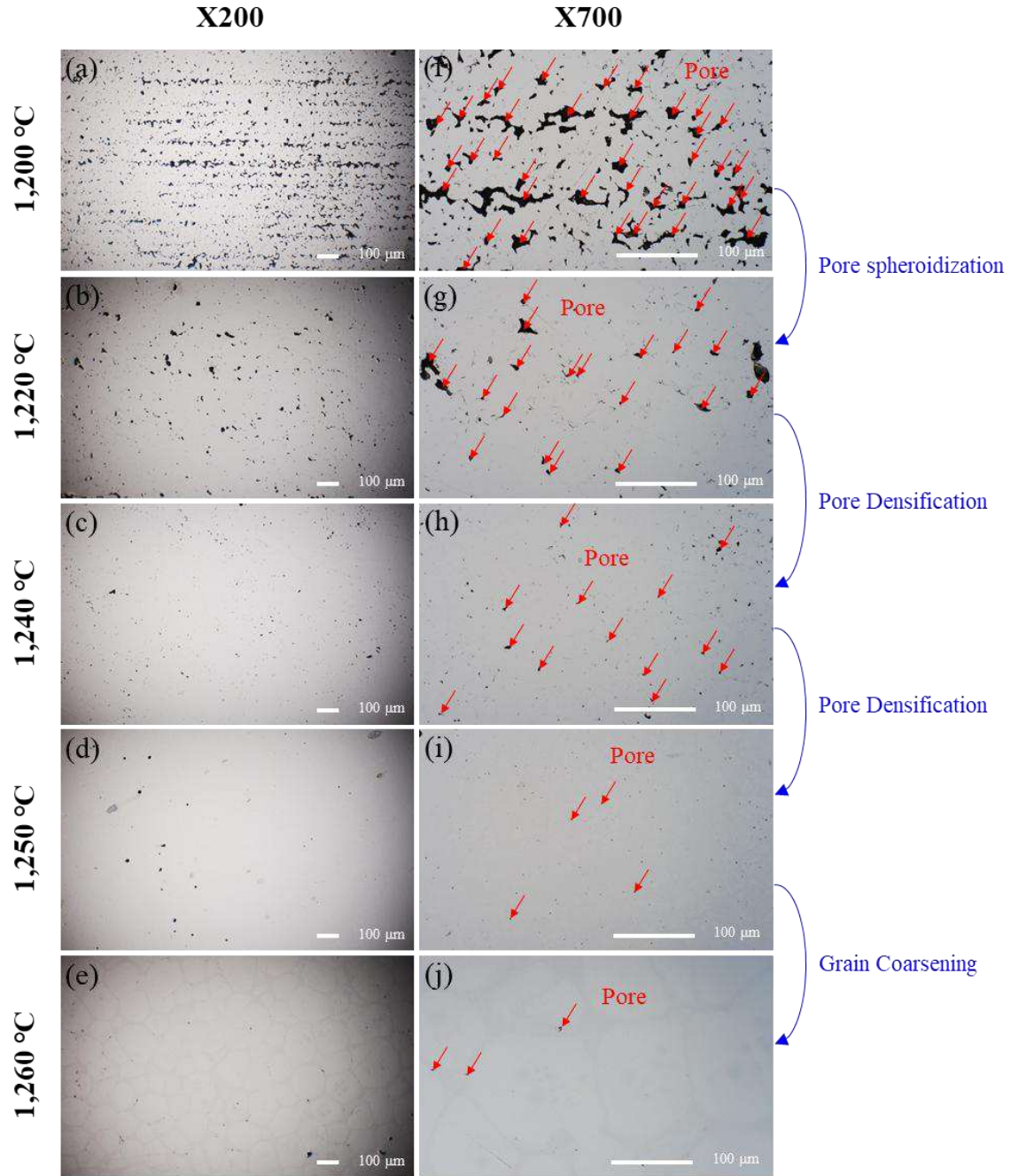


Fig. 9 Optical micrographs of the surfaces of the sintered specimens: taken at 200x magnification (a-e) and 700x magnification (f-j)

Mechanical properties are insufficiently explained by densification behavior alone but are not solely governed by relative density. Instead, a distinct peak is observed at a specific temperature (1,250 °C), as shown in Fig. 10. This implies that the final strength and hardness are governed by complex metallurgical mechanisms, as supported by the microstructural changes illustrated in Fig. 11. The fluctuation in mechanical properties with respect to sintering temperature can be interpreted as the combined effect of carbide dissolution behavior (M_7C_3), modifications in austenite stability, and variations in the fraction of martensite formed during final cooling, extending beyond mere densification. At 1,200 °C, sintering predominantly occurs through solid-state diffusion, with the Cr-rich primary carbide M_7C_3 remaining predominantly intact. Given that carbide dissolution at this temperature is minimal, the level of carbon saturation remains low, thereby restricting the amount of martensite

formed during cooling. Moreover, high pore connectivity combined with inadequate load-transfer pathways results in comparatively low hardness and tensile strength, recorded at 304.5 HV and 555.8 MPa, respectively. At 1,220 °C, thermodynamic calculations suggest the onset of partial carbide dissolution and liquid-phase formation. Throughout this process, the enrichment of carbon and chromium within the matrix increases, and subsequent cooling induces the formation of martensite, which leads to enhanced hardness and tensile strength values of 349.83 HV and 645.48 MPa, respectively. Additionally, partial carbide dissolution facilitates increased grain boundary mobility, thereby promoting pore spheroidization and neck growth, as evidenced by the homogenized pore morphology depicted in Fig. 11 (b). At 1,240 °C, progressive dissolution of M_7C_3 carbides increases the carbon and alloying-element content of the austenitic matrix. Although this enrichment may enhance austenite stability, it is considered insufficient at this stage to strongly suppress martensitic transformation during subsequent cooling. Consequently, the combined effects of increased matrix solute content, martensitic transformation, and reduced porosity contribute to the increase in hardness and tensile strength to 445.78 HV and 741.73 MPa, respectively.

The ascending segments observed in Fig. 10 correspond to this metallurgical transition. Moreover, the diffusion pathways are established through the localized melting of M_7C_3 carbide, facilitating rapid densification and a significant reduction in porosity. Among the investigated conditions, 1,250 °C exhibited the best mechanical properties in this study. At this temperature, austenite is effectively stabilized under high-carbon, high-chromium conditions. The martensite formed during cooling exhibits a relatively uniform composition, resulting in high hardness and strength of 692.5 HV and 860.63 MPa, respectively. Simultaneously, a small amount of liquid phase formed during SLPS significantly reduces residual porosity within the structure. A relatively homogeneous and densified microstructure was observed, as shown in Fig. 11(d). At 1,260 °C, the increased liquid-phase fraction promotes grain-boundary wetting and microstructural coarsening during SLPS. Concurrent grain growth may redistribute residual pores toward grain boundaries, resulting in the formation or enlargement of intergranular pores [31,32]. These pores can disrupt effective load-transfer paths, thereby contributing to the reduction in mechanical strength. Additionally, the excessive dissolution of M_7C_3 carbides and their subsequent reprecipitation as a carbide network along grain boundaries may deteriorate grain boundary strengthening, resulting in a hardness of 515.94 HV and a tensile strength of 697.71 MPa, both lower than those observed at 1,250 °C. These phenomena are consistent with the coarsened grain structure depicted in Fig. 11(e).

Based on these findings, this study establishes that the sintering densification of AISI D2 tool steel produced via BJT is most effective within a temperature range of 1,240 °C to 1,250 °C. Notably, 1,250 °C provides optimal sintering conditions, as demonstrated by measurements of relative density, shrinkage, and microstructural uniformity. Conversely, at 1,260 °C, excessive SLPS induces grain coarsening and carbide network formation; therefore, precise temperature regulation is essential [28,29,31,33].

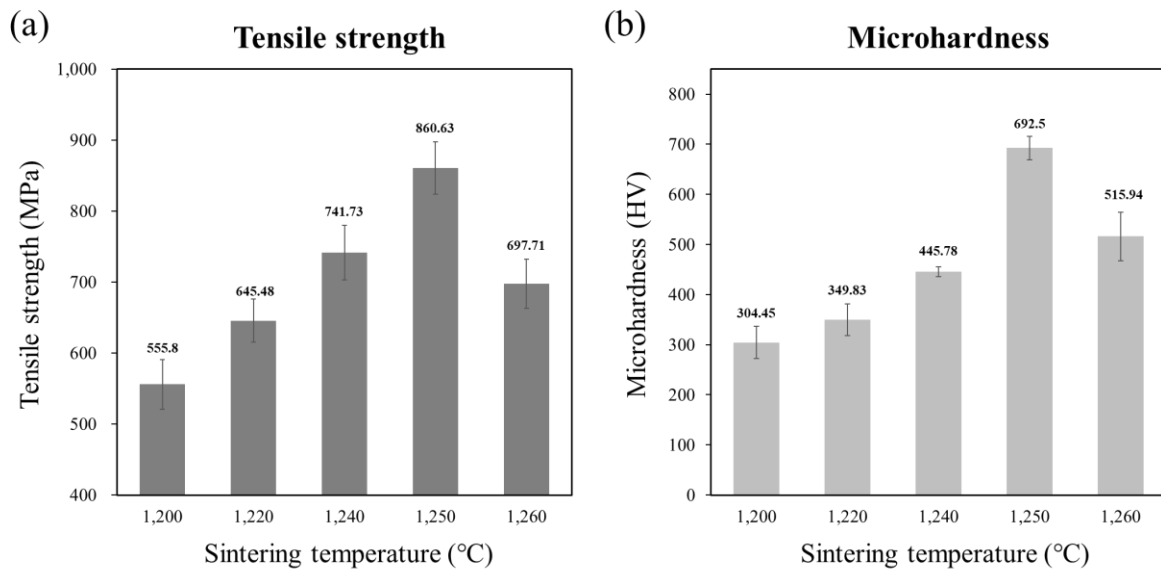


Fig. 10 Effects of sintering temperatures on mechanical behavior of AISI D2 tool steel: (a) tensile strength, (b) microhardness at different sintering temperatures

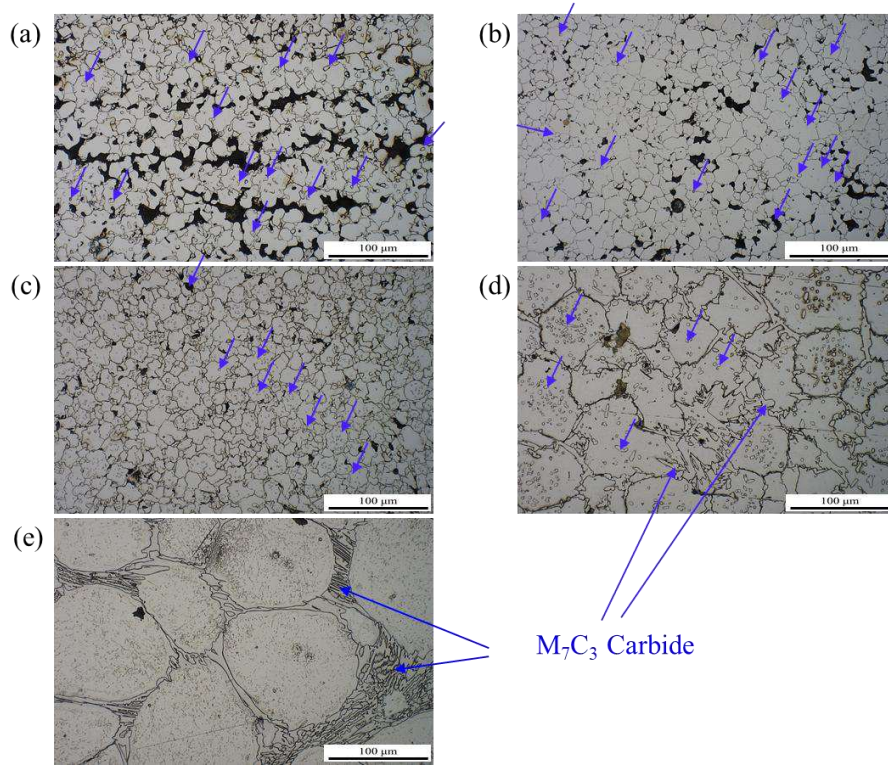


Fig. 11 Microstructures observed at 700x magnification according to sintering temperatures: (a) 1,200 °C, (b) 1,220 °C, (c) 1,240 °C, (d) 1,250 °C, (e) 1,260 °C

3.2 Heat treatment behavior

Fig. 12 compares the stress-strain curves of specimen sintered at 1,250 °C before and after heat treatment. Table 3 summarizes the mechanical properties of the as-sintered and heat-treated specimens. The as-sintered specimen exhibited a tensile strength of 860.63 MPa, an elongation of 1.28%, and a hardness of 692.5 HV. It showed little plastic deformation after yielding and rapidly fractured, indicative of brittle behavior. In contrast, the heat-treated specimen exhibited significantly improved tensile strength of 1,300.23 MPa, an elongation of 6.81%, and a hardness of 720.85 HV, demonstrating typical plastic deformation behavior with sustained work hardening after yielding. The measured hardness was comparable to literature values reported for heat-treated wrought D2, whereas the tensile strength remained lower. Nevertheless, the tensile strength and hardness achieved in this study fall within the range required for industrial cold-work applications, where high hardness and wear resistance are the primary performance criteria.

This variability in mechanical properties can be ascribed to differences in phase composition and microstructural stability. Initially, during the SLPS at 1,250 °C, a substantial portion of the M_7C_3 carbides dissolves into the austenite matrix, resulting in increased levels of carbon and chromium within the matrix. However, due to the absence of rapid cooling during sintering, the transformation of the austenite matrix into martensite is delayed, rendering it stable at ambient temperature [30,34-36]. Although this austenitic structure exhibits a relatively low dislocation density and high ductility, the segregation of alloying elements at grain boundaries may weaken grain boundary strength and facilitate localized brittle behavior. Furthermore, although the sintering process effectively reduces the majority of pores, residual micropores that have not achieved full densification can act as initiation sites for cracks, aligning with the observed rapid failure after yielding [37]. Hence, the limited elongation noted in the as-sintered specimens suggests that intergranular embrittlement or stress concentrations around micropores impede strain accumulation.

In contrast, the heat-treated specimens experienced homogenization of the base composition during the austenitizing process, followed by the transformation of the high-carbon, high-chromium austenite matrix into martensite during air cooling, thereby contributing to enhanced strength [38-40]. Additionally, fine ϵ -carbide precipitates are presumed to form during the 200 °C tempering process, stabilizing the internal deformation field of the martensite developed during cooling [41,42]. This stabilization probably contributed to the initiation of work-hardening behavior and the increased elongation. The modest increase in hardness to 720.85 HV after heat treatment can also be regarded as an indirect indicator of the expected martensitic transformation. Overall, the augmented strength and elongation observed in the heat-treated specimens are likely attributable to the combined effects of the phase transformation (austenite to martensite), internal strain stabilization (via tempering), and microstructural homogenization.

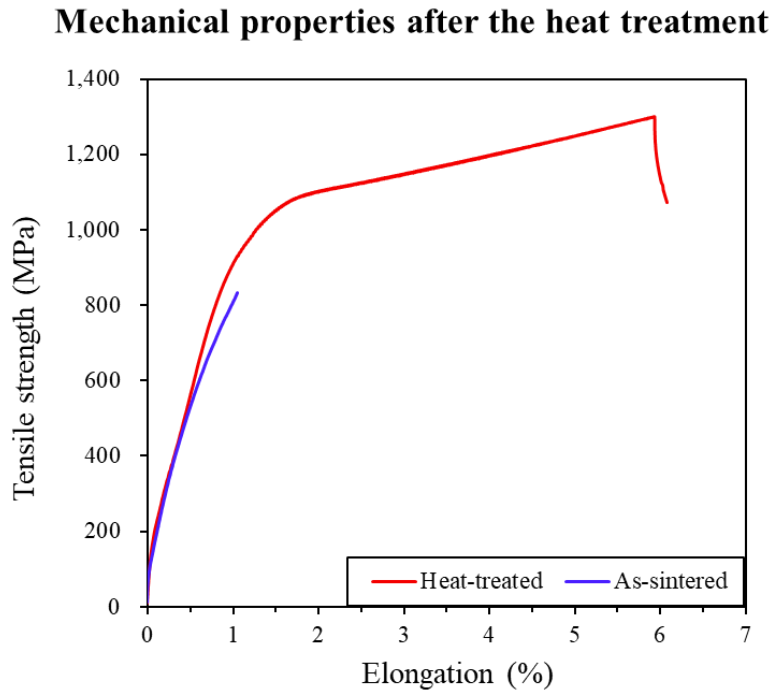


Fig. 12 Tensile stress-strain curve of AISI D2 tool steel fabricated by BJT: as-sintered specimen (sintered at 1,250 °C) versus heat-treated specimen (1,010 °C austenitizing – air cooling – 200 °C tempering)

Table 3. Mechanical properties of AISI D2 tool steel fabricated by BJT and conventional wrought processing before and after heat treatment [2,28]

Specimen	Tensile strength	Elongation	Microhardness
	MPa	%	HV
As-sintered (BJT)	860.63	1.28	692.5
Heat-treated (BJT)	1,300.23	6.81	720.85
Wrought D2 [28]	768.82	20.97	238.51
Heat-treated (Wrought D2) [2]	1,608	5.66	710

Fig. 13 depicts the microstructural differences between the as-sintered and heat-treated specimens, as confirmed through EBSD analyses. The phase map of the as-sintered sample indicates that 41.8% of the matrix consists of

the FCC phase, primarily corresponding to retained austenite (RA). The austenite matrix forms grains with relatively uniform shapes; however, the carbide clusters are heterogeneously distributed. The overall austenitic matrix exhibits a random grain orientation in the inverse pole figure (IPF) map, suggesting that diffusion-driven crystallization dominated during the sintering process. The kernel average misorientation (KAM) map illustrates a heterogeneous intragranular distribution of misorientations, indicative of residual lattice strain within the matrix after sintering. This heterogeneity reflects localized strain fields at carbide-matrix interfaces and microscale strains originating from concentration gradients of alloying elements.

Conversely, the heat-treated specimen exhibits a distinct microstructural transformation. The phase map demonstrates that 95.6% of the matrix adopts a BCC structure, consistent with a near-complete austenite to martensite transformation following austenitizing; however, since EBSD phase mapping alone does not distinguish ferrite from martensite, this interpretation is further supported by the IPF map, which displays a characteristic mosaic orientation pattern typical of martensitic lath or block structures, together with the concurrent increase in hardness. Taken together, these observations suggest that the equiaxed morphology of the original austenitic matrix was disrupted and reorganized into a martensite-like structure. Additionally, analysis of the grain boundary map reveals an increase in low-angle grain boundaries (LAGB) from 16.9% to 45.6%, while high-angle grain boundaries (HAGB) decreased from 78.4% to 50.5%. This substantial shift implies that the initial austenitic grain boundaries were largely eliminated, and new boundaries associated with martensite packets and blocks were formed during heat treatment. The KAM map of the heat-treated specimen exhibits a more homogeneous distribution of KAM values, suggesting that internal stresses were homogenized following phase transformation and subsequent tempering.

The redistribution of carbides represents a noteworthy microstructural modification induced by the heat treatment. The Cr-rich regions, which predominantly existed as coarse carbide clusters in the as-sintered state, were effectively fragmented and redistributed. Following heat treatment, these carbides were discontinuously reprecipitated along the grain boundaries and within the grains as fine, globular particles, rather than continuous films. This redistribution is attributed to metallurgical process involving the partial dissolution of carbides during the austenitizing stage, followed by their reprecipitation into more stable sizes and morphologies during the subsequent quenching and 200 °C tempering [41,42]. The 200 °C tempering stage is essential, as it facilitates the partial relaxation of internal residual stresses resulting from martensite formation, thereby reducing lattice strain. The effect of this stress relaxation was evident in the KAM maps, where the intragranular KAM value distribution became markedly more uniform compared to the as-sintered condition. This uniformity indicates that the internal deformation field was homogenized rather than remaining localized in specific regions. Concurrently, fine carbon precipitation, likely in the form of ϵ -carbides, is hypothesized. Since ϵ -carbides are typically nanoscale and fall below the resolution limit of EBSD analysis, this interpretation relies on observed changes in bulk mechanical properties and the aforementioned uniform KAM distribution.

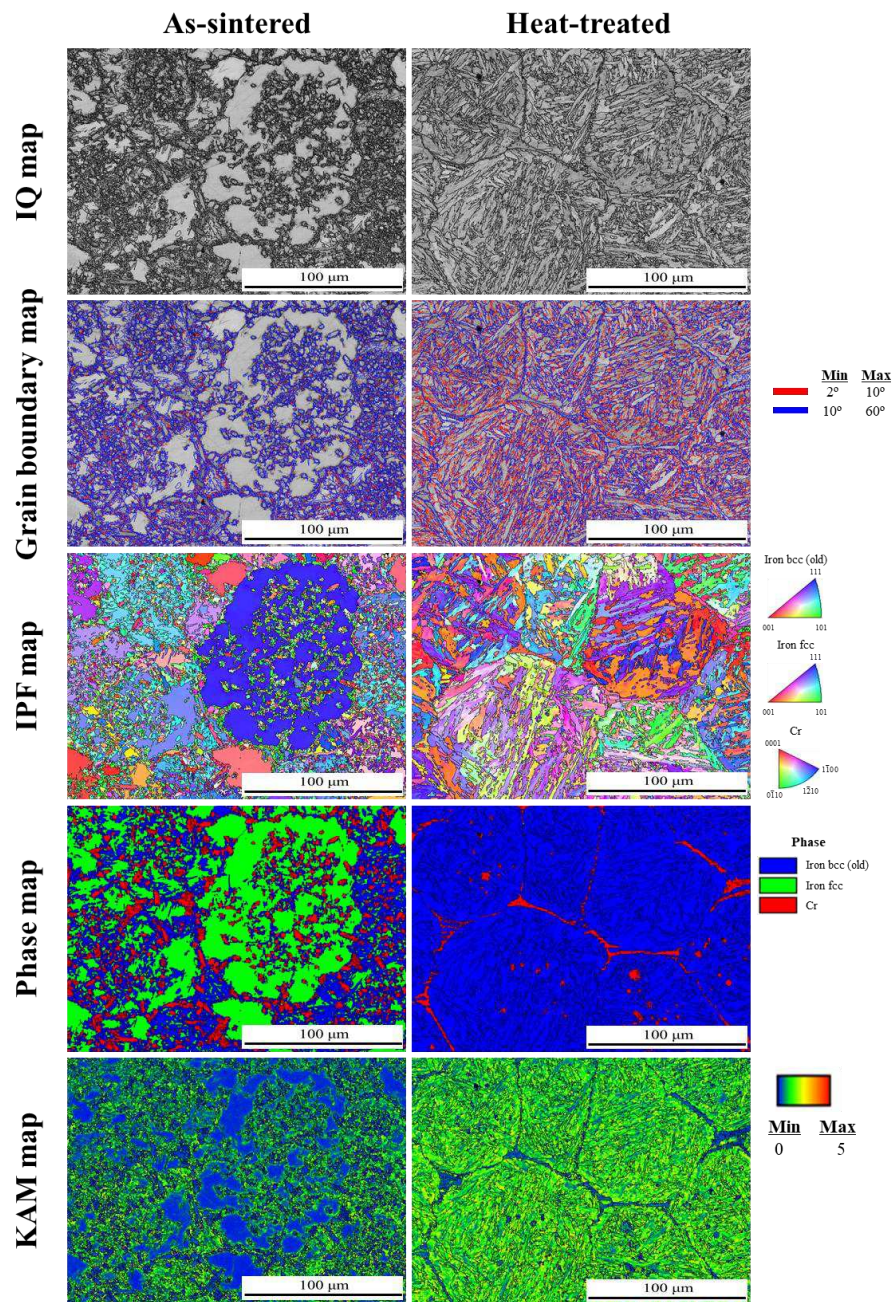


Fig. 13 EBSD analysis of as-sintered(upper) and heat-treated(lower): IQ map, Grain boundary map, IPF map, Phase map, KAM map

4. Conclusion

This investigation systematically examined the effects of sintering temperature (1,200-1,260 °C) and subsequent heat treatment, including austenitizing at 1,010 °C, air-cooling, and tempering at 200 °C, on the microstructure and mechanical properties of AISI D2 tool steel produced via BJT. The sintering temperature plays a crucial role in governing the dissolution behavior of M_7C_3 carbides, austenite stability, and pore architecture, all of which directly affect the final mechanical properties. Sintering temperatures ranging from 1,200 °C to 1,240 °C resulted in incomplete carbide dissolution, leading to microstructural heterogeneity, reduced hardness, and limited tensile strength. Among the investigated conditions, 1,250 °C was identified as the optimal sintering temperature, facilitating substantial dissolution and redistribution of M_7C_3 , stabilization of the high-carbon, high-chromium austenitic matrix, rapid achievement of near full densification, and substantial reduction in residual porosity. Accordingly, the specimen sintered at this temperature demonstrated superior mechanical properties, including a hardness of 692.5 HV and a tensile strength of 860.63 MPa. In contrast, sintering at 1,260 °C was identified as an over-sintering condition, as excessive liquid phase formation promoted grain coarsening and the evolution of resolidified carbide films and micropores along grain boundaries, thereby degrading strength and hardness relative to the 1,250 °C condition.

Subsequent heat treatment significantly enhanced mechanical performance, with tensile strength increasing to 1,300.23 MPa and elongation improving to 6.81%. EBSD analysis, together with the observed increase in hardness, indicated that austenitizing at 1,010 °C followed by air cooling results in a near-complete transformation from austenite to martensite, with the BCC fraction reaching 95.6%. Grain boundary mapping revealed the formation of a characteristic martensitic microstructure, with LAGB fraction increasing from 16.9% to 45.6%, and HAGB fraction decreasing from 78.4% to 50.5%. Furthermore, tempering at 200 °C homogenized the intragranular KAM distribution, indicating a reduced internal lattice strain and stabilization of tempered martensite. This study quantitatively elucidated the process-structure-property relationship of regulating AISI D2 tool steel fabricated via BJT.

In conclusion, the most effective processing route involves sintering at 1,250 °C followed by conventional heat treatment, achieving optimal balance between strength and ductility. These findings underscore the importance of precise temperature control for managing SLPS reactions and ensuring reliable mechanical performance of AISI D2 tool steels.

Data Availability

The data supporting the findings of this study are available from the corresponding author upon reasonable request.

Ethics Declarations

Conflict Interests The authors declare no competing interests.

Funding Declaration

This research did receive fundings.

References

- [1] bin Abdul Rahim MAS, bin Minhat M, Hussein NISB, bin Salleh MS (2018) A comprehensive review on cold work of AISI D2 tool steel. *Metall Res Technol* 115(1):104. <https://doi.org/10.1051/metal/2017048>
- [2] Torkamani H, Raygan S, Rassizadehghani J (2014) Comparing microstructure and mechanical properties of AISI D2 steel after bright hardening and oil quenching. *Mater Des* 54:1049-1055. <https://doi.org/10.1016/j.matdes.2013.09.043>
- [3] Uddeholms AB (2019) Uddeholm Sverker 21 technical datasheet edition 13. Uddeholms AB. https://www.uddeholm.com/app/uploads/sites/216/productdb/api/tech_uddeholm-sverker-21_en.pdf. Accessed 13 October 2025
- [4] Moheimani K et al (2025) Enhancing the mechanical properties and corrosion resistance of AISI 440C martensitic stainless steel via laser powder bed fusion and post-processing treatment. *Prog Addit Manuf* 10:11405-11424. <https://doi.org/10.1007/s40964-025-01294-z>
- [5] Gaitonde VN, Karnik SR, Maciel CHA, Rubio JCC, Abrão AM (2016) Machinability evaluation in hard milling of AISI D2 steel. *Mater Res*. 19(2):360-369. <https://doi.org/10.1590/1980-5373-MR-2015-0263>
- [6] Hosseini A, Hussein HM, Kishawy HA (2016) On the machinability of die/mold D2 steel material. *Int J Adv Manuf Technol* 85(1):735-740. <https://doi.org/10.1007/s00170-015-7971-8>
- [7] Koshy P, Dewes RC, Aspinwall DK (2002) High speed end milling of hardened AISI D2 tool steel (~58 HRC). *J Mater Process Technol* 127(2):266-273. [https://doi.org/10.1016/S0924-0136\(02\)00155-3](https://doi.org/10.1016/S0924-0136(02)00155-3)
- [8] Tang L, Gao C, Huang J, Zhang H, Chang W (2013) Dry sliding friction and wear behaviour of hardened AISI D2 tool steel with different hardness levels. *Tribol Int* 66:165-173. <https://doi.org/10.1016/j.triboint.2013.05.006>
- [9] Gibson I, Rosen D, Stucker B, Khorasani M (2021) Additive manufacturing technologies. Springer
- [10] Ankara A, Ari HB (1996) Determination of hot crack susceptibility in various kinds of steels. *Mater Des* 17(5-6):261-265. [https://doi.org/10.1016/S0261-3069\(97\)00019-8](https://doi.org/10.1016/S0261-3069(97)00019-8)
- [11] Odebiyi OS, Adedayo SM, Tunji LA, Onuorah MO (2019) A review of weldability of carbon steel in arc-based welding processes. *Cogent Eng* 6(1):1609180. <https://doi.org/10.1080/23311916.2019.1609180>
- [12] Lim S-H et al (2021) Hot cracking behavior of additively manufactured D2 steel. *Mater Charact* 178:111217. <https://doi.org/10.1016/j.matchar.2021.111217>
- [13] Yadav S et al (2023) Laser directed energy deposition of high-carbon high-chromium D2 tool steel structures: processing, heat treatment and material behavior. *J Mater Eng Perform* 32(11):4881-4891. <https://doi.org/10.1007/s11665-022-07470-4>
- [14] Omar SMT, Plucknett KP (2022) The influence of DED process parameters and heat-treatment cycle on the microstructure and hardness of AISI D2 tool steel. *J Manuf Process* 81:655-671. <https://doi.org/10.1016/j.jmapro.2022.06.069>
- [15] Ziaee M, Crane NB (2019) Binder jetting: A review of process, materials, and methods. *Addit Manuf* 28:781-801. <https://doi.org/10.1016/j.addma.2019.05.031>
- [16] Li M, Du W, Elwany A, Pei Z, Ma C (2020) Metal binder jetting additive manufacturing: a literature review. *J Manuf Sci Eng* 142(9):090801. <https://doi.org/10.1115/1.4047430>
- [17] Mostafaei A et al (2021) Binder jet 3D printing—Process parameters, materials, properties, modeling, and challenges. *Prog Mater Sci* 119:100707. <https://doi.org/10.1016/j.pmatsci.2020.100707>
- [18] Tancogne-Dejean T, Roth CC, Mohr D (2021) Rate-dependent strength and ductility of binder jetting 3D-printed stainless steel 316L: experiments and modeling. *Int J Mech Sci* 207:106647. <https://doi.org/10.1016/j.ijmecsci.2021.106647>

- [19] Nandwana P, Kannan R, Siddel D (2020) Microstructure evolution during binder jet additive manufacturing of H13 tool steel. *Addit Manuf* 36:101534. <https://doi.org/10.1016/j.addma.2020.101534>
- [20] Choudhari A et al (2024) Additive manufacturing of AISI M2 tool steel by binder jetting (BJ): Investigation of microstructural and mechanical properties. *J Manuf Process* 132:686-711. <https://doi.org/10.1016/j.jmapro.2024.11.008>
- [21] Rodríguez V, Shang N, Dahmen T, Somers MAJ, Nadimpalli VK (2025) Effect of sintering parameters on additively manufactured D2 cold work tool steel via binder jetting of powder feedstock obtained from spray forming. *Euro PM2025 Congress & Exhibition. European Powder Metallurgy Association*. <https://doi.org/10.59499/EP256779530>
- [22] Verdiere A et al (2020) Effect of the austenitizing parameters on the microstructure and mechanical properties of 75Cr1 tool steel. *Mater Sci Eng A* 785:139331. <https://doi.org/10.1016/j.msea.2020.139331>
- [23] Ley R (2022) Binder jet printing of a low-cost tool steel powder. Dissertation, Dalhousie University
- [24] Bombac D, Fazarinc M, Podder AS, Kugler G (2013) Study of carbide evolution during thermo-mechanical processing of AISI D2 tool steel. *J Mater Eng Perform* 22(3):742-747. <https://doi.org/10.1007/s11665-012-0340-y>
- [25] Farayibi PK, Blüm M, Weber S (2020) Hard cladding by supersolidus liquid phase sintering: An experimental and simulation study on martensitic stainless steels. *Metall Mater Trans A* 51:5818-5835. <https://doi.org/10.1007/s11661-020-05953-4>
- [26] Huber D, Vogel L, Fischer A (2021) The effects of sintering temperature and hold time on densification, mechanical properties and microstructural characteristics of binder jet 3D printed 17-4 PH stainless steel. *Addit Manuf* 46:102114. <https://doi.org/10.1016/j.addma.2021.102114>
- [27] Zago M, Lecis N, Mariani M, Cristofolini I (2024) Analysis of the causes determining dimensional and geometrical errors in 316L and 17-4PH stainless steel parts fabricated by metal binder jetting. *Int J Adv Manuf Technol* 132(1):835-851. <https://doi.org/10.1007/s00170-024-13437-7>
- [28] Jeon M-S et al (2025) High-strength D2 tool steel via material extrusion additive manufacturing and post-heat treatment. *Mater Des* 254:114070. <https://doi.org/10.1016/j.matdes.2025.114070>
- [29] Röttger A et al (2025) Additive manufacturing of carbon-martensitic hardening ledeburitic cold work tool steels using fused filament fabrication and subsequent supersolidus liquid-phase sintering. *Prog Addit Manuf* 10(1):679-700. <https://doi.org/10.1007/s40964-024-00650-9>
- [30] Grgac P et al (2004) Thermal stability of metastable austenite in rapidly solidified chromium–molybdenum–vanadium tool steel powder. *Mater Sci Eng A* 375:581-584. <https://doi.org/10.1016/j.msea.2003.10.036>
- [31] German RM (1997) Supersolidus liquid-phase sintering of prealloyed powders. *Metall Mater Trans A* 28(7):1553-1567. <https://doi.org/10.1007/s11661-997-0217-0>
- [32] Ghasemi S et al (2020) The role of pore evolution during supersolidus liquid phase sintering of prealloyed brass powder. *Powder Metall* 63(3):187-196. <https://doi.org/10.1080/00325899.2020.1785155>
- [33] Muro P, Gimenez S, Iturriza I (2002) Sintering behaviour and fracture toughness characterization of D2 matrix tool steel, comparison with wrought and PM D2. *Scripta Mater* 46(5):369-373. [https://doi.org/10.1016/S1359-6462\(01\)01253-2](https://doi.org/10.1016/S1359-6462(01)01253-2)
- [34] Yang M, Keshavarz MK, Vlasea M, Molavi-Kakhki A, Laher M (2023) Supersolidus liquid phase sintering of water-atomized low-alloy steel in binder jetting additive manufacturing. *Heliyon* 9(3):e13882. <https://doi.org/10.1016/j.heliyon.2023.e13882>
- [35] Liu Z-B et al (2020) Carbide dissolution and austenite grain growth behavior of a new ultrahigh-strength stainless steel. *J Iron Steel Res Int* 27(6):732-741. <https://doi.org/10.1007/s42243-020-00429-6>

- [36] Koistinen DP, Marburger RE (1959) A general equation prescribing the extent of the austenite-martensite transformation in pure iron-carbon alloys and plain carbon steels. *Acta Metall* 7:59-60.
[https://doi.org/10.1016/0001-6160\(59\)90170-1](https://doi.org/10.1016/0001-6160(59)90170-1)
- [37] German RM (1996) *Sintering theory and practice*. Wiley
- [38] Algarni M (2019) Mechanical properties and microstructure characterization of AISI “D2” and “O1” cold work tool steels. *Metals* 9(11):1169. <https://doi.org/10.3390/met9111169>
- [39] Salunkhe S, Fabijanic D, Nayak J, Hodgson P (2015) Effect of single and double austenitization treatments on the microstructure and hardness of AISI D2 tool steel. *Mater Today Proc* 2(4-5):1901-1906.
<https://doi.org/10.1016/j.matpr.2015.07.145>
- [40] Stamenković U, Marković I, Manasijević D, Milanović P, Nedeljković M (2024) The influence of heat treatment on mechanical, thermal, and structural properties of AISI D2 steel. *Kovove Mater* 62(6).
<https://doi.org/10.31577/km.2024.6.327>
- [41] Speich GR, Leslie WC (1972) Tempering of steel. *Metall Trans* 3(5):1043-1054.
<https://doi.org/10.1007/BF02642436>
- [42] Ohmori Y, Tamura I (1992) Epsilon carbide precipitation during tempering of plain carbon martensite. *Metall Trans A* 23(10):2737-2751. <https://doi.org/10.1007/BF02651753>