

Review

LPBF AlSi10Mg at the Nanoscale: A Critical Review of Processing–Microstructure–Property Correlations via Nanoindentation

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Abstract

Laser Powder Bed Fusion (LPBF)-processed AlSi10Mg produces highly heterogeneous microstructures, where fine α -Al cells, Si-rich networks, and melt-pool boundaries govern local mechanical behavior. Nanoindentation has emerged as a key tool for probing these variations, yet systematic understanding of the links between processing parameters, microstructure, and nano-mechanical response remains limited. This critical review examines how laser processing parameters influence local mechanical response through their impact on microstructural features. Key challenges in interpreting nanoindentation are highlighted, alongside inconsistencies in experimental protocols and reporting practices that hinder cross-study comparisons. Beyond summarizing existing findings, underexplored aspects of nanoindentation in LPBF AlSi10Mg are identified, including spatially correlated microstructure-mechanical mapping, depth-resolved measurements, and integration with advanced characterization and data-driven approaches. By synthesizing current knowledge and clarifying methodological constraints, this review positions nanoindentation not merely as a descriptive tool, but as a mechanistically informed approach for linking processing conditions, microstructural heterogeneity, and local mechanical response. These insights aim to support more rigorous interpretation of small-scale mechanical data and to guide future studies toward predictive understanding and rational process optimization in additively manufactured aluminum alloys.

Keywords: AlSi10Mg; Laser Powder Bed Fusion (LPBF); nanoindentation; mechanical properties; process–structure–property; hardness; elastic modulus; heat treatment; volumetric energy density (VED); microstructural heterogeneity



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1. Introduction

Lightweight aluminum alloys play a key role in modern industrial sectors such as aerospace and automotive because they offer high strength, corrosion resistance, and recyclability [1–3]. Among these alloys, Laser Powder Bed Fusion (LPBF)-processed AlSi10Mg has emerged as a benchmark material, as it effectively combines the advantages of the alloy's intrinsically well-balanced composition with the unique benefits offered by the LPBF process. The LPBF technique enables near-net-shape fabrication of geometrically complex components with minimal material waste, while the Si addition in AlSi10Mg composition

improves laser absorptivity and melt-pool fluidity, facilitating stable melting and reducing solidification cracking [4–9].

Rapid solidification during LPBF, with cooling rates as high as 10^6 K/s, suppresses segregation and produces a distinctive nonequilibrium microstructure characterized by a refined cellular α -Al matrix surrounded by a continuous Si-rich eutectic network [9–11]. This hierarchical microstructure, defined by melt-pools, grains, and sub-grains, makes LPBF-processed AlSi10Mg a model material for studying process–structure–property relationships in metallic additive manufacturing (AM).

This refined architecture critically influences mechanical behavior, resulting in high yield strength and hardness through microstructural constraint, solid-solution strengthening, and Si network reinforcement [12,13]. Yet, the same thermal and microstructural gradients that enable superior bulk performance also introduce pronounced local heterogeneity [14]. Variations in melt-pool morphology, residual stress state, and Si network continuity generate spatial differences in mechanical response at the micro- and nano-scale [15]. At these length scales, local phase distribution, grain boundary structure, and dislocation behavior facilitate the evolution of thermal stresses during fabrication, ultimately influencing the mechanical reliability of the printed component [16–20]. Conventional mechanical testing averages responses over these fluctuations, hiding the distinct roles of the α -Al cells, Si-enriched networks, and melt-pool boundaries on the localized mechanical response [21,22]. Consequently, methods capable of probing mechanical behavior with sub-micrometer resolution are required to resolve these localized effects and to establish meaningful links between processing conditions and mechanical performance.

Nanoindentation offers a quantitative way of probing the local mechanical response in LPBF-processed AlSi10Mg. By applying controlled loads through a sharp indenter while continuously recording penetration depth, the technique provides high spatial resolution data on hardness, elastic modulus, and time-dependent deformation within volumes comparable to individual microstructural features [23]. This localized approach has been successfully applied to a variety of Al alloys, enabling resolution of mechanical contrasts between distinct regions and assessment of how spatial variations in solidification conditions or phase distribution influence alloy performance [24–26].

Over the past decade, nanoindentation studies on LPBF-processed AlSi10Mg have provided valuable insight into how microstructural refinement, post-processing heat treatments, and localized compositional variations influence the measured hardness and modulus [15,17,27–29]. However, due to substantial variation in processing conditions, testing protocols, and analytical methodologies across studies, it remains challenging to identify consistent trends that link processing parameters, microstructure, and small-scale mechanical response. Although several reviews have addressed the bulk mechanical behavior of AlSi10Mg and other additively manufactured Al alloys fabricated, none has focused specifically on the nanoscale mechanical response revealed through nanoindentation [30–34]. Existing surveys predominantly emphasize tensile, fatigue, and fracture properties, which represent the bulk performance of the alloys but provide limited insight into the localized mechanical phenomena [35–38].

This review therefore addresses this gap by providing the first focused and comprehensive assessment of nanoindentation-based research for LPBF-processed AlSi10Mg. Its central objective is to establish a coherent framework connecting LPBF processing conditions, resulting structural characteristics, and small-scale mechanical behavior. To place these observations on a physically meaningful basis, particular attention is given to how LPBF processing conditions affect defect formation and microstructural development. By systematically examining reported methodologies, experimental results, and interpretative approaches, this work identifies consistent trends, highlights unresolved discrepancies, and

clarifies the capabilities and limitations of nanoindentation for probing small-scale mechanical behavior in LPBF AlSi10Mg alloys. Through the integration of established findings and underexplored aspects of nanoscale mechanics, this review advances nanoindentation from a primarily descriptive technique toward a mechanistically grounded method for process optimization, materials qualification, and the design of next-generation lightweight components for AM.

2. LPBF Processing of AlSi10Mg

The microstructure and density of LPBF-processed AlSi10Mg is defined by the processing conditions under which it is fabricated and subsequently treated. Variations in parameter selection and thermal exposure lead to distinct microstructural configurations and varied densification, which form the basis for the mechanical behavior examined in later sections. Understanding these processing–microstructure relationships is a necessary foundation before considering nanoscale mechanical response.

2.1. Processing Parameters, Densification, and Defect Formation

The outcome of LPBF builds is governed by a combination of primary process parameters, such as laser power, scanning speed, hatch spacing, and layer thickness, which directly determine densification and microstructure evolution [39,40]. Secondary factors such as scan strategy, spot size, and powder characteristics further influence melt-pool behavior and defect formation [41,42]. Together, these parameters define the thermal conditions during fabrication, making an understanding of their roles and interactions essential. A summary of the processing parameters influencing the LPBF-manufactured AlSi10Mg parts quality is schematically illustrated in Figure 1.

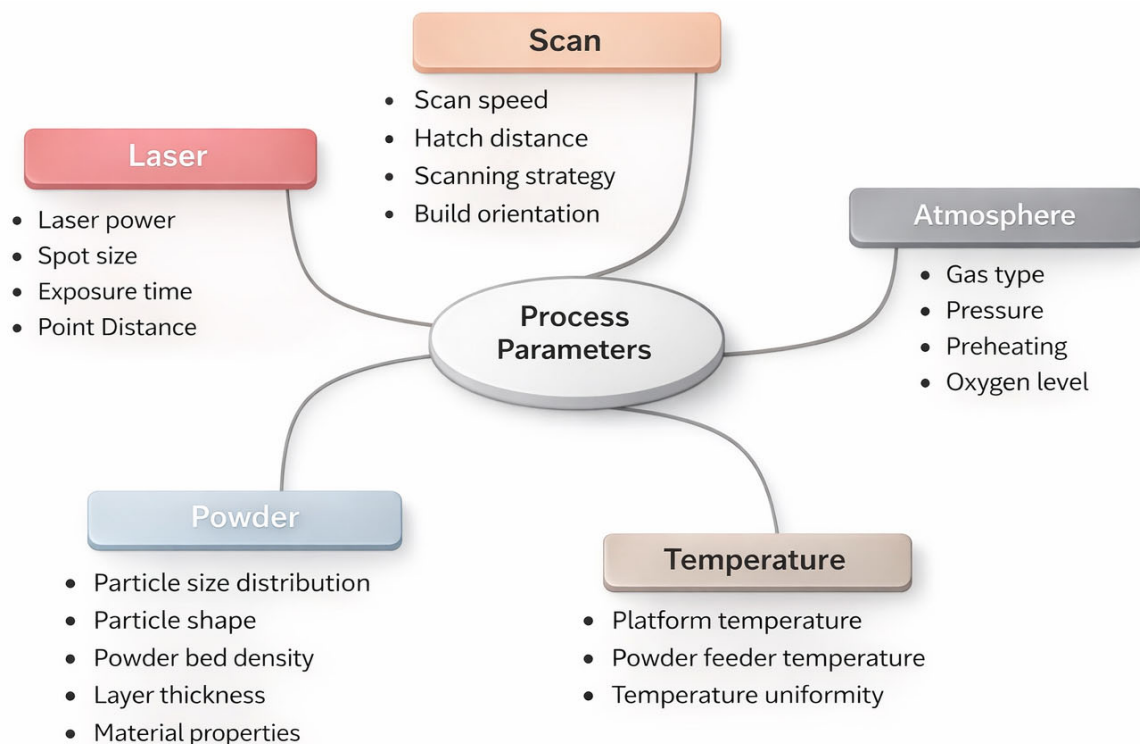


Figure 1. Schematic illustration of the LPBF process parameters and powder characteristics governing melt-pool behavior, densification, and microstructure evolution.

To rationalize the combined effect of the primary LPBF parameters, many studies employ the volumetric energy density (VED) as a single processing descriptor, defined as

$$VED = \frac{P}{v h t}$$

where P (W) is the laser power, v (mm/s) is the scanning speed, h (mm) is the hatching space, and t (mm) is the layer thickness [43,44]. VED represents the total thermal energy input per unit volume of material melted, making it a convenient descriptor for comparing process conditions across different systems [45,46].

The majority of AlSi10Mg LPBF studies identify optimal processing conditions in processing regimes that are consistently associated with a minimal concentration of manufacturing defects, yielding relative densities exceeding 99% [15,47–60]. To illustrate this trend, literature data for LPBF-manufactured AlSi10Mg with relative densities above 95% were collected and analyzed based on the reported processing parameters, and the calculated VED values are plotted as a function of relative density (Figure 2). As can be seen, the high-density (>99%) points lie within a VED range of approximately 40–70 J/mm³, while only a limited number of high-density data points appear at higher (>70 J/mm³) or lower VED values (<40 J/mm³). Outside this range, insufficient energy input leads to lack-of-fusion porosity (LoF) porosity due to incomplete melting and poor track bonding, whereas excessive energy input promoted deep and unstable melt pools, increasing the likelihood of keyhole-induced porosity [61–63].

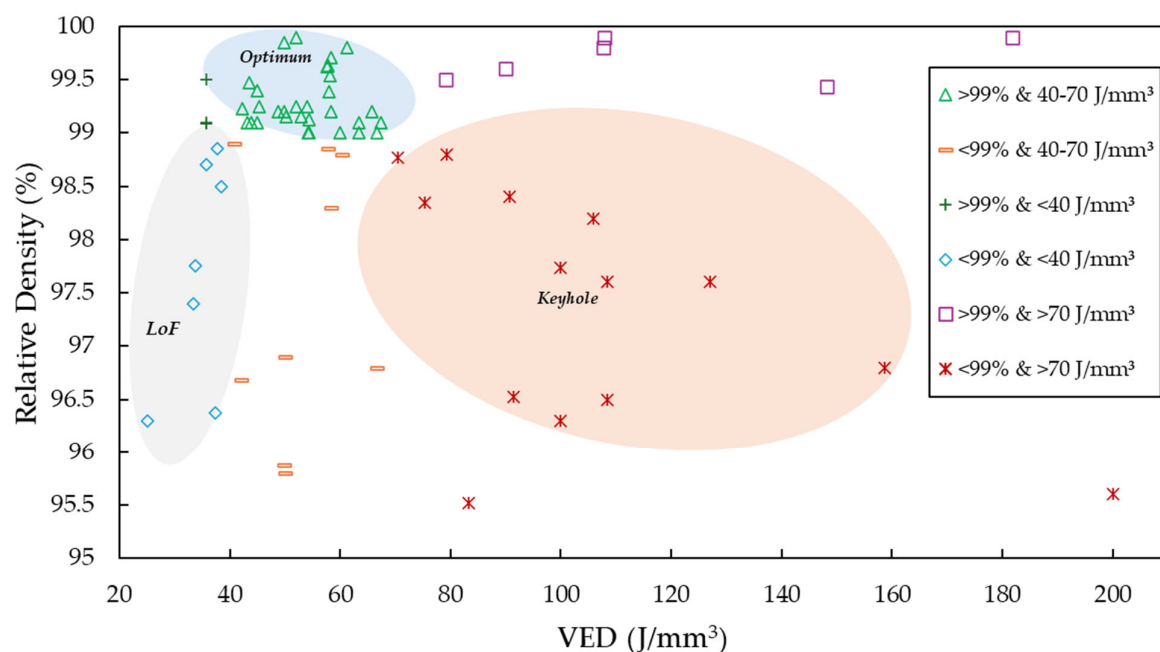


Figure 2. Relative density as function of calculated VED for LPBF-processed AlSi10Mg. Data are grouped by relative density (>99% or <99%) and VED range: >99% and 40–70 J/mm³ (green triangles [14,15,48–50,55,56,59,64–68]), <99% and 40–70 J/mm³ (orange dashes [50,64,69–71]), >99% and <40 J/mm³ (green plus symbols [54,57,58]), <99% and <40 J/mm³ (blue diamonds [64,69,72]), >99% and >70 J/mm³ (purple squares [47,51–54,60]), and <99% and >70 J/mm³ (red crosses [50,64,69,71,73]).

Although VED provides a convenient framework for comparing processing conditions across different studies and machines, it does not uniquely define the thermal history of the material. Different combinations of laser power, scanning speed, hatch spacing, and layer thickness can yield the same nominal VED while producing distinct melt-pool geometries.

As a result, identical VED values may correspond to either fully dense material or significant porosity, depending on the specific parameter set employed. For example, parameter sets with low laser power and slow scan speed (100 W, 500 mm/s, 0.1 mm hatch, and 0.04 mm layer) have been shown to yield relative densities of only 95.8%, while high-power, high-speed (370 W, 1300 mm/s, 0.19 mm hatch, and 0.03 mm layer) combinations at the same nominal VED can produce more stable melt pools and near-full density [51,69,74,75].

Additional variations in reported density values at similar VEDs arise from differences in powder characteristics and density measurement techniques. For example, Mohan et al. demonstrated that density values obtained using the Archimedes method are systematically lower than those measured by micro-XCT, primarily because micro-XCT cannot resolve fine gas porosity below 20 μm , which is typical of near-fully dense LPBF AlSi10Mg [76,77]. Furthermore, Riener et al. demonstrated that powders with different particle size distributions and morphologies, processed using identical LPBF parameters, produced AlSi10Mg samples with relative densities ranging from 99.84% to 99.99%, underscoring the influence of powder characteristics even at constant VED [78].

Beyond densification, VED also exerts a first-order influence on microstructural scale and chemical partitioning in LPBF-processed AlSi10Mg. During LPBF, AlSi10Mg experiences steep thermal gradients, directional heat transfer, and repeated remelting as successive layers are deposited. Solidification occurs at extremely high cooling rates [79], leading to the formation of fine cellular or cellular-dendritic microstructures with metastable phases [80]. Increasing VED generally reduces cooling rates and thermal gradients, promoting grain growth [80,81]. Systematic evolution from fine cellular microstructures at low VED to progressively coarser cellular-dendritic morphologies has been widely reported and attributed to changes in local solidification conditions [82].

Changes in VED also affect silicon distribution within the AlSi10Mg alloy. At lower and intermediate VED values (40–80 J/mm³), rapid solidification leads to supersaturation of Si within the α -Al matrix, while excess Si segregates to form fine eutectic networks along cellular boundaries [15]. With increasing VED, higher Si solubility in the α -Al matrix has been observed, particularly in the XY plane, whereas at the highest energy densities a larger fraction of Si precipitates, resulting in a thicker and more continuous Si network [83]. These trends confirm that VED plays a critical role in controlling both grain morphology and Si partitioning during LPBF processing.

2.2. Effect of Heat Treatments on Microstructure

While LPBF parameters establish the initial cellular architecture and solute distribution through rapid solidification, subsequent heat treatments modify these features through diffusion-controlled mechanisms. Post-processing heat treatments are routinely applied to LPBF-processed AlSi10Mg to relieve residual stresses, alter silicon morphology, and control precipitation behavior [47,84,85]. Owing to the unique thermal history imposed by LPBF, the response of AlSi10Mg to heat treatment differs substantially from that of conventionally processed material [86,87]. As a result, several variations in traditional aluminum heat treatments have been explored with the primary aim of tailoring microstructural features specific to LPBF.

Current industrial practice is largely guided by standards originally developed for aluminum castings. In this context, ASTM F3318 specifies recommended post-processing routes for LPBF AlSi10Mg components based on AMS 2771, including stress-relief treatments, full T6 heat treatment, and combined HIP-T6 routes [88]. These standardized protocols provide a reference framework; however, their direct application to LPBF AlSi10Mg often leads to microstructural changes that differ from those observed in cast alloys, reflecting the distinct as-built microstructure produced by LPBF [85,89].

As already mentioned, as-built AlSi10Mg exhibits a fine cellular-dendritic α -Al matrix surrounded by a continuous, supersaturated Si-rich network formed under rapid solidification. Conventional T6 processing, involving high-temperature solution treatment followed by artificial aging, disrupts this metastable cellular structure by dissolving the interconnected Si network and promoting subsequent re-precipitation of coarser Si particles during aging [58,90,91]. Such T6 treatments lead to significant Si coarsening, accompanied by homogenization of the α -Al matrix and a reduction in microstructural anisotropy [90,92,93]. While effective in relieving residual stresses and stabilizing the microstructure, such treatments also eliminate features responsible for the characteristic as-built condition mechanical response.

To mitigate excessive coarsening, stress-relief annealing has emerged as a baseline post-processing treatment for LPBF AlSi10Mg [94,95]. Heat treatments conducted at intermediate temperatures typically around 300 °C for 2–3 h significantly reduce residual stresses while largely preserving the fine cellular structure and Si network morphology [94,96]. These treatments promote limited diffusion of solute atoms and partial precipitation of Mg-containing phases without fully dissolving the eutectic network and are often considered analogous to a T5-type temper for LPBF materials.

Recognizing the high solute supersaturation retained in the as-built α -Al matrix, several studies have explored direct aging treatments at relatively low temperatures [97,98]. Direct aging (T5-like) at 150–170 °C for 4–8 h has been shown to exploit the LPBF-induced nonequilibrium state to promote controlled precipitation while maintaining the original cellular framework [13,86,99]. However, attention should be paid when applying such treatments as higher aging temperatures or longer durations accelerate precipitation and may lead to over-aging, resulting in coarsening of the Si network [100–102].

In contrast to these lower-temperature cycles, hot isostatic pressing (HIP) and HIP-based routes are employed primarily to reduce porosity. HIP at elevated temperatures (~500–515 °C) and high pressure significantly closes internal pores and yields near-fully dense material, but the elevated temperature also erases much of the LPBF cellular structure, leading to coarse α -Al grains and micron-scale Si precipitates similar to or more homogenized than conventional solutionized structures [103,104]. Combined HIP-T6 treatments aim to both densify and then strengthen the alloy, although HIP alone already significantly alters microstructure and mechanical balance.

3. Methodological Aspects and Challenges of Nanoindentation in LPBF AlSi10Mg

In LPBF-processed AlSi10Mg, features such as melt-pool boundaries, sub-grain cellular structures, and silicon networks can influence mechanical response locally [105]. Nanoindentation allows these spatial effects to be probed with high resolution, offering insight into how processing conditions translate into variations in mechanical properties across the microstructure. However, the same LPBF-induced AlSi10Mg features that motivate the use of nanoindentation can also introduce several experimental complexities. For example, as-built surfaces often exhibit high roughness that can distort indenter contact and increase measurement scatter [106]. Consequently, several experimental factors must be carefully selected to ensure meaningful and reproducible measurements.

Among these factors, indenter geometry is worth noting, as it governs the stress volume and deformation mode during indentation [107]. The Berkovich indenter, a three-sided pyramidal diamond tip, is the most used configuration owing to its well-defined geometry and compatibility with standard Oliver and Pharr analysis [108]. Accordingly, the majority of studies have utilized this indenter type [11,17,29,58,109–114]. Modified Berkovich tips have also been introduced in certain works as they can provide additional insight into the local stress state that a conventional Berkovich tip cannot capture [115–118]. In contrast, cube-corner indenters are typically used when the objective is to examine localized fracture phenomena at low penetration depths [119]. The selection of indenter geometry is not just procedural [120]. Sharper tips induce more localized regions, whereas blunter geometries probe larger material volumes and are more likely to initiate plasticity through the motion of pre-existing dislocations [121]. As a result, differences in indenter geometry may contribute to discrepancies in hardness and modulus reported across studies, even when nominally similar testing conditions are employed.

The applied loading protocol is another factor that varies over studies. Reported maximum loads span over two orders of magnitude, from a few mN to several hundred mN, depending on the targeted scale of investigation. When probing individual phases or microstructural constituents, low loads in the range of 0.5–2 mN are applied to isolate the response from the Al matrix or eutectic Si particles [84,122]. Conversely, loads between 5 and 50 mN are more commonly applied when the goal is to capture statistically representative or bulk properties [27,47,123,124]. High-load protocols, extending up to 200–400 mN, are frequently used in creep or mapping experiments [28,125]. Continuous stiffness measurement (CSM) techniques are integrated into these protocols to acquire depth-dependent hardness and modulus data for precise evaluation of local mechanical behavior [27,115]. The wide spread of applied loads, together with the variation in other nanoindentation testing conditions reported in the literature, is summarized in Table 1, highlighting not only the diversity of testing approaches but also the lack of standardized protocols. Additionally, although not explicitly stated in all cases, it can be assumed that these measurements were conducted under ambient conditions.

Table 1. Nanoindentation testing parameters across literature.

No.	Nanoindenter Model	Indenter Type	Maximum Load (mN)	Load Rate (mN/s)	No. of Indents/Spacing (μm)	Penetration Depth (nm)	Hold Time (s)	Ref.
1	MicroMaterials LTD Nanotest NTX Platform3	Berkovich	7.5	N/A ¹	N/A/10 (horizontal) and 15 (vertical)	N/A	N/A	[47]
2	Anton Paar NHT ³ Nanoindentation	Berkovich	0.7 (AB), 0.5 (HT)	N/A	N/A	N/A	5	[122]
3	Hysitron Ti950 Triboindenter	Berkovich (for indentation), cube corner (for nano-hardness)	N/A	N/A	N/A	400 (Berkovich), 70 (cube corner)	N/A	[119]
4	Hysitron Ubi-1 Nanoindenter	Self-similar pyramidal Berkovich	9.5	1	25/400	N/A	2	[124]
5	iMicro Nanoindenter (KLA Instruments, USA)	Self-similar pyramidal Berkovich	200 (Creep test)	10	N/A	N/A	400	[125]
6	Nanoindenter (GSM)	Berkovich	30	N/A	5/N/A	N/A	15 (for indentation), 600 (for creep)	[126]
7	Keysight Nano-indenter G200	N/A	N/A	N/A	N/A	N/A	N/A	[127]
8	Micro Materials LTD. (Wrexham, UK) Nanotest TM NTX Platform 2	Berkovich	7.5	N/A	N/A/15	N/A	N/A	[110]
9	iMicro Nanoindenter (KLA Instruments, Milpitas, CA, USA)	Self-similar pyramidal Berkovich	400	10	5/N/A	N/A	2	[28]
10	NanoTest system (Micro Materials Ltd., Wrexham, UK)	Berkovich	1–200	N/A	10 for each load along a set of vertical and horizontal parallel lines with a span of 50 μm	N/A	60	[112]
11	Keysight Nano-indenter G200	Berkovich	50	N/A	N/A	1300	N/A	[27]
12	MicroMaterials LTD Nanotest NTX Platform3 nanoindenter (UK)	Berkovich	7.5	N/A	5/N/A	N/A	N/A	[29]
13	NanoTest Vantage system	Berkovich	N/A	2	36/N/A	1500	N/A	[11]
14	N/A	Three-sided pyramid Berkovich	70 mN (for the general microstructure), 2 mN (for the phase hardness)	0.5	30	N/A	N/A	[84]
15	Hysitron Ti950	Berkovich	N/A	N/A	N/A	400	N/A	[17]

Table 1. Cont.

No.	Nanoindenter Model	Indenter Type	Maximum Load (mN)	Load Rate (mN/s)	No. of Indents/Spacing (μm)	Penetration Depth (nm)	Hold Time (s)	Ref.
16	Hysitron Ubi-1 Nanoindenter	Self-similar pyramidal Berkovich	9.5	1	N/A	N/A	N/A	[85]
17	iNano instrument (Nanomechanics, Inc., USA)	N/A	5	1 indent/s	625 (in a $15\text{ }\mu\text{m} \times 15\text{ }\mu\text{m}$ area)	N/A	N/A	[128]
18	Hysitron TI 950	Berkovich	5	N/A	121/N/A	N/A	N/A	[113]
19	NanoTest Vantage (Micromaterials)	Berkovich	10	1	36 (in a $200\text{ }\mu\text{m} \times 200\text{ }\mu\text{m}$ area)	N/A	5	[58]
20	MicroMaterials Nanotest platform 3 (Wrexham, UK)	Three-sided pyramid Berkovich	7.5	0.375	308/10 (horizontal) and 15 (vertical)	N/A	5	[111]
21	PICODENTOR HM500 (Helmut Fischer, Germany)	Vickers	7.5	N/A	N/A	N/A	N/A	[129]
22	N/A	N/A	5	0.033	5/N/A	N/A	N/A	[123]
23	Anton Paar NHT ²	Berkovich	30	1	3 per zone (non-banding and banding) $\times$ 3 regions (top, middle, bottom) $\times$ 2 cooling conditions	50	2	[130]
24	NanoTest Vantage (MicroMaterials Ltd.)	Berkovich	10	1	$400 \times 200\text{ }\mu\text{m}$ mapped area with $40\text{ }\mu\text{m}$ spacing	N/A	5	[114]
25	ITT-MTS-XP	Modified Berkovich	N/A	N/A	N/A/3 (horizontal) and 10 (vertical)	250	N/A	[115]
26	Anton Paar NHT 2	Modified Berkovich	N/A	0.083	N/A	N/A	10	[116]
27	Hysitron TI980	N/A	N/A	N/A	$>100/8$	400	N/A	[131]
28	TriboScope Ti950 (Hysitron Inc., USA)	Berkovich	N/A	N/A	100/5	150	5	[132]
29	iMicro TM Nanoindenter (Nanomechanics Inc.)	Berkovich	50	N/A	81 over $240 \times 240\text{ }\mu\text{m}^2$ /N/A	N/A	N/A	[133]
30	Hysitron TS 77 Select, Bruker (Minneapolis, MN, USA)	Berkovich	N/A	N/A	100/20	1000	3	[15]
31	Hysitron TI 950 TriboIndenter	Berkovich	8	N/A	$50\text{ }\mu\text{m}$ spacing	N/A	N/A	[134]
32	Hysitron TI 950 TriboIndenter	Berkovich	5	1	5/N/A	N/A	5	[135]
33	Hysitron TI980, USA	N/A	1.4	N/A	N/A/N/A	N/A	N/A	[136]
34	Hysitron 950 TriboIndenter	Berkovich	N/A	N/A	10/N/A	2000	N/A	[137]

¹ N/A = Not available.

Because the loading protocol determines the achieved penetration depth, depth selection is crucial in defining spatial resolution and the mechanical interpretability of indentation results. Several investigations report maximum penetration depths between 200 and 1000 nm for bulk mechanical characterization [15,119,130,131]. Shallow indentation depths are advantageous for resolving gradients between melt-pool centers and boundaries, but they are also more sensitive to surface roughness and local heterogeneity that could potentially compromise measurement accuracy [132]. Deeper indentations integrate multiple microstructural features, yielding averaged mechanical responses but potentially masking local anisotropy. Consequently, proper surface finishing is an essential step for reliable data. Most studies employ multi-step mechanical polishing, followed by colloidal silica finishing, and occasionally etching to reveal microstructural features [27,47,85,129]. Even minor deviations in surface quality can significantly distort the load–displacement curve, particularly at low penetration depths, where the relative magnitude of topographical irregularities becomes comparable to indentation depth [138,139].

Time-dependent deformation introduces an additional dimension of variability that is directly governed by the chosen testing protocol. Reported hold times, which range from seconds to minutes, reflect fundamentally different experimental objectives, and therefore yield responses that are not equivalent across studies (Table 1). Hold times at peak load are routinely incorporated to account for thermal drift and to stabilize measurements, with short hold periods typically ranging from 2 to 15 s [58,124,126]. Longer hold durations, extending to several minutes, have also been reported to enable detailed analysis of indentation creep behavior [125]. Such creep-type protocols are particularly valuable for assessing time-dependent plasticity and viscoplastic recovery in AlSi10Mg alloys, as they capture the influence of microstructural stability and cooling rate on deformation kinetics [125,126]. Accurate quantification of this response requires strict control of test conditions, as prolonged hold periods increase susceptibility to thermal drift, potentially leading to artificially inflated measurements if not corrected [140].

Despite variations in procedural details, the Oliver–Pharr method remains the standard analytical approach for extracting hardness and reduced modulus from unloading segments of the load–displacement curve [108]. Its widespread adoption ensures a degree of methodological consistency across studies, even though the heterogeneous microstructure of LPBF AlSi10Mg may occasionally challenge the underlying assumptions of the model. Residual stresses generated during LPBF processing may further complicate interpretation, as tensile and compressive stress fields across melt pools and build layers have been shown to alter apparent hardness and modulus values [18,141–143]. Although post-processing methods are applied to mitigate these stresses, associated microstructural changes affect the nanoindentation response, complicating direct comparison between as-built and heat-treated conditions [48,144,145].

In addition to residual stresses, porosity and internal defects present major sources of uncertainty. The formation of gas pores and lack-of-fusion voids is strongly influenced by laser power, scan speed, and hatch spacing [146]. Voids located close to indentation sites can cause apparent softening, abnormal displacement bursts, known as pop-ins, and increased scatter in measured properties [147,148].

Comparison across the literature is further hindered by incomplete reporting of processing parameters. Many studies provide limited detail on LPBF conditions such as build platform temperature, exposure time, and scan strategy, despite their known influence on thermal history, residual stress development, and microstructural evolution. Similarly, as previously mentioned, critical nanoindentation parameters are often reported without sufficient contextual information. Inconsistent documentation of maximum load or penetration depth, indentation spacing, indent location relative to melt-pool features, and environ-

mental conditions (Table 1) makes it difficult to determine whether reported differences in hardness or modulus arise from intrinsic microstructural effects or from procedural variability. This lack of standardized reporting reduces reproducibility and complicates meaningful comparison of nanoindentation data across studies.

4. Summary of Reported Nanoindentation Results

Building upon the processing-induced and thermally modified microstructures discussed above, the key issue that follows is how these microstructural states translate into a measurable local mechanical response. Resolving this link requires examination at the scale where cellular refinement, Si network morphology, and melt-pool architecture control deformation.

Nanoindentation has become an essential tool for elucidating the local mechanical response of AlSi10Mg, particularly because its cellular-dendritic microstructure exhibits pronounced spatial heterogeneity [149]. The mechanical properties of this alloy are highly sensitive to microstructural features such as the distribution of Si-rich networks, Al-rich matrices, and the morphology of melt pools, as well as processing parameters, post-processing treatments, alloying additions, and external influences such as irradiation. Collectively, these factors define a complex environment in which hardness varies not only between samples but also locally within a single melt pool or strut, highlighting the need for nanoscale resolution in mechanical characterization.

4.1. Hardness Trends

Understanding nano-hardness distributions in AM AlSi10Mg is essential, as they provide direct insight into the influence of microstructural heterogeneity on local mechanical performance. In as-built samples, hardness varies strongly between melt pools and heat-affected zones (HAZs) [150]. Within individual melt pools, fine eutectic regions consistently exhibit higher hardness than the coarser boundary regions. Delahaye and co-workers demonstrated that hardness reaches its maximum at the melt-pool core, where fine α -Al cells and eutectic Si dominate, and decreases toward the HAZ due to Si precipitation coarsening and partial breakdown of the eutectic network (Figure 3a) [17,119]. Subsequent nano-hardness mapping and zoning analyses quantitatively validate this trend (Figure 3c) [58,131]. This lower hardness observed in HAZ arises from wider Si spacing, which allows easier dislocation slip and reduces the effectiveness of Orowan strengthening, a mechanism known to operate in fine nanoscale Si networks in LPBF AlSi10Mg (Figure 3b) [17,119,151]. Extending this observation to ultrathin AlSi10Mg walls, Lu et al. reported that local nano-hardness fluctuates between approximately 1.4 and 1.9 GPa along the cross-section of the walls, reflecting the alternating distribution of finer and coarser microstructural regions [130].

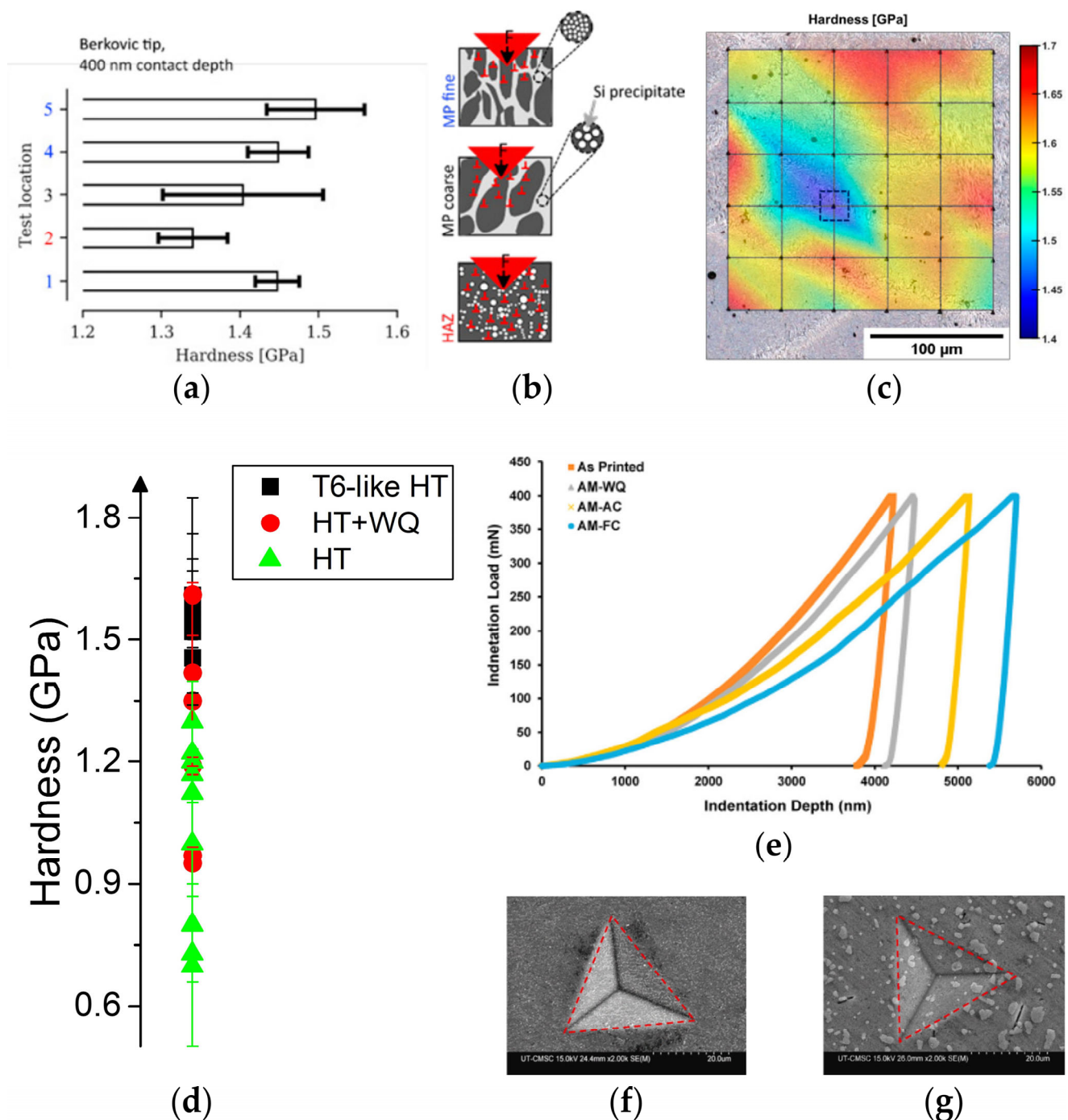


Figure 3. (a) Hardness values measured in distinct microstructural regions of LPBF AlSi10Mg, along with the (b) schematic illustration of force (F) application at these regions, reproduced from [119] under the CC BY-NC-ND 4.0 license. (c) Optical micrograph of LPBF AlSi10Mg with the nanoindentation hardness map superimposed on the microstructure, reproduced from [58] under the CC BY 4.0 license. (d) Hardness values after T6-like heat treatments (black squares [15,47,124,127]), heat treatments followed by water quenching (red circles [15,27,28,127]), and heat treatments performed at specific temperatures (green triangles [29,84,116]). (e) Load–displacement curves for as-built and heat-treated AlSi10Mg, accompanied by SEM images of representative indents (f) on as-built and (g) water quenched samples, reproduced from [28] under the CC BY-NC-ND 4.0 license.

In addition to spatial heterogeneity, post-processing significantly influences hardness, primarily through the spheroidization and coarsening of the eutectic Si network and the reduction in dislocation density in the Al matrix [28]. As-built samples typically show the

highest nano-hardness, for example, 2.71 ± 0.12 GPa was measured prior to T6 treatment, dropping to 1.56 ± 0.11 GPa after the standard T6 cycle [124]. Similar reductions, ranging from around 20 to 40%, have been reported across various thermal treatments, including HIP, solution annealing, and artificial aging [47,51,58,123,127,129]. When nano-hardness values are viewed collectively for different heat-treatment routes, T6-like treatments exhibit a comparatively narrow hardness range compared to heat treatments with or without water quenching (Figure 3d). This reduced variability suggests that the T6 thermal path imposes a more uniform and stable Si morphology, leading to a more consistent local mechanical response. In contrast, equal channel angular pressing (ECAP) followed by aging has been shown to increase nano-hardness by up to 37% compared to the as-built condition, while producing a more uniform hardness distribution across the sample [70]. Cooling rate also plays a critical role with rapid quenching preserving finer Si distributions and higher hardness, whereas furnace cooling promotes coarsening and the softest response (Figure 3e–g) [28,125,152,153]. Post-treatment hardness mapping further reveals increased scatter, reflecting local variations in Si morphology and microstructural evolution [15,47].

Alloying and reinforcement strategies provide effective means to tune hardness. TiN particles in AlSi10Mg induce a pronounced gradient in hardness, from ~ 2.3 GPa in the matrix to ~ 7.3 GPa in the graded layer zone, reaching ~ 20 GPa within the TiN particle zone. This increase is attributed to the formation of a graded interfacial layer containing $\text{Al}_{3.21}\text{Si}_{0.47}$ and (Ti,Al)N phases through in situ reactions and mutual diffusion, combined with dislocation accumulation caused by the thermal expansion mismatch between TiN and the Al matrix [154]. Similarly, SiC nanoparticles increased nano-hardness from ~ 2.09 to 2.46 GPa, with laser power adjustments affecting densification process and eutectic network formation [155]. The addition of 4 wt% Cu also increased hardness across several processing routes, highlighting the role of Cu-rich compounds in expanding the eutectic-like fraction of the microstructure [156]. Conversely, certain morphological modifiers, such as Er, Sr, and nano- TiB_2 , reduce the hardness of the AlSi10Mg compound by breaking the continuous Si network into larger, isolated particles, reducing solid solution strengthening and dislocation blocking while leaving grain refinement largely unaffected [157]. These findings demonstrate that microstructural engineering via reinforcement can offset or even exceed the softening effects of post-processing.

External influences such as irradiation also impact hardness. Helium ion implantation has been shown to produce $\sim 25\%$ hardening within the irradiated layer, resulting from the formation of helium bubbles that act as barriers to dislocation motion [115]. Ungarish et al. reported that irradiation-induced hardening ranges from 12% after stress-relief heat treatment to 24% after solution heat treatment, with the increase attributed mainly to Frenkel-pair defect formation and the lower sink density of coarse silicon precipitates in the solution heat-treated condition [116]. These results highlight the interplay between microstructure, defect formation, and local mechanical response in determining hardness.

Comparisons with cast AlSi10Mg underscore the advantages of LPBF in producing consistently hard materials and mechanically reliable materials. Conventional cast alloys typically exhibit nano-hardness below 1 GPa, with extreme hardness of ~ 9 GPa in Si flakes, whereas LPBF-produced AlSi10Mg consistently reaches 1.8–2.2 GPa with uniform distributions. This improvement is attributed to rapid solidification, refined cellular structures, and uniform Si segregation at α -Al boundaries, which enhance the overall load-bearing capacity of the alloy [28,111,158,159]. Other AM techniques, such as L-DED, demonstrate that processing parameters can influence the α -Al cell size, promote Si precipitation, and retain residual solid solution elements, resulting in higher nano-hardness and improved yield strength [160]. These observations further indicate that AM provides more controllable and predictable hardness profiles compared with conventional casting.

4.2. Elastic Modulus

Unlike nano-hardness, the elastic modulus of LPBF AlSi10Mg exhibits less dramatic variations but remains sensitive to microstructural characteristics, build orientation, and post-processing conditions. At the local scale, mapping shows a strong correlation with Si distribution and crystallographic orientation, with reported values ranging from 76 to 102 GPa [112]. Higher modulus corresponds to regions of finer eutectic structures or specific grain orientations with theoretical modeling, further confirming an increase of ~ 11 GPa when the grain orientation changed from [100] to [110]. This observation highlights the anisotropic response of the cellular-dendritic microstructure. Similarly, Xu et al. showed a higher elastic modulus within melt-pool centers compared to the boundaries, attributing the difference to refined eutectic structures [150].

Build orientation introduces another layer of anisotropy, with samples fabricated parallel or perpendicular to the build direction exhibiting a significantly different modulus. Dvorak and co-workers reported that parallel-oriented samples exhibit values around 70 GPa, whereas samples built perpendicular to the build direction can reach 103.9 GPa, reflecting the influence of printing-induced microstructural features such as grain size variations, partial amorphous phases, and localized porosity [161]. Cross-sectional analysis supports this trend, with XY plane measurements of 75.4 ± 1.6 GPa and XZ plane values of 65.9 ± 2.5 GPa [15]. Such differences reflect the microstructural anisotropy and residual stress distributions characteristic of LPBF.

However, when build direction and post-processing are considered together, more complex trends emerge. Solution annealing slightly increases the modulus in the XY plane but reduces it in the XZ plane, while subsequent artificial aging causes a marked decrease in the XY plane modulus to ~ 55 GPa with minimal change in XZ values (Figure 4a) [15]. This divergence was attributed to the breakdown of the continuous Si network, which shifted the load-bearing role more heavily onto the softer Al matrix.

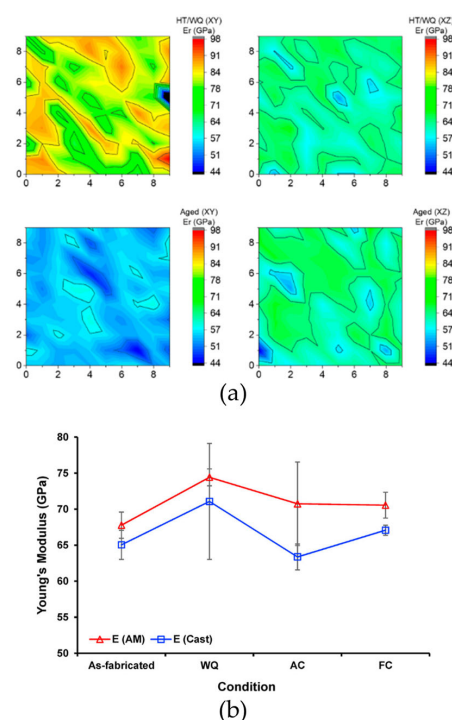


Figure 4. (a) Contour maps of the elastic modulus for AlSi10Mg after solution annealing (**top**) and artificial aging (**bottom**), across XY and XZ planes, reproduced from [15] under the CC BY 4.0 license. (b) Evolution of reduced modulus for cast (blue) and LPBF (red) conditions, reproduced from [28] under the CC BY-NC-ND 4.0 license.

Independent from orientation effects, post-processing introduces another dimension of variability, not through large shifts in the average modulus but through changes in its distribution. Solution treatments, for example, slightly increase the elastic modulus from ~100 to 105 GPa, likely due to spheroidization of Si phases and more uniform stress transfer across the microstructure [47]. Consistent with this, standard heat treatments generally maintain elastic modulus within the standard deviation (Figure 4b) [28]. More intensive treatments, such as friction stir processing, show that the α -Al matrix achieves a modulus of 79.8 GPa, closely matching macroscopic tensile tests, while large Si particles reach 167 GPa [132]. Finite element simulations confirmed that the matrix dominates the composite response, with Si particles exerting only a minor effect. Meanwhile, HIP has been reported to increase elastic modulus by ~10%, primarily through porosity reduction causing fewer void/matrix interfaces [123,162].

Beyond conventional post-processing procedures, external influences such as irradiation cause negligible modulus change (<0.2%), regardless of pre-treatment condition, even though defect sensitivity was altered [116]. This behavior contrasts with hardness, which showed significant degradation under identical conditions.

4.3. Load–Displacement Behavior and Microstructural Correlations

Nanoindentation imprints in LPBF AlSi10Mg correlate directly with the local mechanical response, revealing how heterogeneous microstructures govern deformation. As already established, melt-pool cores exhibit higher hardness than boundaries and HAZs. This heterogeneity manifests in load–displacement behavior with indents in fine eutectic zones being shallower, while coarser Si-rich or HAZ regions allow deeper penetration [17].

Load–displacement (P-h) curves provide critical insight into the local mechanical response of AlSi10Mg. Across multiple studies, smooth and continuous P-h curves are commonly reported in as-built AlSi10Mg, indicating gradual elastic–plastic transitions without discrete pop-in events or cracking [71,124,126]. Where discontinuities occur, they are typically associated with brittle Si-rich regions or large inclusions, whereas Al-rich areas exhibit more homogeneous plastic flow [15,132]. This contrast highlights the central role of local composition and phase distribution in controlling the onset of irreversible deformation.

Building on this observation, several studies have focused on discontinuities such as pop-in events. These pop-ins occur at the beginning of the indentation cycle and are caused by the indenter encountering brittle eutectic silicon, resulting in localized nano-fracture [153]. Complementary mechanical studies combined with localized elemental distribution report similar localization discontinuities in Si-rich zones, while Al-rich areas displayed continuous plastic flow [15]. Pop-in events were further linked with the presence of micro- and nano-voids in the AlSi10Mg microstructure (Figure 5a) [70], indicating that pop-ins are conditional on local brittle phases or defects, rather than intrinsic to the indentation process.

Beyond discrete pop-ins, Isil et al. also identified multi-stage transitions in hardness–depth curves, particularly in ECAP and ECAP-aged samples (Figure 5b) [70]. These stages were directly related to dislocation activity: initial easy glide at the surface, followed by pile-up at grain boundaries, and finally dislocation transfer across boundaries once a critical stress was reached. The presence of (Mg,Si)-rich phases was shown to intensify hardening during these stages, producing higher peak heights in mechanical response.

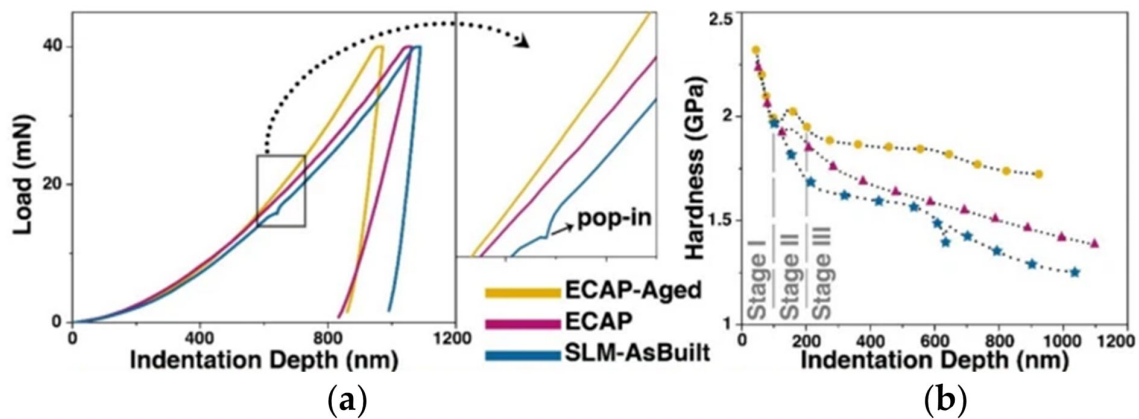


Figure 5. (a) Melt-pool displacement curves of AlSi10Mg samples illustrating pop-in behavior and (b) hardness as a function of indentation depth exhibiting multi-stage response during nanoindentation, reproduced from [70] under the CC BY 4.0 license.

Indentation size effect (ISE) provides further insight into nanoscale strengthening mechanisms. At shallow indentation depths, hardness increases due to strain-gradient plasticity and dislocation starvation, a phenomenon that is more pronounced in as-built samples [124]. Subsequent studies confirmed that this depth-dependent hardness persists across various post-fabrication heat treatments, with hardness decreasing systematically as indentation depth increases [85]. The Nix–Gao model was used to describe this behavior, showing a linear relation between hardness squared and the inverse of indentation depth at large depths, while shallow regions deviated due to the dominance of geometrically necessary dislocations. These findings further highlight the connection between microstructure and local mechanical response, showing that even modest post-processing can significantly alter the load-bearing response of the matrix.

Time-dependent deformation measured during hold segments of indentation further distinguishes microstructural states. Wu et al. reported displacement–time creep traces with an initially linear increase in displacement that stabilized at longer hold times [126]. From these tests the authors extracted unusually high creep stress exponents for LPBF material, and they noted pronounced anisotropy. Other indentation creep studies describe a two-stage response consisting of a transient stage followed by a steady-state stage [125]. High creep resistance in the as-built condition was linked to the refined cellular network and strong dislocation–Si interactions, whereas post-processing coarsened the network and reduced resistance. Together, these observations show that indentation creep measurements are a sensitive probe of microstructural stability and dislocation–particle interactions in AM AlSi10Mg.

5. Correlating Processing Parameters with Nanoindentation Results

Establishing correlations between nanoindentation-derived properties and LPBF parameters is often approached from a microstructural perspective, but a complete process–structure–property linkage requires considering whether measurable nano-mechanical variations can be systematically traced back to controllable fabrication parameters. This task is intrinsically challenging, as changes in laser power, scan speed, layer thickness, and hatch distance simultaneously affect solidification rate, solute redistribution, and phase morphology [11,163]. As a result, isolating the effect of every single processing variable is rarely possible based on the available literature.

To facilitate cross-study comparisons, VED values were calculated where the necessary processing parameters were reported and corresponding nanoindentation values were

available. The resulting compilation is presented in Figure 6, where it can be observed that neither nano-hardness nor elastic modulus exhibits a clear monotonic dependence on VED over the broad range of approximately 30 to 130 J/mm³. Instead, both properties display significant scatter, underscoring the limitations of VED as a unique descriptor of local mechanical response. Additionally, it should be noted that nanoindentation values in the literature are commonly reported without associated statistical uncertainty, and therefore, Figure 6 primarily enables a qualitative assessment of data scatter and relative tendencies rather than a rigorous evaluation of correlations.

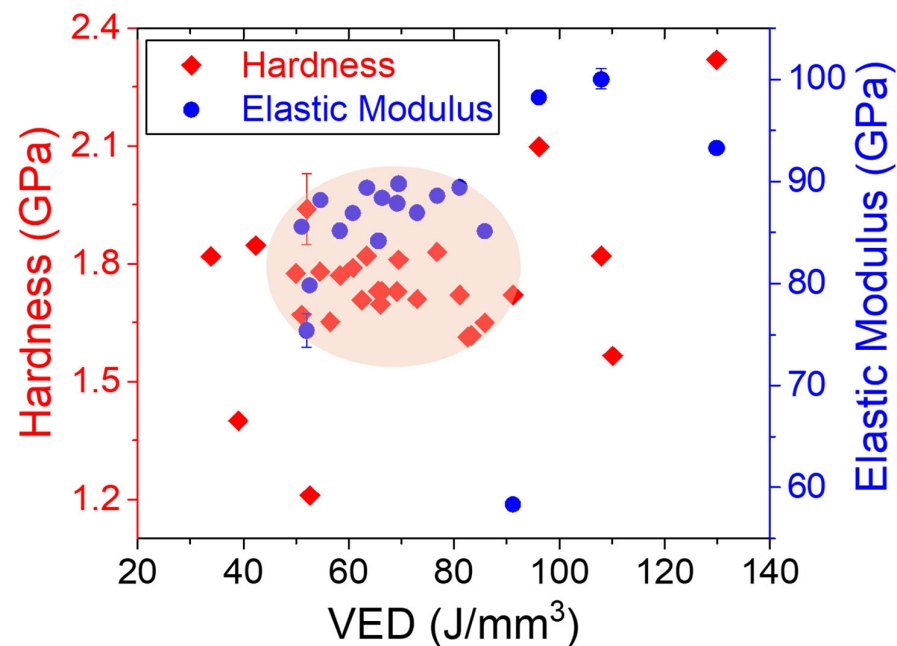


Figure 6. Nano-hardness (red diamonds [15,27,47,123,127,128,131,134,164]) and elastic modulus (blue circle [15,47,127,128,132,134]) as a function of calculated VED for as-built LPBF AlSi10Mg. The highlighted region indicates the VED range where both hardness and elastic modulus predominantly fall.

Within this VED range, reported nanoindentation hardness values for as-built LPBF AlSi10Mg predominantly fall between around 1.6 and 1.9 GPa, with deviations occurring at both lower and higher energy inputs. Increased scatter is evident at low (<50 J/mm³) and high (>90 J/mm³) VED values, whereas the intermediate VED values tend to be associated with more consistent hardness levels. This reduced scatter at an intermediate VED window is consistent with the processing regimes that most frequently produce near-full dense AlSi10Mg parts (Figure 2). Deviations toward insufficient or excessive energy input introduce additional heterogeneity that is reflected in the mechanical response. In this sense, the VED range identified as optimal for densification provides a baseline condition for achieving more reproducible mechanical properties.

Consistent with this interpretation, several studies have reported non-monotonic hardness trends with VED [130,164,165]. For instance, in AlSi10Mg ultrathin walls nano-hardness tends to be reduced at low energy density, whereas at intermediate energy densities it increases and reaches a maximum, before decreasing again at higher inputs [130]. The initial increase reflects a balance of strengthening mechanisms active during microstructure evolution, while the subsequent decrease is attributed to the formation of softer melt-pool boundary regions under higher inputs. Separately, it has been shown that increasing VED from 37 to 92 J/mm³ reduced the relative density of AlSi10Mg specimens from $97.1 \pm 0.05\%$ to $95.2 \pm 0.15\%$, due to the formation of gas pores and deep melt-pool defects at higher

inputs [126]. Nanoindentation measurements then showed that regions formed by these deeper melt pools exhibited higher hardness compared to shallower pools, and that longitudinal sections were harder than transverse ones. These differences were attributed to local variations in the balance between α -Al and Si phases.

Elastic modulus values derived from the available nanoindentation data exhibit a comparatively clearer dependence on VED. As shown in Figure 6, elastic modulus displays substantial scatter at lower VED values ($< \sim 60 \text{ J/mm}^3$), followed by a general increase with increasing energy density and a tendency toward saturation at higher VED. Similar to nano-hardness, fluctuations persist across the entire VED range, reflecting sensitivity to local phase arrangement, porosity, and crystallographic texture rather than to energy density alone.

Beyond VED-based considerations, individual processing parameters exert distinct influences on nanoindentation response, although direct one-to-one comparisons are often hindered by the lack of systematically controlled datasets (Table 2). For example, when comparing bulk samples with varied scan speed, produced at 800 mm/s with lattice struts fabricated at 2500 mm/s, a slight increase in hardness is observed; however, differences in geometry, thermal history, and unreported processing details complicate direct attribution [27,127]. Microstructural analysis revealed the presence of supersaturated α -Al in bulk samples, contributing to solid solution strengthening, whereas the lattice struts display heterogeneous features, including variations in Si spacing and grain orientation. Nevertheless, the increase in nano-hardness with increasing scan speed was further observed by Gökdağ et al. [164]. These observations suggest that scan speed influences hardness indirectly through its effect on local solidification conditions rather than through a simple parametric relationship.

Laser power is another influential factor as it governs the heat flow, cooling rate, and ultimately the microstructure of the alloy. At lower power, the microstructure is characterized by a relatively uniform spacing of Si precipitates and a corresponding hardness of about $1.67 \pm 0.1 \text{ GPa}$ [27]. Increasing the power has been shown to promote the development of additional crystallographic texture orientations, including [110,111] parallel to the build direction, together with a reduction in grain size and a refinement of silicon spacing (Figure 7a). In these conditions, hardness values range from approximately 1.4 GPa in regions with coarser Si spacing to 1.99 GPa in regions with finer Si distribution (Figure 7b). Interestingly, another report showed a systematic decrease in nano-hardness with increasing laser power, in otherwise similar conditions [164]. This could be associated with the reduced cooling rates that promote coarser eutectic cell structures and larger Si precipitates, thereby weakening Hall–Petch and Orowan strengthening contributions.

Table 2. Summary of laser processing parameters, post-processing treatments, and resulting nano-mechanical properties of LPBF AlSi10Mg.

No.	Laser Power (W)	Scan Speed (mm/s)	Layer Thickness (μm)	Point Distance (μm)	Exposure Time (μs)	Hatch Distance (μm)	Scan Strategy	Post Processing/Surface Finish	Hardness (GPa)	Elastic Modulus (GPa)	P-h Characteristics	Effect of Build Orientation	Effect of Heat Treatment	Ref.
1	300	800	30	N/A ¹	N/A	130	N/A	A1: 300 °C/2 h + WQ ² ; A2: 535 °C/1 h + WQ + 190 °C/10 h + FQ ³ ; chemically etched in 25 mL HNO ₃ + 75 mL H ₂ O at 70 °C for 40 s; rinsed in water	AB ⁴ : 2.098; A1: 1.61; A2: 1.456	AB: 98.21; A1: 103.67; A2: 99.38	Higher load in AB than HT ⁵ ; smooth curves observed	N/A	A1 eliminates residual strength and improves mechanical properties	[127]
2	200–300	2500	30	N/A	N/A	N/A	Single contour exposure	HT at 300–500 °C/2 h + WQ; ground and polished, etched 25 s with Keller’s reagent	AB (200 W): 1.67 $\pm$ 0.1; AB (300 W): 1.4–1.99; HT 300 °C (200 W): 1.19 $\pm$ 0.02; HT 350 °C (200 W): 0.97 $\pm$ 0.02	N/A	N/A	N/A	Average hardness decreases with HT	[27]
3	200	570	25	80	140	130	Checkerboard	T6 HT: 520 °C/1 h + AA ⁶ 160 °C/6 h; cross-sectioned, polished, etched with Keller’s reagent	AB: 1.82 $\pm$ 0.01; HT: 1.52 $\pm$ 0.02	AB: 100 $\pm$ 1; HT: 105 $\pm$ 1	N/A	N/A	Hardness decreases after HT	[47]
4	200	1100	20	N/A	N/A	70	Zigzag & Checkerboard	Polished and etched with 0.5% HNO ₃ for 5 s	2.32	93.26	N/A	N/A	N/A	[128]
5	N/A	N/A	30 & 60	N/A	N/A	100 & 190	67° and 90° rotation (0° and 45° orientations)	Mechanically ground with SiC paper, polished and etched with Weck’s reagent	~0.5 to ~1.0 (depending on zone and process parameters)	N/A	N/A	N/A	N/A	[135]
6	259–370	1300	30	N/A	N/A	104–156	67° rotation	Polished and etched with Keller’s reagent	1.65–1.83	~84–90	Smooth loading–unloading curves; slight variation with laser power/hatch distance; consistent elastic recovery	N/A	N/A	[134]
7	400	4717.97 393.16	30 & 60	N/A	N/A	N/A	Single-track	Polished and etched 10–15 s with Keller’s reagent	1.4–1.9	N/A	Shallower indentation at moderate energy density; deeper indentations at high energy density	N/A	N/A	[130]

Table 2. Cont.

No.	Laser Power (W)	Scan Speed (mm/s)	Layer Thickness (μm)	Point Distance (μm)	Exposure Time (μs)	Hatch Distance (μm)	Scan Strategy	Post Processing/Surface Finish	Hardness (GPa)	Elastic Modulus (GPa)	P-h Characteristics	Effect of Build Orientation	Effect of Heat Treatment	Ref.
8	175–275	500–780	50	N/A	N/A	120	N/A	None	Higher hardness in refined microstructure than coarse microstructure	N/A	Increased displacement for transverse sections compared to longitudinal; Different response at different regions	Anisotropic behavior observed	N/A	[126]
9	400	1200	40	N/A	N/A	160	N/A	SA ⁷ : 450 °C/2 h + WQ; SA&AA: 165 °C/6 h + AC ⁸ ; polished and etched for 10–15 s with Kroll's reagent	XY: AB 1.94 ± 0.09 ; SA 1.42 ± 0.22 ; SA&AA: 1.55 ± 0.21 ; XZ: AB 2.01 ± 0.08 ; SA 1.35 ± 0.16 ; SA&AA 1.61 ± 0.24	XY: AB 75.4 ± 1.6 ; SA 83.2 ± 6.0 ; SA&AA 55.2 ± 2.7 ; XZ: AB 65.9 ± 2.5 ; SA 64.6 ± 3.3 ; SA&AA 64.4 ± 3.8	AB: smooth, homogeneous curves; SA: steeper slopes in Si-rich zones; SA&AA: irregular, shallower curves with fracture events	AB: Minor; near isotropic hardness; XY slightly higher modulus; SA: Minimal difference; isotropy maintained; SA&AA: Minimal difference; XZ shows slightly more uniform modulus distribution	AB: Highest hardness; SA: Hardness drop due to Si coarsening, stress relief; SA&AA: Partial hardness recovery, modulus decreases in XY, heterogeneity increases	[15]
10	N/A	N/A	N/A	N/A	N/A	N/A	N/A	HT: 540 °C/2 h in a fluidized bed (alumina particles); AC/FQ/WQ; samples were ground, polished, and etched with Keller's reagent	AB: ~1; WQ: ~0.95; AC: ~0.75; FQ: ~0.6	AB: ~65; WQ: ~70; AC: ~62; FQ: ~68	Higher displacement for FC sample under constant load conditions	N/A	Hardness decreases with slower cooling	[28]
11	N/A	N/A	N/A	N/A	N/A	N/A	N/A	T6 HT: 520 °C/1 h + WQ + AA 170 °C/4 h; polished, and etched with Keller's reagent	YZ: AB 2.71 ± 0.12 ; HT 1.56 ± 0.11 ; XY: AB 2.58 ± 0.12 ; HT 1.59 ± 0.11	N/A	Lower indentation displacement for AB	XY and YZ planes show different hardness values	Hardness decreases after HT; softening observed in both planes	[124]
12	175	195	60	N/A	N/A	N/A	N/A	Polished, and etched with Keller's reagent	Melt-pool fine: 1.52 ± 0.06 ; Melt-pool coarse: 1.44 ± 0.10 ; HAZ: 1.35 ± 0.04	N/A	N/A	N/A	N/A	[17]

Table 2. Cont.

No.	Laser Power (W)	Scan Speed (mm/s)	Layer Thickness (μm)	Point Distance (μm)	Exposure Time (μs)	Hatch Distance (μm)	Scan Strategy	Post Processing/Surface Finish	Hardness (GPa)	Elastic Modulus (GPa)	P-h Characteristics	Effect of Build Orientation	Effect of Heat Treatment	Ref.
13	400	1350	50	N/A	N/A	150	Stripe scan with 45° rotation	In situ HT at 220, 300, 400, 450 and 500 °C	HT (220 °C): 1.3 ± 0.1; HT (300 °C): 1.2 ± 0.1; HT (400 °C): 1.0 ± 0.1; HT (450 °C): 0.7 ± 0.2; HT (500 °C): 0.73 ± 0.07	N/A	N/A	N/A	Average nano-hardness values decreased by ~43% with increasing temperature	[29]
14	275	2000	30	80	40	80	67° rotation, counterclockwise	Polished	Maximum value at melt-pool center	N/A	N/A	N/A	N/A	[11]
15	200	1200	30	N/A	N/A	110	90° rotation, unidirectionally	Polished	N/A	Large scatter in the range of small loads	N/A	N/A	N/A	[112]
16	350	1150	50	40	N/A	170	67° rotation, counterclockwise bidirectional stripes	T5 HT: 160 °C/4 h + AC; T6R HT: 510 °C/10 min + WQ + 160 °C/6 h + AC; T6B HT: 540 °C/1 h + WQ + 160 °C/4 h + AC; polished and etched with Weck's reagent	AB: 1.4–1.7; T6R HT: 1.3–1.4; T6B: 1.1–1.8	N/A	N/A	N/A	T6 HT led to more homogenous Si particle distribution; better mechanical properties	[58]
17	350	1150	50	40	N/A	170	67° rotation, counterclockwise bidirectional stripes	T5 HT: 160 °C/4 h + AC; T6R HT: 510 °C/10 min + WQ + 160 °C/6 h + AC; T6B HT: 540 °C/1 h + WQ + 160 °C/4 h + AC; polished and etched with Weck's reagent	AB: 1.6–1.7; T5 HT: 1.6–1.7; T6R HT: 1.3–1.4; T6B: 1–1.7	N/A	N/A	N/A	T5 HT showed similar hardness to AB; T6 HT decreased hardness with T6R showing more uniform Si particle distribution	[114]
18	350	1400	50	N/A	N/A	100	N/A	polished, and etched with Keller's reagent	~1.16–1.58	68.35–87.24	Slight variations in penetration depths	N/A	N/A	[136]
19	400	1000	N/A	N/A	N/A	N/A	Hatching technique, continuous change in scanning direction	HT: 300 °C/2 h; polished, He implantation	25% increase upon irradiation	N/A	N/A	N/A	Used for stress relief	[115]

Table 2. Cont.

No.	Laser Power (W)	Scan Speed (mm/s)	Layer Thickness (μm)	Point Distance (μm)	Exposure Time (μs)	Hatch Distance (μm)	Scan Strategy	Post Processing/Surface Finish	Hardness (GPa)	Elastic Modulus (GPa)	P-h Characteristics	Effect of Build Orientation	Effect of Heat Treatment	Ref.
20	400	1000	30	N/A	N/A	N/A	N/A	HT (1): 300 °C/2 h; HT (2): 500 °C/1 h + AC; polished, irradiation	HT (1) unirradiated: 1.17 ± 0.06 ; HT (1) irradiated: 1.31 ± 0.03 ; HT (2) unirradiated: 0.80 ± 0.07 ; HT (2) irradiated: 0.99 ± 0.06	N/A	N/A	N/A	HT (2) significantly affects microstructure and radiation response; increases diffusivity and hardening	[116]
21	175	195	60	N/A	N/A	N/A	N/A	Reheated and resolidified; polished and etched with Keller's reagent	Range depending on microstructure	N/A	N/A	N/A	N/A	[119]
22	400	1100	60	N/A	N/A	200	90° rotation	HT: 300 °C/2 h; T6 HT: 560 °C/2 h + WQ + 180 °C/12 h; polished	N/A	N/A	N/A	N/A	N/A	[122]
23	N/A	N/A	N/A	N/A	N/A	N/A	N/A	HT (1): 300 °C/2 h; HT (2): 300 °C/4 h; polished and etched	HT (1): 1.221; HT (2): 1.125	HT (1): 65; HT (2): 60	Fusion zone is softer than semi-elliptical core; Si particles exhibit lower penetration depth than Al matrix	N/A	Hardness reduction due to stress relief	[84]
24	225	1200	30	N/A	N/A	160	N/A	HIP; polished and etched with Weck's reagent	AB: 1.4; HIP: 0.8	AB: 79.9; HIP: ~88	HIP samples show greater indentation depth	N/A	HIP caused: ~42% decrease in nano- hardness; ~10% increase in modulus; grain growth; larger Si particles, increased roughness, more pores	[123]
25	200	N/A	25	80	140	130	N/A	Solution HT: 540 °C/2 h + AC/FC, polished and etched using Keller's reagent	N/A	N/A	Indentation displacement: AB: 3134 nm; HT/AC: 3758 nm; HT/FC: 4292 nm	N/A	Microstructural and mechanical variations upon HT	[125]

Table 2. Cont.

No.	Laser Power (W)	Scan Speed (mm/s)	Layer Thickness (μm)	Point Distance (μm)	Exposure Time (μs)	Hatch Distance (μm)	Scan Strategy	Post Processing/Surface Finish	Hardness (GPa)	Elastic Modulus (GPa)	P-h Characteristics	Effect of Build Orientation	Effect of Heat Treatment	Ref.
26	390	1300	30	N/A	N/A	190	67° rotation	Build platform temperature: 35 °C or 200 °C	Varies by zone and condition from ~1.21 at HAZ (200 °C) to ~2.02 at FMP (35 °C)	N/A	N/A	Nano-hardness evolution: 35 °C: Rise for both horizontal and vertical orientation; 200 °C: Rise for horizontal, drop for vertical orientation	N/A	[131]
27	390	1300	30	N/A	N/A	190	67° rotation	FSP	N/A	FSP: 79.8	Decrease in influence of matrix when P-h curves are obtained on Si particles	N/A		[132]
28	N/A	N/A	N/A	N/A	N/A	N/A	N/A	HT: 520 °C/1h + WQ/AC/FC, polished and etched with Keller's reagent	AB: 1.2–1.22; WQ: 1–1.05; AC: 0.9–0.94; FC: 0.8–0.83	N/A	Under constant load, displacement increases as follows: FC > AC > WQ > AB	N/A	HT samples are softer than AB	[85]

¹ N/A = Not available. ² WQ = Water-quenched. ³ FQ = Furnace-quenched. ⁴ AB = As-built. ⁵ HT = Heat treatment. ⁶ AA = Artificial aging. ⁷ SA = Solution annealing. ⁸ AC = Air-cooled.

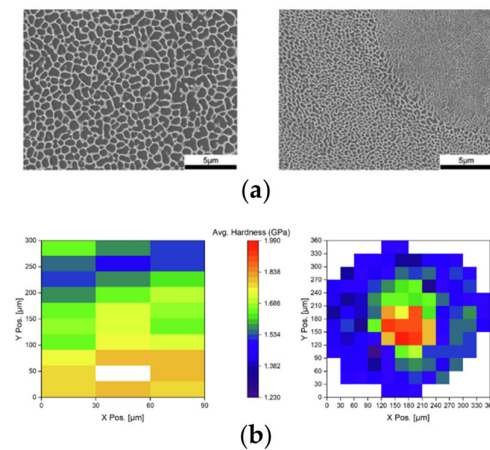


Figure 7. (a) SEM images as-built AlSi10Mg struts fabricated using laser power of 200 W (left) and 300 W (right) together with the (b) corresponding nano-hardness maps of the same vertical struts, reproduced from [27] under the CC BY 4.0 license.

Layer thickness is another parameter that influences the mechanical response of LPBF-fabricated AlSi10Mg alloys. It has been shown to affect both grain morphology and crystallographic texture, with low layer thicknesses promoting columnar grain growth parallel to the build direction and a strong $\langle 100 \rangle$ fiber texture [166]. With increasing layer thickness, grain growth directions become more diversified, the fraction of equiaxed grains rises, and the intensity of the $\langle 100 \rangle$ texture correspondingly decreases. Although there are no systematic studies directly linking layer thickness to nano-mechanical properties, existing reports provide some indications. For example, at a constant laser power of 200 W reported nanoindentation data suggest that higher hardness values are frequently associated with thinner layers (Table 2) [27,47,128,159]. This trend suggests that lower layer thickness may contribute to higher hardness through microstructural refinement. However, concurrent variations in scan speed and laser power complicate direct attribution, indicating that layer thickness primarily influences nano-hardness through its impact on microstructural refinement rather than as an independent control parameter. It should also be recognized that other parameters, such as hatch distance, have been shown to exert a stronger influence on microstructure compared to layer thickness [135]. Such information is not consistently documented; however, Paul et al. observed that thicker layers and wider hatch spacings lead to larger cells, increased mean free path for dislocations, and lower hardness, while finer scan strategies preserve hardness by maintaining microstructural refinement [135]. Complementing these findings, a more recent study revealed that small hatch distances exhibit lower hardness values, whereas intermediate-to-larger hatch distances promote greater Si segregation at α -Al boundaries, resulting in higher nano-hardness [134]. The trend peaks at medium hatch distances and slightly decreases at the largest hatch distance.

In addition to processing parameters, the build orientation imparts anisotropy to the hardness distribution as well. Wu et al. demonstrated that longitudinal sections aligned with the build direction exhibit higher hardness than transverse sections, indicating anisotropy linked to build orientation [126]. This anisotropy is amplified in regions with elongated hatch overlaps and larger melt-pool structures, where microstructural orientation influences local deformation mechanisms. A topological study further confirmed that printing orientation and track direction can produce marked differences in nano-hardness [161].

Taken together, rather than defining explicit processing windows for targeted nano-mechanical properties, the collective evidence suggests that control over local microstructural attributes, such as cellular refinement, Si network morphology, melt-pool geometry,

and residual stress state, represents a more reliable pathway for tailoring nanoindentation response. In this context, LPBF processing parameters should be viewed as means that shape these microstructural features, rather than as independent design variables. Strategies that promote refined cellular structures and controlled thermal gradients are consistently associated with higher local hardness, whereas excessive energy input or thermal accumulation tends to reduce hardness through microstructural coarsening and defect formation. However, given the strong parameter interdependencies associated with LPBF process and the limited availability of systematically controlled datasets, such guidance remains qualitative and microstructure-centric rather than prescriptive.

6. Methodological Considerations for Consistent Nanoindentation of LPBF AlSi10Mg

While VED and individual process parameters offer useful guidance, observed variability in nanoindentation responses highlights the critical role of measurement methodology. Accurate assessment of the process–microstructure–property relationship requires experimental protocols that capture the local heterogeneity created by LPBF and post-processing, ensuring that nanoindentation truly reflects the underlying microstructural state.

Reliable and reproducible nanoindentation of LPBF AlSi10Mg requires careful control of testing parameters, surface preparation, and documentation. Standardized procedures, including those outlined in ISO 14577 and ASTM E2546, provide a general framework for instrument calibration, specifying general principles for load application, displacement measurement, and data analysis [167,168]. While these recommendations ensure that measurements can be compared across different studies and laboratories, additional material-specific considerations are required to address the pronounced microstructural heterogeneity associated with LPBF-processed AlSi10Mg.

Prior to testing, the purpose of the experiment should be defined, as the scale of indentation must reflect the intended analysis. Two distinct indentation regimes should therefore be considered: (i) intrinsic phase-scale indentation, where the indentation depth must remain much smaller than the characteristic microstructural length scale; (ii) composite or microstructure-averaged indentation, where the indentation depth approaches or exceeds the microstructural length scale and the response represents an averaged mechanical behavior.

Shallow penetration depths are required when the goal is to probe individual microstructural constituents, such as α -Al cells or Si-rich eutectic regions. For example, the cellular α -Al structure inside melt pools has been reported to range from below 0.2 μm in very fine regions to approximately 0.4–0.7 μm in typical fine and coarse cellular zones [14,169]. These dimensions correspond to the as-built LPBF condition and may vary with processing parameters or post-processing treatments. To access these intrinsic properties, a common yet conservative adopted guideline for heterogeneous materials is that the maximum indentation depth should not exceed approximately one tenth of the relevant microstructural length scale [170]. Given an average Al cell size of 0.5 μm , a maximum indentation depth of approximately 50 nm is recommended for strict phase isolation. Under these conditions, the characteristic material domain sensed by Berkovich indentation, on the order of three times the depth, remains confined within the phase of interest [170].

Conversely, when the objective is to obtain statistically representative values that approximate bulk-like behavior, deeper penetrations should be intentionally used to average across multiple cells, melt-pool substructures, and local orientations. For HAZs with widths of approximately 2–3 μm [169], indentation depths should be limited to ~200–300 nm to resolve the region-specific mechanical response. In contrast, coarse substructures extending up to 10 μm permit depths approaching ~1 μm without significant scale interference. Accordingly, considering grain diameters ranging from approximately 5 to 12 μm depending

on build orientation and processing conditions [169,171], indentation depths could exceed 500 nm to approach grain-averaged mechanical behavior.

Testing parameters, including indenter type, maximum load, loading rate, hold time, and spacing, should be selected according to the intended indentation depth and microstructural scale, while indentation arrays instead of isolated indents should be used. For example, for each microstructural region of interest, a minimum of 10–15 indents would be sufficient to capture local scatter, while 20–30 indents per region would improve statistical confidence, when mapping spatial gradients in heterogeneous microstructures. In addition to the number of indents, indent spacing must be large enough to avoid interaction between plastic zones. For Berkovich indentation, studies have shown that a minimum spacing of approximately 10 times the indentation depth results in negligible hardness deviation, enabling reliable measurements while allowing high-resolution mapping [172].

Additionally, nanoindentation results should be interpreted in the context of the local microstructure. Whenever possible, measurements should be correlated with high-resolution microstructural characterization, including phase distribution, grain orientation, and defect population. Such correlations enable a more accurate understanding of the observed mechanical response and facilitate consistent comparison between different samples and studies.

Comprehensive reporting is necessary to allow meaningful comparison between studies. Alongside nanoindentation parameters, all relevant LPBF process conditions, including laser power, scan speed, layer thickness, hatch distance, platform temperature, and build orientation, should be provided. Documentation of environmental conditions, surface preparation methods, and any post-processing is also required. Much of the variability observed across studies can be attributed to differences in the experimental scale, the specific microstructural regions being targeted, and the level of detail in reporting, rather than to intrinsic differences in the material response itself. Improving comparability therefore requires not only more consistent measurement practices, but also a clearer alignment between indentation strategy and microstructural intent.

7. Unexplored & Underexplored Aspects

Despite the widespread use of LPBF-processed AlSi10Mg in both research and industrial applications, nanoindentation studies on this alloy remain relatively limited. Existing works predominantly focus on isolated hardness or elastic modulus measurements without sufficient contextualization in terms of grain orientation, phase distribution, or porosity [124,127]. Furthermore, the limited spatial and depth-resolved analyses, together with the minimal use of data-driven or predictive approaches, highlight the need for integrated investigations that combine mechanical measurements, microstructural characterization, and process information to elucidate structure–property relationships in LPBF AlSi10Mg.

7.1. Linking Local Mechanical Response and Microstructure

Understanding how local mechanical properties vary as a function of microstructural features in LPBF AlSi10Mg requires carefully aligned multi-technique characterization. As discussed in the previous chapters, many studies report nanoindentation-derived hardness or modulus values, yet only a subset establish meaningful correlations between these measurements and the underlying microstructure. In many cases, mechanical data are presented without spatially registered structural information, limiting the ability to identify the mechanisms by which crystallographic orientation, sub-grain cell size, Si network morphology, porosity, or grain boundary character influence the local response.

A few studies have tried to bridge this gap. For instance, Everitt et al. systematically mapped nanoindentation hardness and modulus on LPBF AlSi10Mg and related these

values to the distribution of the Si phase [111]. Their findings indicated that when Si is evenly dispersed throughout the microstructure, mechanical properties remain relatively homogeneous despite underlying pronounced solidification textures. This observation hints that nanoscale homogeneity in phase distribution can reduce mechanical anisotropy, an effect highly relevant for LPBF microstructures characterized by melt-pool bands and complex thermal gradients. Despite such promising efforts, the central limitation remains the lack of direct spatial alignment between mechanical measurements and crystallographic characterization. While many works employ techniques such as electron backscatter diffraction (EBSD) or high-resolution imaging [58,85,114,126], it is not common to find indentation arrays in the same area. Without such alignment, it is difficult to determine how crystallographic orientation, cell size, or the continuity of the silicon network influence the local response measured at each indentation.

Recent developments in automated high-density nanoindentation mapping, combined with large-area EBSD acquisition and spatial registration algorithms, enable statistically robust correlation between crystallographic orientation and local mechanical response in AM alloys [173–175]. Such correlative approaches to LPBF AlSi10Mg allow direct comparison of separate datasets and enable a clear distinction between orientation-related effects and strengthening associated with phase distribution. For example, performing dense indentation arrays across melt-pool cores, boundaries, and overlapping regions within a single build, followed by direct correlation with EBSD orientation maps, would allow quantitative assessment of whether hardness variations arise primarily from crystallographic misorientation or from discontinuities in the Si network. Consistent with this limitation, a recent study provided statistical evidence that, despite local correlations, easily accessible tools such as scanning electron microscopy (SEM) alone cannot reliably predict nanoindentation hardness in LPBF AlSi10Mg [137].

In addition, advanced characterization techniques such as transmission electron microscopy (TEM), atom probe tomography, and synchrotron-based imaging could reveal fine-scale heterogeneity in grain boundaries, sub-grain networks, or silicon morphology, enabling more accurate correlations with local mechanical response [176–178]. Complementing these methods, in situ nanoindentation techniques could further provide real-time insight into local deformation mechanisms during loading, offering a direct view of processes that conventional post-test indentation methods are unable to capture [179]. Specifically, in situ TEM nanoindentation would further enable the direct observation of dislocation nucleation and motion at Si/Al interfaces and grain boundaries in nanoscale volume, thereby clarifying how the cellular α -Al/Si network accommodates plastic strain and governs the initiation of pop-ins or indentation size effects observed in melt-pool displacement curves [180].

Another unexplored area is depth-resolved mapping through the build direction. Indentation experiments remain confined to a single polished surface, providing only two-dimensional information [127,128]. Yet the LPBF process produces a microstructure that evolves with depth as a function of heat flow and layer reheating [181]. Repeated thermal cycling modifies cellular spacing, Si redistribution, and residual stress state along the build height [182,183], suggesting that nano-mechanical properties are unlikely to remain uniform through thickness. Understanding how hardness or modulus vary from the top layers toward the base of a build would provide new insights into the development of internal stresses and microstructural gradients that conventional cross-sections cannot capture. A systematic indentation campaign conducted at multiple build heights could therefore quantify reheating-induced strengthening or softening effects, directly linking thermal history to local mechanical response.

Beyond experimental techniques, there is minimal exploration using machine learning or other data-centric tools to extract relationships in what are often complex datasets [109,112,122]. Rajan et al. used micromechanical modeling where nanoindentation data for both the Al matrix and Si-rich regions were used to calibrate a constitutive model capable of predicting tensile behavior [122]. Building upon such efforts, data-driven frameworks are increasingly being adopted in AM research to correlate processing parameters and microstructural descriptors with mechanical performance [184,185]. While such studies focus primarily on bulk properties, the same methodology can be extended to nanoindentation datasets. Quantitative descriptors such as cellular spacing, Si network thickness, and porosity fraction extracted from image analysis can serve as input features, with hardness and modulus as outputs. Given the strong interdependency of LPBF processing parameters, multivariate statistical or machine-learning approaches may be particularly useful for separating the combined influence of multiple processing parameters that cannot be independently varied in most studies. For instance, a dataset integrating laser power, scan speed, cellular spacing, and locally measured hardness values could be used to train regression or classification models capable of predicting nanoscale strengthening trends under new processing conditions, thereby enabling data-guided optimization of LPBF parameters. Such integration would shift nanoindentation studies from qualitative mapping toward predictive microstructure–mechanics relationships grounded in measurable descriptors rather than isolated parametric trends.

To advance the field, future work should focus on generating datasets that explicitly combine nanoindentation measurements with high-resolution microstructural information and integrate them using advanced statistical or machine-learning frameworks. By linking processing conditions, microstructure, and local mechanical properties in a quantitative framework, it would be possible to move beyond descriptive hardness mapping and develop a genuine microstructure–mechanics understanding, addressing one of the most significant gaps in LPBF AlSi10Mg research.

7.2. Scope and Limitations of Relating Nanoindentation to Bulk Mechanical Behavior

An important but often insufficiently addressed aspect of nanoindentation studies on LPBF AlSi10Mg is the distinction between local mechanical response and bulk mechanical performance. While nanoindentation provides localized insight into hardness and elastic modulus, these quantities cannot be directly translated into macroscopic properties such as yield strength, ductility, or fracture resistance. The small-probed volume, combined with the pronounced microstructural and defect heterogeneity of LPBF AlSi10Mg, limits the representativeness of individual indentation measurements with respect to bulk behavior.

Several studies have nonetheless attempted to compare trends observed by nanoindentation with results from tensile or compressive testing, primarily to assess the consistency of microstructural evolution across length scales. For instance, reductions in nano-hardness following heat treatment have been reported alongside decreases in calculated yield strength, reflecting the common underlying softening mechanism caused by microstructural coarsening [85]. Another study compared nano-, micro-, and macro-mechanical data and found that the decrease in nano-hardness from 1.82 GPa to 1.52 GPa was accompanied by a similar reduction in yield and ultimate tensile strength and a concurrent increase in elongation [47]. Such findings underscore that the evolution of microstructure governs both local and global mechanical properties.

Importantly, bulk mechanical behavior in LPBF AlSi10Mg emerges from the collective response of multiple microstructural regions, which interact under complex stress states [30,105]. In contrast, nanoindentation probes a highly localized deformation field beneath the indenter, where the plastic zone beneath is shaped by microstructural ori-

entation [186], and grain boundaries [187], all of which also control macroscopic yield strength and strain-hardening behavior in LPBF AlSi10Mg [188]. As a consequence, when nanoindentation is interpreted in conjunction with bulk testing and detailed microstructural characterization, it can help to rationalize observed macroscopic trends and identify the microstructural origins of strengthening or softening.

8. Conclusions

LPBF-processed AlSi10Mg remains one of the most extensively studied systems in Al-based AM, yet its nanoscale mechanical behavior remains an area of active investigation. Nanoindentation has proven to be a powerful method for exploring these small-scale phenomena, enabling the assessment of local hardness, elastic modulus, and time-dependent deformation that are inaccessible through conventional bulk testing. More importantly, it enables discrimination between distinct microstructural states that arise from LPBF-specific thermal histories, a capability that is particularly relevant for materials processed under rapid solidification and cyclic reheating conditions.

The existing literature consistently shows that the nanomechanical response of LPBF-AlSi10Mg is strongly linked to its processing history. Variations in laser power, scanning speed, energy density, and scan strategy alter solidification rate, Si segregation, and cellular refinement, which in turn control local hardness and modulus. Additional parameters such as layer thickness, hatch spacing, and build orientation further contribute to anisotropy and spatial variability in mechanical response. In the as-built condition, the refined cellular-dendritic α -Al network reinforced by Si segregation gives rise to elevated hardness, whereas post-processing treatments relieve residual stresses and promote coarsening or spheroidization of the Si network, generally leading to mechanical softening at the nanoscale. The melt-pool architecture introduces distinct anisotropy with varying hardness and modulus between core, boundary, HAZ, reflecting the thermal gradients developed during laser scanning. Within this framework, nanoindentation functions not only as a testing technique, but as a sensitive probe of local thermal history, microstructural evolution, and phase distribution.

Despite these advances, research in this area remains limited. The small-probed volume restricts statistical representativeness and necessitates careful selection of indentation depth and array spacing relative to microstructural length scales. Variations in surface preparation and experimental methodology further contribute to the wide scatter reported in the literature. Meaningful structure–property correlations therefore require integration with complementary characterization techniques capable of resolving crystallographic orientation, phase morphology, and defect populations. The lack of standardized testing practices and incomplete reporting of processing parameters continue to hinder direct comparison across studies and limit the ability to generalize nano-mechanical trends to larger length scales.

Overall, nanoindentation plays a critical role in advancing the understanding of LPBF AlSi10Mg by enabling direct correlations between local composition, microstructural features, and mechanical response. Beyond its role as a characterization tool, nanoindentation provides a pathway for informed process optimization, heat-treatment design, and alloy development tailored to AM. For researchers, it offers mechanistic insight into solidification and phase evolution. For process engineers, it enables evaluation of scan strategies and thermal histories, and for alloy designers, it provides a rapid screening method for new compositions and AM-specific alloy concepts. Future progress will depend on the integration of nanoindentation with multiscale microstructural characterization and computational modeling, ultimately supporting the development of more reliable and application-ready AM aluminum components.

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Abbreviations

The following abbreviations are used in this manuscript:

AA	Artificial aging
AB	As-built
AC	Air-cooled
AM	Additive manufacturing
CSM	Continuous stiffness measurement
ECAP	Equal channel angular pressing
EBSD	Electron backscatter diffraction
FQ	Furnace-quenched
HAZ	Heat-affected zones
HIP	Hot isostatic pressing
HT	Heat treatment
ISE	Indentation size effect
L-DED	Laser Directed Energy Deposition
LoF	Lack-of-fusion
LPBF	Laser Powder Bed Fusion
N/A	Not available
SA	Solution annealing
SEM	Scanning electron microscopy
TEM	Transmission electron microscopy
VED	Volumetric energy density
WQ	Water-quenched

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