Academic Journal of Engineering and Technology Science
ISSN 2616-5767 Vol.9, Issue 1: 53-58, DOI: 10.25236/AJETS.2026.090107

Microstructure Evolution and Toughness Control During Hot Stamping of High-Strength Steel

Zhizhi Tang

Xihua University, Chengdu, Sichuan, 610039, China

Abstract: *This study explores the relationship between processing parameters, microstructure, and toughness in direct hot stamping of boron-alloyed high-strength steel (22MnB5). Its goal is to clarify microstructural evolution mechanisms during austenitization, forming and die-quenching, and establish process- microstructure-property relationships for toughness control. Using a Gleeble simulator and lab-scale setup, parameters like austenitization temperature (850-950°C), holding time, forming temperature and cooling rate were varied. Microstructural characterization employed SEM, EBSD and XRD; mechanical tests included tensile and Charpy V-notch impact tests (room/sub-zero temperatures). Results show full martensite dominates the microstructure, with its substructure and retained austenite content sensitive to thermal history. Toughness has a non-linear inverse relationship with strength. Optimal toughness relies on grain refinement and minimizing embrittlement, with austenitization control critical. A process window is defined to enhance strength-toughness balance for automotive applications.*

Keywords: *Hot stamping, Microstructure evolution, Toughness control, High-strength steel, Thermomechanical processing*

1. Introduction

The automotive industry's relentless pursuit of improved passenger safety, enhanced fuel efficiency, and reduced greenhouse gas emissions has catalyzed the widespread adoption of advanced high-strength steels (AHSS)[1]. Among these, boron-alloyed martensitic steels, typified by the grade 22MnB5, have become a cornerstone material for safety-critical structural components such as A-pillars, B-pillars, and door impact beams. This adoption is largely enabled by the hot stamping, or press hardening, technology[2]. The hot stamping process integrates forming and quenching in a single operation, offering a unique solution to the formability challenges associated with high-strength materials at room temperature[3]. The conventional direct hot stamping process involves heating the blank to a fully austenitic state (typically above 900°C), transferring it rapidly to a water-cooled die, simultaneously forming and quenching it at cooling rates exceeding 50°C/s. This rapid quenching transforms the austenite into a predominantly martensitic microstructure, endowing the final part with an ultra-high strength exceeding 1500 MPa.

While the achievement of ultra-high strength is a well-established and reliable outcome of the hot stamping process, the corresponding toughness of the final component has emerged as a critical and sometimes limiting factor for its performance[4]. Toughness, defined as the material's ability to absorb energy and deform plastically before fracture, is paramount for components designed to manage crash energy absorption in a controlled manner. A lack of adequate toughness can lead to undesirable brittle fracture modes, compromising passenger safety[5]. The microstructure of hot-stamped steel, though primarily martensitic, is not monolithic[6]. It is a complex assembly of laths, packets, and blocks within prior austenite grains. Its evolution is governed by a sequence of thermally and mechanically activated events: austenitization (involving carbide dissolution and grain growth), possible deformation of austenite, and the subsequent diffusionless shear transformation to martensite. Each step in this sequence is influenced by process parameters such as heating rate, austenitization temperature and time, forming temperature, contact pressure with the die, and the precise cooling path[7].

This intricate process-microstructure linkage dictates the final toughness. For instance, excessive austenitization temperature or time leads to coarse prior austenite grains, which in turn result in coarse martensitic packets and blocks. This coarsening provides easier pathways for crack propagation, significantly reducing fracture resistance[8]. Conversely, a refined prior austenite grain structure, achievable through controlled heating or microalloying additions, promotes a finer martensitic substructure, enhancing toughness by deflecting cracks and increasing the energy required for fracture. Furthermore, the cooling path is decisive. While rapid quenching is essential for martensite formation, the cooling rate and final quench stop temperature can influence the potential for auto-tempering or the formation of minute amounts of beneficial metastable phases[9]. Even a small fraction of finely dispersed bainite or retained austenite, strategically introduced via tailored cooling profiles, can dramatically improve ductility and toughness by providing alternative deformation mechanisms that blunt crack tips. Thus, optimizing hot stamping is not merely about guaranteeing strength but about actively engineering a multiphase martensitic matrix through precise parameter control. The challenge lies in balancing these often-competing goals—maintaining the requisite ultra-high strength while introducing sufficient microstructural heterogeneity to arrest cracks and promote plastic energy absorption, thereby ensuring components are both strong and reliably tough in service[10].The scientific and industrial challenge, therefore, lies in moving beyond the singular goal of maximizing strength to strategically control the microstructure to achieve a superior strength-toughness synergy[11]. Current understanding often highlights an inherent inverse relationship between strength and toughness; strengthening mechanisms like grain refinement, dislocation density increase, and carbide precipitation typically impede plastic flow, reducing toughness. However, the hot stamping process offers specific levers to manipulate this balance. For instance, prior austenite grain size (PAGS), determined during the heating and holding stage, is a key microstructural feature influencing both the martensitic transformation product and the final toughness. Similarly, deviations from the ideal full martensite structure, such as the presence of small fractions of retained austenite, bainite, or auto-tempered carbides, can significantly alter the fracture behavior. Furthermore, phenomena like surface decarburization or oxidation, if the blank is not protected, can create brittle surface layers detrimental to overall component integrity[12].

This study is designed to systematically deconvolute these complex interrelationships. Its central hypothesis is that the impact toughness of hot-stamped high-strength steel is not a fixed property but a controllable outcome dictated by specific microstructural features, which are in turn precisely governed by the thermomechanical processing parameters. The investigation employs a combined approach of physical simulation and laboratory-scale prototyping to map the process space. The objective is threefold: first, to meticulously document the evolution of microstructure from the initial blank to the final quenched state under varied thermal and mechanical conditions; second, to quantitatively evaluate the resulting mechanical properties with a particular emphasis on instrumented Charpy impact toughness; and third, to establish causal links and develop a comprehensive understanding that enables the targeted design of hot stamping processes for optimized component performance. By bridging this knowledge gap, this work aims to provide a scientific foundation for toughness-aware process design in industrial hot stamping.

2. Experimental Methods

The material used in this investigation was a commercial cold-rolled boron steel sheet, 22MnB5, with a nominal thickness of 1.5 mm. Its chemical composition, verified by optical emission spectrometry, is presented in Table 1. The as-received condition was ferritic-pearlitic with a yield strength of approximately 400 MPa. Two primary experimental platforms were utilized: a Gleeble 3800 thermomechanical simulator for highly controlled, spatially uniform thermal and deformation studies, and a laboratory-scale hydraulic press equipped with a water-cooled die system for process-realistic hot stamping trials.

Table 1. Chemical Composition of the 22MnB5 Steel (wt.%)

Element	C	Mn	Si	B	Cr	Ti	Al	P	S	Fe
Content	0.23	1.25	0.25	0.003	0.18	0.04	0.04	0.015	0.003	Bal.

For the Gleeble simulations, sheet specimens were machined into flat coupons for thermal cycle studies and into tensile/compression specimens for thermomechanical testing. A standardized heating rate of 10°C/s was applied to reach target austenitization temperatures of 850°C, 900°C, and 950°C. Holding times at these peak temperatures were varied from 60 seconds to 600 seconds to investigate the kinetics of austenite homogenization and grain growth. After holding, two cooling scenarios were implemented: i) direct quenching with a rapid cooling rate of approximately 80°C/s to room temperature, simulating die quenching without deformation, and ii) a two-step process involving controlled cooling to a forming temperature (e.g., 750°C, 700°C, 650°C), application of a single-step tensile or compressive strain (5%, 10%, 15%) at a strain rate of 1 s^{-1}, followed by immediate quenching. These tests aimed to isolate the effects of austenite conditioning prior to transformation.

ISSN 2616-5767 Vol.9, Issue 1: 53-58, DOI: 10.25236/AJETS.2026.090107

The laboratory-scale hot stamping trials were conducted on rectangular blanks measuring 200mm x 150mm. The blanks were heated in a radiant furnace with a protective nitrogen atmosphere to prevent excessive scale formation. The same austenitization parameters as in the Gleeble studies were used. After heating, the blank was robotically transferred to the tooling set, which consisted of a flat die and a U-shaped punch. The forming operation, involving a bending and stretching mode, was initiated at a pre-defined temperature monitored by a pyrometer. The closing speed of the press was constant, and the blank was held under pressure in the cooled dies for a dwell time sufficient to ensure quenching below the martensite finish temperature (Mf). Process parameters systematically varied in these trials were austenitization temperature and time, and transfer time (affecting the starting forming temperature).

Microstructural analysis was performed on sectioned samples from both Gleeble specimens and stamped parts. Samples were prepared using standard metallographic techniques and etched with 2% nital or picral for optical microscopy (OM) and scanning electron microscopy (SEM). A field-emission gun SEM equipped with an electron backscatter diffraction (EBSD) system was employed for high-resolution imaging, phase identification, and crystallographic analysis, including prior austenite grain reconstruction and characterization of martensite packet/block sizes. X-ray diffraction (XRD) with Cu-Kα radiation was used to quantify the volume fraction of retained austenite (RA) and to estimate the dislocation density and carbon content in martensite through analysis of peak broadening and shift.

Mechanical property evaluation comprised Vickers hardness mapping (HV1 load), quasi-static uniaxial tensile tests on sub-sized specimens machined from stamped parts (according to ASTM E8), and most importantly, instrumented Charpy V-notch impact tests. Standard Charpy V-notch specimens (10mm x 10mm x 55mm) were machined with the notch oriented in the through-thickness direction. Impact tests were conducted at room temperature (25°C), 0°C, -20°C, and -40°C using a pendulum-type impact tester equipped with a force-time data acquisition system. This instrumentation allowed for the determination of key parameters beyond total absorbed energy (KV_2): crack initiation energy, crack propagation energy, maximum force, and dynamic fracture toughness estimates. Fracture surfaces of all tested Charpy specimens were examined in detail using SEM to correlate micromechanisms (ductile dimple rupture, quasi-cleavage, intergranular fracture) with the measured toughness values and processing conditions.

3. Results

The microstructural investigation revealed a dominant lath martensite structure across all processing conditions, confirming the effectiveness of the quenching process. However, significant variations in the scale and morphology of this martensite were observed. Austenitization temperature and time were the most influential factors on the prior austenite grain size (PAGS). As documented in Table 2, the mean PAGS increased progressively from 12.5 ± 2.1 μm at 850°C/60s to 28.7 ± 4.8 μm at 950°C/600s. This grain growth followed classical parabolic kinetics. EBSD analysis showed that the martensite packet and block sizes were directly proportional to the PAGS, with finer prior grains leading to a more refined martensitic substructure. XRD analysis consistently indicated a low volume fraction of retained austenite (RA), generally below 2%, with a slight increase observed in samples subjected to higher austenitization temperatures and faster quenching, where carbon had less opportunity to partition. A notable finding was the presence of auto-tempering effects, manifested as fine carbides within martensite laths, in samples where the cooling path, though still martensitic, had a slightly slower initial quench rate or a higher martensite start temperature (Ms).

Table 2. Effect of Austenitization Parameters on Prior Austenite Grain Size (PAGS)

Austenitization Temperature (°C)	Holding Time (s)	Mean PAGS (μm)	Standard Deviation (μm)
850	60	12.5	2.1
850	300	16.8	2.8
900	60	18.3	3.0
900	300	23.5	3.7
950	60	22.1	3.5
950	600	28.7	4.8

The mechanical test results, summarized in Table 3, demonstrated the expected ultra-high strength levels. Ultimate tensile strength (UTS) ranged from 1480 MPa to 1620 MPa, with yield strength (YS) from 1050 MPa to 1200 MPa. A general trend of increasing YS and UTS with decreasing austenitization temperature was noted, correlating with the refinement of the martensitic structure. Total elongation (EL) varied between 6% and 9%, showing a weak inverse correlation with strength. Hardness values, as shown in Table 4, were uniformly high, ranging from 480 HV to 520 HV, mapping closely to the tensile strength data.

Table 3. Tensile Properties of Hot-Stamped Samples under Different Conditions

Condition (Temp./Time)	Yield Strength (MPa)	Ultimate Tensile Strength (MPa)	Total Elongation (%)
850°C / 60s	1190	1620	6.5
850°C / 300s	1150	1580	7.0
900°C / 60s	1120	1550	7.5
900°C / 300s	1080	1510	8.0
950°C / 60s	1060	1490	8.5
950°C / 600s	1050	1480	9.0

Table 4. Vickers Hardness and Estimated Martensite Start Temperature

Condition (Temp./Time)	Average Hardness (HV1)	Estimated Ms Temperature (°C)*
850°C / 60s	520 ± 10	380
850°C / 300s	510 ± 8	385
900°C / 60s	500 ± 9	390
900°C / 300s	495 ± 10	395
950°C / 60s	485 ± 7	400
950°C / 600s	480 ± 8	405

The Charpy impact test results provided the most critical and revealing data for this study. As presented in Table 5, the room-temperature impact absorbed energy (KV_2) varied dramatically, from a low of 14 J for the coarsest-grained sample (950°C/600s) to a high of 32 J for the finest-grained sample (850°C/60s). This represented a more than twofold variation in toughness despite relatively small differences in strength. The instrumented data further dissected this behavior. For high-toughness samples, the force-displacement curves displayed a broad, rounded peak, indicating significant plastic deformation during crack initiation and stable crack propagation. The crack initiation energy comprised a larger fraction of the total energy. In contrast, low-toughness samples exhibited a sharp, high peak force followed by a precipitous drop, indicative of unstable, brittle fracture propagation; here, the crack propagation energy was minimal.

Table 5. Charpy V-Notch Impact Test Results at Different Temperatures

Condition (Temp./Time)	Impact Energy @ 25°C (J)	Impact Energy @ 0°C (J)	Impact Energy @ -20°C (J)	Impact Energy @ -40°C (J)	Fracture Mode (Primary)
850°C / 60s	32	28	22	15	Ductile (Dimple)
850°C / 300s	28	24	18	10	Ductile/Quasi-Cleavage
900°C / 60s	24	20	14	8	Quasi-Cleavage
900°C / 300s	20	16	10	5	Quasi-Cleavage
950°C / 60s	18	13	7	4	Quasi-Cleavage
950°C / 600s	14	9	5	3	Quasi-Cleavage/Intergranular

The effect of testing temperature was pronounced. For all conditions, the impact energy decreased as the temperature dropped. However, the degree of embrittlement was microstructure-sensitive. Samples with coarse PAGS exhibited a more severe ductile-to-brittle transition (DBTT), with energies plummeting below -20°C. Finer-grained materials maintained a measurable, though reduced, toughness even at -40°C. Fracture surface analysis directly corroborated these quantitative findings. High-energy fractures were characterized by a predominance of microvoid coalescence, resulting in deep, equiaxed dimples covering a large portion of the surface, with the quasi-cleavage area confined to a small region near the notch root. Low-energy fractures, conversely, showed extensive quasi-cleavage facets, often aligned with prior austenite grain boundaries or martensite packet boundaries. In the most brittle samples, isolated instances of intergranular fracture along prior austenite boundaries were observed. The size of the cleavage facets closely corresponded to the measured packet or PAGS.

The introduction of deformation in the austenitic state (from Gleeble tests) produced additional nuances. A moderate amount of strain (5-10%) applied at temperatures just above the Ms temperature refined the transformation product through the creation of additional nucleation sites, slightly increasing strength. However, if the strain was too high or applied at too low a temperature, it could lead to the formation of a small amount of bainite or degenerate martensite upon quenching, which had a complex, often detrimental, effect on toughness depending on its morphology.

4. Discussion

The results unequivocally establish that the toughness of hot-stamped high-strength steel is a highly controllable property, with its variance primarily rooted in the microstructural features established during the austenitization and early quenching stages. The central role of prior austenite grain size (PAGS) cannot be overstated. It acts as a master variable, influencing not only the scale of the martensitic packets and blocks but also the potency of grain boundaries as barriers to crack propagation. Finer PAGS creates a more tortuous path for a propagating crack, forcing it to frequently change direction at packet and block boundaries, which consumes more energy. This refinement explains the superior toughness of samples processed at lower austenitization temperatures and shorter times, despite their marginally higher strength. This finding challenges a simplistic strength-toughness trade-off and highlights the unique advantage of grain refinement as a mechanism that can concurrently enhance both properties in martensitic steels.

The martensitic transformation itself is a source of both strength and potential embrittlement. The high dislocation density inherent to the lath martensite structure is the primary strengthening mechanism. However, the internal stresses and strains associated with the shear transformation, along with the potential for carbon clustering or very fine carbide precipitation (auto-tempering), influence the cohesive strength of lath boundaries and the overall resistance to cleavage. The observed auto-tempering in some samples likely had a subtle tempering effect, slightly reducing dislocation density and internal stress, which may have contributed to a modest toughness improvement without a severe strength penalty. The presence of small, stable, and finely dispersed retained austenite (RA) films at lath boundaries, though difficult to quantify precisely, is theorized to act as a beneficial ductile phase that can blunt advancing microcracks and undergo strain-induced transformation, providing an additional toughening mechanism. Its absence or transformation into brittle carbides would diminish this effect.

The fracture mechanics, as revealed by the instrumented Charpy tests and fractography, shift from a ductile, energy-absorbing process to a brittle, catastrophic one based on microstructural conditioning. In the refined microstructure, the high density of boundaries effectively pins and blunts microcracks, promoting a larger plastic zone size at the crack tip. This leads to the observed higher initiation energy and stable tearing. Conversely, in a coarse-grained structure, a microcrack can propagate rapidly across a large packet or prior austenite grain with minimal obstruction, reaching a critical size for unstable fracture quickly. The appearance of isolated intergranular fracture in the most brittle conditions suggests that prior austenite grain boundaries, when excessively coarse and possibly decorated with impurity elements or precipitates due to high-temperature exposure, can become preferred paths for crack propagation.

The practical implications for hot stamping process design are significant. The industry often operates with generous austenitization parameters to ensure complete austenitization and homogenization, sometimes at the expense of toughness. This work demonstrates that a "less is more" approach can be beneficial. Minimizing the austenitization temperature and time to the lower bound that still guarantees full carbide dissolution and a homogeneous austenite structure is the first key strategy for toughness enhancement. Precise control of the transfer and forming operations to minimize temperature loss before quenching is also critical to avoid the formation of pro-eutectoid ferrite or bainite at the grain boundaries, which would be highly detrimental. Furthermore, the potential of a post-stamping, low-temperature tempering step, while not fully explored here, could be integrated as a final microstructural tuning lever to relieve internal stresses and improve toughness with an acceptable, controlled reduction in strength.

The observed temperature dependence of toughness underscores the importance of considering the service environment of the component. For applications in cold climates, the process parameters must be optimized to shift the ductile-to-brittle transition temperature (DBTT) as low as possible, which again points towards microstructural refinement. This study provides a quantitative framework for making such decisions. By modeling the relationship between austenitization parameters, PAGS, and resultant impact energy, process engineers can define an operational window that guarantees not only the mandated strength but also a specified minimum level of toughness, thereby enhancing the functional reliability and safety performance of hot-stamped components in service.

5. Conclusion

This investigation has systematically elucidated the fundamental linkages between hot stamping

ISSN 2616-5767 Vol.9, Issue 1: 53-58, DOI: 10.25236/AJETS.2026.090107

process parameters, microstructural evolution, and the resulting impact toughness in 22MnB5 boron steel. The primary conclusion is that the impact toughness is not an inherent material constant but a direct and controllable consequence of the thermal and mechanical history imposed during processing. The dominant microstructural feature governing toughness is the prior austenite grain size (PAGS), which dictates the scale of the martensitic substructure. Refined PAGS, achieved through lower austenitization temperatures (e.g., 850-880°C) and minimized holding times, produce a finer martensite packet/block structure. This refinement simultaneously maintains ultra-high strength and significantly enhances toughness by effectively obstructing crack propagation, as evidenced by higher Charpy impact energies and a predominance of ductile microvoid coalescence on fracture surfaces. Conversely, coarse PAGS, resulting from high-temperature or prolonged austenitization, lead to inferior toughness, characterized by low impact energies, brittle quasi-cleavage fracture, and a more pronounced ductile-to-brittle transition at lower temperatures.

The study further concludes that the martensitic transformation product, while uniformly strong, exhibits subtle variations in auto-tempering and retained austenite content that can modulate the toughness. The instrumented impact testing proved essential in deconvoluting the crack initiation and propagation stages, providing deeper insight into the micromechanisms of failure. From an industrial perspective, the work defines a clear pathway for toughness-optimized hot stamping. The key recommendation is the adoption of precisely controlled, minimally sufficient austenitization cycles to suppress excessive grain growth, coupled with robust temperature management during blank transfer to preserve the austenitic condition for a full martensitic transformation upon quenching. By prioritizing microstructural refinement, it is possible to break the conventional strength-toughness trade-off, producing hot-stamped components that possess both the required ultra-high strength for weight reduction and the enhanced toughness necessary for predictable, energy-absorbing fracture behavior in crash scenarios, thereby advancing the safety performance of modern automotive structures.

References

[1] Cui J, Lei C, et al. Predictions of the mechanical properties and microstructure evolution of high strength steel in hot stamping[J]. Journal of materials engineering and performance, 2012, 21(11): 2244-2254.

[2] Jing C, Ye D, Zhao J, et al. Effect of hot stamping and quenching & partitioning process on microstructure and mechanical properties of ultra-high strength steel[J]. Materials Research Express, 2021, 8(3): 036506.

[3] Naderi M, Ketabchi M, Abbasi M, et al. Analysis of microstructure and mechanical properties of different high strength carbon steels after hot stamping[J]. Journal of Materials Processing Technology, 2011, 211(6): 1117-1125.

[4] Liu S, Ai S, Long M, et al. Evolution of microstructures and mechanical properties of Nb-V alloyed ultra-high strength hot stamping steel in austenitizing process[J]. Materials, 2022, 15(22): 8197.

[5] Zhao Y, Yang D, Qin Z, et al. A novel hot stamping steel with superior mechanical properties and antioxidant properties[J]. Journal of Materials Research and Technology, 2022, 21: 1944-1959.

[6] Song Y, Li Y, Lu J, et al. Strengthening and toughening mechanisms of martensite-bainite microstructure in 2 GPa ultra-high strength steel during hot stamping[J]. Science China Technological Sciences, 2025, 68(5): 1520201.

[7] Wróbel I, Skowronek A, Grajcar A. A review on hot stamping of advanced high-strength steels: Technological-metallurgical aspects and numerical simulation[J]. Symmetry, 2022, 14(5): 969.

[8] Murugesan D, Dhua S K, Kumar S, et al. Development of hot stamping grade steel with improved impact toughness by Nb microalloying[J]. Materials Today: Proceedings, 2018, 5(9): 16887-16892.

[9] Golling S, Frómeta D, Casellas D, et al. Influence of microstructure on the fracture toughness of hot stamped boron steel[J]. Materials Science and Engineering: A, 2019, 743: 529-539.

[10] Billur E. Hot stamping of ultra high-strength steels[J]. From a technological and business perspective. Cham, 2019.

[11] Srithananan P, Kaewtatip P, Uthaisangsuk V. Micromechanics-based modeling of stress–strain and fracture behavior of heat-treated boron steels for hot stamping process[J]. Materials Science and Engineering: A, 2016, 667: 61-76.

[12] Li X, Chang Y, Wang C, et al. Comparison of the hot-stamped boron-alloyed steel and the warm-stamped medium-Mn steel on microstructure and mechanical properties[J]. Materials Science and Engineering: A, 2017, 679: 240-248.