

Effects of Nitrogen and Vanadium Additions on Microstructures, Tensile and Impact Properties of a Cast Austenitic High-Manganese Steel for Cryogenic Application

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Research Article

Keywords: cast high Mn steel, nitrogen & vanadium additions, mechanical properties, TWIP

Posted Date: August 20th, 2026

DOI: <https://doi.org/10.21203/rs.3.rs-10452903/v1>

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Additional Declarations:

No competing interests reported.

Tables 1 to 4 are available in the supplementary files section.

Version of Record: A version of this preprint was published at International Journal of Metalcasting on October 14th, 2026. See the published version at <https://doi.org/10.1007/s40962-026-02154-4>.

**EFFECTS OF NITROGEN AND VANADIUM ADDITIONS ON
MICROSTRUCTURES, TENSILE AND IMPACT PROPERTIES OF A
CAST AUSTENITIC HIGH-MANGANESE STEEL FOR CRYOGENIC
APPLICATION**

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Abstract

The effects of nitrogen (N) and vanadium (V) additions on the microstructures and tensile and impact properties of a cast austenitic high-manganese (Mn) steel for cryogenic applications were investigated using optical microscopy (OM), electron backscatter diffraction (EBSD), field-emission scanning electron microscopy-energy-dispersive spectroscopy (FE-SEM-EDS), X-ray diffraction (XRD), and martensite measurement via feritscope. The enhanced strength, ductility, and impact toughness of the HMN015V alloy containing N and V, which exhibits a

much higher stacking fault energy (SFE) of 32 mJ/m² than the experimental HMBASE alloy (25.2 mJ/m²) and the commercial CF3 alloy (16.2 mJ/m²), can be attributed to three primary factors: (1) interstitial solid-solution strengthening from its medium N content (0.13 wt.%) and high carbon (C) content (0.48 wt.%); (2) precipitation hardening due to vanadium nitride (VN); and (3) Hall–Petch-type strengthening resulting from numerous twin grain boundaries and fine twin grains formed via the twinning-induced plasticity (TWIP) mechanism during tensile and impact deformations.

Keywords: *cast high Mn steel, nitrogen & vanadium additions, mechanical properties, TWIP*

I n t r o d u c t i o n

Recently, austenitic high-Mn steels (AHMSs) have been widely used because of their exceptional mechanical properties, i.e., an outstanding combination of high strength and substantial ductility at cryogenic temperature (−196°C).^{1–3} These unique characteristics have led to their widespread adoption across various industries, including automotive manufacturing and cryogenic facilities such as liquefied natural gas (LNG) ships, LNG storage tanks and containers, and pipelines, in response to the global demands of LNG, N, oxygen (O), and hydrogen (H).

As of 2019, the International Maritime Organization (IMO) had approved only five materials for LNG tanks: a 36 wt.% Nickel (Ni)-containing alloy steel (ASTM F1684-K93604), an 8.5 wt.% Ni-containing 304L austenitic stainless steel (ASS) (ASTM A240-S30403), a 9 wt.% Ni-containing alloy steel (ASTM A553-Type 1), an aluminum (Al)-magnesium (Mg) alloy (ASTM B209M-5083), and a 24 wt.% Mn-containing AHMS⁴ (A1106/A1106M⁵).

The excellent combination of high strength and substantial ductility in AHMSs is attributed to the TWIP effect. During deformation, mechanical twinning occurs gradually in the austenite (γ) matrix, with the twinned crystal planes acting as sub-boundary-like obstacles to dislocation motion.⁶⁻¹⁰ This phenomenon results in Hall-Petch-type strengthening of the material as strain progresses.¹¹⁻¹⁴

The deformation mechanisms and mechanical properties of the AHMSs are strongly influenced by their SFE,¹⁵ a crucial nucleation parameter that determines whether twinning, martensitic transformation, or dislocation glide predominates during deformation. It is well established that deformation martensite via transformation-induced plasticity (TRIP), deformation twinning by the TWIP effect, and dislocation gliding occur within specific SFE ranges of the AHMSs: $< 20 \text{ mJ/m}^2$, $20\text{--}50 \text{ mJ/m}^2$, and $> 50 \text{ mJ/m}^2$, respectively.^{1, 14, 16-19}

It is well known that V microalloying can effectively enhance the yield strength of wrought AHMS strip through grain refinement and precipitation strengthening.²⁰ An apparent increment of ~247 MPa in the yield strength of Fe-17Mn-0.9C-0.45V steel has been reported as a result of V addition.²¹ A greater amount of V addition is required to attain the effective strengthening effect of V in the wrought AHMSs hot-rolled plate. However, limited literature is available on enhancing the yield strength of cast AHMSs using the V alloying method, especially regarding the strengthening mechanism, its effect on cryogenic impact toughness, and the optimal V addition for improving mechanical performance.

Solid-solution strengthening effect has been reported to be much more pronounced for interstitial elements, such as C²² and N,²³ than for substitutional elements, such as silicon (Si)²⁴ and Al.²⁵ Specifically, N exhibits the most pronounced interstitial solid-solution strengthening effect, exceeding that of C. Rawers et al.²³ proposed a statistical analysis of the yield stress (σ) as a function of N concentration for interstitial solid-solution strengthening of wrought ASS sheet and plate. The following Eqn. 1 was used for statistical analysis at 27°C:

$$\sigma \text{ (YS)} = 215 + 939 \times \%N + 61 \times \%N^{0.5} \quad \text{Eqn. 1}$$

where %N represents the weight percentage of N in the steel. This relationship indicates that the yield stress increases as the N content, which contributes to interstitial solid-solution strengthening, increases.

The background and objectives of this study are as follows. First, investigations of AHMSs have focused primarily on wrought products, such as cold-rolled sheets and hot-rolled plates, rather than castings with greater thicknesses. Considering the mass effect, there is a paucity of studies that quantitatively elucidate the room- and cryogenic-temperature deformation behaviors of AHMS castings associated with microstructural changes. Second, a previous study by the present authors,²⁶ published in this journal, reported that the HMBASE base alloy, without V and N additions, exhibited superior tensile properties compared to CF3 ASS, with yield and tensile strengths 1.31 and 1.28 times higher, respectively.

However, because studies elucidating the effects of N and V additions on the tensile and impact properties of the microstructures of cast AHMSs for cryogenic applications are difficult to locate, further in-depth study is essential. Hence, it is necessary to improve the tensile and impact properties substantially and to quantitatively verify the deformation mechanisms of a cast N- and V-containing AHMS compared with the cast HMBASE alloy and the CF3 cast ASS. The CF3 alloy, rather than an alloy such as the 36 wt.% high Ni-containing steel, was chosen for the comparison are for the following reasons: ASSs are generally less sensitive, more inexpensive, and more applicable to cryogenic environments than the high-Ni-containing steel, and, thus, they are recommended to apply to liquid H and N environments as well as LNG environment.²⁷⁻²⁸ Temperature of liquid H is -253°C , considerably lower than that of LNG (-162°C) and liquid N (-196°C). At present, 316L ASS is

commonly used for liquid H storage vessels; however, Kawasaki Heavy Industries in Japan is reported to have applied 8.5 wt.% Ni-containing 304L ASS to Suiso Frontier, a liquid H carrier.²⁹

This study compares the tensile and impact properties of a cast N- and V-containing AHMS with a cast AHMS (base alloy) and CF3 cast ASS through various tests and analyses: thermodynamic simulation of phase diagrams, solidification modes, equilibrium phase fractions, and SFE; room-temperature tensile testing; cryogenic-temperature Charpy V-notch (CVN) impact testing; metallographic examination; ferrite and martensite measurements using XRD test and feritscope; EBSD; and fracture surface and inclusion analyses via FE–SEM–EDS.

E x p e r i m e n t a l

Thermodynamic Calculation of Phase Diagram, Solidification Mode, Equilibrium Phase Fractions, and Stacking Fault Energy

The phase diagram, solidification mode, equilibrium phase fractions, and SFE were calculated against the temperature for the experimental alloys using a commercial JMatPro 7.0 (Sente Software Ltd., United Kingdom) software package.

Materials and Test Methods

Austenitic high-Mn alloys with different alloying elements were prepared by melting 300 kg in a high-frequency induction furnace at 1545 °C. For convenience, the experimental alloys are designated as HMBASE²⁶ (no additions of N and V), HMN015 (N addition) and HMN015V (additions of N and V), respectively. The melt was deoxidized with 300 gr. of a calcium (Ca) bearing Si (CaSi) in the furnace and was poured into a ladle. A double keel block (Figure 1) was cast by pouring the deoxidized melt at 1473 °C into an artificial silica (SiO₂) mold manufactured using the carbon dioxide (CO₂) molding process according to the American Society for Testing and Materials (ASTM) E 8.³⁰ The sound portion of 31.75 mm (1 ¼ in) thickness, obtained after removing the riser, underwent solution heat treatment at 950 °C for 1.5 h, followed by water cooling. Table 1 presents the chemical compositions of the experimental HMBASE²⁶, HMN015, HMN015V and commercial CF3 cast ASS (CF3 alloy) registered to ASTM A 351 (UNS J92500),³¹ which served as a reference specimen. The CF3 alloy is equivalent to AISI 304L ASS.

A tensile test was conducted at room temperature on the specimens with a gauge length of 50.8 mm, gauge diameter of 12.7 mm, and reduced section length of 57.15 mm at a crosshead speed of 1.5 mm/min, according to ASTM E 8. Charpy V-Notch impact tests were performed on 10 × 10 × 55 mm³ at -196 °C (liquid N₂) using a 300 J capacity test machine, following ASTM E 23.³² Both tensile and Charpy impact tests were repeated thrice for each data point.

The metallographic examination was conducted in compliance with ASTM E 3³³ (metallographic sample preparation) and ASTM E 407³⁴ (metal microetching) standards. The experimental HMBASE, HMN015, HMN015V alloys were ground, polished, and etched with 4 wt.% Nital solution for 10 s.^{35,36} The solution-annealed specimens' optical microstructures (OMs) were characterized using an optical microscope.

XRD analysis (Cu K α radiation, scan range: 30°–100° in 2 θ , scan rate: 2° min⁻¹, scan step: 0.02°) test was performed to identify the phases of the fcc γ -austenite (00-052-0512 PDF), bcc α -ferrite (03-065-4899 PDF), hcp- ϵ martensite (01-071-8285 PDF), and bcc α' -martensite (01-071-8285 PDF) before and after the tensile test. Integrated intensities of (111) γ , (200) γ , (220) γ , (311) γ , and (222) γ as the fcc γ -phase; (110) α , (200) α , (211) α , and (220) α as the bcc α -phase; (101) ϵ as the hcp ϵ -phase; and (110) α' , (200) α' , (211) α' , and (220) α' as the bcc α' -phase peaks were utilized for this XRD method.

Quantification of bcc α -ferrite and bcc α' -martensite content before and after tensile and impact tests was accomplished using a feritscope (FMP30 and a set of standards from Fischer®).

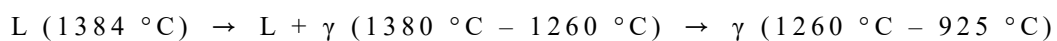
To verify the V precipitate, elemental mapping was carried out using FE-SEM. EBSD analyses (step size: 0.1 μ m) were conducted using FE-SEM-EDS before and after the tensile and impact tests and the resulting data were interpreted using orientation imaging microscopy (OIM) analysis

software provided by TexSEM Laboratories, Inc. (United States). Furthermore, fracture surfaces resulting from tensile and impact tests were analyzed using FE–SEM–EDS to determine whether the fracture mode was ductile or brittle.

Results and Discussions

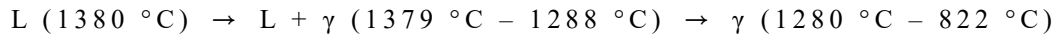
Thermodynamic Calculation of Phase Diagram, Solidification Mode, Equilibrium Phase Fractions, and Stacking Fault Energy

Figure 2 illustrates the phase diagrams of the HMBASE (Figure 2a ²⁶), HMN015 (Figure 2b) and HMN015V alloys (Figure 2c) calculated using a thermodynamic simulation software (JMatPro 7.0). As shown in Figure 2a, the HMBASE alloy exhibits a single γ -phase extending from 1260 °C to 925 °C, following primary solidification primarily as γ -phase from the liquid state. This behavior indicates that the solidification mode corresponds to the austenite (A) mode (Eqn. 2). The optimal solution annealing temperature shall exceed 925°C to avoid chromium (Cr) carbide formation ($M_{23}C_6$). This precaution is crucial as the presence of such carbides could potentially deteriorate the alloy's toughness and corrosion resistance.



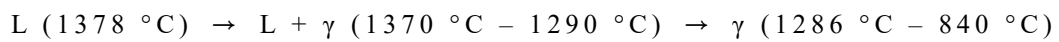
Eqn.2

Figure 2b illustrates that the HMN015 alloy containing N as a powerful austenitizing element exhibits much wider region for a single γ -phase extending from 1280 °C to 822 °C compared with the HMBASE alloy, following primary solidification primarily as γ -phase from the liquid state. This behavior indicates that the solidification mode corresponds to the austenite (A) mode (Eqn. 3). The behavior of Cr carbide precipitation of the HMN015 alloy is similar to the HMBASE alloy. However, carbonitride of $M_2(C,N)$ due to N addition to the HMN015 alloy can precipitate below 780 °C.



Eqn. 3

Figure 2c illustrates that the HMN015V alloy containing N and V exhibits a single γ -phase extending from 1286 °C to 840 °C which is similar to the HMN015 alloy, following primary solidification primarily as γ -phase from the liquid state. This behavior indicates that the solidification mode corresponds to the austenite (A) mode (Eqn. 4). The behavior of Cr carbide precipitation of the HMN015V alloy is similar to the HMBASE alloy. However, carbonitride of $M_2(C,N)$ can precipitate below 590 °C due to additions of N and V to the HMN015V alloy.



Eqn. 4

The SFE of the experimental HMBASE, HMN015 and HMN015V alloys were calculated using the thermodynamic simulation software (JMatPro 7.0), yielding a value of 25.2, 35.2, 32.0 mJ/m², respectively (Figure 3). It was reported that the formation of V carbide (VC) carbide consumed the C content in austenite matrix, resulting in a decrease of the SFE.^{37,38} Hence, It seems that the SFE of the HMN015V alloy containing N and V decreased by 3.2 due to V precipitate, compared with the HMN015 alloy containing N. V precipitate was explained detailly in the section of “Tensile Property at Room Temperature Associated with Microstructures”.

For the CF3 alloy, the SFE was determined following an empirical formula⁵³⁹ that incorporates the content (wt.%) of Ni, Cr, Mn, and molybdenum (Mo). The calculated SFE value for CF3 alloy was 16.2 mJ/m². However, the SFE of the CF3 alloy calculated using the thermodynamic simulation software (JMatPro 7.0) and composition of all elements in Table 1 increased to 23.3 mJ/m² due to the effect of Si, Cu, N, and C.

$$\Gamma(\text{SFE}) = -53 + 6.2 \times \text{Ni} + 0.7 \times \text{Cr} + 3.2 \times \text{Mn} + 9.3 \times \text{Mo}$$

Eqn. 5

The experimental HMBASE, HMN015 and HMN015V alloys with the SFE of 25.2, 35.2, 32.0 mJ/m², respectively, belongs to the deformation twinning region^{1, 14, 16-18} because it exists in the range of $20 \leq \Gamma(\text{SFE}) \leq 50$ due to the TWIP effect (Figure 3). As explained in our previous study²⁶, the CF3 alloy with insufficient γ -austenite stabilizers such as Ni, Mn,

and C can result in a metastable structure with small amounts of α -ferrite after casting. Especially, the CF3 alloy can form deformation-induced martensite in the γ -phase during room-temperature tensile and cryogenic-temperature impact test,¹⁹ irrespective of the SFEs calculated by the equation 4 and the thermodynamic simulation software.

Tensile Property at Room Temperature Associated with Microstructures

Figure 4 and Table 2 show the engineering tensile stress-strain behavior at room temperature. The yield strength of the HMN015 alloy due to the interstitial solid solution strengthening effect by N addition was 1.14 times higher than that of the HMBASE alloy whereas the tensile strength and elongation were similar to the HMBASE alloy. Furthermore, the yield strength of the HMN015 alloy was 1.49 times higher than that of the commercial CF3 alloy. Although the elongation of the HMN015 alloy was lower by 6 % than that of the CF3 alloy, it still demonstrated a high elongation of 57 %. Consequently, the overall tensile properties of the HMN015 alloy were superior to those of the experimental HMBASE and commercial CF3 alloys.

The HMN015V alloy due to synergistic effects of the interstitial solid solution strengthening effect by N addition and precipitation hardening by V addition exhibited yield and tensile strengths 1.24 and 1.12 times higher than those of the HMN015 alloy, respectively whereas the elongation were lower than the HMN015 alloy. Especially, the HMN015V

alloy exhibited yield and tensile strengths 1.85 and 1.42 times much higher than those of the CF3 alloy, respectively. Although the elongation of the HMN015V alloy was lower by 15 % than that of the CF3 alloy, it still demonstrated a high elongation of 48%. Consequently, the overall tensile properties of the HMN015V alloy were superior to those of the experimental HMBASE, HMN015 and commercial CF3 alloys.

In summary, Figure 5 illustrates the effects of C, N and V additions on the tensile properties on the tensile properties of the experimental HMBASE, HMN015, HMN015V and commercial CF3 alloys. As C and N contents as the alloying elements for the interstitial solid solution strengthening and V content as alloying element for precipitation hardening increase, the overall tensile properties increase.

Figure 6 presents the OMs of the HMBASE alloy before and after the tensile test at room temperature, conducted according to ASTM E 8. Prior to testing, the HMBASE showed a single γ -phase structure with large grain boundaries and a few inclusions (Figure 6a1), whereas post-tensile test showed a significant amount of mechanical twins formed within the γ -phase during the tensile deformation (Figure 6b1). The microstructure of the HMN015 alloy containing N was similar to that of the HMBASE alloy (Figure 6a2 and 6b2). In contrast, the HMN015V alloy containing N and V showed a single γ -phase structure with small grain boundaries and a few inclusions (Figure 6a3 and 6a3), irrespective of prior to testing and post-tensile test, compared with the HMBASE and HMN015 alloys. However,

as the V precipitate, mechanical twins and martensite were not discernable through OMs, these were subsequently analyzed using FE-SEM elemental mapping, EBSD analysis, XRD test, and the α' -martensite measurement using the feritscope.

Figure 7 presents FE-SEM elemental mapping of the solution heat-treated HMN015V alloy: (a) analysis at location of $\frac{1}{4}$ thickness and (b) analysis at center of double keel bock for test specimens (Figure 1). It was elucidated that V precipitates are V nitrides (VNs), irrespective of the different analysis locations. VNs, which form at grain boundaries, pin those and act as obstacles to grain growth. Consequently, VNs result in grain refinement, thereby precipitation hardening.

Figure 8 presents EBSD IPF and phase maps analyzing the formation of hcp ϵ - and bcc α' -martensite in the HMN015V alloy after the tensile testing. The HMN015V alloy revealed very small quantities of the ϵ - (yellow) and α' -martensite (green) within the γ -austenite matrix (red), irrespective of the different analysis locations.

Figure 9 shows EBSD grain boundaries maps after the tensile test at room temperature of (a) the HMBASE, (b) HMN015 and (c) HMN015V alloys according to ASTM E 8. The maps reveal numerous mechanical twin boundaries (blue) formed within the original grain boundaries of the γ -austenite due to the TWIP effect during the tensile deformation, irrespective of the additions of alloying element to the alloys. These

mechanical twin boundaries act as sub-boundary-like obstacles to dislocation motion,⁷⁻⁹ resulting in a favorable combination of high strength (via Hall-Petch-type strengthening)¹¹⁻¹⁴ and high ductility.

Figure 10 presents XRD patterns of fractured tensile specimens and Table 3 shows the content of each phase calculated using the equations investigated by Y. Onukil et al.⁴⁰ The HMBASE alloy revealed a full γ -austenite single phase before the tensile test (Figure 10a1), whereas it revealed a small amount of ϵ - and α' -martensite within the γ -austenite matrix after the tensile test (Figure 10a2). In contrast, the N-added HMN015 and N- and V-added HMN015V alloys with much higher SFEs showed a full γ -austenite single phase after the tensile test (HMN015: Figure 10b2, HMN015V: Figure 10c2), compared with the HMBASE alloy. These XRD-based phase identifications corroborate well with the EBSD analyses (Figure 8) though the quantity of each phase measured through XRD test is slightly different from that measured by EBSD analyses.

Table 4 present the α' -martensite content, calculated by Eqs. 5 and 6, following feritscope measurements of α -ferrite content before and after the room-temperature tensile test of the experimental and commercial alloys. Previous research has shown that feritscope readings (F) can be converted to actual α' -martensite contents ($C\alpha'$) using a calibration curve represented by Eqn. 6.⁴¹ However, it should be noted that ϵ -martensite content cannot be quantified using this equation due to its nonmagnetic nature.

$$C\alpha' \text{ (wt.\%)} = 1.7 \times F \quad \text{Eqn. 6}$$

For the CF3 alloy containing 4.3 wt.% ferrite before the tensile test, the α' -martensite content can be quantified using Eqn. 7.

$$C\alpha' \text{ (wt.\%)} = 1.7 \times (F - 4.3) \quad \text{Eqn. 7}$$

The experimental HMBASE, HMN015 and HMN015V alloys with the SFE of 25.2, 35.2, 32.0 mJ/m², respectively, did not form α' -martensite after the tensile test, which is consistent with the TWIP effect expected at these SFEs. In contrast, the CF3 alloy exhibited a significant increase to 44.5 wt.%, indicative of the TRIP effect.

Figure 11 presents the FE-SEM-EDS fractographs of the HMN015 and HMN015V alloys after room-temperature tensile test according to ASTM E 8. The HMN015 alloy's fracture surfaces predominantly exhibited large, shallow dimples (Figure 11a) and slightly exhibited cleavage plane, while those of the HMN015V alloy displayed exhibited large, shallow dimples (Figure 11b). Both alloys demonstrated ductile-dimpled fracture, irrespective of the chemical compositions. The fracture surfaces of the HMN015 alloy contained oxysulfide inclusions with composition (Mn, Si, Al, Fe, Cr)OS (Figure 11a1 and 11a2). The HMN015V alloy contained oxide inclusions with compositions (Mn, Fe, Cr, Si)O (Figure 11b1 and 11b2). These inclusions are attributed to the use of Ca-Si and Al as deoxidizers during the alloy production process.

In summary, based on microstructure analyses via OM and EBSD, as well as martensite identification through XRD test and feritscope measurements, the overall enhanced tensile properties of the HMN015V alloy containing N and V, compared with the experimental HMBASE and commercial CF3 alloys, can be attributed to three primary factors: (1) interstitial solid solution strengthening effect due to N addition, (2) precipitation hardening effect due to V nitride (VN) and (3) Hall–Petch type strengthening effect resulting from numerous twin grain boundaries, which act as sub-boundary-like obstacles to dislocation motion, formed via TWIP mechanism during the tensile deformation. These characteristics contrast with the CF3 alloy, which contains significantly lower C content (0.02 wt.%). Moreover, the high Mn–low Ni–low Cr–medium N–low V (24.8Mn–0.5Ni–3.6Cr–0.13N–0.19V, wt.%) composition of the HMN015V alloy demonstrates superior cost-effectiveness compared to the low Mn–high Ni–high Cr–low C (1.0Mn–8.61Ni–18.27Cr–0.02C, wt.%) CF3 alloy. This economic advantage is primarily due to the HMN015V alloy’s lower content of expensive alloying elements, such as Ni and Cr.

Charpy Impact Property at Cryogenic Temperature of -196°C Associated with Microstructure

Table 2 presents the Charpy V-Notch impact test data at a cryogenic temperature of -196°C . Superiority of the absorbed impact energy of the experimental and commercial alloys can be arranged in the following order: CF3 > HMN015 $\approx$ HMBASE > HMN015V. Although the absorbed impact

energy of the HMN015V alloy containing N and V was lower by 53 J than that of the CF3 alloy, it still demonstrated high impact energy of 114 J at cryogenic temperature of $-196\text{ }^{\circ}\text{C}$, which is attributable to its high ductility of high elongation (48 %).

Figure 12 illustrates (a) EBSD IPF and (b) phase maps to analyze the hcp ϵ - and bcc α' -martensite formed in the HMN015V alloy containing N and V after the Charpy V-notch impact test. The HMN015V alloy after the impact test revealed very small quantities of the ϵ - (yellow) and α' -martensite (green) within the γ -austenite matrix (red), irrespective of the different analysis locations (Figure 12b2 and 12b3). The phase transformation behavior of the HMN015V alloy after the impact test is very similar to that after the tensile test (Figure 8b2 and 8b3).

Figure 13 and Table 4 show the α' -martensite content, calculated using Eqn. 6 and 7 based on feritscope measurements, after the Charpy V-notch impact test of the experimental HMBASE, HMN015, HMN015V and commercial CF3 alloys at -196°C , according to ASTM E 23, compared with that after the tensile test at room temperature.

The SFE decreases by $\sim 30\%$ as the temperature decreases from room temperature to cryogenic temperature of $-196\text{ }^{\circ}\text{C}$, which often induces a martensitic transformation.⁴²⁻⁴⁷ This results in microstructures complicatedly mixed with martensite and twins, and influences the strength and toughness.⁴²⁻⁴⁷ The toughness is generally evaluated by the Charpy impact tests under dynamic loading, where changes in deformation

mechanisms occurring in the Charpy-notch-tip area become more important than in quasi-static tensile tests.

The HMBASE alloy showed a marginal increase in α' -martensite content from 0 wt.% (tensile test) to 0.29 wt.% (impact test), consistent with the TWIP effect whereas the HMN015 and HMN015V alloys didn't reveal the α' -martensite due to much higher SFE than the HMBASE alloy. In contrast, the CF3 alloy exhibited a significant increase from 44.5 wt.% (tensile test) to 58 wt.% (impact test), indicative of the TRIP effect. These results were compared with those obtained after room-temperature tensile testing.

In summary, based on the microstructural analyses by the EBSD and feritscope martensite measurements after the Charpy impact test at a cryogenic temperature of -196°C , the high impact toughness of the HMN015V alloy is attributable to the TWIP effect.

C o n c l u s i o n s

In this study, the effects of N and V additions on microstructures, tensile and impact properties of a cast austenitic high-Mn steel for cryogenic application were investigated in relation to their microstructures using OM, EBSD, FE SEM-EDS, XRD, and martensite measurement via feritescope. The following conclusions were drawn.

1. The HMN015V alloy containing N and V exhibited the yield and tensile strengths 1.41 and 1.11 times higher than those of the HMBASE alloy, respectively whereas the elongation were lower than the HMBASE alloy. Especially, The HMN015V alloy exhibited yield and tensile strengths 1.85 and 1.42 times much higher than those of the CF3 alloy, respectively. Although the elongation of the HMN015V alloy was lower by 15 % than that of the CF3 alloy, it still demonstrated a high elongation of 48%.
2. The HMBASE alloy showed a marginal increase in α' -martensite content from 0 wt.% (tensile test) to 0.29 wt.% (impact test), consistent with the TWIP effect whereas the HMN015V alloys didn't reveal the α' -martensite due to much higher SFE than the HMBASE alloy. In contrast, the CF3 alloy exhibited a significant increase from 44.5 wt.% (tensile test) to 58 wt.%, (impact test) indicative of the TRIP effect. These results were compared with those obtained after room-temperature tensile testing and cryogenic temperature (-196 °C) impact testing.
3. Based on the microstructural analyses by the EBSD and feritscope martensite measurements after the Charpy impact test, the high impact toughness of the HMN015V alloy is attributable to the TWIP effect despite its higher yield and tensile strengths.
4. The enhanced strength, ductility, and impact toughness of the HMN015V alloy containing N and V, which has a high SFE of 32 mJ/m², compared with the experimental HMBASE with 25.2 mJ/m² SFE and commercial CF3 alloys with 16.2 mJ/m² SFE, can be attributed to three primary factors: a) interstitial solid solution strengthening from its medium N content (0.13 wt.%) and high C content (0.48 wt.%), b) precipitation hardening due to

V nitride (VN) and c) Hall–Petch type strengthening resulting from numerous twin grain boundaries and fine twin grains formed via the TWIP mechanism during tensile and impact deformations.

A c k n o w l e d g m e n t s

This work has been supported under “Material Components Technology Development Project” by the Ministry of Trade, Industry, and Energy of the Republic of Korea (Project No.:2410017893/No.:KT224866, Title: Alloy design of crack-resistant high manganese steel with improved yield stress above 345 MPa and alloy development for casting and forging parts)

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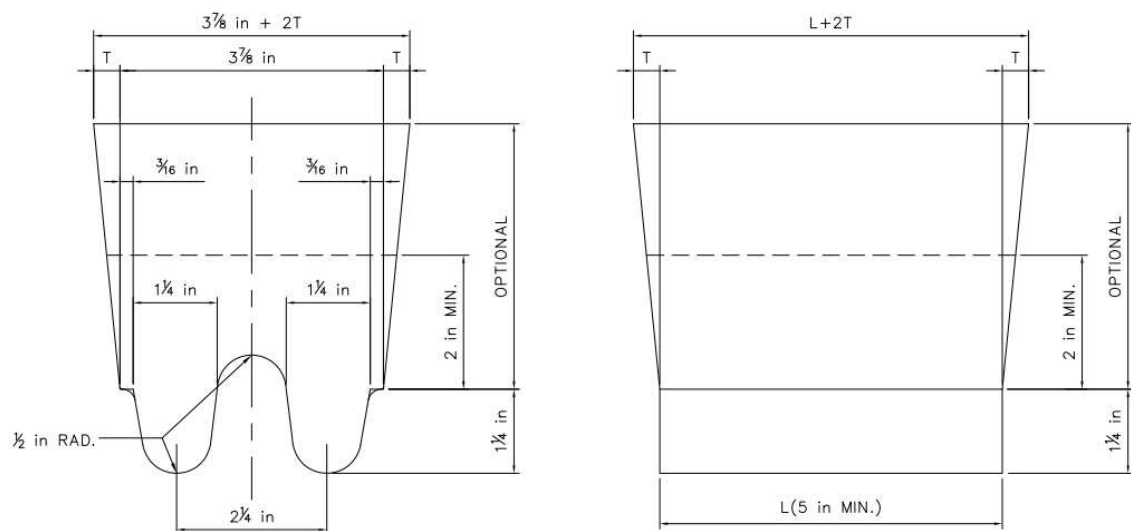
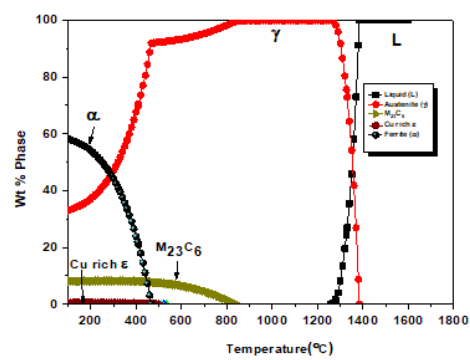
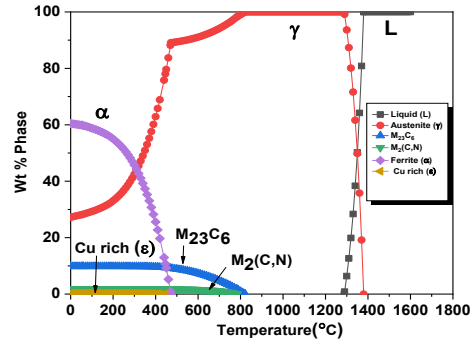


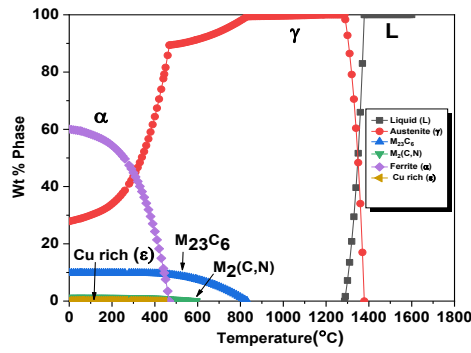
Figure 1. Dimension (mm) of the double keel block for test specimens according to ASTM E 8.



(a)



(b)



(c)

Figure 2. Phase diagrams calculated using a thermodynamic simulation software (JMatPro 7.0): (a) the HMBASE ²⁶, (b) HMN015 and (C) HMN015V alloys.

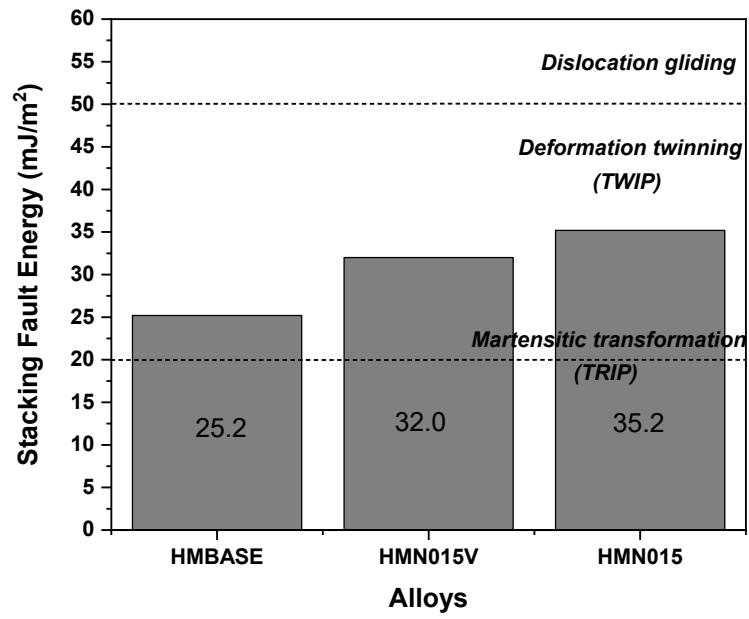
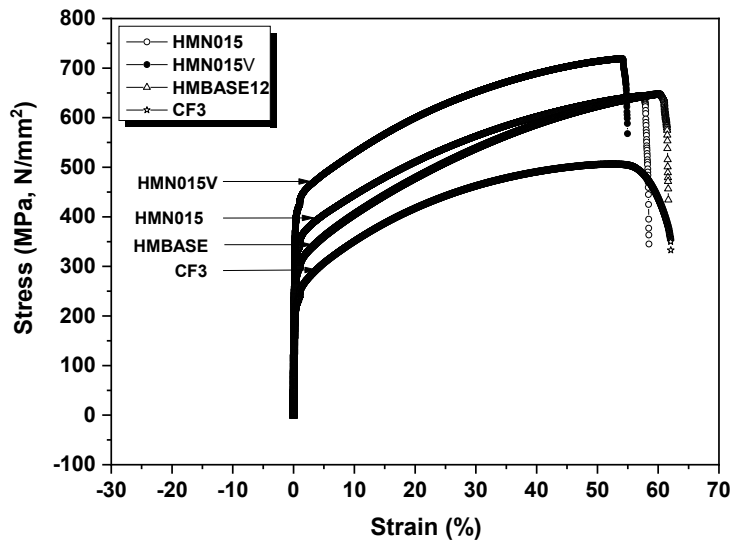
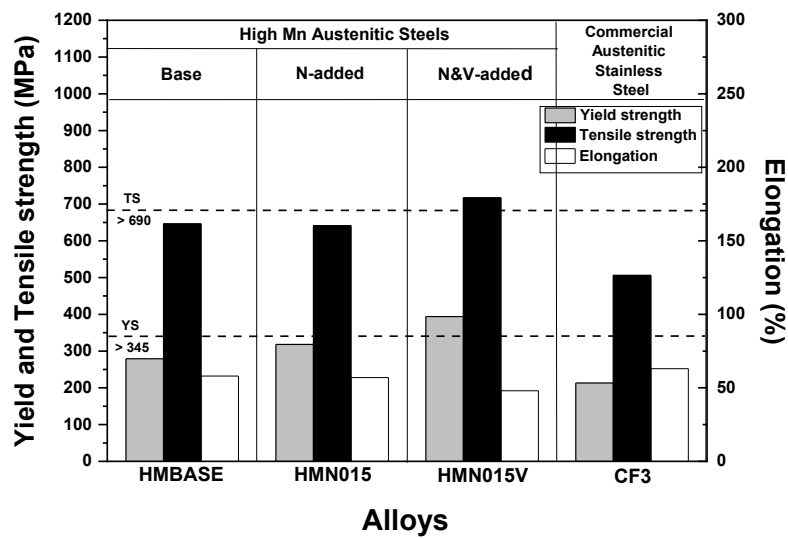


Figure 3. Stacking fault energies of the HMBASE ²⁶, HMN015 and HMN015V alloys calculated using a thermodynamic simulation software (JMatPro 7.0).



(a)



(b)

Figure 4. Engineering tensile stress–strain behavior at room temperature of the experimental HMBASE ²⁶, HMN015, HMN015V and commercial CF3 alloys according to ASTM E 8: (a) tensile curves and (b) summary of tensile data.

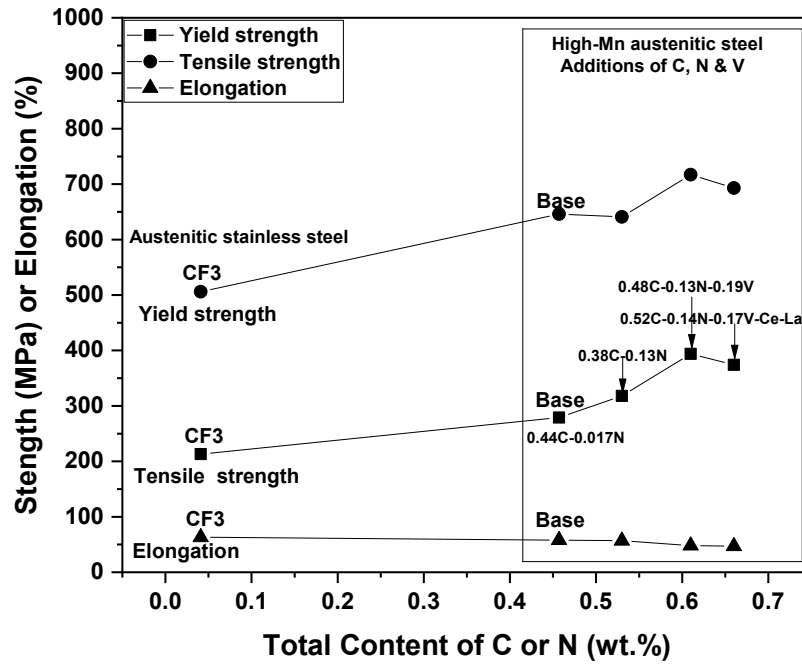


Figure 5. Effects of C, N and V additions on the tensile properties on the tensile properties of the experimental HMBASE ²⁶, HMN015, HMN015V and commercial CF3 alloys.

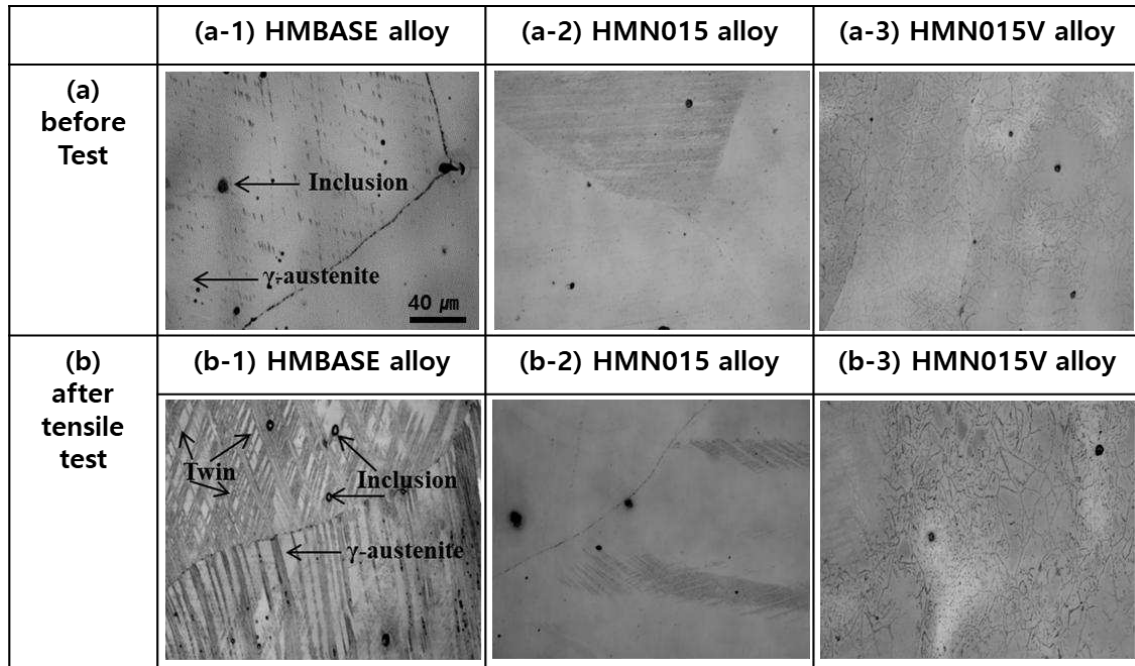
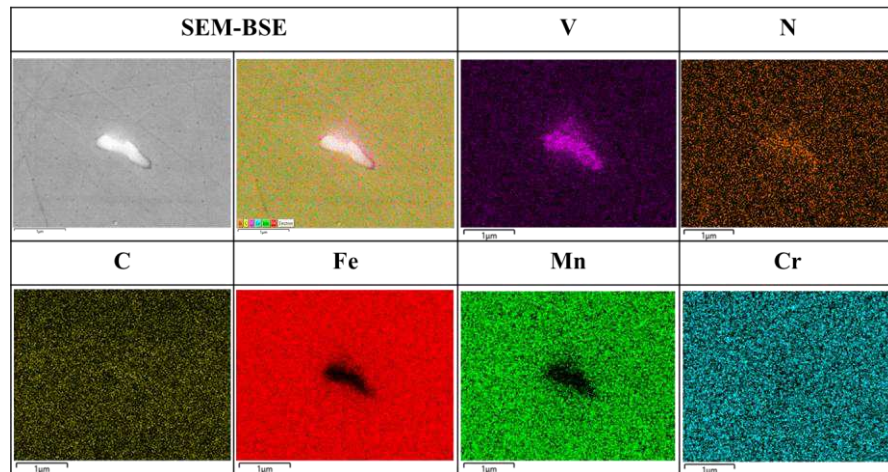
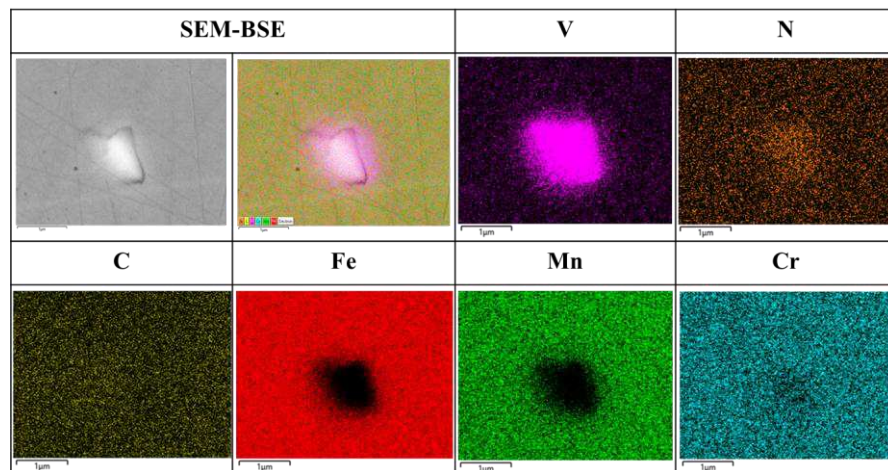


Figure 6. Optical microstructures (a) before and (b) after the tensile test at room temperature of the HMBASE ²⁶, HMN015 and HMN015V alloys according to ASTM E 8.



(a)



(b)

Figure 7. FE-SEM elemental mapping of the solution heat-treated HMN015V alloy: (a) analysis at location of $\frac{1}{4}$ thickness and (b) analysis at center of double keel bock for test specimens (Figure 1) .

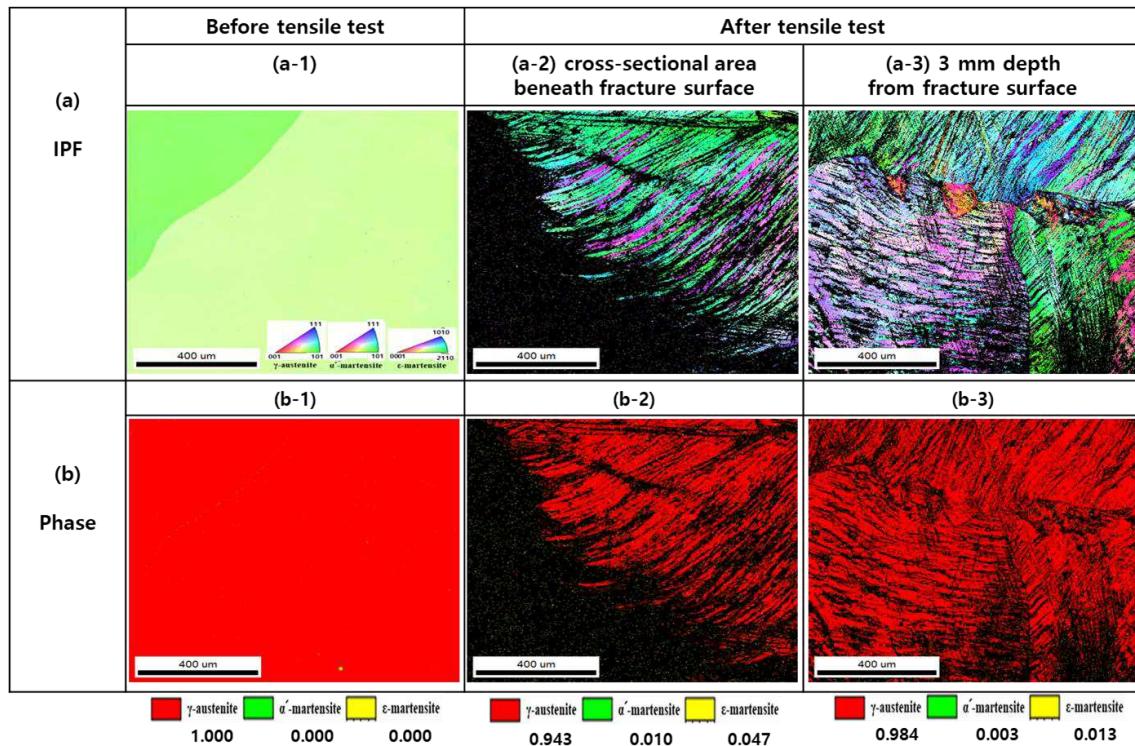


Figure 8. EBSD analysis after the tensile test at room temperature of the HMN015V alloy according to ASTM E 8: (a) IPF and (b) phase maps.

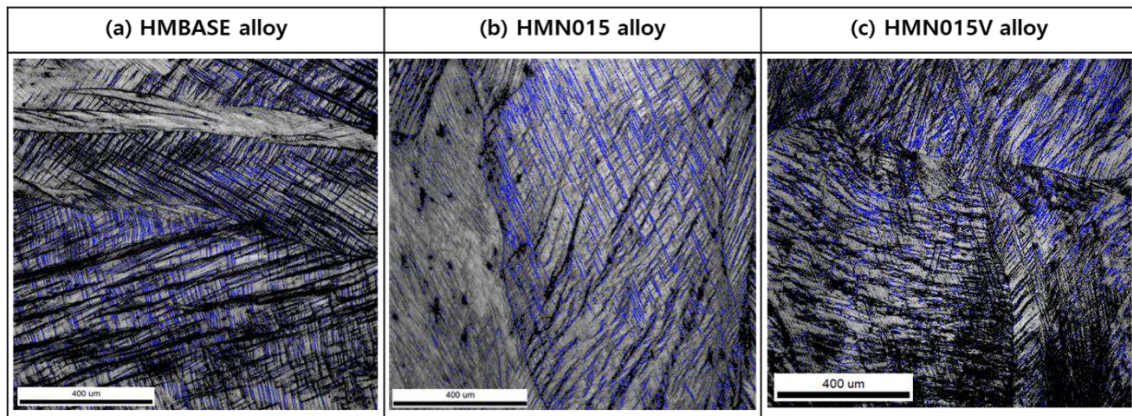


Figure 9. EBSD grain boundaries maps after the tensile test at room temperature of (a) the HMBASE ²⁶, (b) HMN015 and (c) HMN015V alloys according to ASTM E 8.

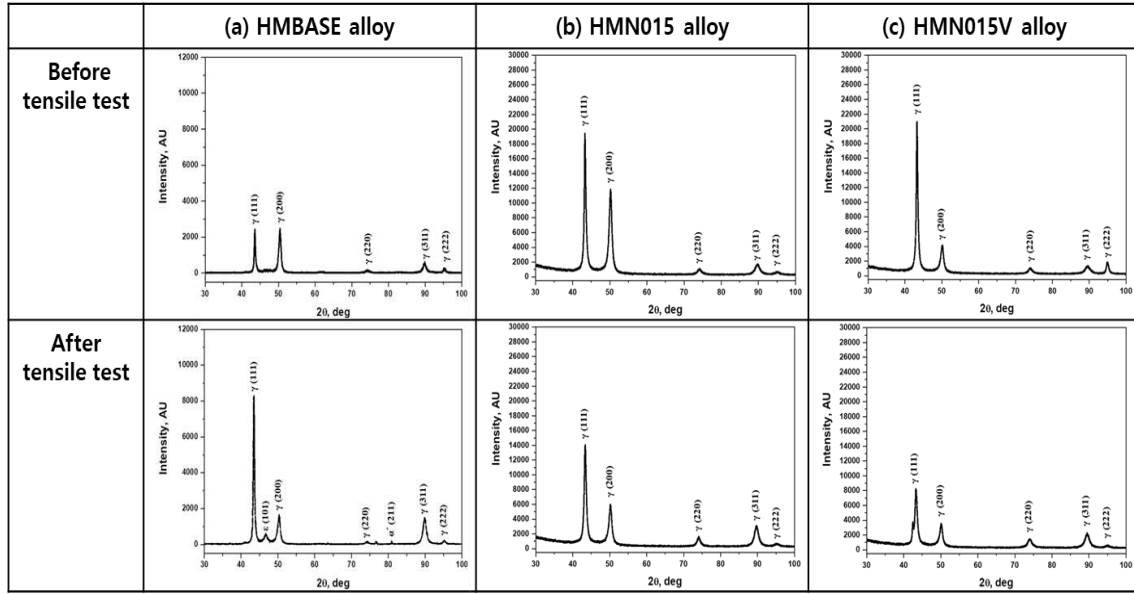
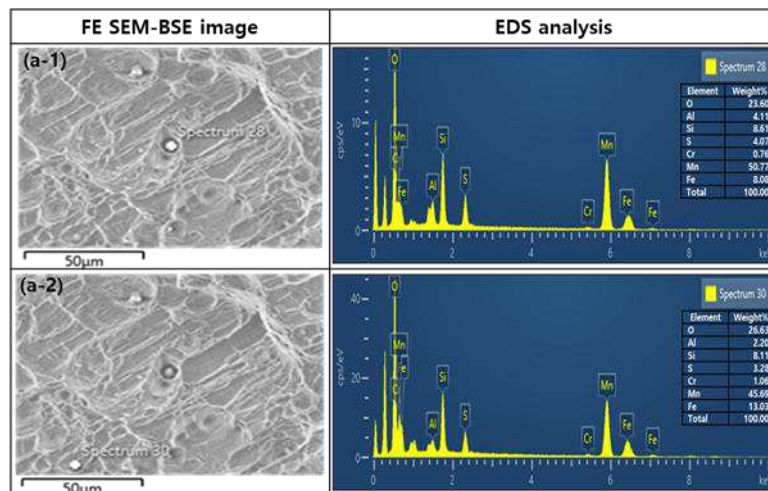
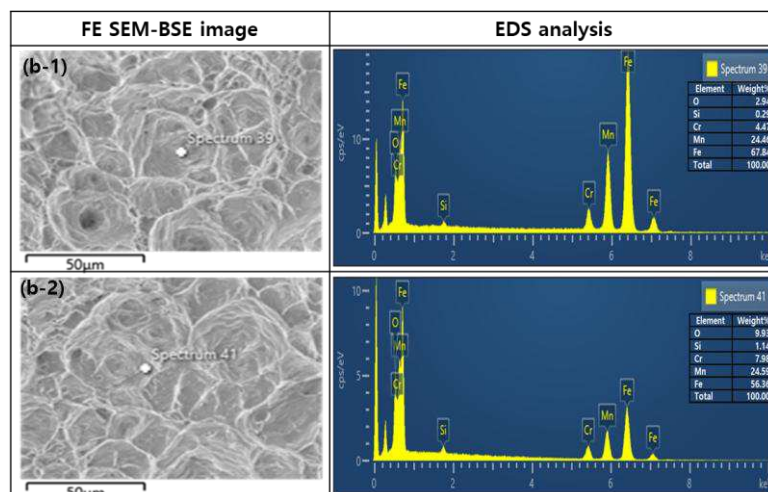


Figure 10. XRD patterns before and after the tensile test at room temperature of (a) the HMBASE ²⁶, (b) HMN015 and (C) HMN015V alloys according to ASTM E 8.



(a)



(b)

Figure 11. FE SEM-EDS fractographs after the tensile test at room temperature of (a) the HMN015 and HMN015V alloys according to ASTM E 8.

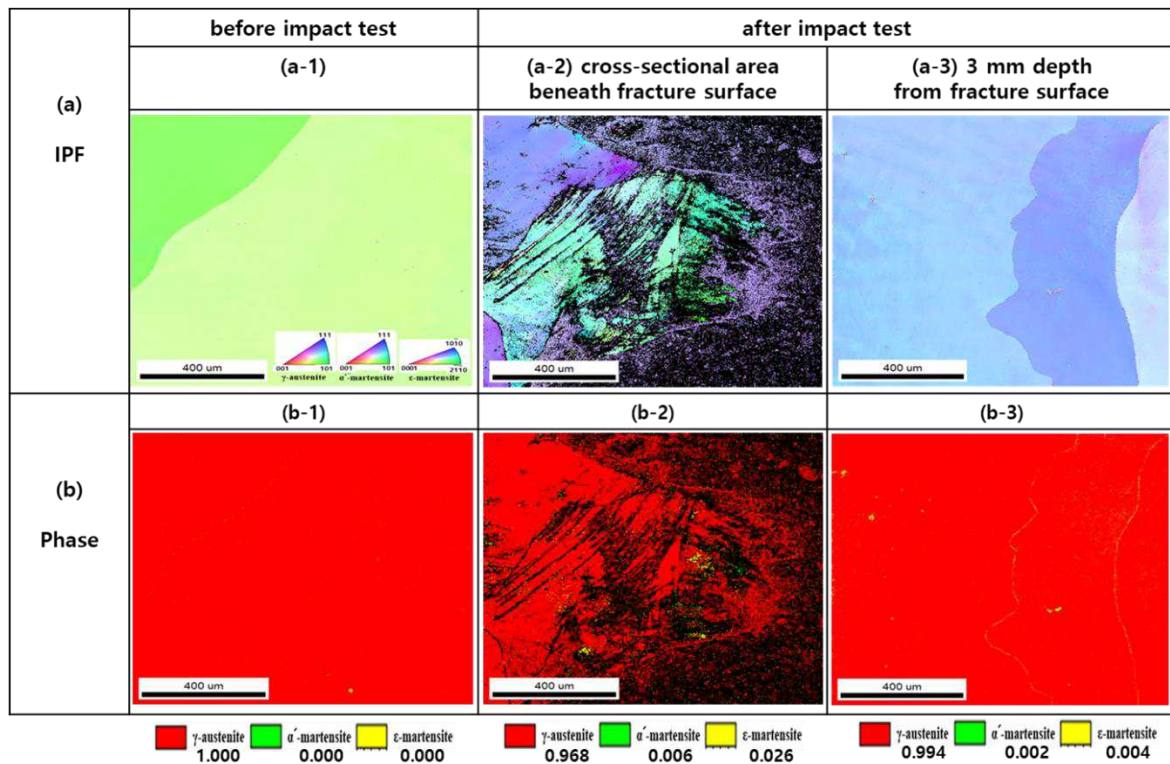


Figure 12. EBSD analysis of after the Charpy V-notch impact test of the HMN015V alloy at - 196 °C according to ASTM E 23: (a) IPF and (b) phase maps.

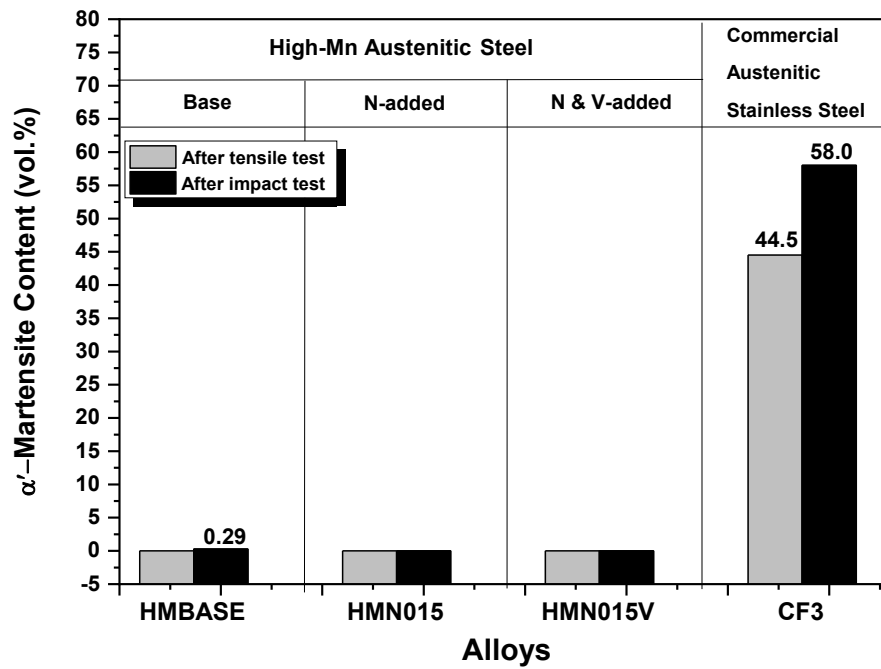


Figure 13. α' -martensite content calculated by the equations 4 and 5 after the measurement of α -ferrite content using the ferrite-scope after the Charpy V-notch impact test of the HMBASE²⁶, HMN015, HMN015V and CF3 alloys at - 196 °C according to ASTM E 23, compared with that after the tensile test at room temperature (Table 3).

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