

CHAPTER IV

RESULTS AND DISCUSSION

4.1 Effect of Nb content in as-cast condition

In this section, the influence of Nb on the as-cast microstructure and hardness was investigated because changes in carbide and eutectic structures during solidification directly affect the heat treatment behavior and abrasive wear resistance. Accordingly, this section discusses the influence of Nb content on the primary phase, eutectic morphology, matrix structure and hardness of the alloy.

4.1.1 Effect of Nb content on microstructure

4.1.1.1 Effect of Nb content on primary phase and eutectic structure

To investigate the primary phase and eutectic structure, the polished specimens were etched with Groesbeck's reagent to reveal the differences in carbide types and morphologies, as shown in Figure 4.1. In the Nb-free alloy, the microstructures exhibited a hypoeutectic composition, consisting of primary austenite dendrites (γ_p) and eutectic colonies of austenite with carbides. Based on their morphology and elemental distribution, the eutectic carbides were interpreted as MC and M_2C types, consistent with previous reports by Wu et al. (1997), and Hashimoto et al. (2004b) for multi-alloyed white cast irons of similar composition. The solidification path starts with the crystallization of primary austenite dendrite directly from the liquid melt as schematically illustrated in Figure 4.2. During dendritic growth, carbon and carbide-forming elements are rejected into the residual interdendritic liquid, which induces the eutectic reactions as the temperature drops. The eutectic reactions begin with the formation of a $\gamma + MC$ eutectic, which are V-rich carbides, followed by the eutectic of $\gamma + M_2C$, mainly containing Mo and W. Morphologically, eutectic MC carbides exhibited a petal-like structure, whereas M_2C carbides displayed a plate-like morphology (Hashimoto et al., 2004b; Wu et al., 1996; Wu et al., 1995; Wu et al., 1997).

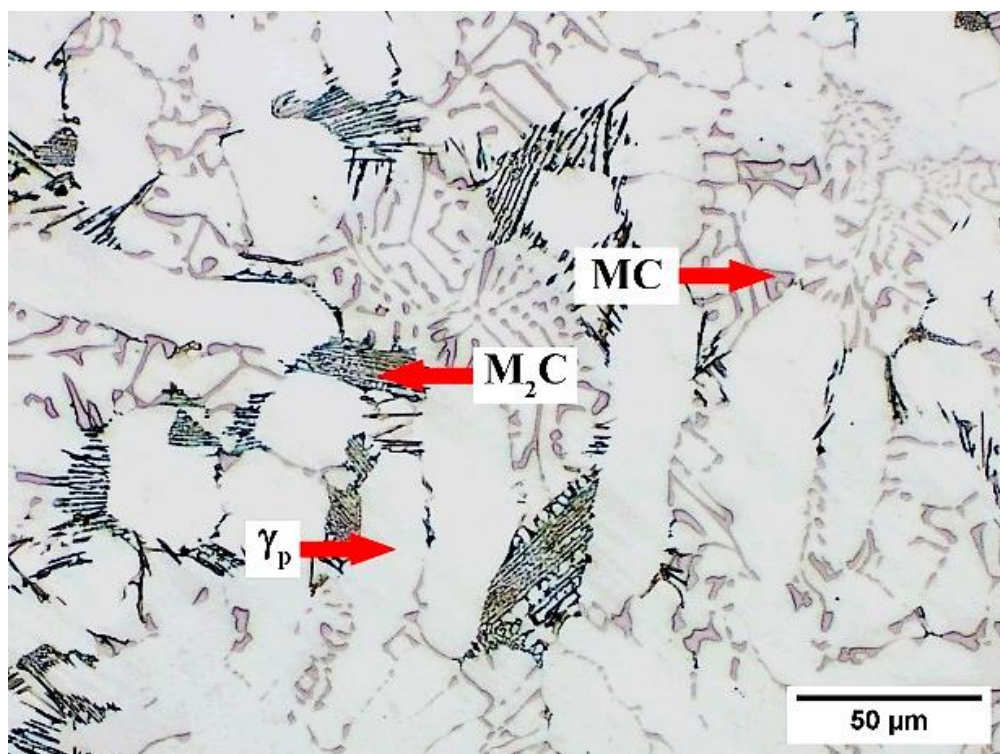
With the addition of 0.70%Nb, the microstructure consisted of primary austenite dendrites along with MC and M_2C eutectic carbides, similar to the

specimen No.1 (Nb-free). However, the 0.70%Nb specimen exhibited a finer dendrite size and irregular eutectic MC morphology with roughened boundaries. The addition of Nb effectively shifts the chemical composition of the alloy closer to the eutectic point as demonstrated in the calculation isopleth plotted by Boccalini et al. (2008). This thermodynamic shift narrows the solidification temperature range between the liquidus and solidus line, which significantly reduces the time for the growth of primary austenite dendrites. As a result, the dendrites are unable to expand to a large size, leading to a finer and more refined dendritic structure. Furthermore, the roughening of the boundaries of the eutectic MC carbides observed upon the addition of 0.70%Nb is attributed to the substitution of Nb atoms within the MC carbide lattice. This phenomenon is further driven by the existence of a miscibility gap in the VC-NbC system, which triggers phase decomposition during post-solidification cooling. Consequently, the morphology of the eutectic carbides evolves from smooth, petal-like shape into a roughened structure (Adnane et al., 1992; Boccalini et al., 2008).

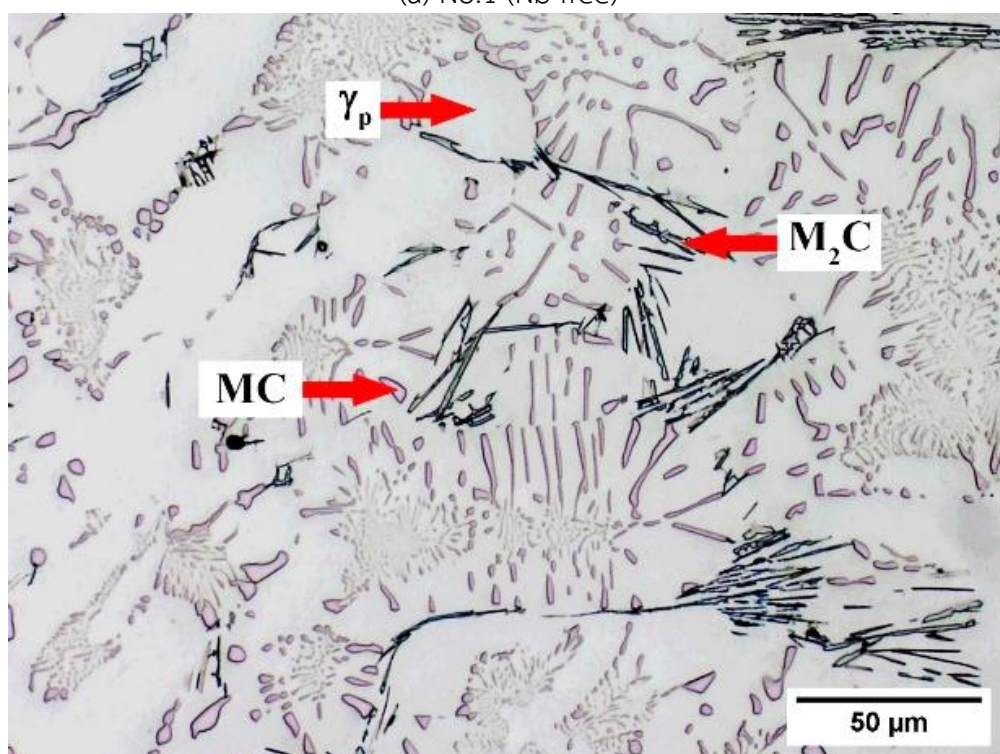
The microstructure of specimen No.3 (1.05%Nb) consisted of primary carbides, proeutectic austenite dendrite and eutectic structures of (γ + MC) and (γ + M_2C). According to the solidification path shown in Figure 4.3, the solidification starts with the crystallization of primary carbide directly from the liquid melt. These primary carbides act as heterogeneous nucleation sites for the crystallization of austenite dendrite. As temperature decreases and the residual interdendritic liquid becomes enriched with carbon and other carbide-forming elements, eutectic reaction similar to those observed in specimen No.1 and No.2. The presence of primary carbide is due to the addition of Nb shifting the alloy from hypoeutectic to hypereutectic composition. According to Boccalini et al. (2008), the primary carbide occurs as an MC type (MC_p) with a granular (or blocky angular) morphology. This reflects the strong carbide-forming tendencies of Nb and V. Owing to its high affinity for carbon, Nb promoted the early precipitation of (V, Nb)-rich MC carbides during the solidification. The occurrence of austenite dendrites in the hypereutectic alloys can be attributed to the coupled zone theory (Boccalini et al., 2008; Kurz & Fisher, 1979). In systems consisting of a non-faceted phase (austenite) and a faceted phase (MC carbide), the coupled zone is often asymmetric (skewed toward the faceted phase). As the melt undercools below the primary MC precipitation temperature, the liquid composition and the temperature may fall outside the skewed coupled zone. In this region, the

growth velocity of austenite dendrite becomes higher than that of the cooperative eutectic growth, allowing austenite to crystallize as a second phase before the eutectic reaction of austenite + MC carbide. In the case of eutectic carbides, the eutectic MC carbide morphology exhibits a more pronounced petal-like morphology with increasing Nb content, which can be attributed to the depletion of C within the liquid melt during the crystallization of primary MC_p . The crystallization of MC_p depletes the C content in the residual melt, thereby promoting the formation of MC carbides with a petal-like morphology during eutectic reaction. This observation aligns with the findings of Wu et al. (1995), who found that in multi-alloyed white cast irons with low C content, the eutectic MC carbide crystallizes into a petal-like morphology. The morphology of the M_2C eutectic carbides does not change with the addition of Nb, due to the fact that Nb primarily forms MC-type carbides (Shabalin, 2019).

In the case of specimen No.4 (2.58%Nb), the microstructure exhibited MC_p , proeutectic austenite dendrite and eutectic structures of (γ + MC) and (γ + M_2C) as in specimen No.3. The change from 1.05%Nb to the 2.58%Nb alloy resulted in larger MC_p carbides, a higher amount of these carbides and a more pronounced granular (or blocky angular) morphology.

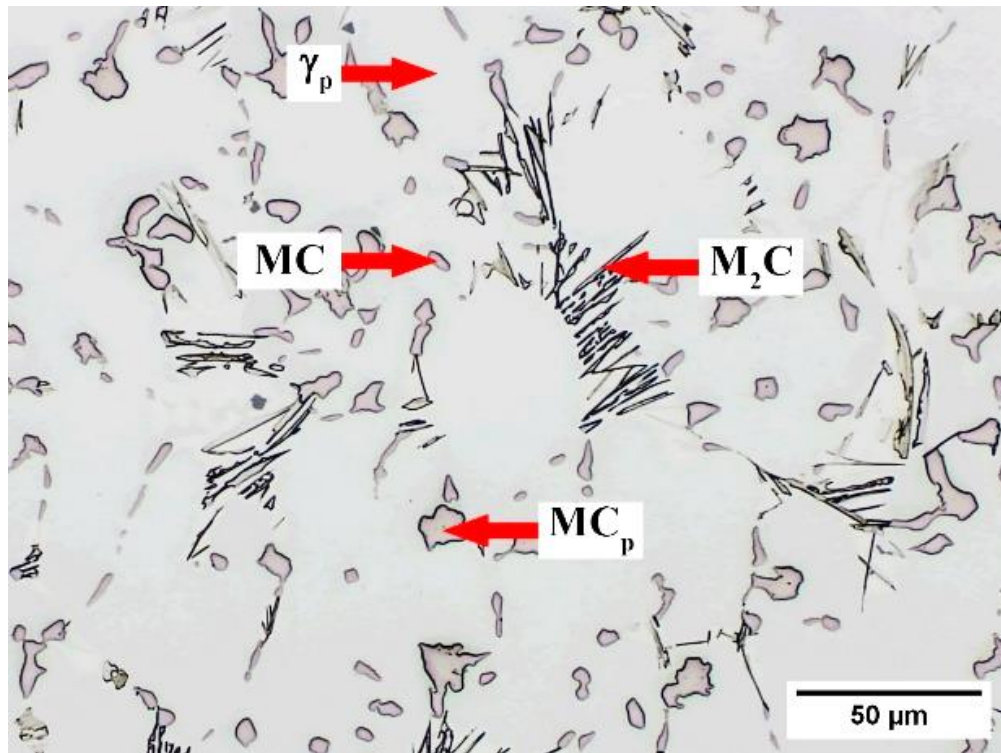


(a) No.1 (Nb-free)

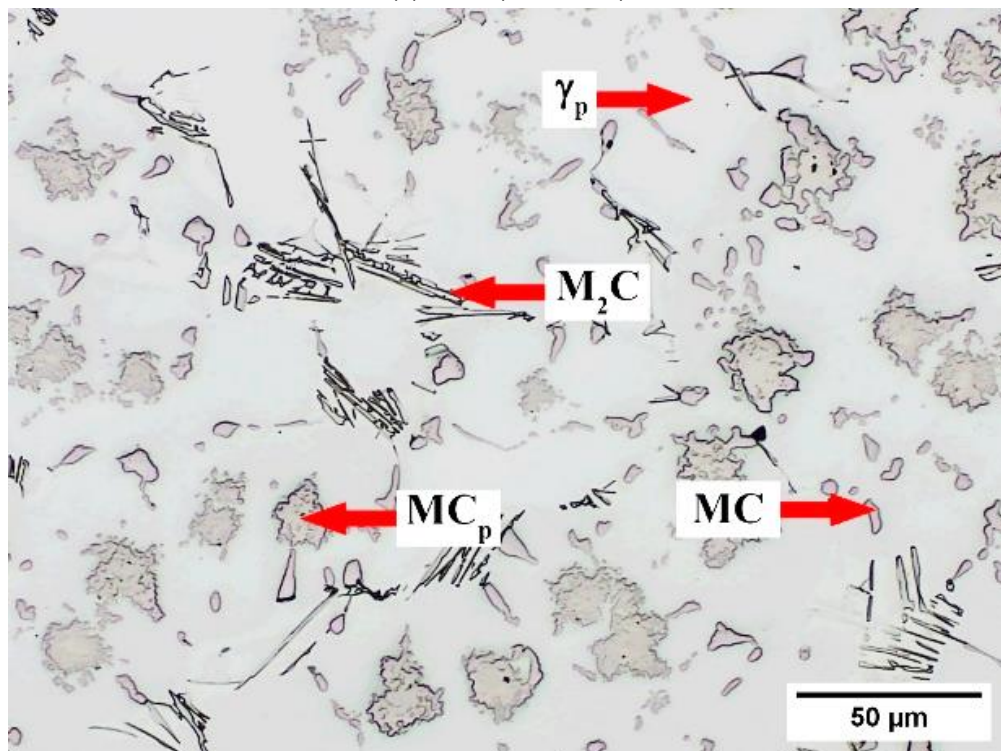


(b) No.2 (0.70%Nb)

Figure 4.1 Microstructural features of as-cast specimens with different Nb; (a) No.1 (Nb-free), (b) No.2 (0.70%Nb).



(c) No.3 (1.05%Nb)



(d) No.4 (2.58%Nb)

Figure 4.1 Microstructural features of as-cast specimens with different Nb; (c) No.3 (1.05%Nb) and (d) No.4 (2.58%Nb).

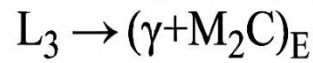
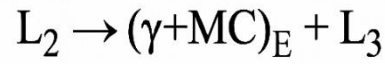
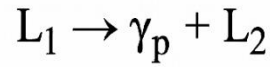


Figure 4.2 Schematic representation of the solidification path for the specimen No.1 (Nb-free) and No.2 (0.70%Nb).

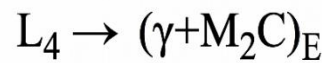
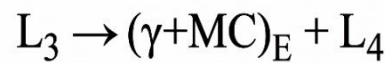
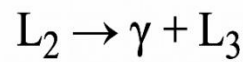


Figure 4.3 Schematic representation of the solidification path for the specimen No.3 (1.05%Nb) and No.4 (2.58%Nb).

The distribution of alloying elements was further characterized by electron probe microanalysis with characteristic X-ray mapping, as shown in Figure 4.4 to 4.7. The element maps revealed that Cr, Mo and W were predominantly concentrated in the M_2C carbides, whereas V and Nb were enriched within MC and MC_p carbides (Hashimoto et al., 2004b). These results are consistent with those reported by Matsubara et al. and Wu et al. (Matsubara et al., 1994; Wu et al., 1995), who showed that in multi-alloyed white cast iron, eutectic MC carbides consist predominantly of V, while eutectic M_2C carbides are composed mainly of Mo and W with some Cr. In the present work, Nb was found to partition predominantly into both eutectic and primary MC carbides, rather than into M_2C , which is attributed to the strong thermodynamic tendency of Nb to form MC-type carbides (Coelho et al., 2003). In the case of Co, the element predominantly dissolved into the matrix (Yu & Matsubara, 2001; Yokomizo et al., 2000).

The alloy concentrations of eutectic MC and M_2C carbides are shown in Figure 4.8 and Figure 4.9 respectively. In the case of eutectic MC carbides, V acted as the primary alloying element across all specimens. However, at 0.70%Nb, the V concentration in the eutectic MC carbides decreased from 50.87% to 31.26%. This is attributed to the fact that Nb additions reduce the solid solubility of V within the MC crystal lattice during solidification. Consequently, as the Nb content increased, the concentration of Nb within the eutectic MC carbides increased. Regarding eutectic M_2C carbide, Nb is present only in trace amounts due to its negligible solid solubility in the eutectic M_2C carbides. For the MC_p carbides, energy-dispersive spectrometer (EDS) line scans of representative particles from the 1.05% and 2.58%Nb specimens are shown in Figure 4.10. In the alloy with 1.05%Nb, the MC_p carbides were primarily V-rich with minor niobium enrichment. Conversely, in the 2.58%Nb specimen, the MC_p exhibited a Nb-rich core surrounded by a V-rich eutectic MC rim. These findings agree with the observation by Boccalini et al. (2008), who reported that Nb-rich primary carbides act as effective nucleation sites for the subsequent growth of eutectic MC carbides during solidification. Moreover, there was a progressive reduction in the V content of MC_p carbide with increasing Nb. This trend is attributed to the higher carbide affinity of Nb relative to V, which favors the preferential formation of NbC (MC-type) carbides during solidification, thereby limiting the incorporation of V into the MC_p phase. The observation that Nb partitions predominantly into MC carbides with negligible presence in M_2C carbides has important implications for the matrix composition and transformation behavior, as discussed in the following section.

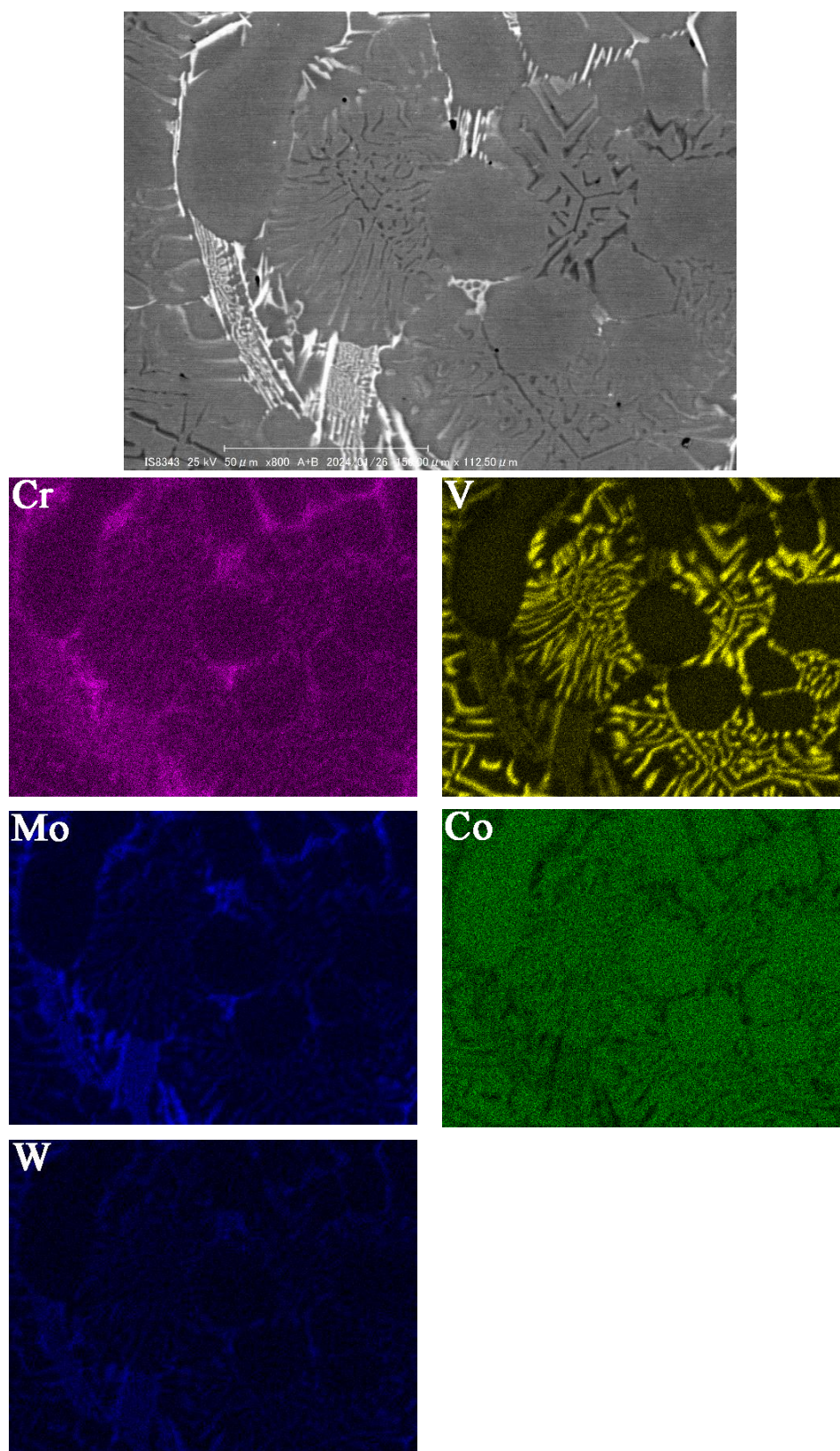


Figure 4.4 EDS mapping analysis results of specimen No.1 (Nb-free) in as-cast condition.

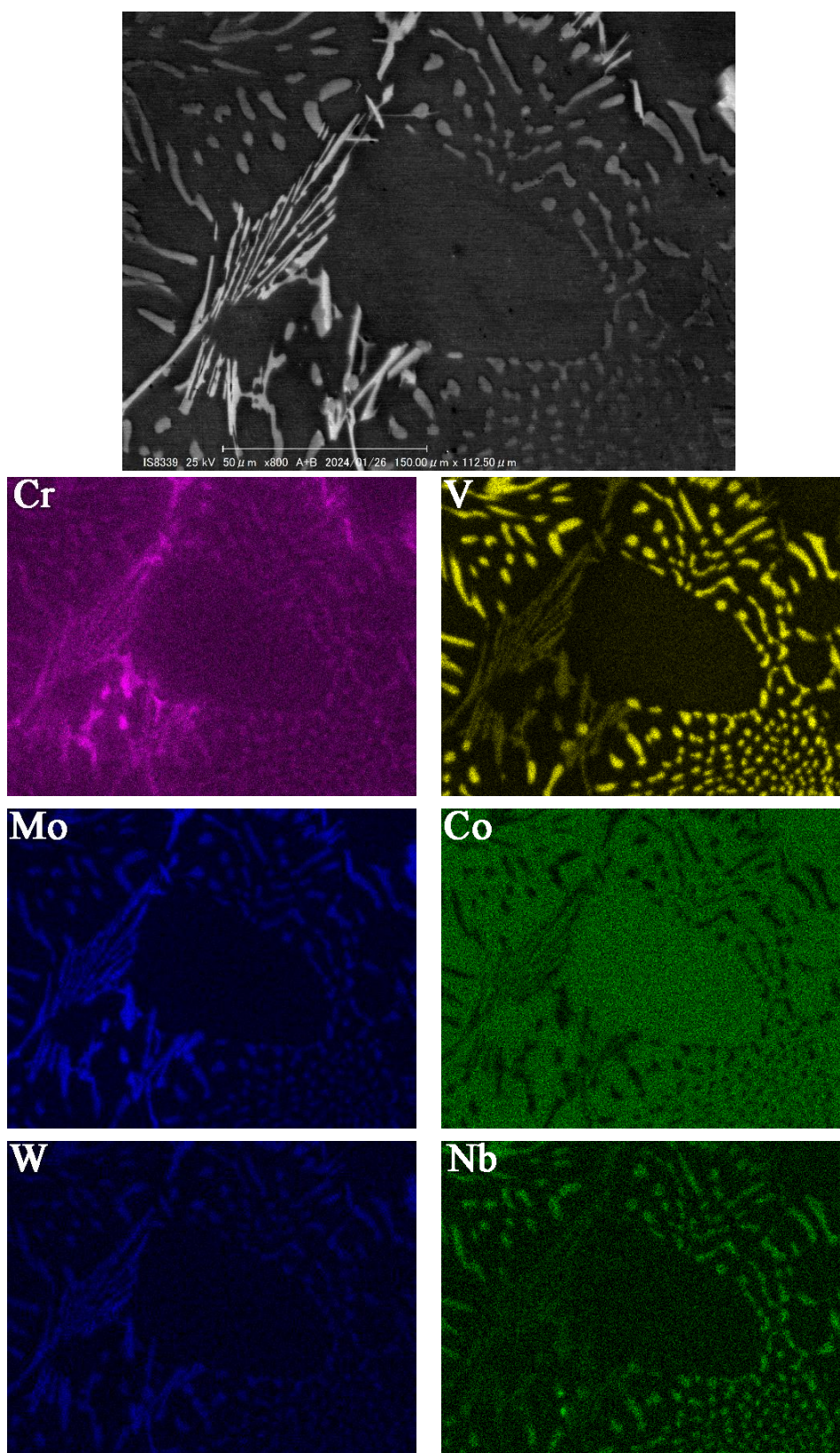


Figure 4.5 EDS mapping analysis results of specimen No.2 (0.70%Nb) in as-cast condition.

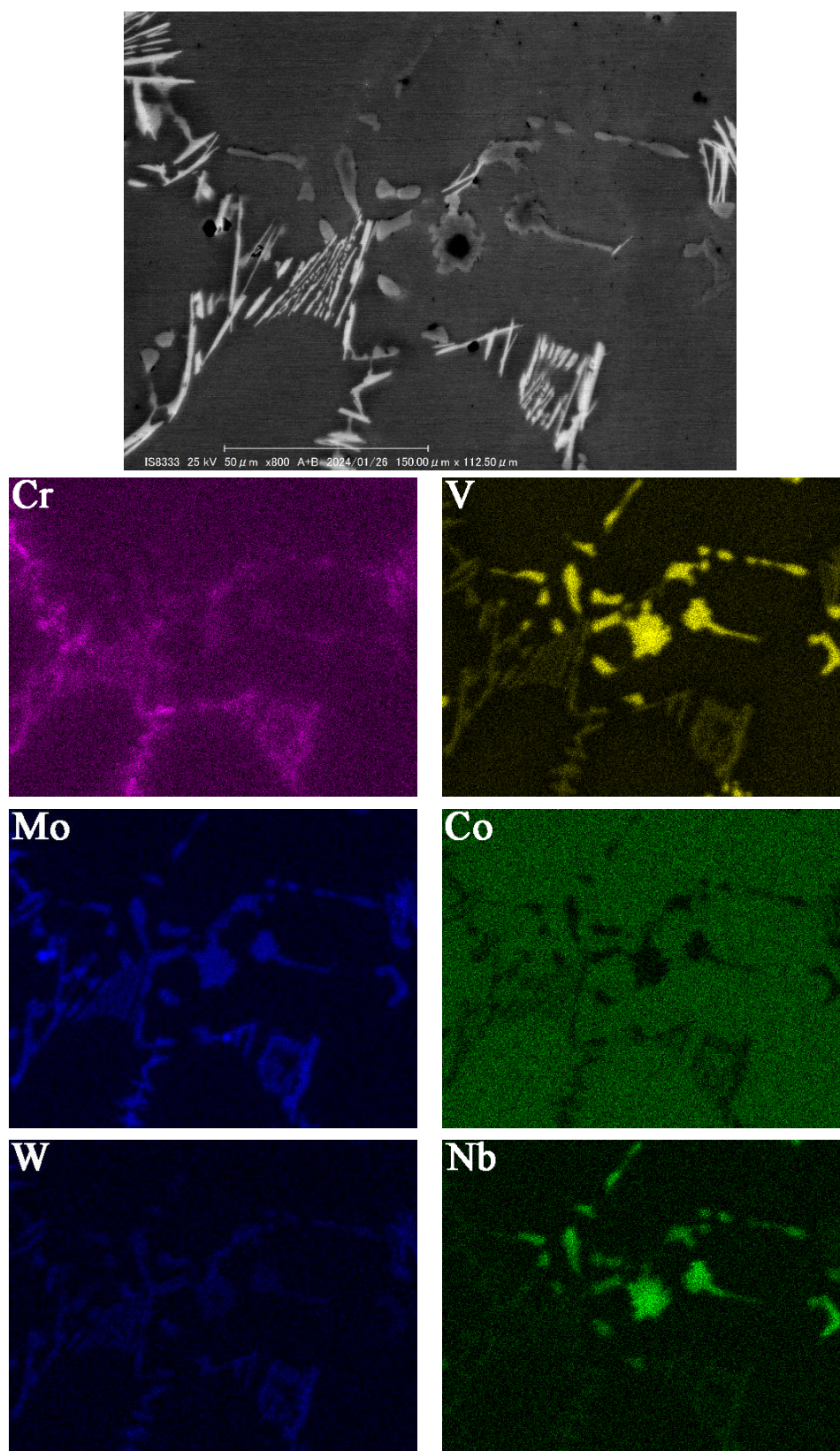


Figure 4.6 EDS mapping analysis results of specimen No.3 (1.05%Nb) in as-cast condition.

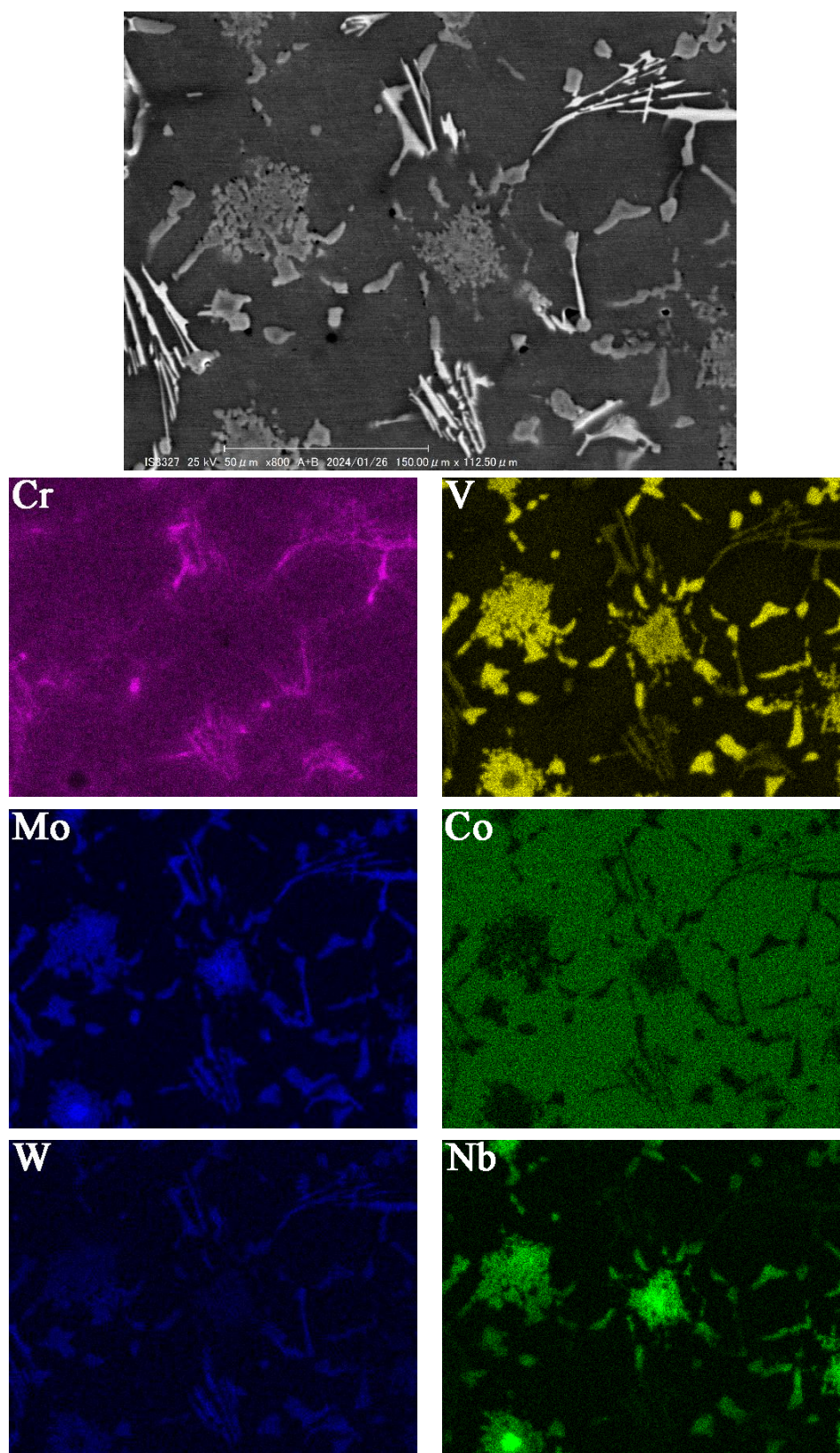


Figure 4.7 EDS mapping analysis results of specimen No.4 (2.58%Nb) in as-cast condition.

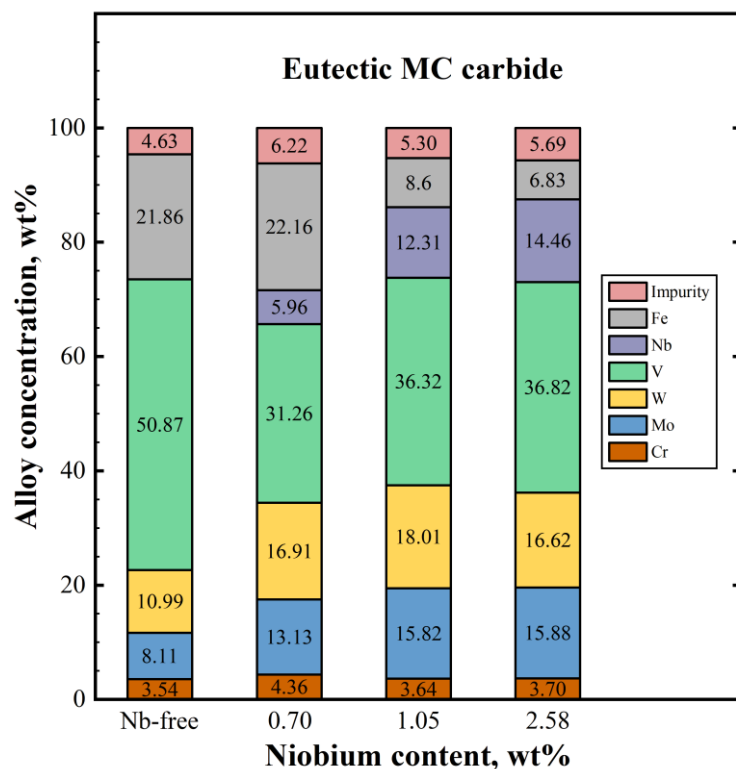


Figure 4.8 Alloy concentration of eutectic MC carbides as a function of Nb content.

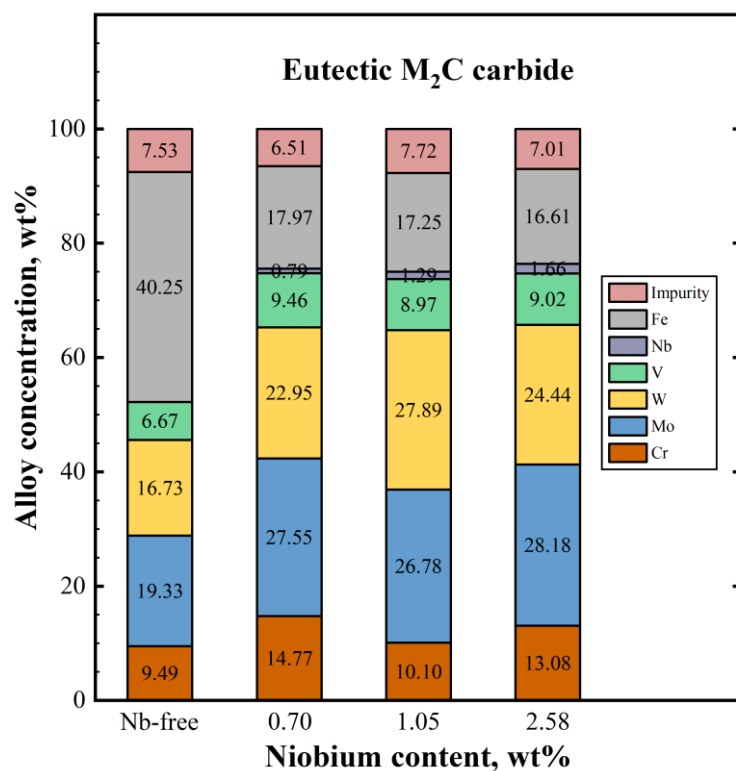


Figure 4.9 Alloy concentration of eutectic M_2C carbides as a function of Nb content.

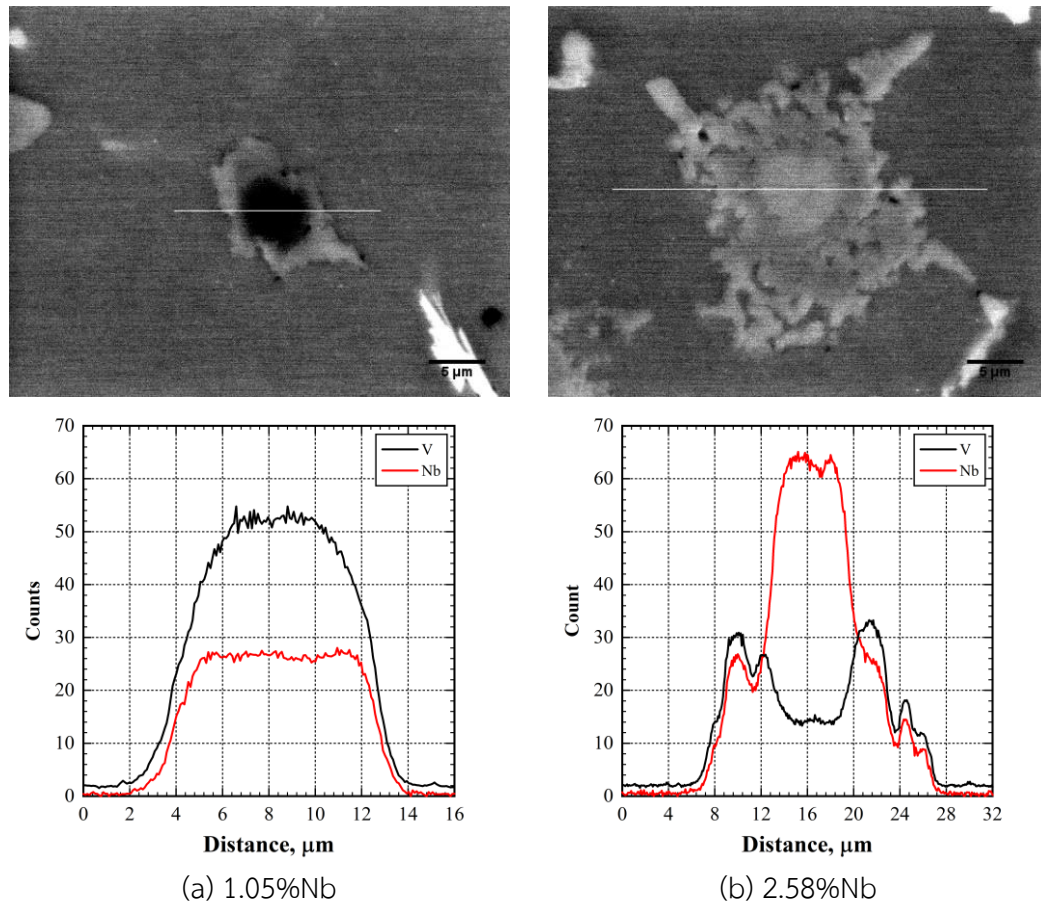


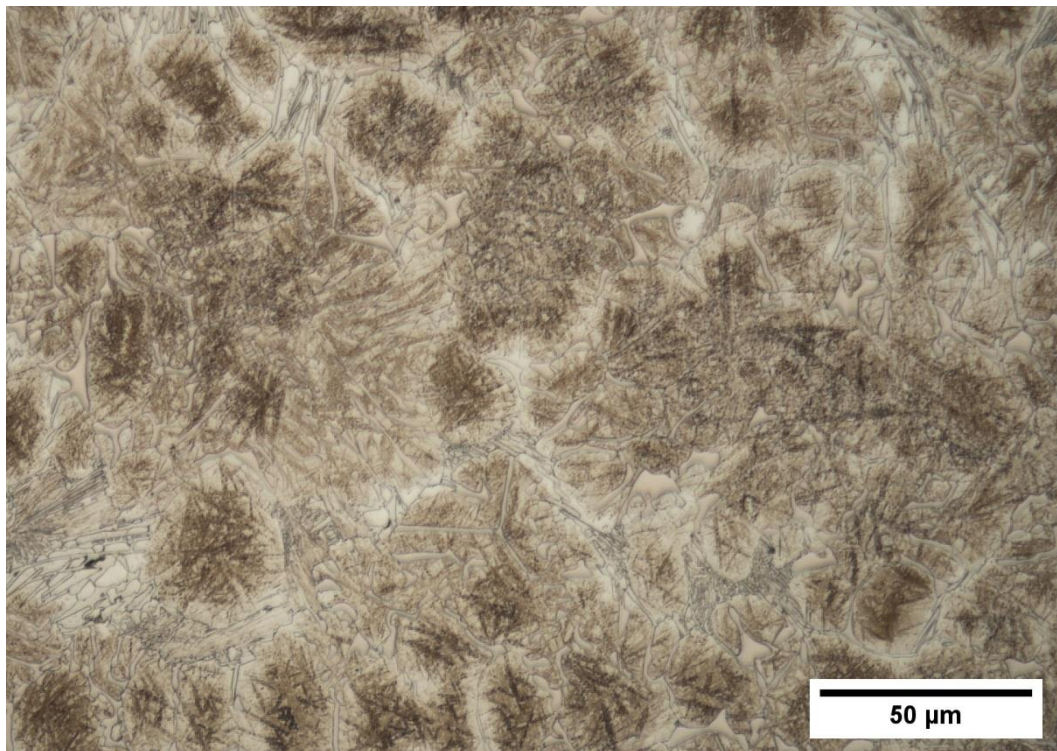
Figure 4.10 EDS line scanned profile of primary MC carbide (MC_p) in specimen; (a) 1.05%Nb and (b) 2.58%Nb.

4.1.1.2 Effect of Nb content on matrix structure

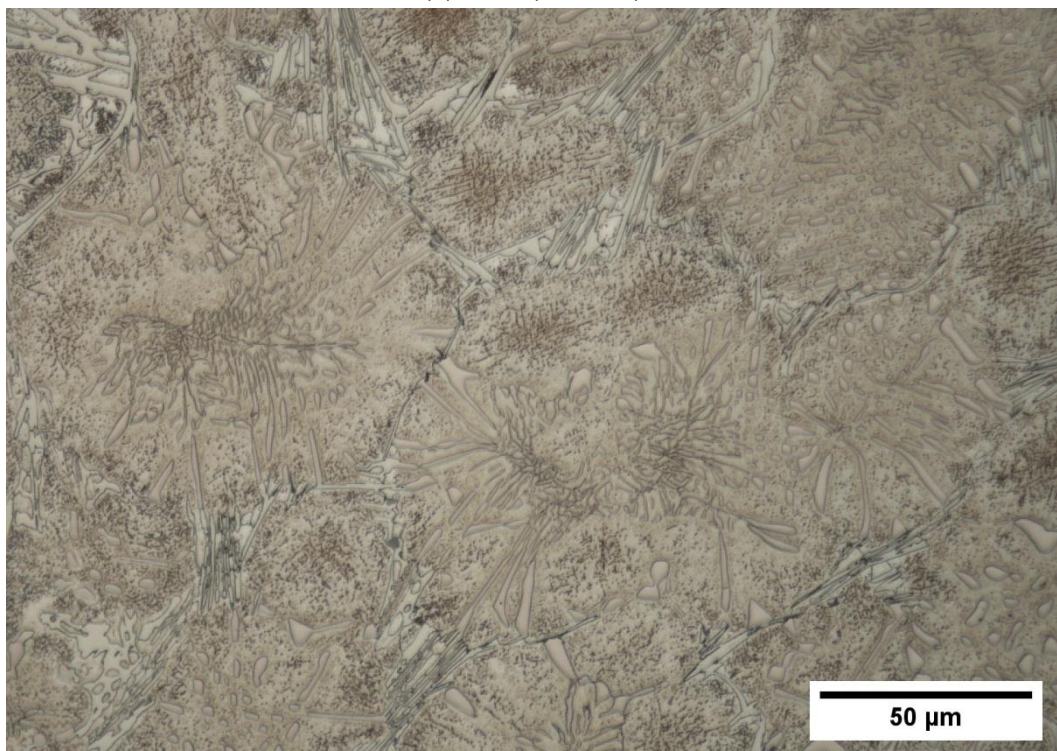
The as-cast matrix microstructures of multi-alloyed white cast iron with varying Nb content are presented in Figure 4.11, all etched with Vilella's reagent. In all alloys, the matrix consists of bright-etching retained austenite and dark-etching regions corresponding to martensite and/or bainite, consistent with the typical as-cast matrix constituents reported for multi-alloyed white cast iron (Opapaiboon et al., 2024). In the Nb-free alloy (Figure 4.11(a)), dark-etching regions are predominant, while bright-etching austenite occupies a relatively small portion of the matrix, indicating that a considerable amount of austenite transformed during cooling from the solidification temperature. With increasing Nb content to 0.70%Nb and 1.05%Nb (Figure 4.11(b, c)), the bright-etching austenite appears to increase, while the dark-etching transformed regions decrease. This observation should be regarded as a qualitative optical trend because the matrix constituents were identified from etched

optical microstructures. The apparent increase in bright-etching austenite is interpreted in terms of changes in matrix morphology and carbon redistribution during solidification. At 0.70%Nb, the dendritic/matrix structure became finer, which reduced the scale and continuity of dark-etching transformed regions and made the bright-etching austenitic regions more apparent under optical observation. At 1.05%Nb, the formation of MC_p carbides before the eutectic reaction modified the solidification path and increased the matrix fraction. As a result, the bright-etching matrix regions became more prominent. Therefore, the apparent increase in bright-etching austenite is interpreted mainly in terms of matrix/carbide morphology and possible carbon redistribution during solidification.

In the specimen No.4 with 2.58%Nb (Figure 4.11(d)), the matrix appears to contain the largest amount of bright-etching austenite, with only a small fraction of dark-etching transformed regions remaining. This pronounced apparent increase in bright-etching austenite should be interpreted as the combined effect of elevated Nb and carbon content. In particular, the higher nominal carbon content is expected to increase austenite stability by suppressing the M_s temperature (Ishida, 1995; Platl et al., 2020).

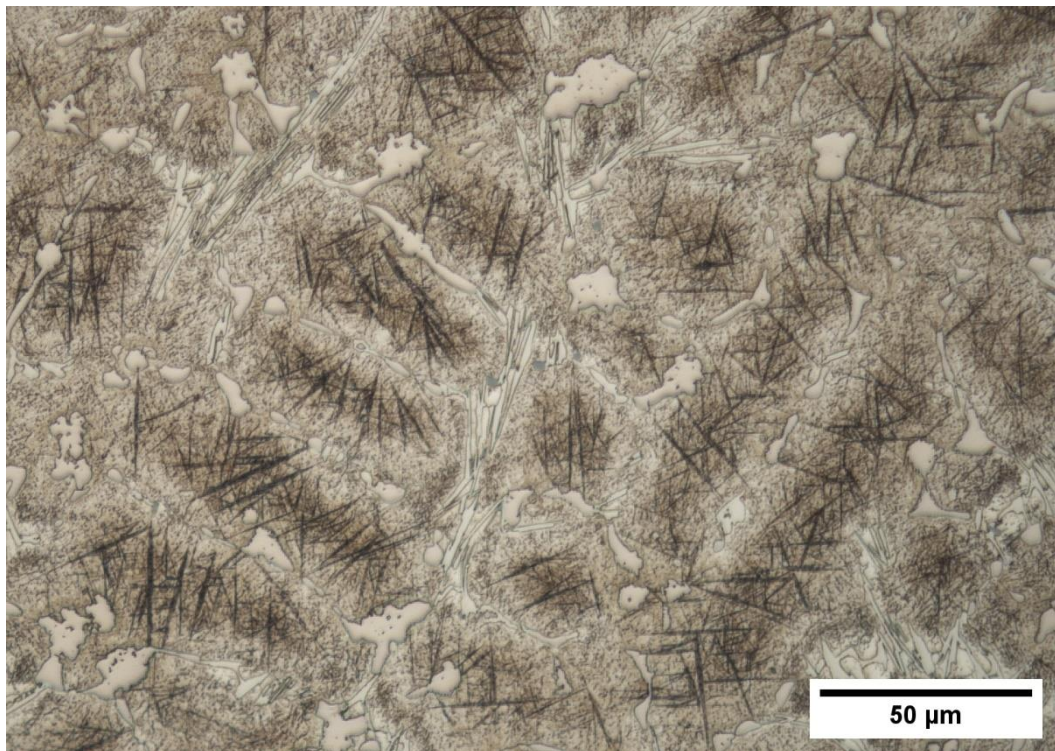


(a) No.1 (Nb-free)

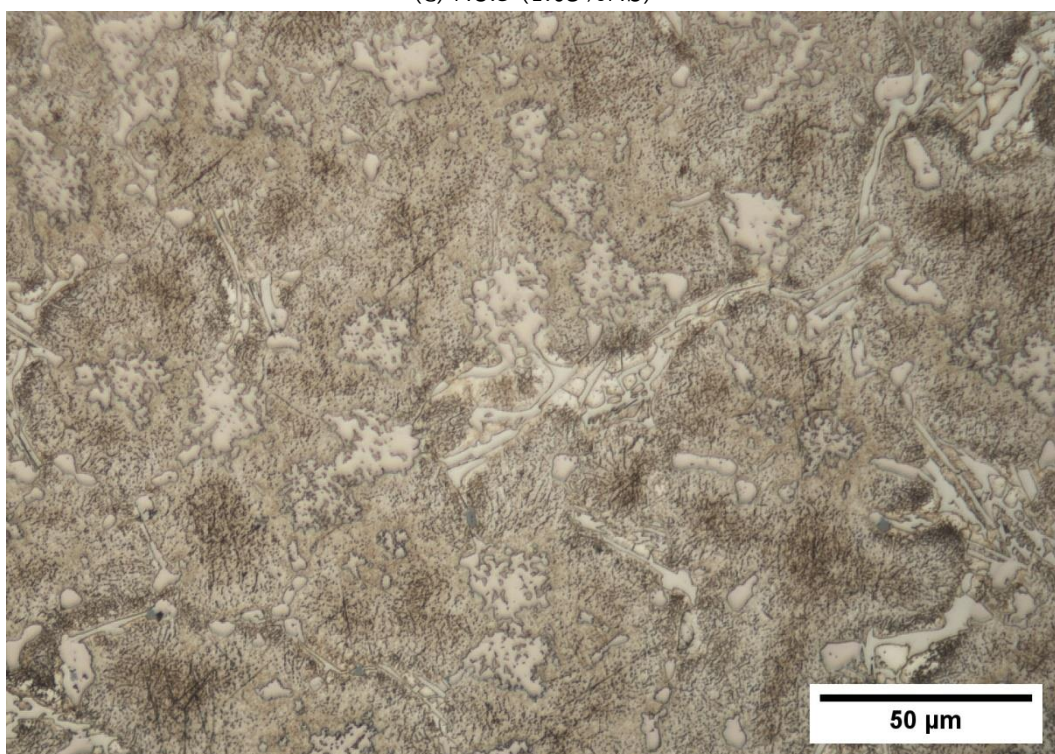


(b) No.2 (0.70%Nb)

Figure 4.11 Matrix microstructure of as-cast specimens with different Nb; (a) No.1 (Nb-free), (b) No.2 (0.70%Nb).



(c) No.3 (1.05%Nb)



(d) No.4 (2.58%Nb)

Figure 4.11 Matrix microstructure of as-cast specimens with different Nb; (c) No.3 (1.05%Nb) and (d) No.4 (2.58%Nb).

4.1.2 Effect of Nb content on area fraction of carbides and matrix structure

To clarify the influence of Nb on solidification behavior and phase transformation, quantitative image analysis was performed to determine the area fractions of carbides and matrix phases, as summarized in Table 4.1 and Figure 4.12. The area fraction of eutectic MC carbides in the Nb-free specimen is 11.39% and initially increased, reaching a maximum of 17.25% at 0.70%Nb, followed by a sharp decline to 6.85% at 1.05%Nb, and then exhibited a slight increase to 7.19% at 2.58%Nb. The initial increase is attributed to enhanced nucleation of the (γ + MC) eutectic by Nb addition. At 1.05%Nb, however, the formation of MC_p carbides consumed substantial amounts of Nb, V, and C from the liquid, thereby promoting the growth of austenite dendrite and suppressing eutectic formation (Zhi et al., 2008). In specimen No.4 with 2.58%Nb alloy, the higher carbon content was sufficient to sustain eutectic (γ + MC) formation despite the earlier precipitation of MC_p carbides. The MC_p carbides were first detected at 1.05%Nb and their fraction increased with increasing Nb content.

In contrast, the area fraction of eutectic M_2C carbides slightly decreased with increasing Nb content from 9.97% to 4.61%. This trend suggests that Nb additions preferentially stabilize MC carbides while suppressing the (γ + M_2C) eutectic reaction, most likely due to competitive carbon consumption. These observations agree with the findings of Boccalini et al. (2008). The total carbide fraction increased from 21.36% in the Nb-free specimen to the highest at 24.54% in specimen No.2 with 0.70%Nb due to the increase in area fraction of eutectic MC carbide. When the Nb content rose to 1.05%, although the crystallization of MC_p carbide occurred, the area fraction of total carbides decreased to the lowest at 15.97%. This decrease is attributed to carbon consumption during MC_p formation, which reduced the carbon content in the residual liquid and suppressed eutectic carbide formation. After that, the area fraction of total carbides increased to 22.09% due to the increased carbon content, which was sufficient to sustain the crystallization of both MC_p and eutectic carbides. The matrix fraction showed an inverse trend to the total carbide fraction.

Table 4.1 Area fraction of matrix and carbide phases in multi-alloyed white cast iron with different Nb content.

Specimen	Area fraction of matrix structure, (%)	Area fraction of total carbide, (%)	Area fraction of carbides		
			Primary MC, MC _p (%)	Eutectic MC carbide, (%)	Eutectic M ₂ C carbide, (%)
No.1 (Nb-free)	78.64	21.36	-	11.39	9.97
No.2 (0.70%Nb)	75.46	24.54	-	17.25	7.29
No.3 (1.05%Nb)	84.03	15.97	1.98	6.85	7.14
No.4 (2.58%Nb)	77.91	22.09	10.29	7.19	4.61

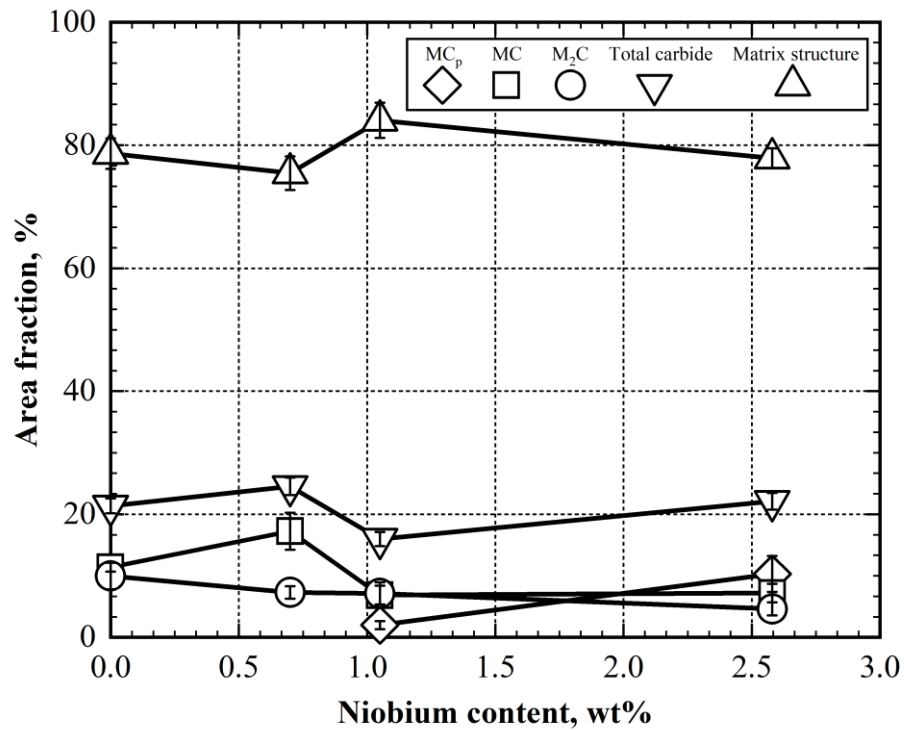


Figure 4.12 Effect of Nb content on the area fractions of the matrix and carbide phases in as-cast specimens.

4.1.3 Effect of Nb addition on macro-hardness and micro-hardness in as-cast condition

In the as-cast condition, the macro-hardness and micro-hardness values of specimens with different Nb contents are shown in Table 4.2 and Figure 4.13 respectively. The macro-hardness reflects the combined contribution of carbides and matrix structure, whereas the micro-hardness mainly reflects the hardness of the selected matrix region. Therefore, the hardness variation should be discussed separately in terms of the overall carbide/matrix structure and the local matrix response.

In the case of macro-hardness, the increase from 672.2 HV30 in the Nb-free specimen to 846.3 HV30 at 0.70%Nb is attributed to the increase in eutectic MC carbide fraction from 11.39% to 17.25%. Since MC carbides are hard phases and Nb promotes MC-type carbide formation during solidification (Ibrahim et al., 2020; Zhi et al., 2008), when the Nb content increased to 1.05%, the macro-hardness decreased to 769.6 HV30, corresponding to the decrease in eutectic MC and M_2C carbide fractions and reduction in total carbide fraction to 15.97%. In the 2.58%Nb alloy, the macro-hardness increased again to 860.9 HV30. This increase is mainly due to the pronounced increase in MC_p , together with the higher carbon content, which promoted carbide formation.

For micro-hardness, the value also increased from 580.6 HV0.1 in the Nb-free specimen to 772.8 HV0.1 at 0.70%Nb due to the refined matrix/carbide morphology. At 1.05%Nb, the micro-hardness decreased to 701.8 HV0.1, which is consistent with the reduction in eutectic carbide fraction and the increased matrix fraction. In the 2.58%Nb alloy, the micro-hardness increased again to 792.3 HV0.1. This increase may be related to the local influence of the increased MC_p carbides fraction and the elevated carbon content.

Table 4.2 Macro-hardness and micro-hardness of multi-alloyed white cast iron with different Nb content in as-cast condition.

Specimen	Macro-hardness, (HV30)	Micro-hardness, (HV0.1)
No.1 (Nb-free)	672.2	580.6
No.2 (0.70%Nb)	846.3	772.8
No.3 (1.05%Nb)	769.6	701.8
No.4 (2.58%Nb)	860.9	792.3

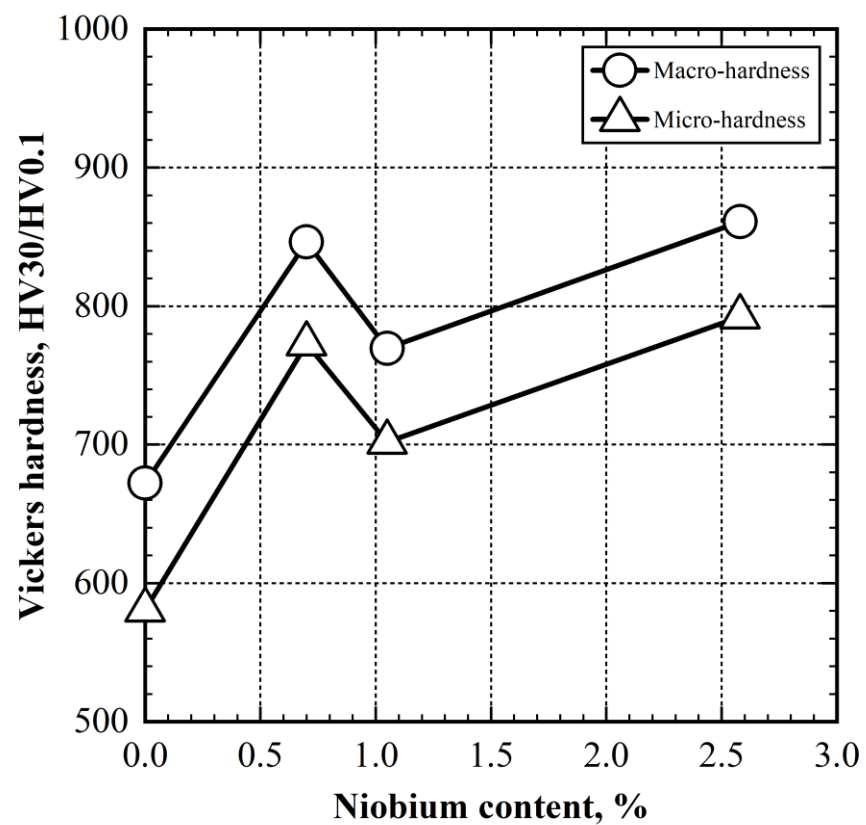


Figure 4.13 Relationship between Macro- and Micro-hardness and Nb content in as-cast condition.

4.2 Effect of Nb addition on heat treatment behavior

In practical applications, multi-alloyed white cast irons are commonly used in the heat-treated condition in order to obtain suitable mechanical properties for service. Previous studies have shown that the matrix structure and final properties of these alloys are strongly influenced by heat treatment particularly through secondary carbide precipitation and the transformation behavior of retained austenite (Inthidech & Matsubara, 2020b; Meebupha et al., 2017). Therefore, clarifying the heat treatment behavior of multi-alloyed white cast irons with different Nb content is important for understanding and controlling their final properties.

4.2.1 Relationship between heat treatment conditions and variation of matrix structure in as-hardened specimens

The type and morphology of eutectic carbides in multi-alloyed white cast irons are minimally affected by heat treatment as reported by several studies (Hashimoto et al., 2004b; Inthidech & Matsubara, 2020; Inthidech et al., 2021b; Meebupha et al., 2017; Opapaiboon et al., 2019a; Opapaiboon et al., 2024; Opapaiboon et al., 2015). In contrast, heat treatment conditions significantly affected the matrix microstructure. The matrix structures in the as-hardened state, examined by scanning electron microscopy (SEM), are presented in Figure 4.14 and 4.15. All specimens consist of secondary carbide (Cs), martensite (M), and retained austenite (γ_R). Consistent with the findings of Hashimoto et al. (2004a), the secondary carbides were identified primarily as MC, M_6C , and M_7C_3 . These results indicate that austenite destabilization occurs during austenitizing through the precipitation of secondary carbides, which subsequently promotes the martensitic transformation upon quenching.

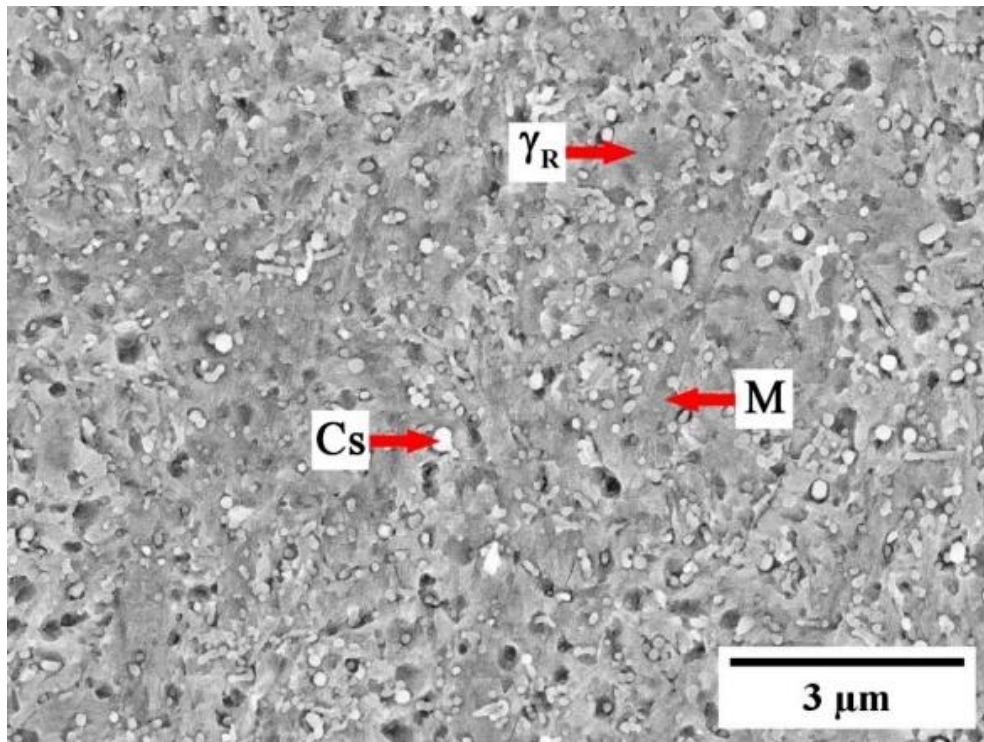
The area fraction of secondary carbides was quantified from back-scattered electron images using image analysis software, as summarized in Table 4.3 and Figure 4.16. After austenitizing at 1050°C, the secondary carbide area fraction decreased from 10.38% in the Nb-free specimen to 8.81% in the 0.70%Nb specimen. This decrease is attributed to the increase in eutectic MC carbides in the 0.70%Nb specimen, which consumed carbon and strong carbide-forming elements during solidification and reduced the carbon available in the matrix for secondary carbide precipitation during austenitizing. In the 1.05%Nb specimen, the secondary carbide area fraction reached a minimum value of 8.03%. This minimum value is associated with

the crystallization of primary MC carbides (MC_p), which further consumed C and V in the matrix for the formation of secondary carbides. After reaching the minimum value, the area fraction of secondary carbide increased to 10.23% in the 2.58%Nb specimen. This increase is attributed to the higher carbon content of specimen No.4, which may provide sufficient carbon for both carbide formation during solidification and secondary carbide precipitation during austenitizing.

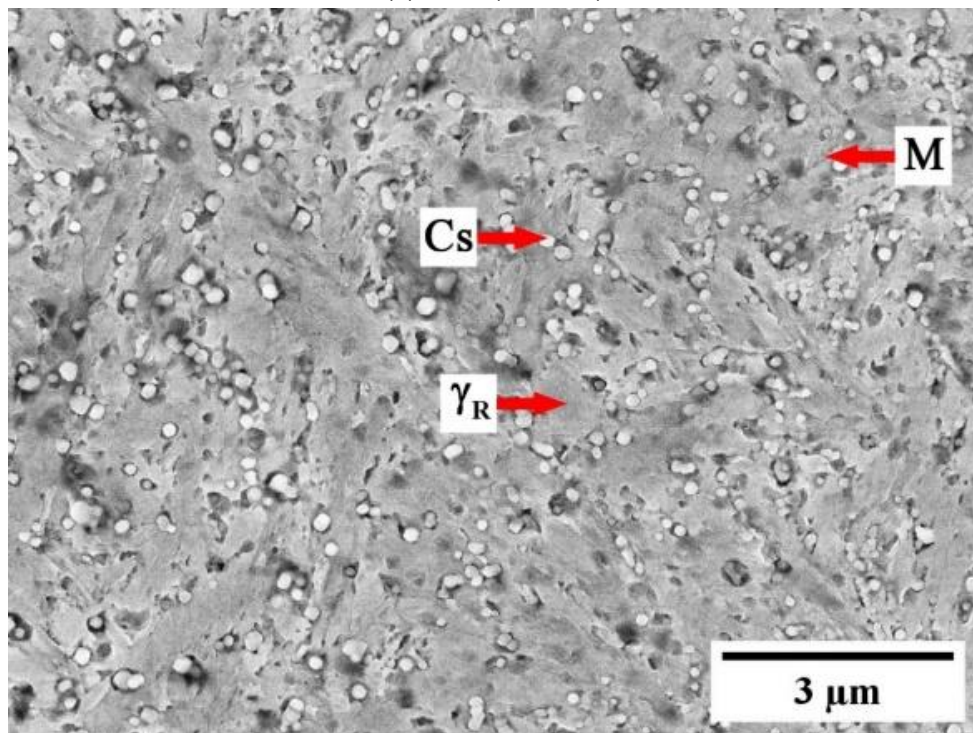
A similar trend was observed after austenitizing at 1100°C. The area fraction of secondary carbide decreased from 7.48% in the Nb-free specimen to 7.02% in the 0.70%Nb specimen, reaching a minimum value of 6.13% in the 1.05%Nb specimen and increased to 8.55% in the specimen with 2.58%Nb. Compared with austenitizing at 1050°C, the 1100°C treatment produced a lower secondary carbide fraction in all specimens. This result suggests that a higher austenitizing temperature increases the dissolution of carbon and alloying elements into austenite, thereby increasing austenite stability. A similar effect of higher austenitizing temperature on retained austenite has been reported by Inthidech and Matsubara (2020) and Inthidech et al. (2021b).

Figure 4.17 shows the matrix alloy concentration in as-hardened specimens with different Nb content. After austenitizing at 1050°C, the Cr, Mo and W concentration in the matrix remained approximately constant as the Nb content increased. In contrast, the matrix V concentration tended to decrease in the Nb-containing specimens. This tendency is consistent with the formation of MC-type carbides, as suggested by the increased eutectic MC carbide fraction in specimen No.2 and the crystallization of MC_p in specimen No.3 and No.4. Nb was detected only at a very low concentration in the matrix. This result suggests that Nb preferentially partitioned into the eutectic MC and MC_p carbides rather than remaining in solid solution in the matrix (Melo et al., 2020; Shabalin, 2019; Silva et al., 2021).

Compared with the specimens austenitized at 1050°C, those austenitized at 1100°C generally showed similar or slightly higher matrix concentration for some alloying elements, particularly Cr. The slight increase observed can be associated with the greater solubility of carbon and alloying elements and enhanced dissolution of carbides at the higher austenitizing temperature (Opapaiboon et al., 2019a).

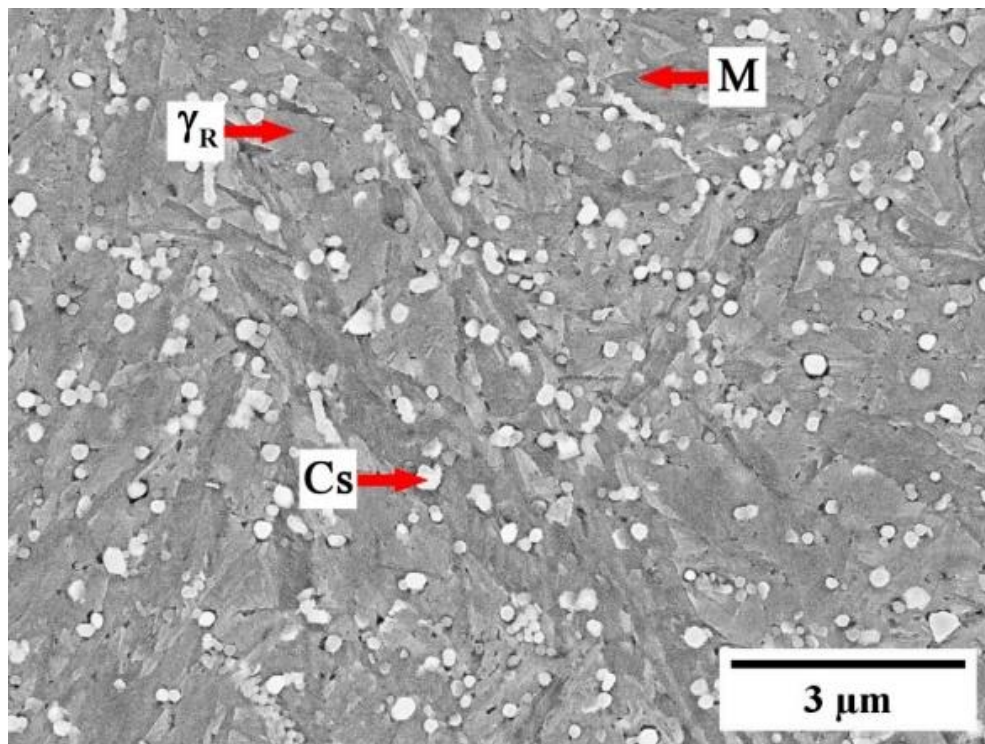


(a) No.1 (Nb-free)

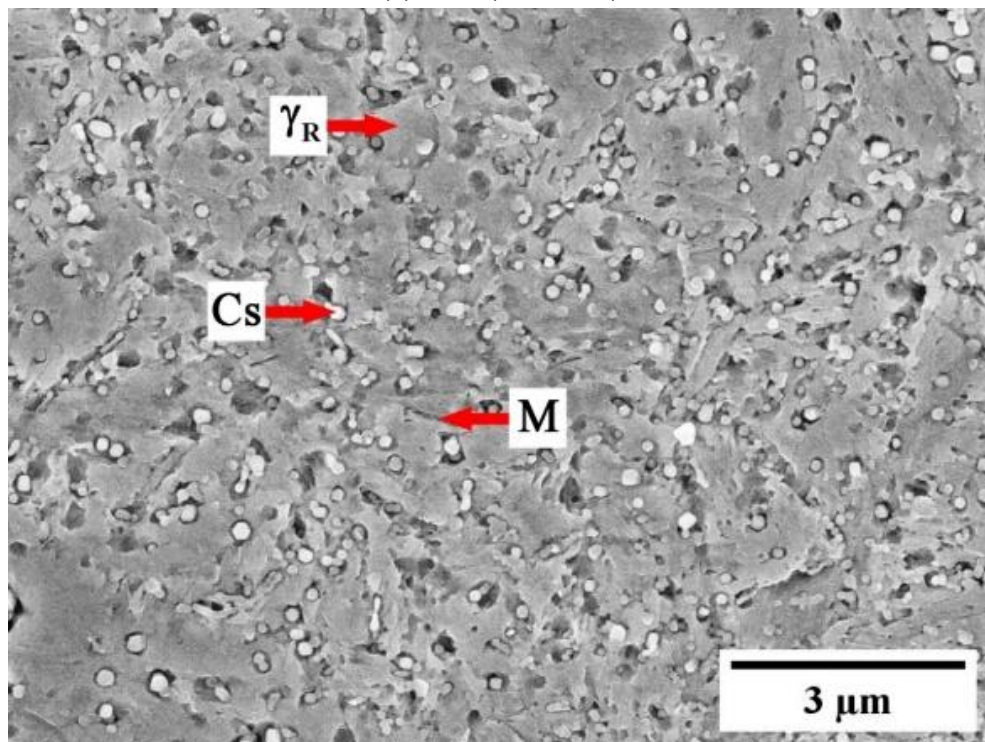


(b) No.2 (0.70%Nb)

Figure 4.14 Matrix microstructure of as-hardened specimens with different Nb; (a) No.1 (Nb-free), (b) No.2 (0.70%Nb).

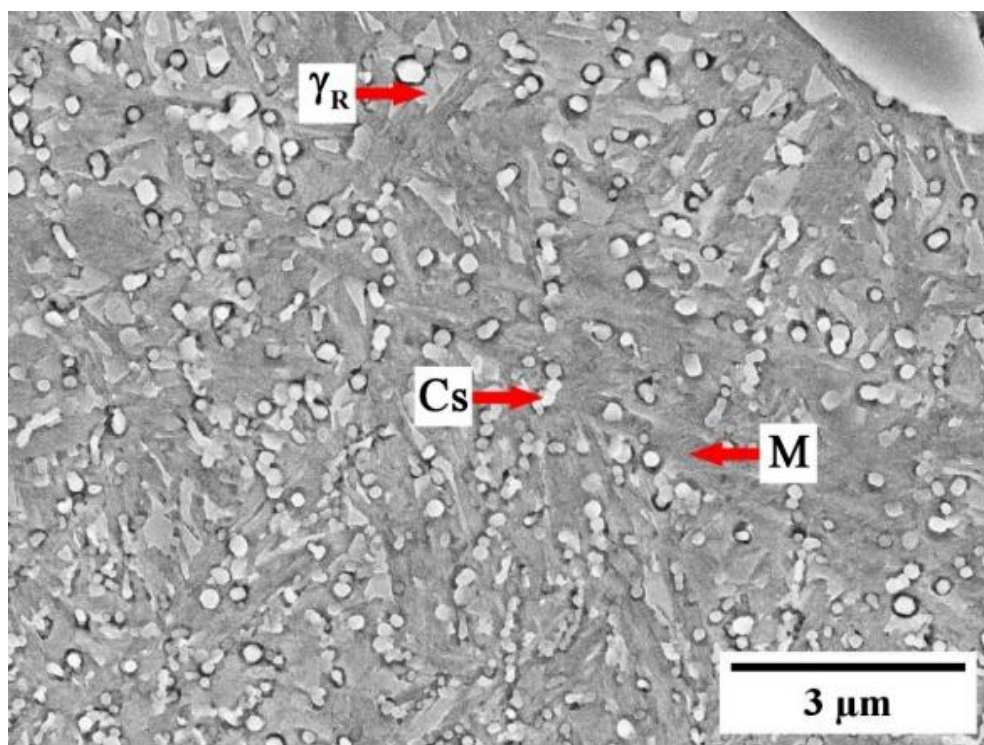


(c) No.3 (1.05%Nb)

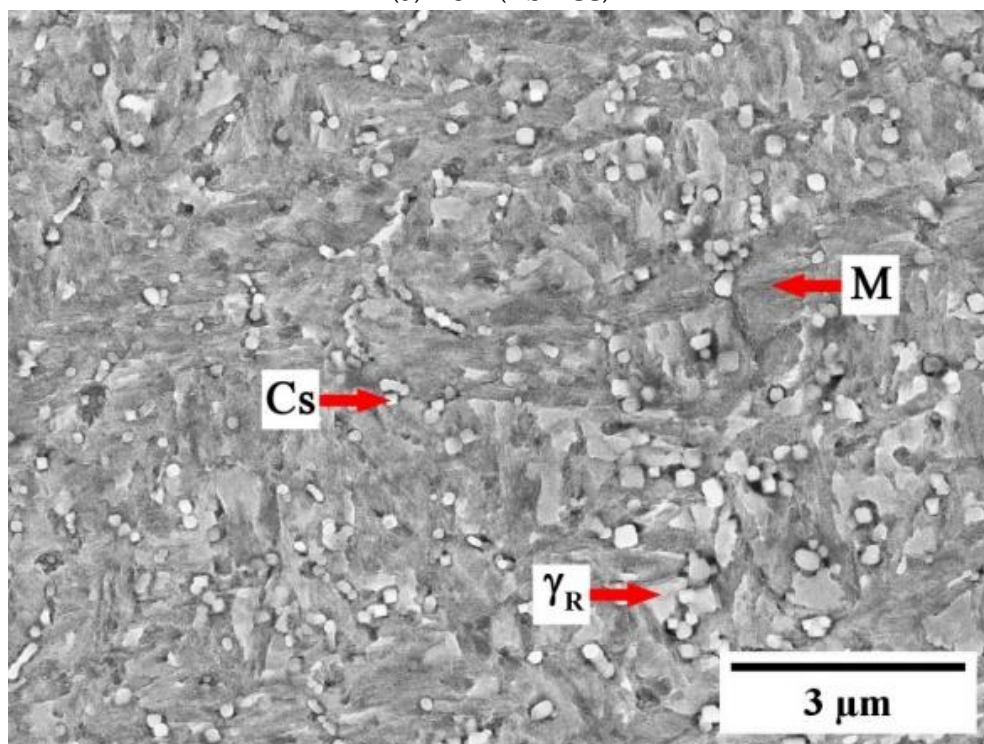


(d) No.4 (2.58%Nb)

Figure 4.14 Matrix microstructure of as-hardened specimens with different Nb; (c) No.3 (1.05%Nb) and (d) No.4 (2.58%Nb). Austenitizing temperature: 1050°C.

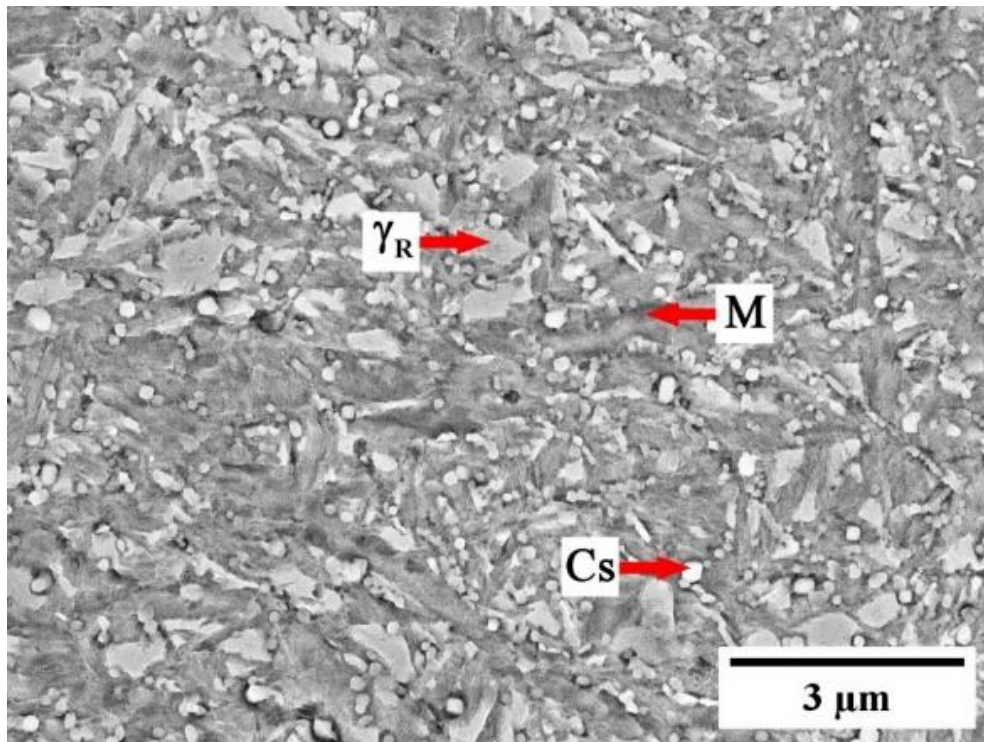


(a) No.1 (Nb-free)

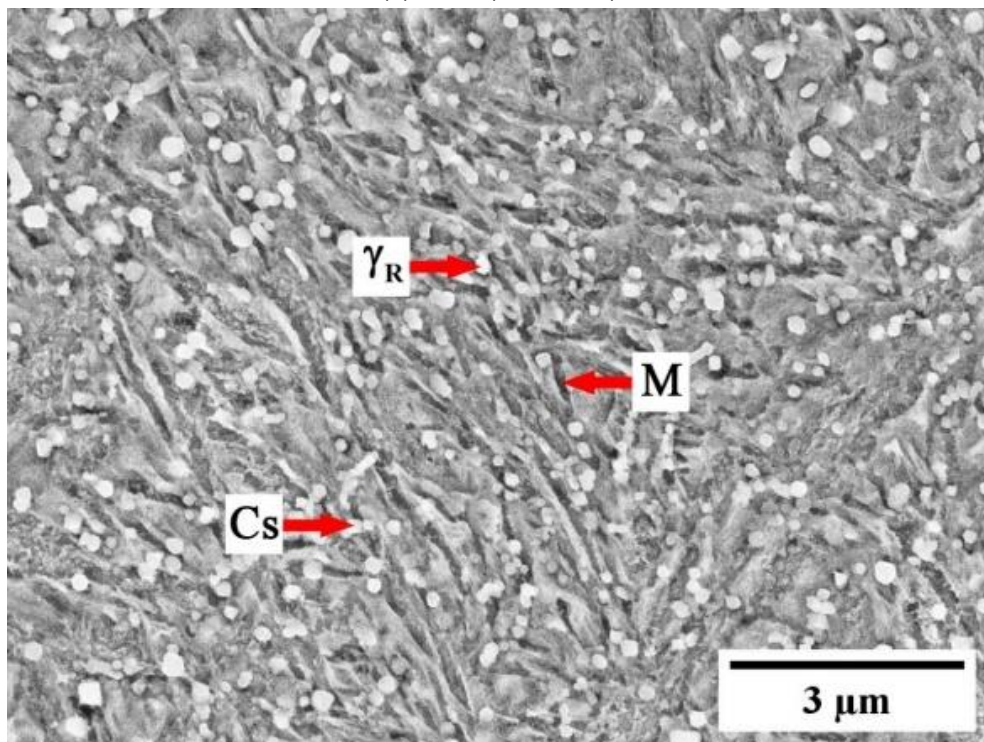


(b) No.2 (0.70%Nb)

Figure 4.15 Matrix microstructures of as-hardened specimens with different Nb; (a) No.1 (Nb-free), (b) No.2 (0.70%Nb).



(c) No.3 (1.05%Nb)



(d) No.4 (2.58%Nb)

Figure 4.15 Matrix microstructures of as-hardened specimens with different Nb; (c) No.3 (1.05%Nb) and (d) No.4 (2.58%Nb). Austenitizing temperature: 1100°C.

Table 4.3 Area fraction of secondary carbide precipitated in the as-hardened specimen.

Specimen	Area fraction of secondary carbide, %	
	1050°C austenitizing	1100°C austenitizing
No.1 (Nb-free)	10.38	7.48
No.2 (0.70%Nb)	8.81	7.02
No.3 (1.05%Nb)	8.03	6.13
No.4 (2.58%Nb)	10.23	8.55

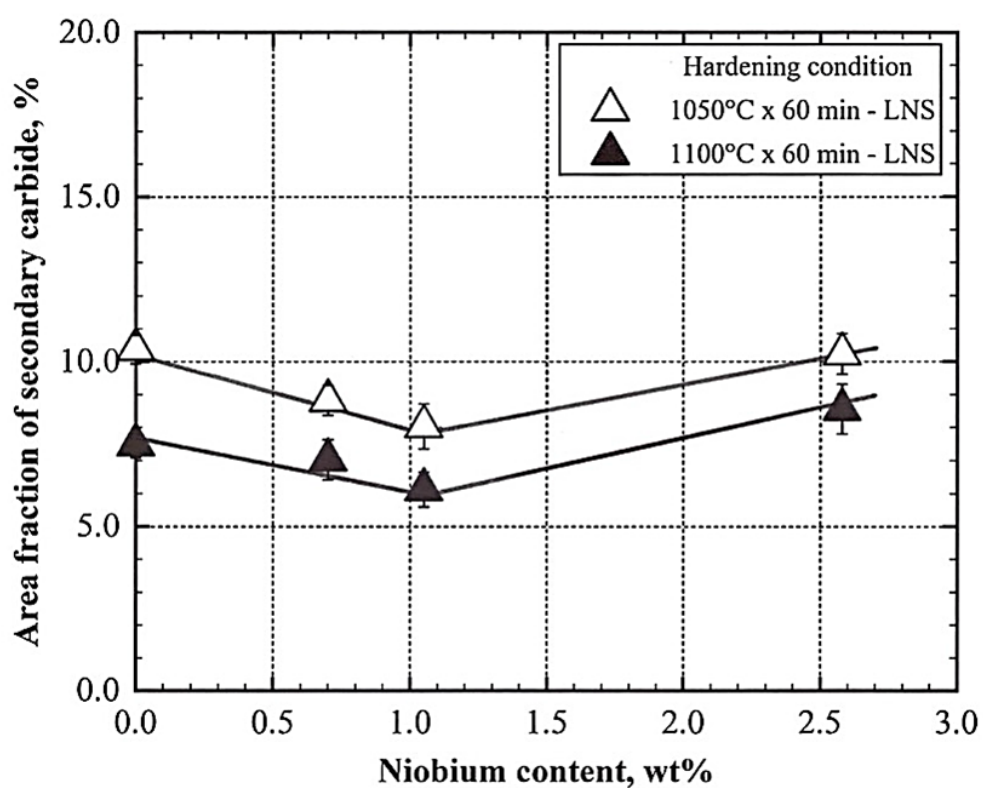
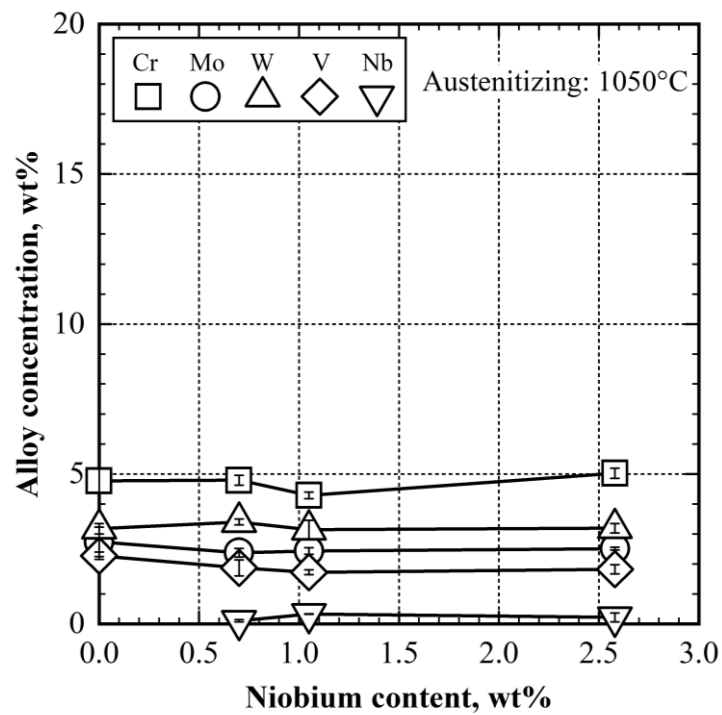
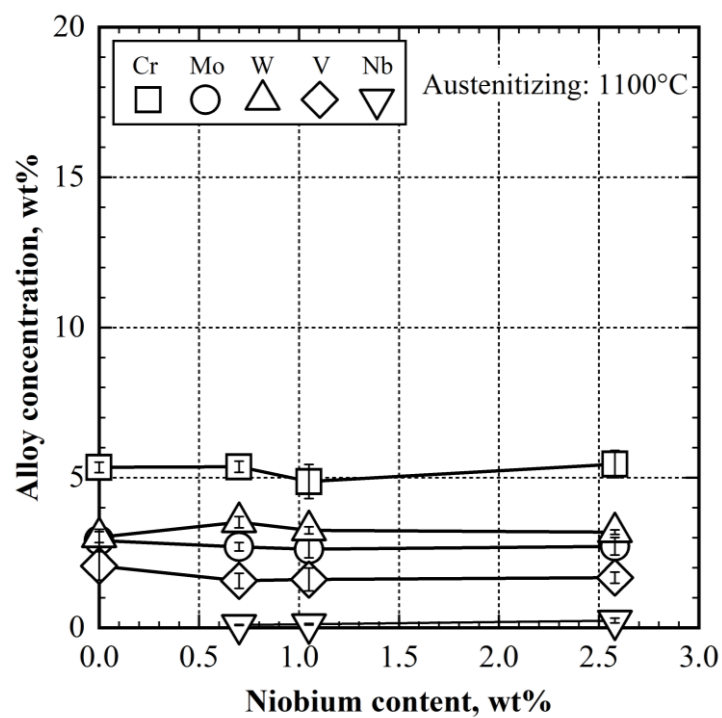


Figure 4.16 Relationship between Nb content and area fraction of secondary carbide precipitated in the as-hardened specimens.



(5) 1050°C austenitizing temperature.



(b) 1100°C austenitizing temperature

Figure 4.17 Matrix alloy concentration in as-hardened specimens with different Nb content.; (a) 1050°C austenitizing and (b) 1100°C austenitizing.

4.2.2 Behavior of hardness and volume fraction of retained austenite (V_γ) in as-hardened specimens

The influence of Nb content on macro-hardness, micro-hardness and the volume fraction of retained austenite (V_γ) in the as-hardened condition is shown in Table 4.4 and Figure 4.18. After austenitizing at 1050°C, the macro-hardness initially decreased from 927.1 HV30 in the Nb-free specimen to 913.5 HV30 in the 0.70%Nb specimen and reached a minimum of 909.5 HV30 in the 1.05%Nb specimen. After reaching this minimum value, the macro-hardness increased to 960.2 HV30 in 2.58%Nb specimen. Regarding V_γ , increasing the Nb content increased the V_γ from 10.3% in the Nb-free specimen to 15.2% in the 0.70%Nb specimen. Thereafter, V_γ slightly increased to 15.9% in the 1.05%Nb and reached a maximum value of 18.5% in the 2.58%Nb specimen.

In the case of austenitizing at 1100°C, the macro-hardness and V_γ exhibited trends similar to those observed at 1050°C. The macro-hardness decreased from 904.8 HV30 in the Nb-free specimen to 885.7 HV30 in the 0.70%Nb and reached a minimum value of 841.7 HV30 in the 1.05%Nb. Thereafter, macro-hardness increased to 872.0 HV30 in specimen No.4 with 2.58%Nb. The V_γ increased from 14.7% in the Nb-free specimen to 22.1% in the 0.70%Nb, 26.8% in the 1.05%Nb and 28.2% in the 2.58%Nb specimen.

The decrease in the macro-hardness of the 0.70%Nb specimen can be attributed to the reduction in area fraction of secondary carbide. The increased amount of eutectic MC carbide consumed a larger proportion of carbon during solidification, thereby reducing the amount of carbon available in the matrix (Inthidech & Matsubara, 2020). During austenitizing, the low carbon availability limited the precipitation of secondary carbides because their formation requires both carbon and carbide-forming elements (Opapaiboon et al., 2022). Consequently, a larger proportion of the remaining carbide-forming elements could remain dissolved in the austenitic matrix. If the austenite-stabilizing effect of these dissolved alloying elements outweighed the opposing effect of the reduced carbon concentration, the martensite-start temperature (M_s) would decrease, resulting in an increase in the volume fraction of retained austenite (V_γ) (Opapaiboon et al., 2024). In addition, the reduced carbon concentration in the matrix produced martensite with a lower intrinsic hardness, while the increased

amount of relatively soft retained austenite further reduced the matrix hardness. Consequently, the overall hardness of 0.70%Nb specimen decreased.

In the case of 1.05%Nb, where MC_p carbides began to crystallize, a significant amount of V, Nb and C was consumed to form MC_p carbides. This suppressed the formation of both MC and M_2C eutectic carbides and greatly reduced the precipitation of secondary carbides. Although MC_p carbides are intrinsically hard, their strengthening effect was outweighed by the pronounced reduction in total carbide fraction, together with the higher V_γ , resulting in decreased hardness. In addition, the hardness of martensite itself is reduced due to the diminished carbon content in the matrix. These results are consistent with the finding of Kim et al. (2003), who reported that the crystallization of MC_p reduces the dissolution of carbide forming elements in the matrix, thereby lowering the matrix hardness.

In specimen No.4 with 2.58%Nb, a distinct increase in hardness was observed. This improvement can be explained by the addition of carbon required to compensate for the formation of MC_p carbides during solidification. The excess carbon not only promotes the precipitation of eutectic carbides but also enriched the matrix, thereby facilitating secondary carbide precipitation during austenitizing and enhancing martensitic hardness after quenching. These combined mechanisms account for the recovery and subsequent enhancement of hardness at higher Nb and C contents.

Regardless of Nb content, higher austenitizing temperature consistently resulted in lower hardness values due to higher V_γ . At higher austenitizing temperatures, carbon and alloying elements can easily dissolve into the matrix, which decreases the M_s temperature. As a result, the higher V_γ was obtained.

The increase in the amount of retained austenite with rising Nb content can be attributed to the reduced precipitation of secondary carbides in the matrix as shown in Figure 4.16. Thus, although Nb itself is not regarded as a strong austenite stabilizer, its indirect role through the modification of carbide chemistry markedly increases the retained austenite fraction after quenching. This effect becomes more pronounced at elevated austenitizing temperatures, where the enhanced solubility of alloying elements in austenite promotes further stabilization during quenching, ultimately resulting in an increased V_γ value.

From the above results, the highest macro-hardness after austenitizing at 1050°C was obtained in the specimen No.4, with a value of 960.2 HV30. In contrast,

after austenitizing at 1100°C, the highest macro-hardness was obtained in the Nb-free specimen, with a value of 904.8 HV30. As illustrated in Figure 4.19, the micro-hardness exhibited a trend similar to that of the macro-hardness. The highest micro-hardness of 878.8 HV0.1 was obtained in the specimen No.4 after austenitizing at 1050°C. For the specimens austenitized at 1100°C, the Nb-free specimen showed the highest micro-hardness, with a value of 841.5 HV0.1. In contrast, V_γ increased steadily with increasing Nb content and the highest V_γ values of 18.5% and 28.2% were obtained in specimen No. 4 after austenitizing at 1050°C and 1100°C, respectively.

Further evidence supporting this interpretation is provided by the correlation between micro-hardness and the retained austenite fraction (V_γ), as shown in Figure 4.20. Specimens with higher V_γ exhibited lower micro-hardness values up to approximately 16 and 27% V_γ in 1050°C and 1100°C, respectively, thereby confirming both the softening effect of retained austenite relative to martensite and the reduction in secondary carbide precipitation. Notably, a sudden increase in micro-hardness was observed over 16 and 27% V_γ (corresponding to the 2.58%Nb specimen), which suggests the precipitation of additional secondary carbides with exceptionally high intrinsic hardness. This hardening phenomenon markedly enhanced the matrix strength despite the nearly unchanged level of retained austenite, highlighting the complex interplay among secondary carbide precipitation, retained austenite stability, and martensitic transformation.

Table 4.4 Macro-hardness, micro-hardness and volume fraction of retained austenite (V_γ) in as-hardened specimens.

Specimen	1050°C austenitizing			1100°C austenitizing		
	Macro-hardness, HV30	Micro-hardness, HV0.1	V_γ , %	Macro-hardness, HV30	Micro-hardness, HV0.1	V_γ , %
No.1 (Nb-free)	927.1	857.6	10.3	904.8	841.5	14.7
No.2 (0.70%Nb)	913.5	841.5	15.2	885.7	799.6	22.1
No.3 (1.05%Nb)	909.5	836.5	15.9	841.7	779.3	26.8
No.4 (2.58%Nb)	960.2	878.8	18.5	872.0	811.3	28.2

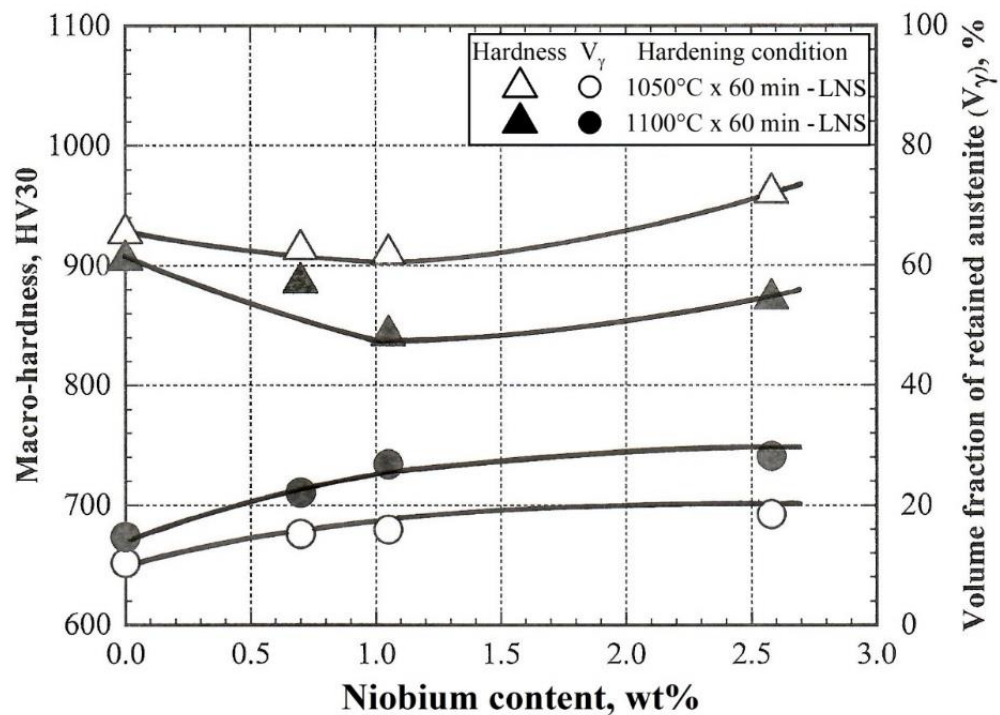


Figure 4.18 Relationship between macro-hardness, volume fraction of retained austenite (V_γ) and Nb content in as-hardened condition.

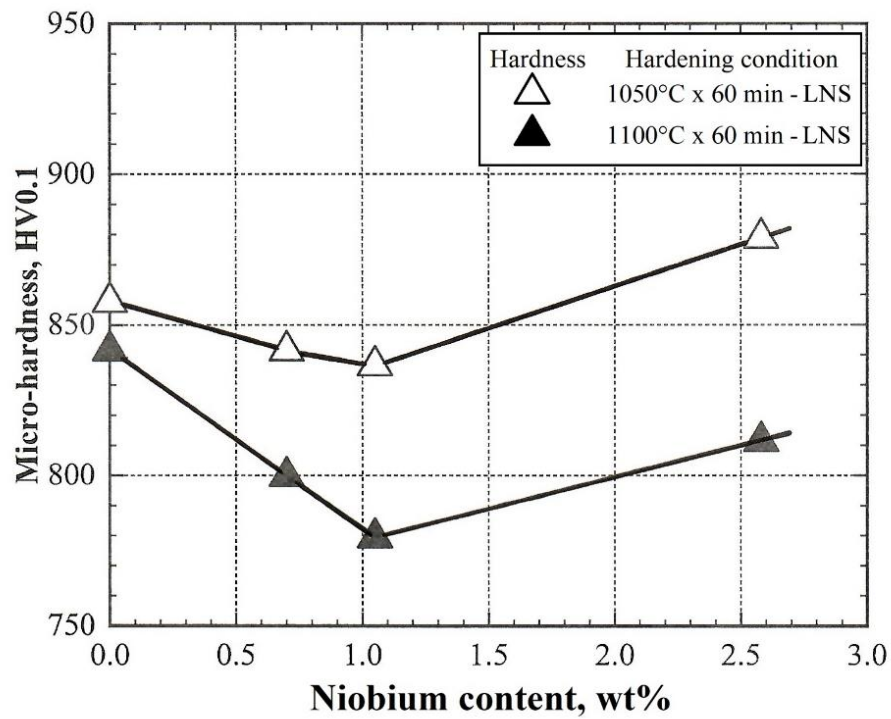


Figure 4.19 Relationship between micro-hardness and Nb content in as-hardened specimens.

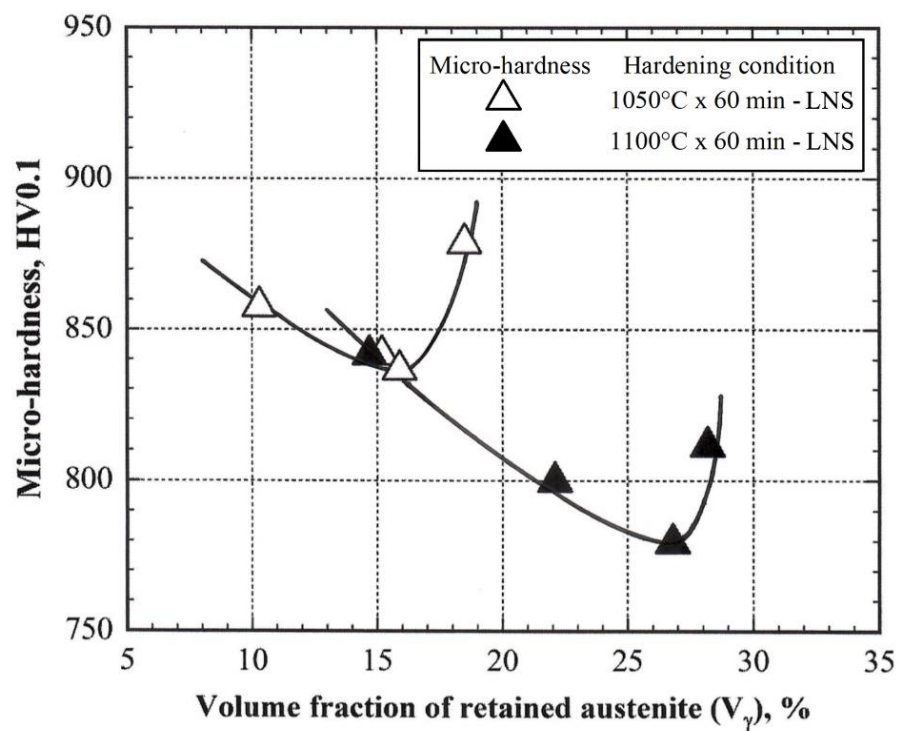


Figure 4.20 Relationship between micro-hardness and volume fraction of retained austenite (V_γ) in as-hardened condition.

4.2.3 Effect of tempering temperature on hardness and volume fraction of retained austenite (V_γ)

After hardening from austenitizing temperature of 1050°C and 1100°C, the specimens were tempered between 400°C and 600°C at 50°C intervals, with additional 25°C intervals applied near the temperature range where the highest hardness was expected. The relationships between macro-hardness, micro-hardness, volume fraction of retained austenite (V_γ) and tempering temperature are shown in Figure 4.21 to Figure 4.24 for specimen No.1 to No.4 respectively. For reference, hardness and V_γ values in the as-hardened condition are also included. All specimens exhibited the characteristic secondary hardening behavior, that is, an initial decrease in hardness at 400°C, followed by an increase to maximum value (H_{Tmax}), and a subsequent sharp decline at higher temperatures. Macro-hardness curves displayed a similar trend to micro-hardness but with consistently higher values, reflecting the fact that macro-hardness measurements incorporate both the matrix and carbide phases, whereas micro-hardness values represent only the matrix. The V_γ decreased only gradually at low tempering temperatures, dropped sharply as the temperature approached H_{Tmax} , and then decreased gradually again at higher temperatures. At H_{Tmax} , V_γ values were consistently below 10%, regardless of austenitizing temperature.

As shown in Figure 4.21, the macro-hardness and micro-hardness of specimen No.1 increased and reached the maximum tempered hardness (H_{Tmax}) at a tempering temperature of 525°C. For the specimen austenitized at 1050°C, the H_{Tmax} values were 894.6 HV30 and 851.5 HV0.1 for macro-hardness and micro-hardness, respectively. For the specimen austenitized at 1100°C, higher H_{Tmax} values of 922.1 HV30 and 869.6 HV0.1 were obtained at the same tempering temperature. Therefore, the higher austenitizing temperature resulted in higher maximum tempered hardness than lower austenitizing temperature. In the case of V_γ , increasing the tempering temperature gradually decreased the volume fraction of retained austenite from that in the as-hardened state.

As presented in Figure 4.22, specimen No.2 also exhibited a secondary hardening response, with the H_{Tmax} appearing at 525°C for both austenitizing temperatures. However, the hardness was lower than that of specimen No.1 throughout most of the tempering range. After austenitizing at 1050°C, the H_{Tmax} values were 873.8 HV30 and 818.1 HV0.1. In the case of 1100°C austenitizing, the H_{Tmax}

increased to 911.1 HV30 and 848.8 HV0.1. The V_γ decreased slightly at first and gradually decreased where the tempering temperature exceeded 500°C.

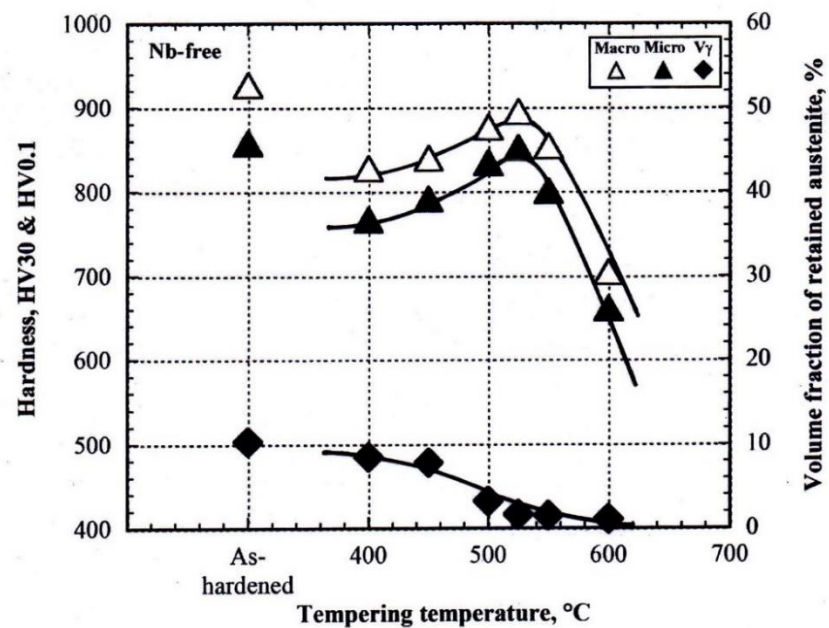
In the case of specimen No.3 as shown in Figure 4.23, the lowest H_{Tmax} values among the investigated specimens were obtained at both austenitizing temperatures. For the specimen austenitized at 1050°C, H_{Tmax} values of 857.2 HV30 and 792.9 HV0.1 were obtained after tempering at 525°C. Similarly, for the specimen austenitized at 1100°C, the H_{Tmax} was also obtained at a tempering temperature of 525°C, with corresponding values of 896.6 HV30 and 845.1 HV0.1. Regarding V_γ , specimen No.3 exhibited a trend similar to that of specimen No.2.

The results for specimen No. 4 are shown in Figure 4.24. Among all specimens, this specimen exhibited the highest H_{Tmax} at both austenitizing temperatures, which were obtained after tempering at 525°C. For the specimen austenitized at 1050°C, the H_{Tmax} values were 907.1 HV30 and 817.5 HV0.1. In the case of specimen austenitized at 1100°C, the H_{Tmax} values were 927.0 HV30 and 854.7 HV0.1. Regarding V_γ , the V_γ slightly decreased at lower tempering temperature and then gradually decreased when the tempering temperature exceeded 500°C.

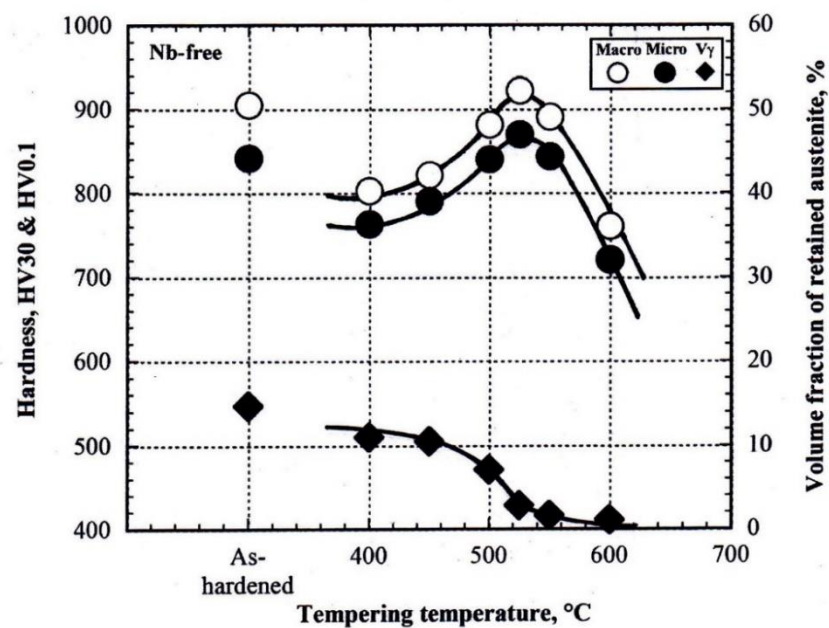
The initial hardness reduction at 400°C relative to the as-hardened state is primarily attributed to the tempering of martensite, specifically the precipitation of very fine $M(Mo, W, V, Cr)_3C$ within the martensitic lattice (Laird et al., 2000; Opapaiboon et al., 2024). At this relatively low temperature, the decomposition of retained austenite is limited because the thermal energy is insufficient to destabilize the austenite phase. As the tempering temperature increases, hardness rises due to secondary carbide precipitation within the retained austenite. This process depletes carbon and alloying elements from the austenite, reducing its stability and promoting its transformation into martensite during post-tempering cooling, thereby enhancing matrix hardness. The maximum hardness was observed at 525°C for specimens austenitized at both 1050 and 1100°C, corresponding to the peak rate of retained austenite transformation into martensite. At this stage, additional carbide reactions within the martensite matrix produce nearly pure secondary carbides (Opapaiboon et al., 2024) such as MC , M_6C , and M_2C , which exhibit high hardness and significantly strengthen the alloy (Hashimoto et al., 2004a; Inthidech & Matsubara, 2020; Inthidech et al., 2021b; Matsubara et al., 1999; Meebupha et al., 2017; Opapaiboon et al., 2019a;

Opapaiboon et al., 2022; Opapaiboon et al., 2024; Opapaiboon et al., 2015). The schematic diagram of carbide reactions during tempering is shown in Figure 4.25.

When the tempering temperature exceeds 525°C, however, coarsening and agglomeration of secondary carbides occur via Ostwald ripening, while excessively high temperatures may also promote the transformation of retained austenite into softer phases such as pearlite or ferrite. These microstructural changes collectively lead to a pronounced reduction in matrix hardness. Furthermore, specimens hardened at higher austenitizing temperatures consistently exhibited higher H_{Tmax} compared to those treated at lower austenitizing temperatures. This behavior is attributed to the increased dissolution of carbon and alloying elements in austenite at elevated austenitizing temperatures, which enhances secondary hardening during tempering and results in greater hardness.

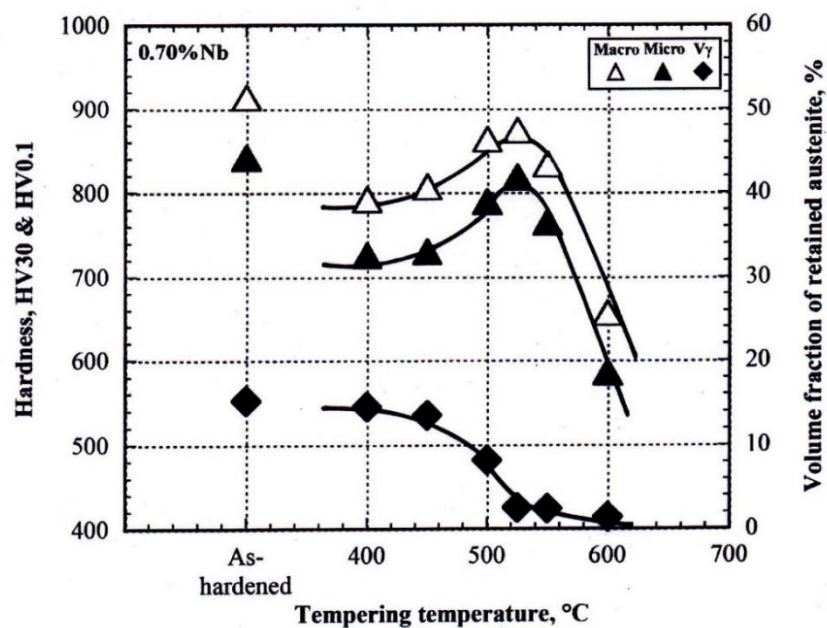


(a) Austenitizing temperature: 1050°C

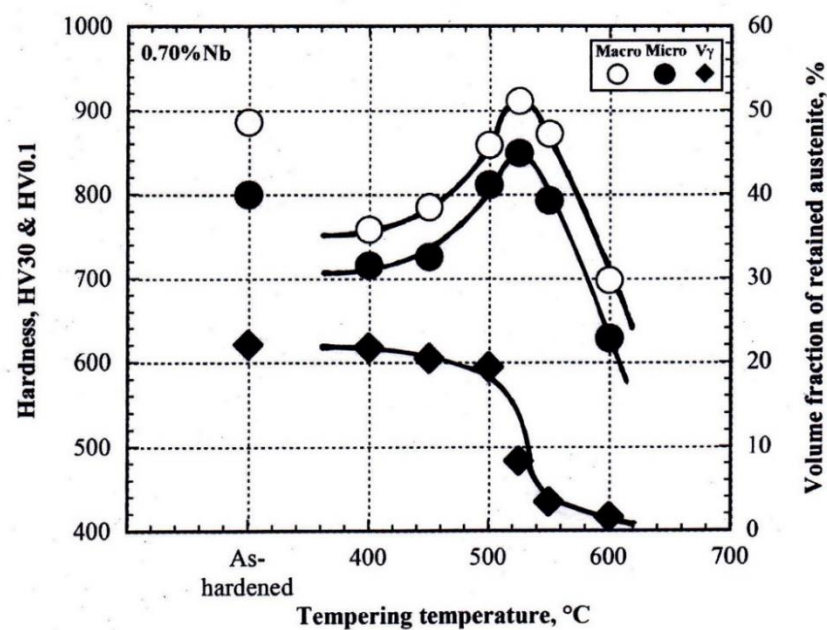


(b) Austenitizing temperature: 1100°C

Figure 4.21 Relationship between macro-hardness, micro-hardness, volume fraction of retained austenite (V_{γ}) and tempering temperature of specimen No.1 (Nb-free); (a) Austenitizing temperature: 1050°C and (b) Austenitizing temperature: 1100°C.

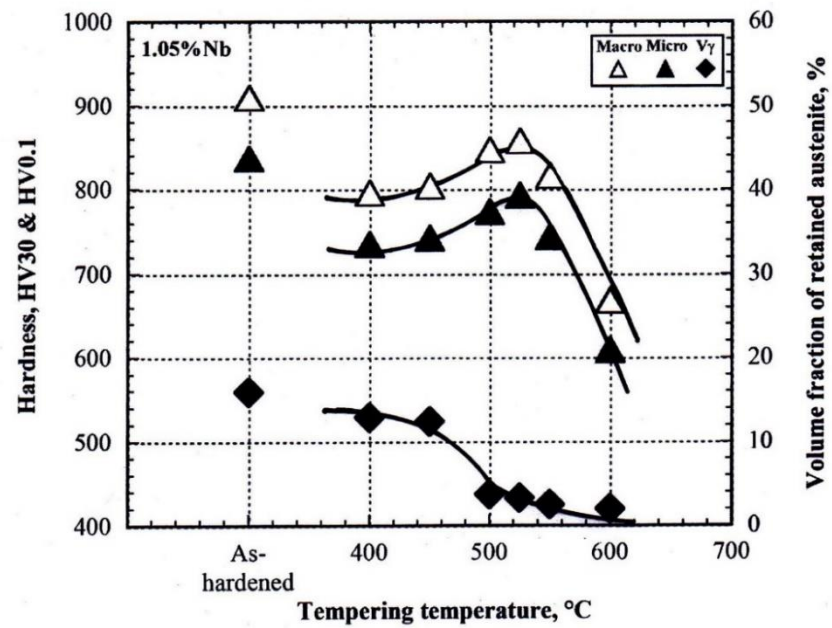


(a) Austenitizing temperature: 1050°C

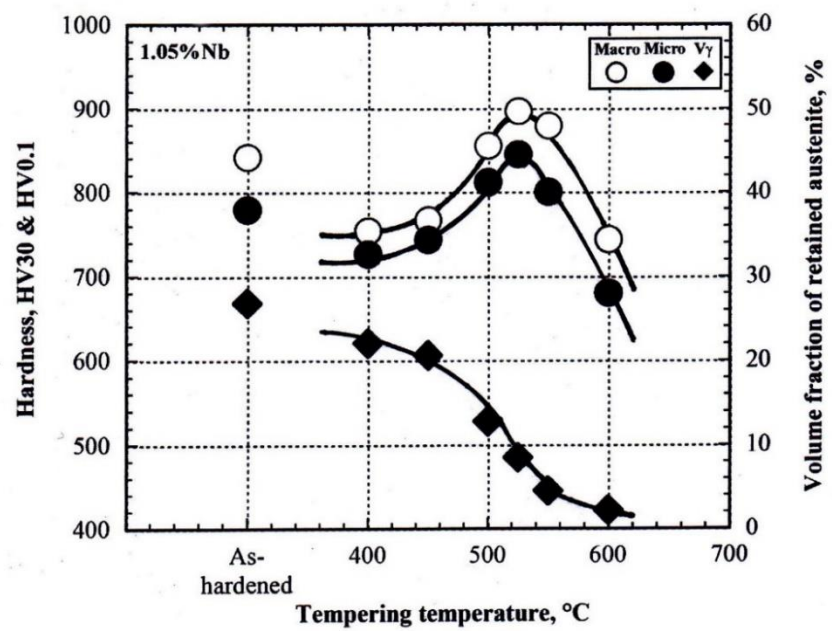


(b) Austenitizing temperature: 1100°C

Figure 4.22 Relationship between macro-hardness, micro-hardness, volume fraction of retained austenite (V_γ) and tempering temperature of specimen No.2 (0.70%Nb); (a) Austenitizing temperature: 1050°C and (b) Austenitizing temperature: 1100°C.

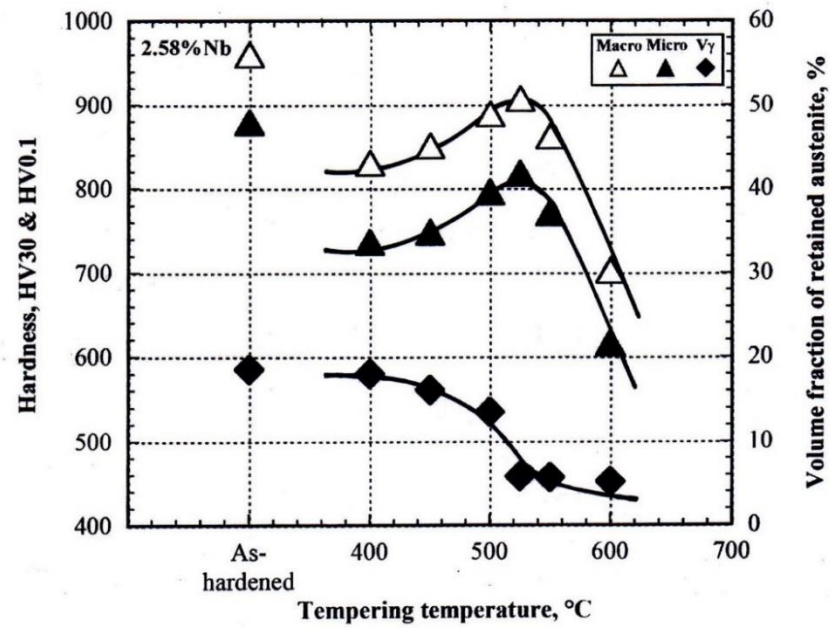


(a) Austenitizing temperature: 1050°C

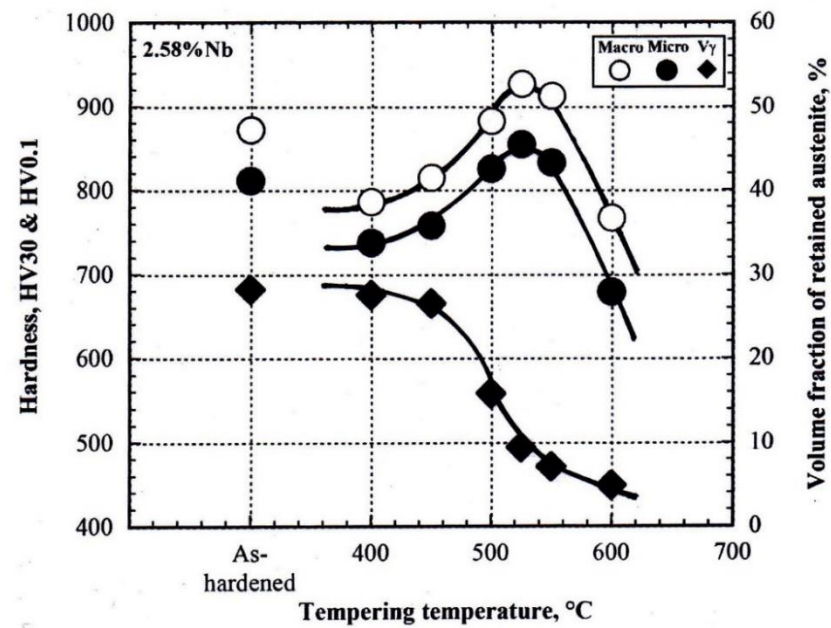


(b) Austenitizing temperature: 1100°C

Figure 4.23 Relationship between macro-hardness, micro-hardness, volume fraction of retained austenite (V_γ) and tempering temperature of specimen No.3 (1.05%Nb); (a) Austenitizing temperature: 1050°C and (b) Austenitizing temperature: 1100°C.



(a) Austenitizing temperature: 1050°C



(b) Austenitizing temperature: 1100°C

Figure 4.24 Relationship between macro-hardness, micro-hardness, volume fraction of retained austenite (V_γ) and tempering temperature of specimen No.4 (2.58%Nb); (a) Austenitizing temperature: 1050°C and (b) Austenitizing temperature: 1100°C.

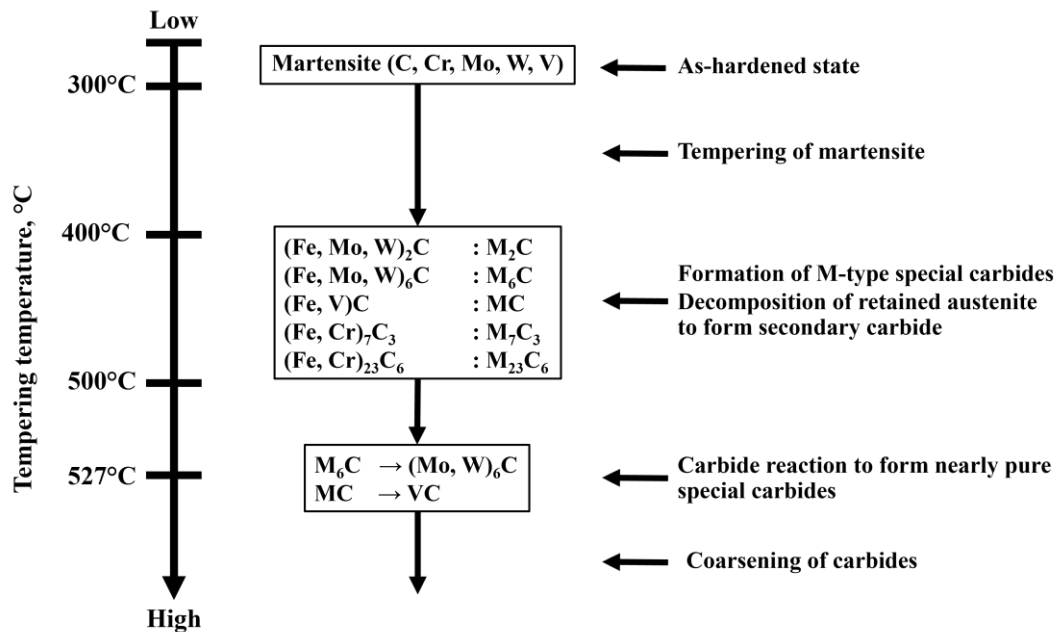


Figure 4.25 Schematic diagram illustrating carbide reactions during tempering from as-hardened state of matrix in multi-alloyed white cast iron (Opapaiboon et al., 2024).

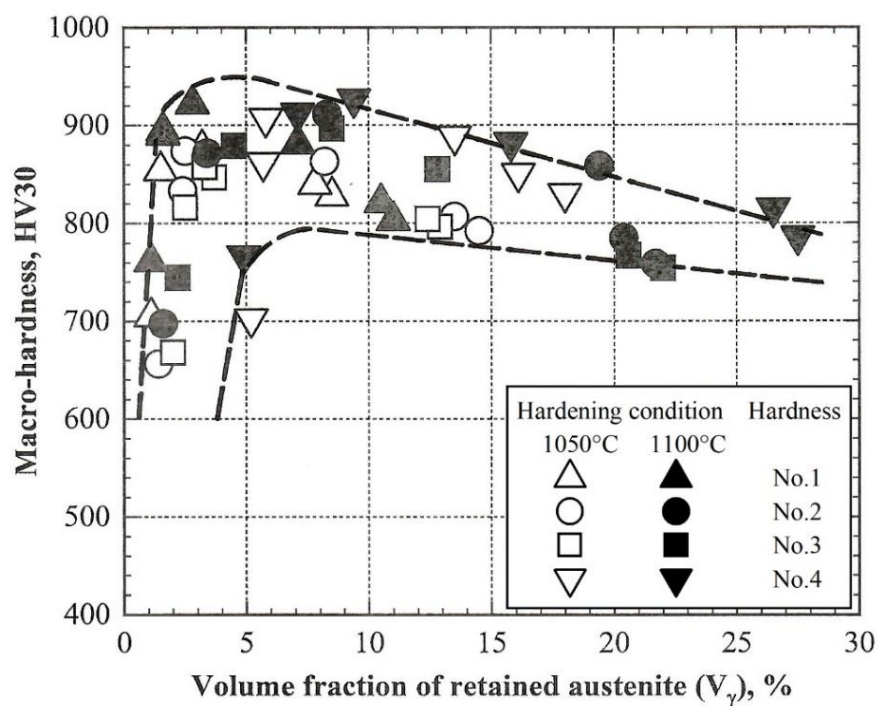
4.2.4 Effect of volume fraction of retained austenite (V_γ) on tempered hardness

As discussed earlier, the hardness of cast iron is largely governed by the presence of eutectic carbides and the characteristics of the matrix microstructure. Tempering minimally affects eutectic carbides; however, it significantly influences the matrix, which dictates the overall hardness. Consequently, the amount of retained austenite in the tempered matrix becomes a critical factor influencing the mechanical properties, particularly hardness. The relationship between macro-hardness, micro-hardness, and the volume fraction of retained austenite (V_γ) for all tempered specimens is presented in Figure 4.26. Both macro-hardness and micro-hardness show similar trends, that is, hardness increases with V_γ sharply, reaching a maximum at approximately 2-3% V_γ , and then decreases with increases in V_γ . Although the overall variation in micro-hardness closely parallels that of macro-hardness, the absolute values of micro-hardness remain consistently lower. At a constant V_γ , specimens quenched from 1100°C exhibited higher hardness than those quenched from 1050°C. This difference can be attributed to the stronger secondary hardening effect observed at higher austenitizing temperatures. Notably, maintaining V_γ within the range of

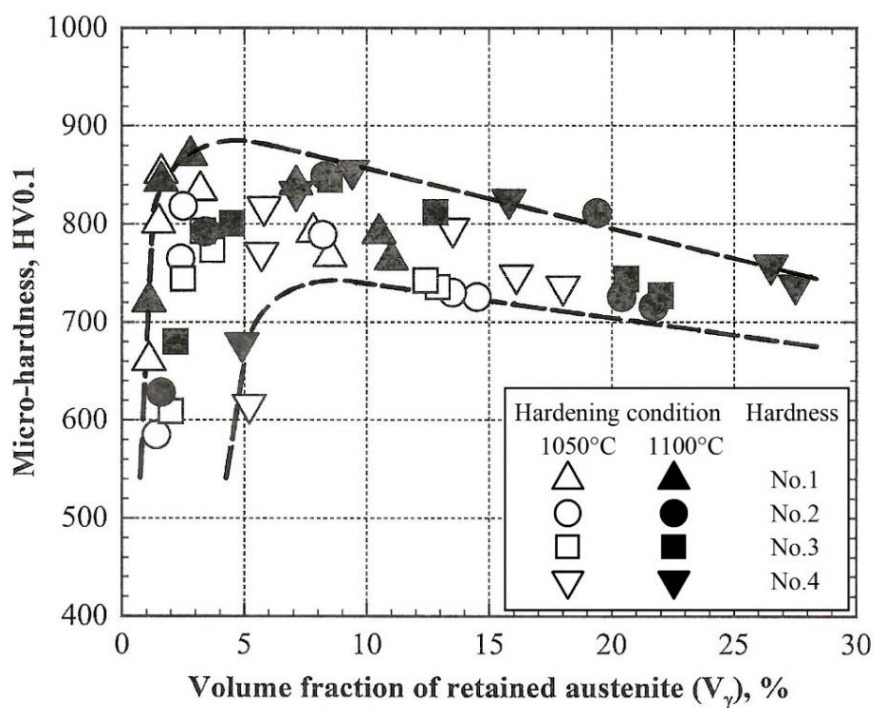
approximately 2-10% was found to be effective for achieving macro-hardness values above 900 HV30.

In the range below 5% V_γ , both macro-hardness and micro-hardness values exhibited considerable scatter. Beyond this threshold, however, the variation in hardness decreased progressively with increasing V_γ . This trend can be explained by the higher retained austenite content and the corresponding reduction in the martensite fraction. The large scatter in hardness values at less than 5% V_γ is primarily attributed to variations in the matrix structure among specimens. Specimens showing relatively low hardness values were typically associated with over-tempering, which softened the matrix. Conversely, specimens with higher hardness values possessed matrices composed predominantly of martensite, together with a uniform distribution of fine secondary carbides, both of which contributed significantly to increased hardness.

To further clarify this behavior, the relationship between matrix hardness (micro-hardness) and tempering temperature was examined for specimens with V_γ below 10%, as illustrated in Figure 4.27. Despite the limited variation in V_γ , micro-hardness decreased progressively with increasing tempering temperature from 525°C to 600°C, although some data scattering was observed. At a tempering temperature of 525°C, the micro-hardness exceeded 800 HV0.1, representing the peak hardness (H_{Tmax}) in the present study. This high hardness can be attributed to a matrix primarily composed of tempered martensite and finely dispersed secondary carbides. The subsequent reduction in hardness at higher tempering temperatures is due to the coarsening of secondary carbides and the transformation of retained austenite into softer phases, such as pearlite or ferrite. Notably, specimens tempered at 600°C exhibited a pronounced decline in micro-hardness, indicating that both carbide coarsening and phase transformation processes were largely completed at this stage.



(a) Macro-hardness



(b) Micro-hardness

Figure 4.26 Relationship between hardness and volume fraction of retained austenite (V_γ) in tempered condition; (a) Macro-hardness and (b) Micro-hardness.

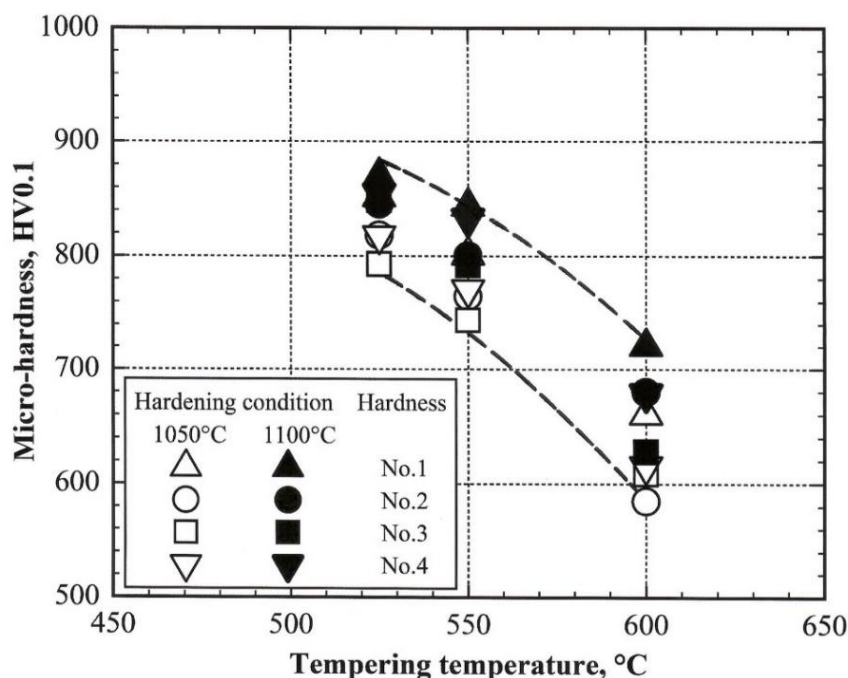
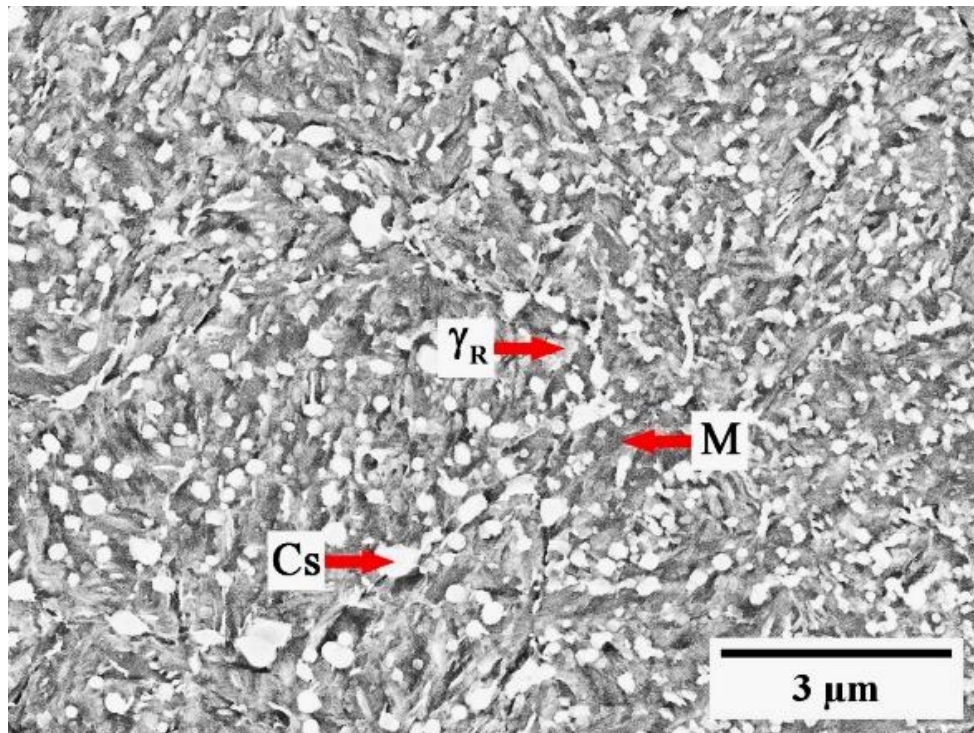


Figure 4.27 Relationship between micro-hardness and tempering temperature of specimens with less than 10%V_γ.

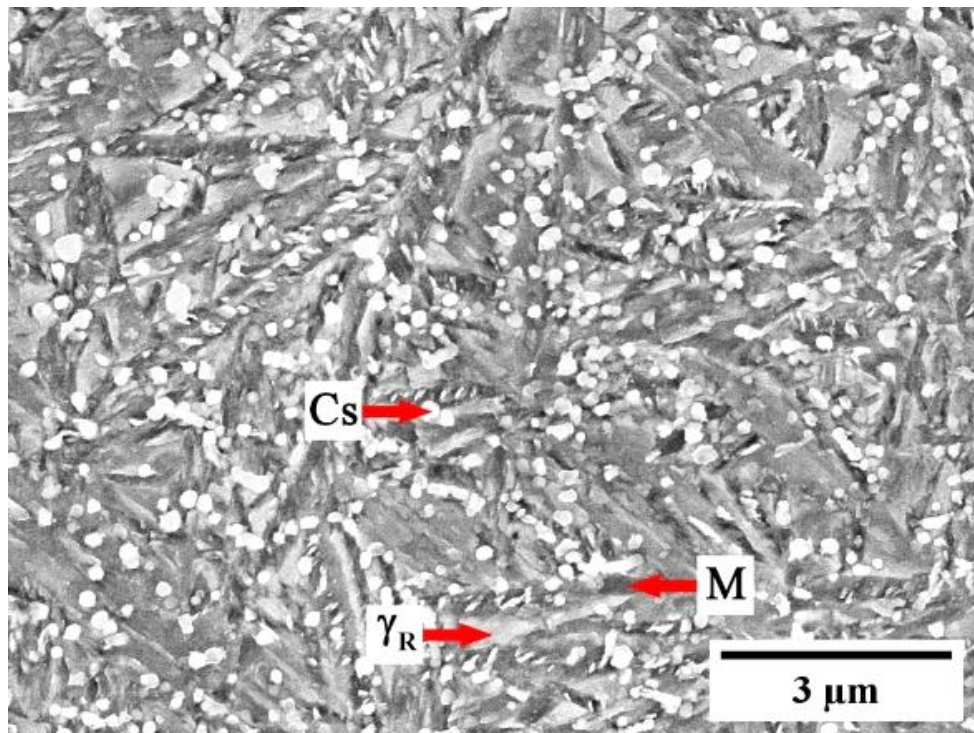
4.2.5 Effect of Nb on the maximum tempered hardness ($H_{T_{max}}$)

The matrix microstructures of specimens tempered to $H_{T_{max}}$ were examined by SEM, with representative micrographs shown in Figure 4.28 and Figure 4.29. The matrices of all specimens consisted of martensite (M), retained austenite (γ_R) and secondary carbides (C_s). While the amount of retained austenite could not be clearly distinguished from SEM observations, notable differences in the size and distribution of secondary carbides were observed as a function of Nb content. The amount of secondary carbides appeared to decrease with increasing Nb content up to 1.05%. In contrast, the specimen containing 2.58%Nb exhibited a larger size of secondary carbides compared with the other specimens. The relationship between Nb content and area fraction of secondary carbide precipitated in the $H_{T_{max}}$ specimens is shown in Table 4.5 and Figure 4.30. For the specimens austenitized at 1050°C, the area fraction of secondary carbides decreased with increasing Nb content, from 11.61% in the Nb-free specimen to 9.72% in the 0.70%Nb specimen. It reached a minimum value of 8.61% in the 1.05%Nb specimen and then increased to 11.16% in specimen No.4 with 2.58%Nb. A similar trend was observed for the specimens austenitized at 1100°C, where the area fraction of secondary carbide decreased from 14.38% in the Nb-free

specimen to 10.47% in the 0.70%Nb specimen, reached a minimum value of 9.94% in the 1.05%Nb specimen and subsequently increased to 13.45% in the specimen No.4. The initial reduction in secondary carbide fraction can be attributed to the formation of eutectic MC and MC_p carbides, which depletes C and V available in the matrix for secondary carbide precipitation during heat treatment. As the Nb content increased from 1.05% to 2.58% (with a corresponding increase in C from 2.00% to 2.40%), the area fraction of secondary carbides increased. This behavior can be rationalized by the higher total C content in the alloy, which leads to a higher C concentration dissolved into the matrix during hardening. Consequently, this promotes a greater precipitation of secondary carbides during tempering. Regarding the austenitizing temperature, specimens hardened at higher temperatures exhibited a greater area fraction of secondary carbides across all Nb contents. This is because at high austenitizing temperatures, alloying elements can easily dissolve into the matrix and result in a higher amount of secondary carbide precipitated during tempering.

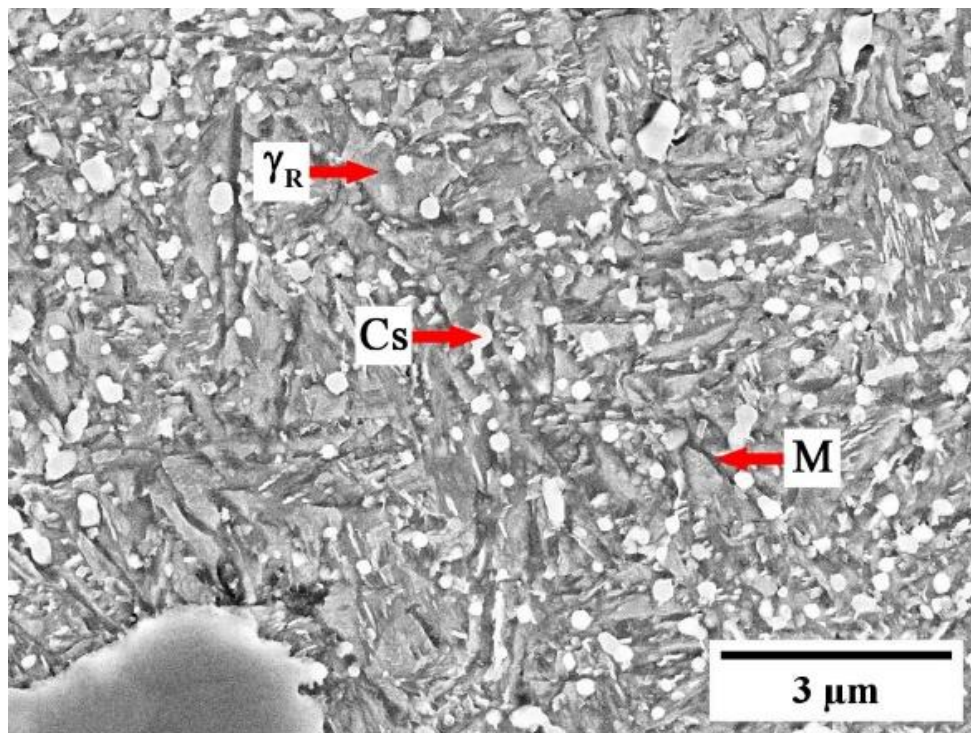


(a) No.1 (Nb-free)

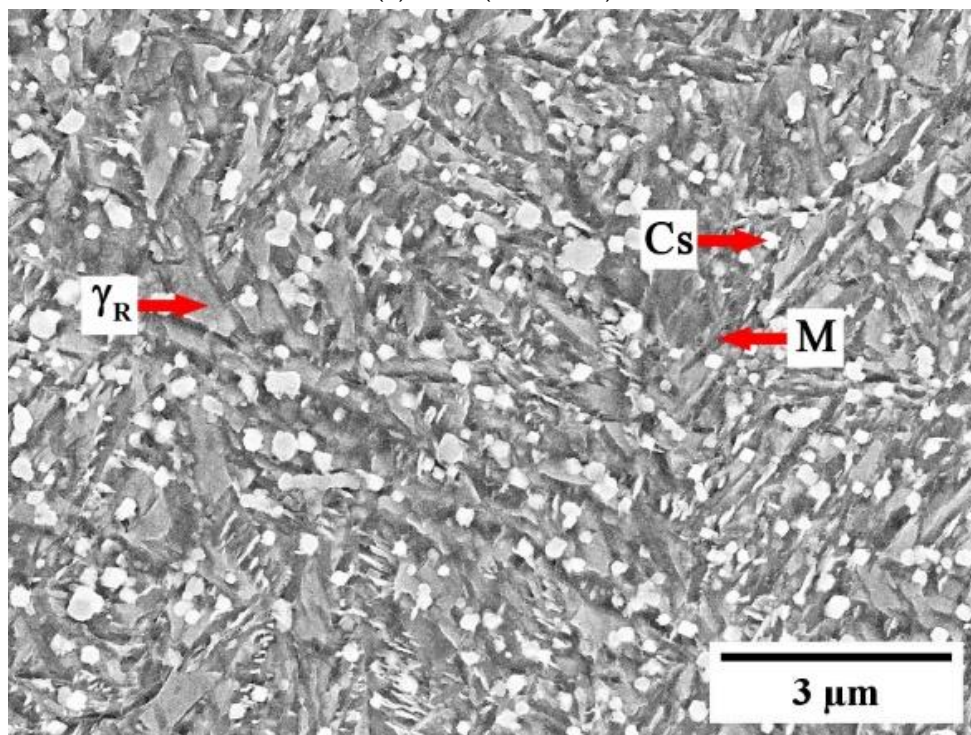


(b) No.2 (0.70%Nb)

Figure 4.28 Matrix microstructures of $H_{T_{max}}$ specimens with different Nb content; (a) No.1 (Nb-free), (b) No.2 (0.70%Nb).

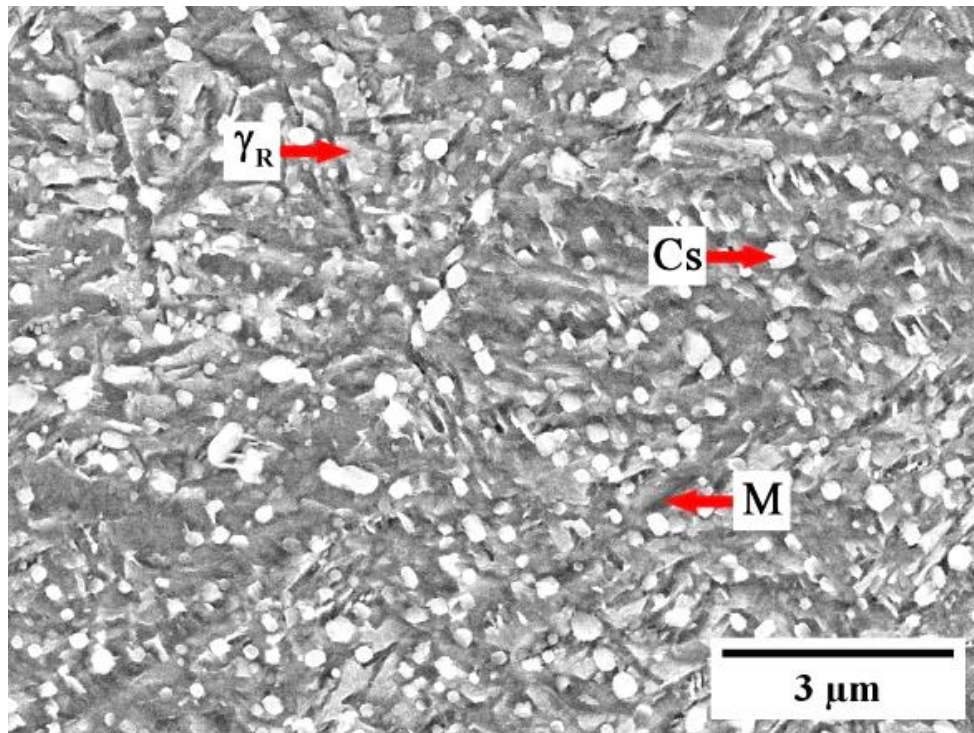


(c) No.3 (1.05%Nb)

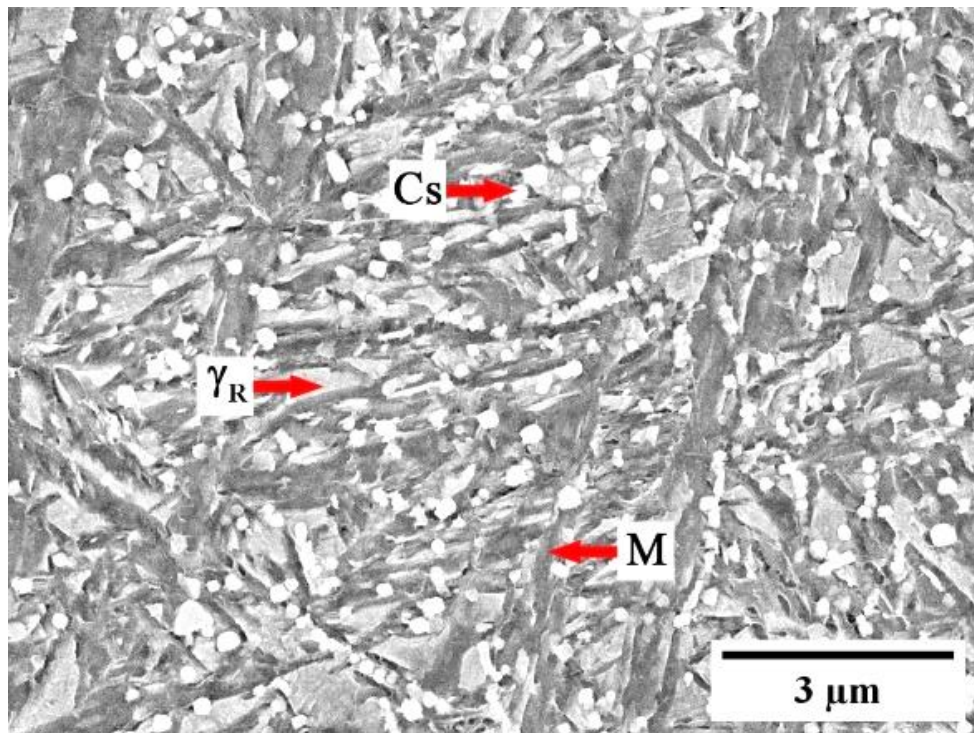


(d) No.4 (2.58%Nb)

Figure 4.28 Matrix microstructures of $H_{T_{max}}$ specimens with different Nb content; (c) No.3 (1.05%Nb) and (d) No.4 (2.58%Nb). Austenitizing temperature: 1050°C.

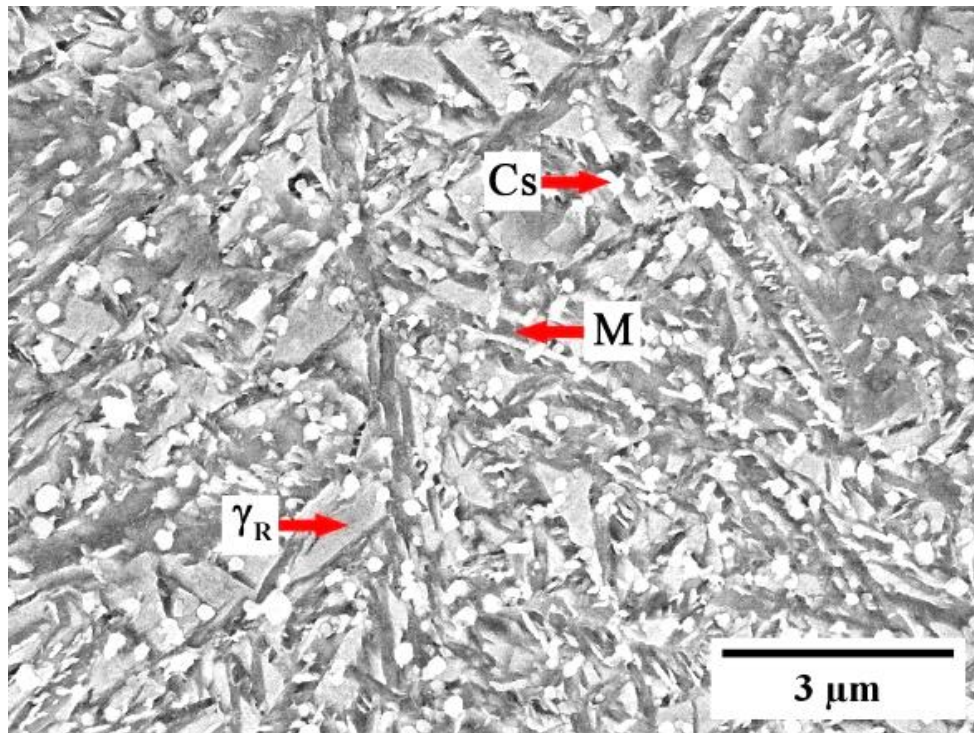


(a) No.1 (Nb-free)

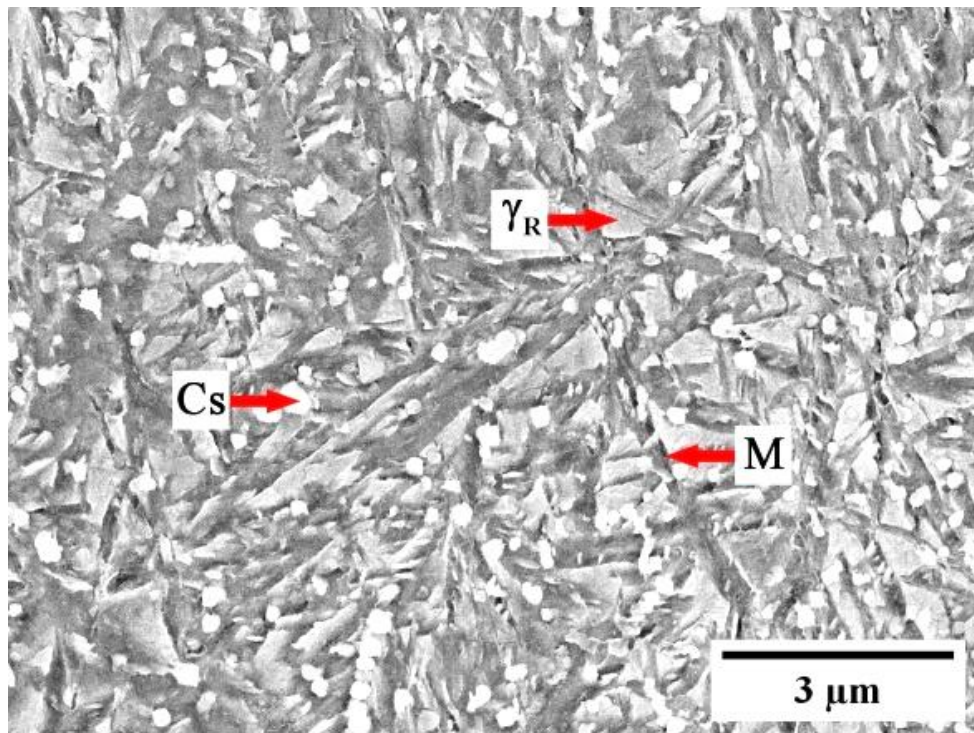


(b) No.2 (0.70%Nb)

Figure 4.29 Matrix microstructures of $H_{T_{max}}$ specimens with different Nb content; (a) No.1 (Nb-free), (b) No.2 (0.70%Nb).



(c) No.3 (1.05%Nb)

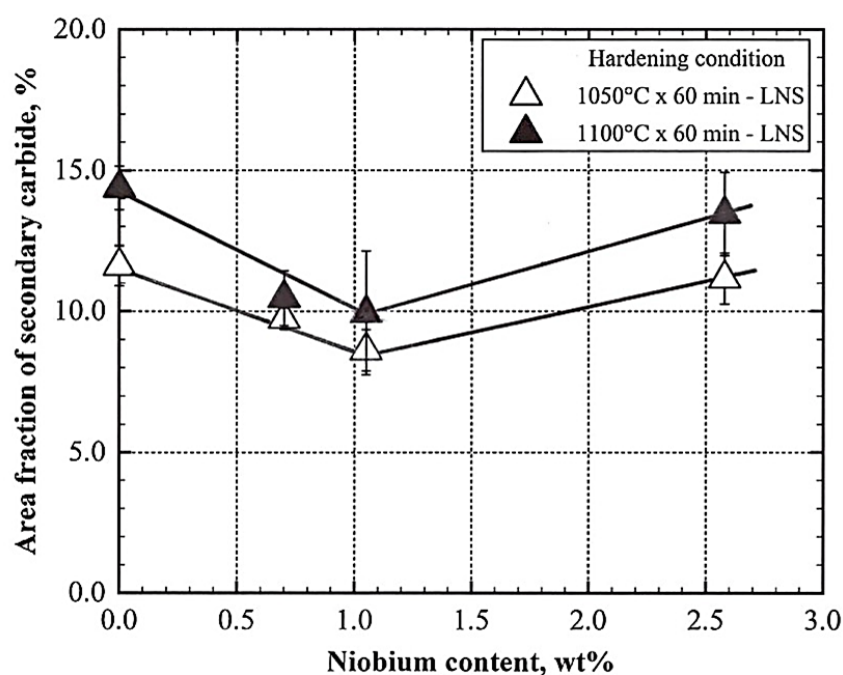


(d) No.4 (2.58%Nb)

Figure 4.29 Matrix microstructures of $H_{T_{max}}$ specimens with different Nb content; (c) No.3 (1.05%Nb) and (d) No.4 (2.58%Nb). Austenitizing temperature: 1100°C.

Table 4.5 Area fraction of secondary carbide precipitated in the $H_{T_{max}}$ specimen.

Specimen	Area fraction of secondary carbide, %	
	1050°C austenitizing	1100°C austenitizing
No.1 (Nb-free)	11.61	14.38
No.2 (0.70%Nb)	9.72	10.47
No.3 (1.05%Nb)	8.61	9.94
No.4 (2.58%Nb)	11.16	13.45

Figure 4.30 Relationship between Nb content and area fraction of secondary carbide precipitated in the $H_{T_{max}}$ specimens.

The relationship between $H_{T_{max}}$, the V_{γ} at $H_{T_{max}}$, and Nb content is presented in Table 4.6 and Figure 4.31. For the specimens austenitized at 1050°C, both macro-hardness and micro-hardness decreased with increasing Nb content, from 894.6 HV30 and 851.5 HV0.1 in the Nb-free specimen to 873.8 HV30 and 818.1 HV0.1 in the 0.70%Nb specimen. The hardness then reached minimum values of 857.2 HV30 and 792.9 HV0.1 in the 1.05%Nb specimen. After reaching these minimum values, the hardness increased to 907.1 HV30 and 817.5 HV0.1 in specimen No.4 with 2.58%Nb. A similar trend was observed for the specimens austenitized at 1100°C. The macro-

hardness and micro-hardness decreased from 922.1 HV30 and 869.6 HV0.1 in the Nb-free specimen to 911.1 HV30 and 848.8 HV0.1 in the 0.70%Nb specimen and then reached minimum values of 896.6 HV30 and 845.1 HV0.1 in the 1.05%Nb specimen. Subsequently, the hardness increased to 927.0 HV30 and 854.7 HV0.1 in specimen No.4.

From the above results, all specimens exhibited H_{Tmax} values exceeding 850 HV30. The H_{Tmax} decreased with Nb addition up to 1.05% and subsequently increased with higher Nb. This trend was observed consistently in specimens hardened from different austenitizing temperatures. For a given Nb content, the H_{Tmax} values obtained after austenitizing at 1100°C were higher than those at 1050°C. The maximum H_{Tmax} was achieved at 2.58%Nb, reaching 907.1 HV30 for 1050°C and 927.0 HV30 for 1100°C austenitizing conditions. As shown in Figure 4.32, the relationship between micro-hardness and Nb content in H_{Tmax} specimens followed a similar trend to the macro-hardness presented in Figure 4.31. However, the highest H_{Tmax} values differed, the maximum H_{Tmax} was obtained for the Nb-free specimens, reaching 851.5 HV0.1 and 869.6 HV0.1 at austenitizing temperature of 1050°C and 1100°C, respectively.

The initial decrease in H_{Tmax} up to 1.05%Nb can be rationalized by the fact that although the V_γ in the as-hardened state increased slightly with increasing Nb content, the extent of secondary hardening, arising from both secondary carbide precipitation and the transformation of retained austenite to martensite, was insufficient to counteract the softening effect. In addition, the crystallization of MC_p carbides consumed significant amounts of C and carbide-forming elements, thereby restricting the precipitation of fine secondary carbides within the martensitic matrix, which is a key mechanism of secondary hardening. In other words, the MC_p carbides, despite their intrinsic hardness, effectively reduce overall hardness by suppressing secondary carbide precipitation in the matrix. In the case of specimen No.4 with 2.58%Nb, H_{Tmax} increased gradually. This improvement can be attributed to enhanced secondary hardening, as the additional C in cast iron provides a sufficient supply for the formation of both MC_p and eutectic carbide, while still retaining enough C dissolved in the matrix to promote secondary hardening.

Regarding the volume fraction of retained austenite (V_γ), V_γ increased with increasing Nb content at both austenitizing temperatures. For the specimens austenitized at 1050°C, V_γ increased from 1.6% in the Nb-free specimen to 2.5% in the

0.70%Nb specimen. It then slightly increased to 3.3% in the 1.05%Nb specimen and reached a maximum value of 5.8% in the specimen No.4. For the specimens austenitized at 1100°C, V_γ increased from 2.8% in the Nb-free specimen to 8.3% in the 0.70%Nb specimen, followed by a slight increase to 8.5% in the 1.05%Nb specimen and reached a maximum value of 9.4% in specimen No.4 with 2.58%Nb.

As a result, V_γ values at $H_{T_{max}}$ were less than 5% in the Nb-free specimen and remained below 10% in the Nb-containing specimens. This indicates that even at maximum hardness achieved through tempering, a certain amount of austenite was retained in the microstructure. The V_γ at $H_{T_{max}}$ was higher in specimens hardened from the higher austenitizing temperature of 1100°C, owing to the greater amount of retained austenite generated in the as-hardened state. Consequently, under identical tempering conditions, a larger fraction of austenite persisted in the matrix when hardening was performed at the higher austenitizing temperature.

As shown in Figure 4.32, the maximum micro-hardness of $H_{T_{max}}$ specimens was obtained in Nb-free specimens at both austenitizing temperatures, this can be attributed to their higher area fraction of secondary carbides as shown in Figure 4.30. Macro-hardness gives higher maximum $H_{T_{max}}$ values than micro-hardness because macro-hardness represents the combined hardness of the matrix and the carbides crystallized during solidification, whereas micro-hardness measures only the matrix. Although the Nb-free specimen exhibited the highest micro-hardness, its macro-hardness was lower than that of the 2.58%Nb specimen. This is attributed to the presence of MC_p and a higher total area fraction of carbides in the 2.58%Nb alloy compared to the Nb-free specimen.

Table 4.6 Macro-hardness, micro-hardness and volume fraction of retained austenite (V_γ) in H_{Tmax} specimens.

Specimen	1050°C austenitizing			1100°C austenitizing		
	Macro-hardness , HV30	Micro-hardness , HV0.1	V_γ , %	Macro-hardness , HV30	Micro-hardness , HV0.1	V_γ , %
No.1 (Nb-free)	894.6	851.5	1.6	922.1	869.6	2.8
No.2 (0.70%Nb)	873.8	818.1	2.5	911.1	848.8	8.3
No.3 (1.05%Nb)	857.2	792.9	3.3	896.6	845.1	8.5
No.4 (2.58%Nb)	907.1	817.5	5.8	927.0	854.7	9.4

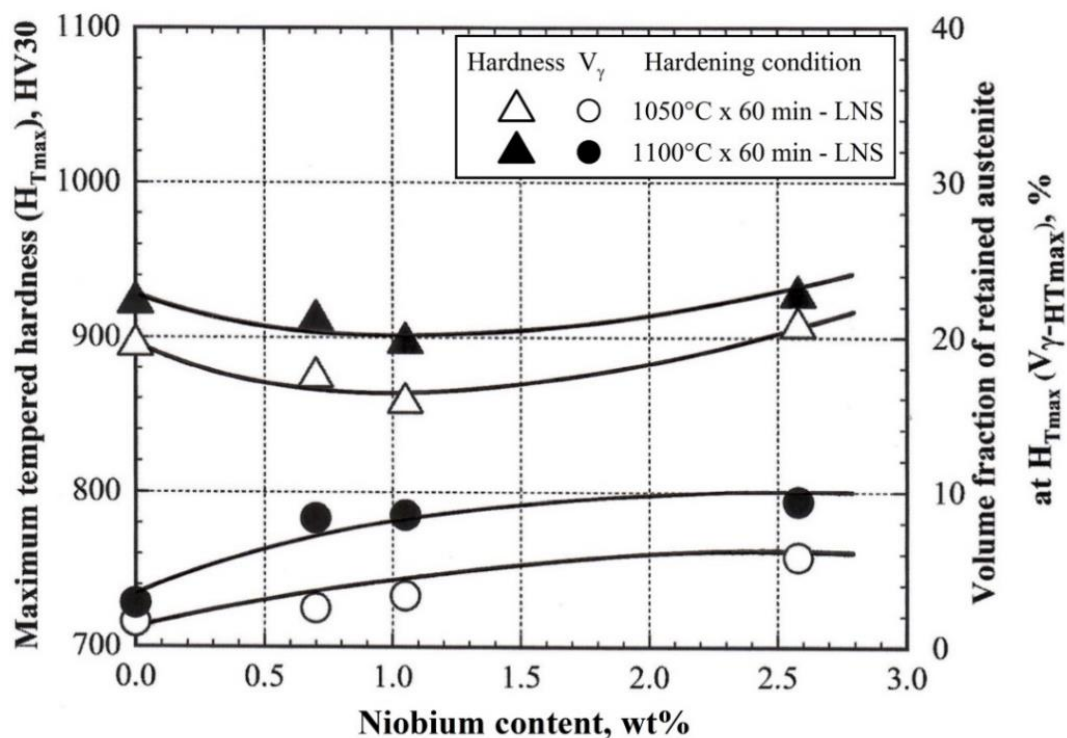


Figure 4.31 Relationship between maximum tempered hardness (H_{Tmax}) volume fraction of retained austenite (V_γ) at H_{Tmax} and Nb content.

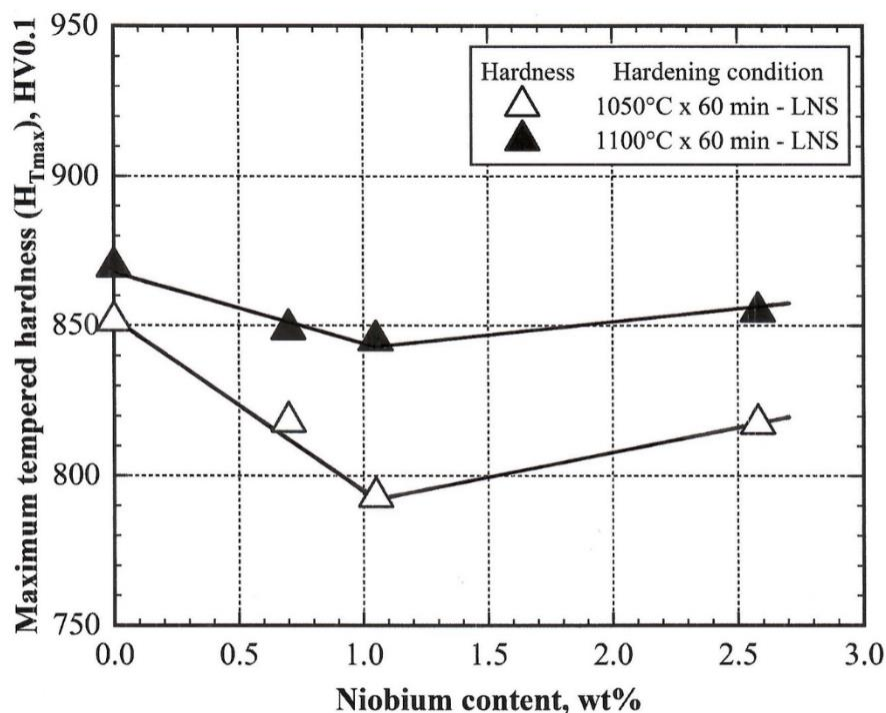


Figure 4.32 Relationship between micro-hardness and Nb content in H_{Tmax} specimens.

From the above results, it is evident that Nb exerts a pronounced influence on H_{Tmax} . To further clarify this effect, the V_γ in the as-hardened condition must be carefully considered to achieve the desired target hardness after tempering. As shown in Figure 4.33, the H_{Tmax} is directly correlated with V_γ in the as-hardened state. With increasing V_γ from 10% to 28%, H_{Tmax} rises proportionally, indicating that retained austenite contributes indirectly to secondary hardening. This effect arises from two mechanisms: (i) the precipitation of secondary carbides and (ii) the transformation of retained austenite into martensite during tempering. Notably, a minimum of $\sim 10\%V_\gamma$ in the as-hardened condition is required to achieve an H_{Tmax} above 900 HV30. However, excessive retained austenite becomes detrimental, as a considerable fraction may remain untransformed after tempering. In such cases, the softening effect of retained austenite on the matrix outweighs the strengthening effect provided by secondary carbide precipitation and martensitic transformation, thereby lowering the overall hardness.

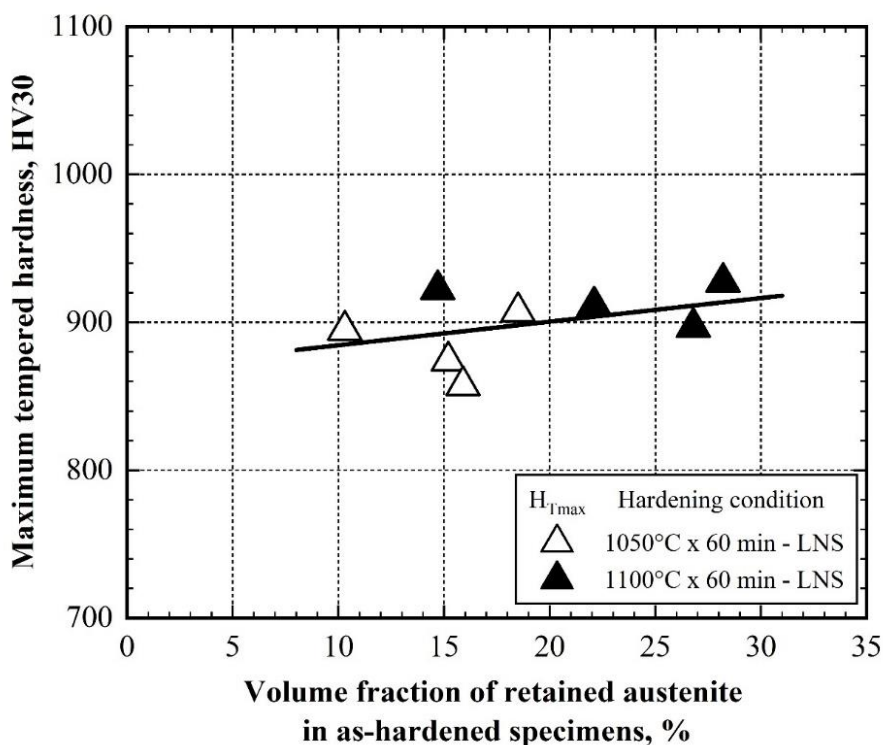


Figure 4.33 Relationship between maximum tempered hardness (H_{Tmax}) and volume fraction of retained austenite (V_γ) in as-hardened condition.

4.2.6 Degree of secondary hardening

In the tempered hardness curves of each specimen (Figure 4.21 to 4.24), secondary hardening is clearly observed. This phenomenon is attributed to three mechanisms: the precipitation of secondary carbides from the retained austenite, the resulting transformation of the destabilized austenite into martensite, and the carbide reaction in the martensitic matrix. The influence of Nb content on the degree of secondary hardening on the matrix (ΔH_{sm}), defined as the difference in micro-hardness between H_{Tmax} and that at the tempering temperature at which secondary hardening begins, is presented in Figure 4.34. At an austenitizing temperature of 1050°C, the ΔH_{sm} value increased from 84.8 HV0.1 in the Nb-free specimen to 92.6 HV0.1 in the 0.70%Nb specimen. Beyond this Nb content, ΔH_{sm} decreased sharply to 57.4 HV0.1 in the 1.05%Nb specimen, followed by a gradual increase to 80.8 HV0.1 in the 2.58%Nb specimen. For the specimens austenitized at 1100°C, ΔH_{sm} increased from 106.5 HV0.1 in the Nb-free specimen to 133.2 HV0.1 in the 0.70%Nb specimen. It then decreased to 118.4 HV0.1 in the 1.05%Nb specimen and further decreased slightly to 116.9 HV0.1 in the 2.58%Nb specimen. The maximum ΔH_{sm} is obtained at 0.70%Nb at both

austenitizing temperatures. At the same Nb, specimens hardened from 1100°C exhibit a greater ΔH_{sm} than those hardened from 1050°C, indicating that higher austenitizing temperatures enhance secondary hardening. The degree of secondary hardening is compared with the volume fraction of retained austenite in the as-hardened condition in Figure 4.35. A larger V_γ in the as-hardened condition tends to raise the potential for secondary hardening, although the higher ΔH_{sm} of the specimens austenitized at 1100°C also reflects the greater amount of carbon and alloying elements dissolved in the matrix at that temperature. This potential is realized only where sufficient secondary carbide is available to precipitate, so that ΔH_{sm} is not determined by V_γ alone.

The increase in ΔH_{sm} up to 0.70%Nb originates from lower matrix hardness at the tempering temperature at which secondary hardening begins rather than from an elevated peak hardness, since $H_{T_{max}}$ itself was reduced by the Nb addition. This indicates that ΔH_{sm} expresses the amount of hardness increase developed during tempering and not the hardness level attained, so that a specimen with a large ΔH_{sm} does not necessarily reach a high tempered hardness. The depressed onset hardness and the enlarged hardness increment share a common origin in the greater fraction of retained austenite in the as-hardened state, which softens the matrix before tempering and supplies the austenite that transforms into martensite during tempering, thereby raising the hardness and enlarging ΔH_{sm} .

The decrease in ΔH_{sm} at 1.05%Nb occurred despite a further increase in the as-hardened V_γ and is therefore attributable to the suppression of secondary carbide precipitation, as the crystallization of MC_p carbides consumed Nb, V and C. The area fraction of secondary carbides had in fact already decreased from Nb-free to 0.70%Nb, yet ΔH_{sm} increased because the accompanying gain in retained austenite outweighed the loss. At 1.05%Nb, where the area fraction fell to the minimum, the larger retained austenite fraction could no longer compensate and ΔH_{sm} decreased.

The recovery of ΔH_{sm} at 2.58%Nb, observed at 1050°C, is attributable to the higher total carbon content of this alloy, which left sufficient carbon dissolved in the matrix after hardening to restore secondary carbide precipitation. At 1100°C, however, ΔH_{sm} decreased only slightly even though secondary carbide precipitation recovered to an even greater extent, because a larger part of the retained austenite

was too stable to transform within the tempering range, 9.4% still remaining at $H_{T_{max}}$ (Table 4.6).

Table 4.7 Degree of secondary hardening (ΔH_{sm}) of tempered specimens.

Specimen	Degree of secondary hardening, HV0.1	
	1050°C austenitizing	1100°C austenitizing
No.1 (Nb-free)	84.8	106.5
No.2 (0.70%Nb)	92.6	133.2
No.3 (1.05%Nb)	57.4	118.4
No.4 (2.58%Nb)	80.8	116.9

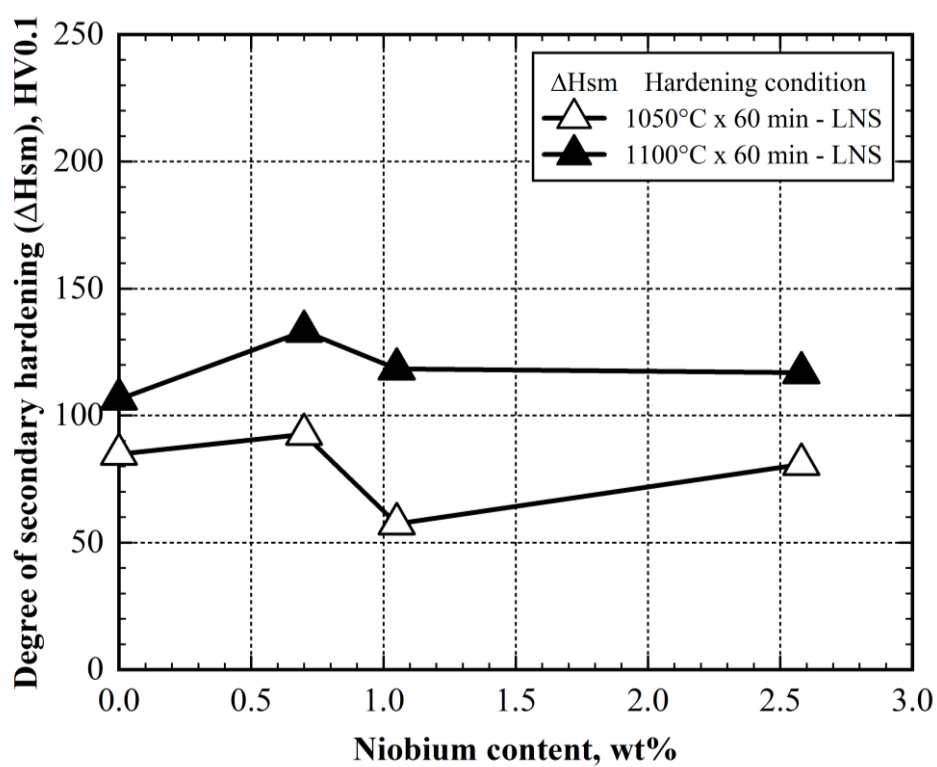


Figure 4.34 Effect of Nb content on degree of secondary hardening of matrix (ΔH_{sm}).

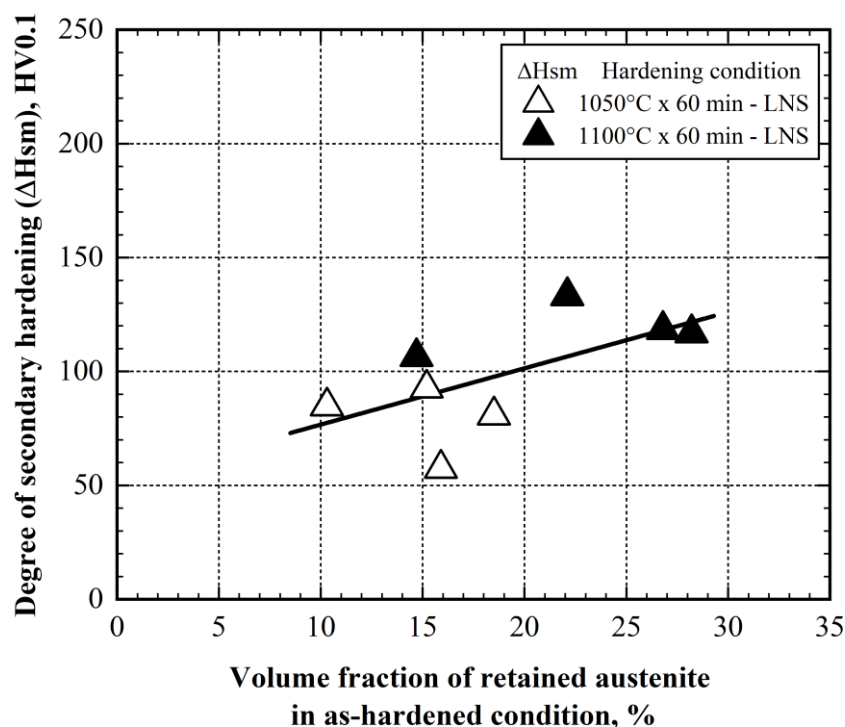


Figure 4.35 Relationship between degree of secondary hardening (ΔH_{sm}) and volume fraction of retained austenite (V_γ) in as-hardened condition.

4.3 Effect of Nb addition and heat treatment conditions on abrasive wear resistance

Under actual service conditions, alloyed white cast irons are typically used in their heat-treated condition. Therefore, the wear resistance of multi-alloyed white cast iron must be evaluated under different heat treatment and wear conditions. As discussed in the earlier section, it was found that the hardness and V_γ of specimens varied significantly depending on the heat treatment conditions. Since the addition of Nb in the cast iron influences the solidification sequence, as well as the matrix transformation and the resulting phases such as retained austenite, martensite and secondary carbides, Nb consequently governs the wear resistance of the cast iron.

In this section, the effect of Nb on the wear behavior under various heat treatment conditions is described. The specimens were hardened at 1050°C and 1100°C (As-H). Subsequently, the hardened specimens were tempered at three different temperatures relative to the maximum tempered hardness point as shown in Figure 3.4, at maximum tempered hardness (H_{Tmax}), below H_{Tmax} ($L-H_{Tmax}$) and above H_{Tmax} ($H-$

$H_{T_{max}}$). The actual tempering temperature at three points is shown in Table 4.8. The hardness and V_γ for each specimen are shown in Table 4.9 and 4.10 for the 1050°C and 1100°C austenitizing temperature respectively. Although the hardness of the L- $H_{T_{max}}$ and H- $H_{T_{max}}$ specimens is lower than that of $H_{T_{max}}$, the V_γ values exhibit significant variation across these three tempering conditions. Various heat-treated specimens were subjected to Suga (two-body type) and rubber wheel (three-body type) abrasive wear tests to evaluate their wear behavior and abrasive wear resistance.

When compared under nominally identical tempering conditions, the wear test specimens consistently showed higher V_γ and lower macro-hardness than the corresponding heat treatment specimens. Micro-hardness showed the same tendency but with considerably more scatter and did not consistently follow the same direction. One plausible explanation is the difference in casting size between the two specimen sets, as the wear test specimens were cast in a Y-block mold considerably larger than the cylindrical sand mold used for the heat treatment study specimens. The larger casting would be expected to solidify at a lower rate, which could result in more pronounced microsegregation and a lower eutectic carbide fraction. If so, segregation-enriched interdendritic region would be expected to locally depress M_s and raise V_γ , while a lower carbide fraction, largely unaffected by subsequent tempering, may partly account for the persistently lower macro-hardness.

One approach to comparing wear resistance to evaluate is the total wear loss of the test specimens at the completion of the abrasive wear test. However, total wear loss represents only the accumulated material loss and may be affected by the total wear distance and the initial surface condition of the specimen. In contrast, when a linear relationship between weight loss or wear loss (W_l) and wear distance (W_d) is obtained, the wear rate (R_w , mg/m), calculated from the slope of the $W_l - W_d$ relationship, provides a clearer assessment of the material loss per unit wear distance and allows the wear behavior to be more reasonably compared under the same testing conditions. Because R_w is derived from the slope of the linear, steady-state portion of the $W_l - W_d$ data, it also excludes the transient wear behavior associated with the running-in period at the early stage of the test, which is often higher and not representative of the material's intrinsic wear resistance (Pellizzari et al., 2005). Therefore, in this study, R_w was used as a key parameter to evaluate the abrasive wear resistance of the specimens. A lower R_w value indicates lower mass loss during abrasion

and consequently, higher wear resistance, which may correspond to better service performance in abrasive environments (Inthidech et al., 2021a).

Table 4.8 Selected tempering temperatures for L-H_{Tmax}, H_{Tmax} and H-H_{Tmax} in each specimen.

Specimen	Tempering temperature (°C)					
	1050°C austenitizing			1100°C austenitizing		
	L-H _{Tmax}	H _{Tmax}	H-H _{Tmax}	L-H _{Tmax}	H _{Tmax}	H-H _{Tmax}
No.1 (Nb-free)	475	525	554	475	525	567
No.2 (0.70%Nb)	400	525	575	450	525	585
No.3 (1.05%Nb)	450	525	567	490	525	575
No.4 (2.58%Nb)	400	525	573	450	525	585

Table 4.9 Macro-hardness, Micro-hardness and volume fraction of retained austenite (V_γ) of heat-treated specimens for abrasive wear test. Austenitizing temperature: 1050°C.

Specimens	Heat treatment condition	Hardness (HV30)	Hardness (HV0.1)	V_γ (%)
No.1 (Nb-free)	As-hardened	925.3	862.0	15.6
	L- $H_{T_{\max}}$ (475°C)	847.6	792.0	10.3
	$H_{T_{\max}}$ (525°C)	883.0	831.0	2.2
	H- $H_{T_{\max}}$ (554°C)	840.8	774.0	1.5
No.2 (0.7%Nb)	As-hardened	895.1	815.3	18.2
	L- $H_{T_{\max}}$ (400°C)	781.0	714.5	17.5
	$H_{T_{\max}}$ (525°C)	875.4	791.1	3.2
	H- $H_{T_{\max}}$ (575°C)	736.7	696.6	1.8
No.3 (1.05%Nb)	As-hardened	867.8	796.8	18.8
	L- $H_{T_{\max}}$ (450°C)	791.6	720.3	18.1
	$H_{T_{\max}}$ (525°C)	863.7	787.1	3.7
	H- $H_{T_{\max}}$ (567°C)	782.1	708.6	2.2
No.4 (2.58%Nb)	As-hardened	917.8	844.2	20.0
	L- $H_{T_{\max}}$ (400°C)	811.1	748.3	18.1
	$H_{T_{\max}}$ (525°C)	901.6	834.4	6.1
	H- $H_{T_{\max}}$ (573°C)	788.6	709.4	6.0

Table 4.10 Macro-hardness, Micro-hardness and volume fraction of retained austenite (V_γ) of heat-treated specimens for abrasive wear test. Austenitizing temperature: 1100°C.

Specimens	Heat treatment condition	Hardness (HV30)	Hardness (HV0.1)	V_γ (%)
No.1 (Nb-free)	As-hardened	909.2	851.0	25.1
	L- $H_{T_{\max}}$ (475°C)	840.8	801.0	15.2
	$H_{T_{\max}}$ (525°C)	906.4	862.0	4.0
	H- $H_{T_{\max}}$ (567°C)	847.6	811.0	1.2
No.2 (0.7%Nb)	As-hardened	815.6	779.4	26.2
	L- $H_{T_{\max}}$ (450°C)	759.2	719.5	20.9
	$H_{T_{\max}}$ (525°C)	893.0	830.3	8.6
	H- $H_{T_{\max}}$ (585°C)	801.3	756.1	3.3
No.3 (1.05%Nb)	As-hardened	786.9	741.0	27.7
	L- $H_{T_{\max}}$ (490°C)	795.2	756.8	19.4
	$H_{T_{\max}}$ (525°C)	869.8	808.4	8.8
	H- $H_{T_{\max}}$ (575°C)	808.0	744.5	4.2
No.4 (2.58%Nb)	As-hardened	864.4	779.6	29.1
	L- $H_{T_{\max}}$ (450°C)	778.1	715.9	26.9
	$H_{T_{\max}}$ (525°C)	932.1	852.8	11.8
	H- $H_{T_{\max}}$ (585°C)	819.8	753.5	5.3

4.3.1 Suga abrasive wear test (two-body type)

The relationship between W_l and W_d was examined for specimens No.1 (Nb-free) to No.4 (2.58%Nb) under two austenitizing conditions. Test results for specimens hardened from 1050°C are presented in Figure 4.36 to 4.39, while those hardened from 1100°C are presented in Figure 4.40 to 4.43 respectively. Each graph is accompanied by a regression equation describing the relationship, where the slope represents the R_w . Across all specimens and both austenitizing temperatures, W_l increased linearly with W_d , indicating that the wear behavior followed a consistent linear trend irrespective of heat treatment condition. The R_w values derived from the slopes of these regression equations are summarized in Table 4.11 for specimens austenitized at 1050°C and in Table 4.12 for those austenitized at 1100°C, allowing direct comparison of wear resistance between two heat treatment conditions.

4.3.1.1 Wear test results of specimens hardened from 1050°C

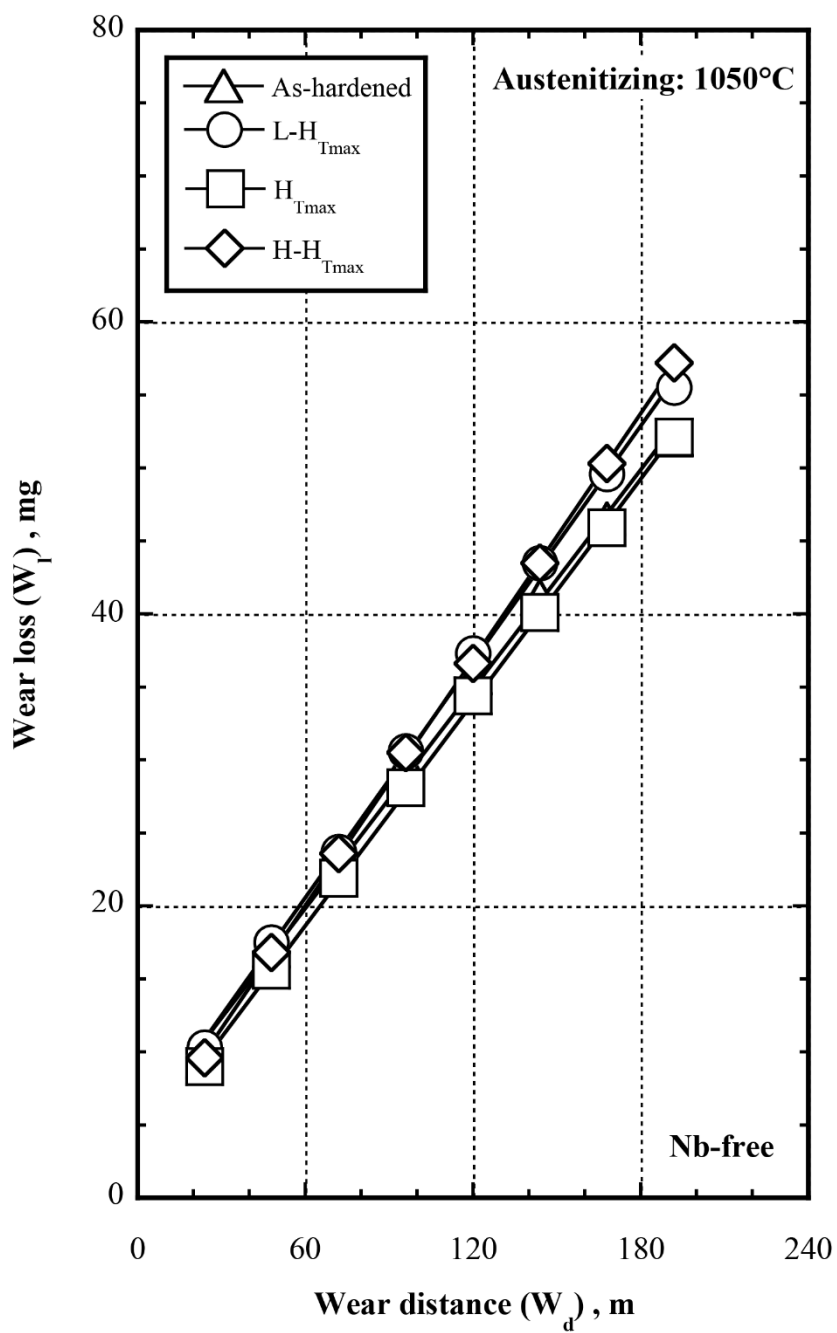
Figure 4.36 shows the relationship between W_l and W_d of specimens No.1 (Nb-free) under different heat treatment conditions. A linear relationship between W_l and W_d was observed, in which W_l increased with increasing W_d for both the As-hardened and tempered specimens. Considering the total wear loss, the As-hardened specimen exhibited the lowest total wear loss, whereas the H- H_{Tmax} specimen exhibited the highest total wear loss. When the trend lines of the W_l – W_d relationship were considered, the wear behavior could be divided into two groups with only a slight difference in slope: the As-hardened and H- H_{Tmax} specimens belonged to the lower wear loss group, while the L- H_{Tmax} and H-H- H_{Tmax} specimens belonged to the higher wear loss group. This indicates that, for the Nb-free specimen, heat treatment had only a slight effect on the wear behavior. In the case of R_w , the results were consistent with the total wear loss. The As-hardened specimen exhibited the lowest R_w of 0.2505 mg/m, while the H- H_{Tmax} specimen showed the highest R_w of 0.2810 mg/m. The R_w values of the L- H_{Tmax} and H- H_{Tmax} specimens were 0.2693 mg/m and 0.2550 mg/m, respectively.

For specimen No.2 with 0.70%Nb, shown in Figure 4.37, a linear relationship between W_l and W_d was also observed. The total wear loss followed the same order as specimen No.1, with the As-hardened specimen giving the lowest value and the H- H_{Tmax} giving the clearly highest total wear loss. Unlike specimen No.1, where the four trend lines separated into only two closely spaced groups, the trend lines of specimen No.2 separated into three distinct levels, with the As-hardened and H- H_{Tmax}

specimens showing the lowest and closely similar slope and the L-H_{Tmax} specimen showing a distinctly steeper slope and the H-H_{Tmax} specimen showing the steepest slope of all. This indicates that the heat treatment had a pronounced effect on the wear behavior of the 0.70%Nb specimen, especially after tempering above H_{Tmax}. The R_w values followed the same order as the total wear loss, increasing from 0.2810 mg/m in the As-hardened specimen to 0.2916 mg/m in the H_{Tmax} specimen, 0.3700 mg/m in the L-H_{Tmax} specimen and 0.4767 mg/m in the H-H_{Tmax} specimen, which exhibited the highest R_w among the four conditions.

Specimen No.3, containing 1.05%Nb, produced the wear results presented in Figure 4.38, again following a linear $W_l - W_d$ relationship. Among the four heat treatment conditions, the As-hardened specimen produced the smallest total wear loss and the H-H_{Tmax} specimen produced the largest. The $W_l - W_d$ trend lines grouped into two sets similar to specimen No.1, with the As-hardened and H_{Tmax} specimens on one side and L-H_{Tmax} and H-H_{Tmax} specimens on the other, but the gap between the two sets was noticeably wider than in specimen No.1 and narrower than the three-tier separation seen in specimen No.2. This places the sensitivity of 1.05%Nb specimen's wear behavior to heat treatment at an intermediate level between specimens No.1 and No.2. A comparable pattern emerged for R_w , which ranged from 0.2870 mg/m for the As-hardened specimen up to 0.3686 mg/m for the H-H_{Tmax} specimen, with the L-H_{Tmax} and H_{Tmax} specimen reaching 0.3470 mg/m and 0.3083 mg/m.

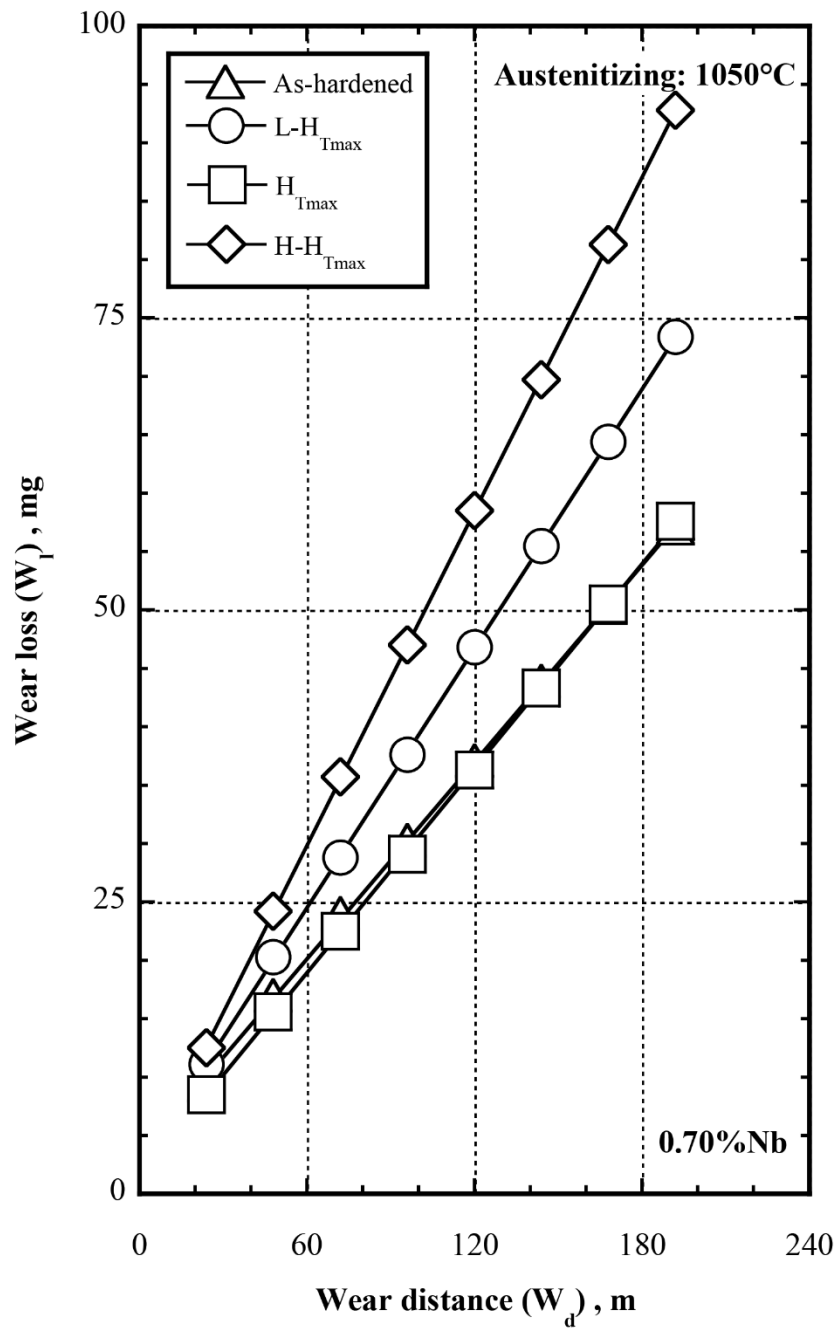
In Figure 4.39, the $W_l - W_d$ trend lines of specimen No.4 with 2.58%Nb formed a pattern distinct from the other three specimens, with the As-hardened specimen sitting clearly apart from the rest with the lowest slope and the lowest R_w of 0.2215 mg/m, the H_{Tmax} specimen occupying an intermediate position with a R_w of 0.2410 mg/m, and the L-H_{Tmax} and H-H_{Tmax} specimens running close together with the steepest slopes and the highest R_w values of 0.2936 mg/m and 0.3028 mg/m, respectively. The R_w of the H-H_{Tmax} specimen was slightly above that of the L-H_{Tmax} specimen. Consequently, the As-hardened specimen exhibited the smallest total wear loss, clearly separated from the other three conditions, while the H-H_{Tmax} specimen exhibited the largest.



As-hardened	△	$W_l = 0.2504 W_d + 4.81$ $R^2 = 0.9963$
L- $H_{T_{max}}$	○	$W_l = 0.2693 W_d + 4.41$ $R^2 = 0.9993$
H- T_{max}	□	$W_l = 0.2550 W_d + 3.35$ $R^2 = 0.9996$
H- $H_{T_{max}}$	◇	$W_l = 0.2810 W_d + 3.16$ $R^2 = 0.9998$

W_l : Wear loss, W_d : Wear distance

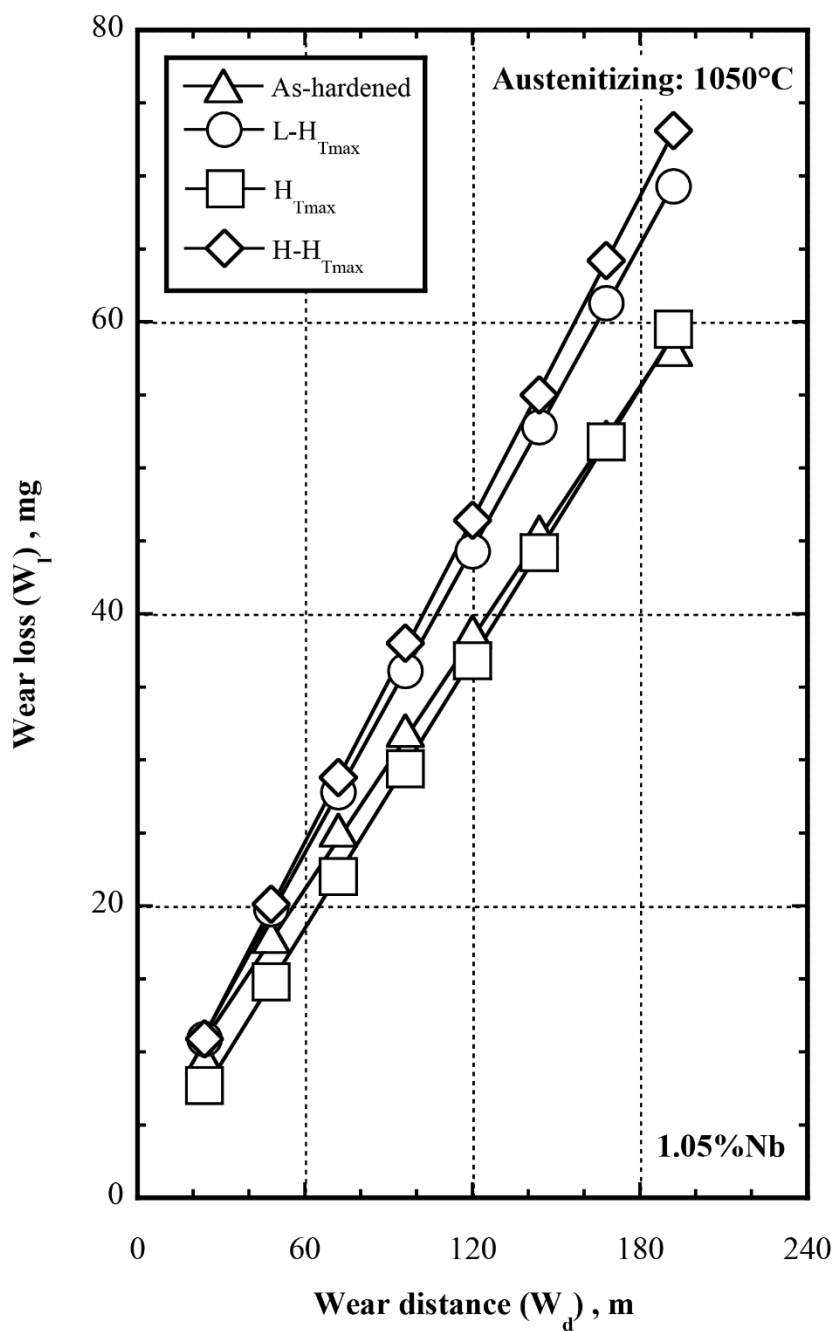
Figure 4.36 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.1 (Nb-free). Suga abrasive wear test. Load: 9.8 N (1 kgf).



As-hardened	△	$W_l = 0.2810 W_d + 3.26$	$R^2 = 0.9995$
L- $H_{T_{max}}$	○	$W_l = 0.3700 W_d + 2.28$	$R^2 = 0.9999$
H- T_{max}	□	$W_l = 0.2916 W_d + 1.44$	$R^2 = 0.9999$
H- $H_{T_{max}}$	◇	$W_l = 0.4767 W_d + 1.23$	$R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

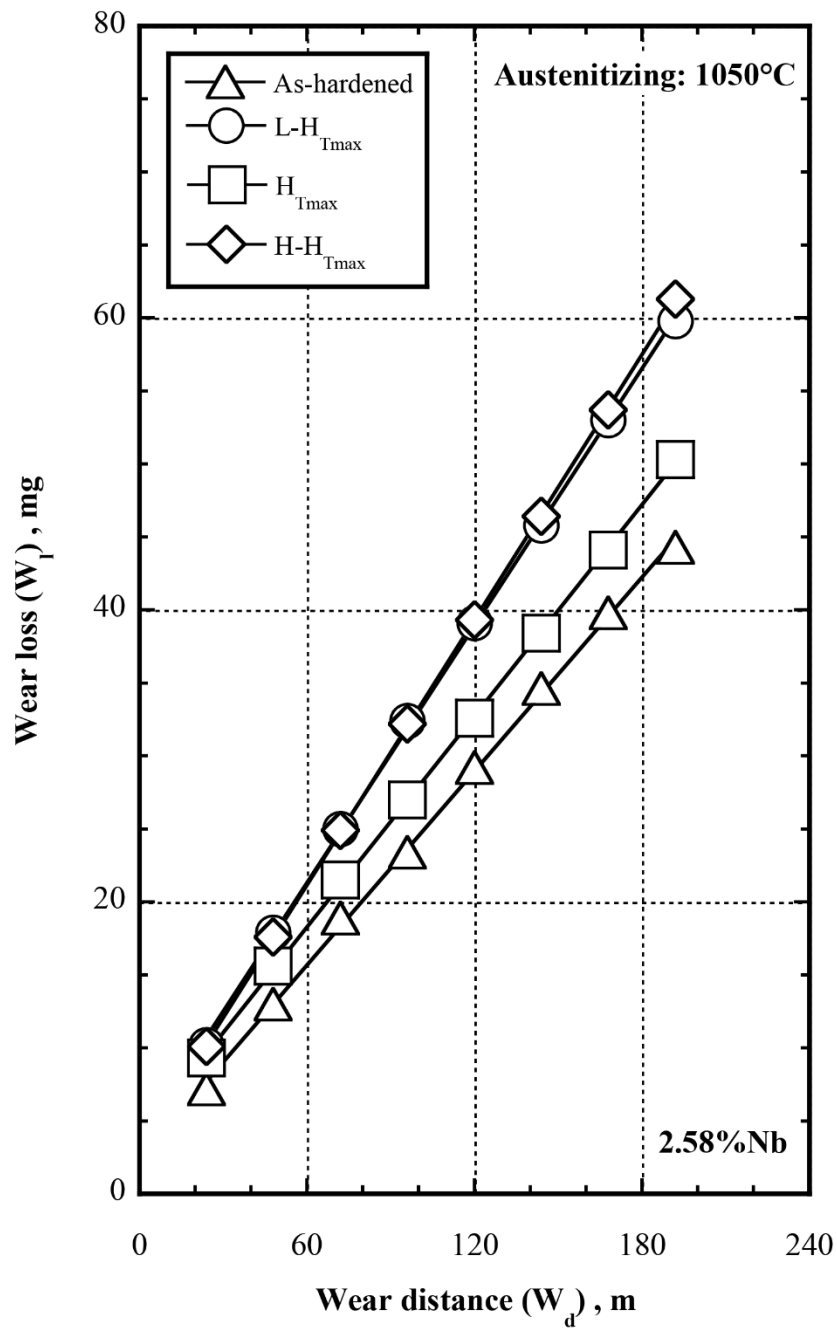
Figure 4.37 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.2 (0.70%Nb). Suga abrasive wear test. Load: 9.8 N (1 kgf).



As-hardened	△	$W_l = 0.2870 W_d + 3.92$	$R^2 = 0.9983$
L- $H_{T_{max}}$	○	$W_l = 0.3470 W_d + 2.81$	$R^2 = 0.9999$
H- T_{max}	□	$W_l = 0.3083 W_d - 0.03$	$R^2 = 0.9999$
H- $H_{T_{max}}$	◇	$W_l = 0.3685 W_d + 2.26$	$R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

Figure 4.38 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.3 (1.05%Nb). Suga abrasive wear test. Load: 9.8 N (1 kgf).



As-hardened	△	$W_l = 0.2215 W_d + 2.38$	$R^2 = 0.9990$
L- $H_{T_{max}}$	○	$W_l = 0.2936 W_d + 3.70$	$R^2 = 0.9996$
$H_{T_{max}}$	□	$W_l = 0.2410 W_d + 3.83$	$R^2 = 0.9997$
H- $H_{T_{max}}$	◇	$W_l = 0.3028 W_d + 2.98$	$R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

Figure 4.39 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.4 (2.58%Nb). Suga abrasive wear test. Load: 9.8 N (1 kgf).

Table 4.11 Wear rate (R_w) of heat-treated specimens. Suga abrasive wear test under a load of 9.8 N (1 kgf). Austenitizing: 1050°C.

Specimen	Wear rate (mg/m)			
	As-H	L-H _{Tmax}	H _{Tmax}	H-H _{Tmax}
No.1 (Nb-free)	0.2504	0.2693	0.2550	0.2810
No.2 (0.70%Nb)	0.2810	0.3700	0.2916	0.4767
No.3 (1.05%Nb)	0.2870	0.3470	0.3083	0.3685
No.4 (2.58%Nb)	0.2215	0.2936	0.2410	0.3028

4.3.1.2 Wear test results of specimens hardened from 1100°C

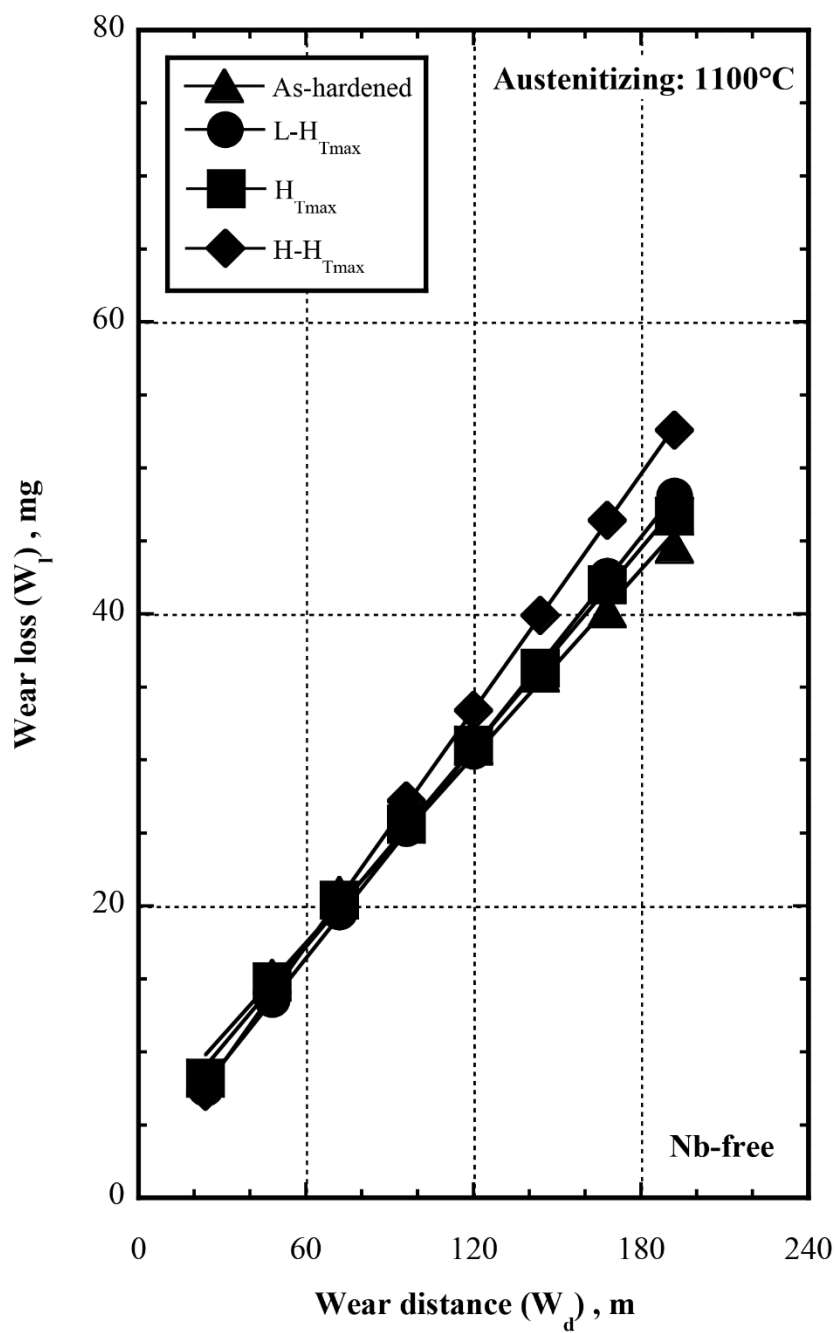
The relationship between wear loss (W_l) and wear distance (W_d) for specimens hardened from 1100°C is illustrated in Figure 4.40 to 4.43. The wear results of the Nb-free specimen shown in Figure 4.40 produced a linear relationship between W_l and W_d . The lowest total wear loss was obtained in the As-hardened specimen, while the highest was obtained in the H-H_{Tmax} specimen. The grouping of the four conditions changed from the two-group split seen at 1050°C to a different pattern in 1100°C, with the As-hardened, H_{Tmax} and L-H_{Tmax} specimens now forming a single closely spaced group, while only the H-H_{Tmax} specimen separated out with a higher wear loss. This indicates that, at the higher austenitizing temperature, under-tempering no longer raised the R_w as it did at 1050°C, and only over-tempering beyond H_{Tmax} continued to noticeably impair the wear resistance of the Nb-free specimen. The R_w values also decreased at every heat treatment condition compared with the 1050°C specimens, reaching 0.2130 mg/m for the As-hardened specimen, 0.2275 mg/m for the H_{Tmax} specimen, 0.2404 mg/m for the L-H_{Tmax} specimen and 0.2694 mg/m for the H-H_{Tmax} specimen.

In the case of specimen No.2 with 0.70%Nb, the wear test results are shown in Figure 4.41. The wear behavior was clearly separated depending on the heat treatment condition. The H_{Tmax} condition showed the lowest total wear loss, corresponding to an R_w value of 0.2568 mg/m, followed by the As-hardened condition at 0.2761 mg/m. The L-H_{Tmax} condition exhibited an R_w value of 0.2962 mg/m, while the highest total wear loss was observed in the H-H_{Tmax} condition with an R_w value of 0.3660 mg/m. Compared with the specimen hardened from 1050°C, the separation between the As-hardened and H_{Tmax} conditions became more evident after hardening

from 1100°C. In the case of specimen were hardened from 1050°C, the lowest R_w was obtained in the As-hardened specimen. By contrast, specimen hardened from 1100°C, the lowest R_w was obtained in the H_{Tmax} specimen.

Figure 4.42 presents the wear results of specimen No.3 with 1.05%Nb. The H_{Tmax} , L- H_{Tmax} and As-hardened conditions showed closely similar wear behavior, while the H- H_{Tmax} condition showed clearly highest total wear loss. The H_{Tmax} condition showed the lowest total wear loss, corresponding to an R_w value of 0.2705 mg/m, followed by the L- H_{Tmax} condition at 0.2809 mg/m. The As-hardened condition exhibited an R_w value of 0.2847 mg/m, while the highest R_w value was obtained in the H- H_{Tmax} condition with an R_w value of 0.3425 mg/m. Relative to the specimen hardened from 1050°C, the range of the R_w values across the four conditions narrowed slightly once the specimen was austenitized at 1100°C. While the As-hardened condition had given the best wear resistance at 1050°C, this position shifted to the H_{Tmax} condition at 1100°C.

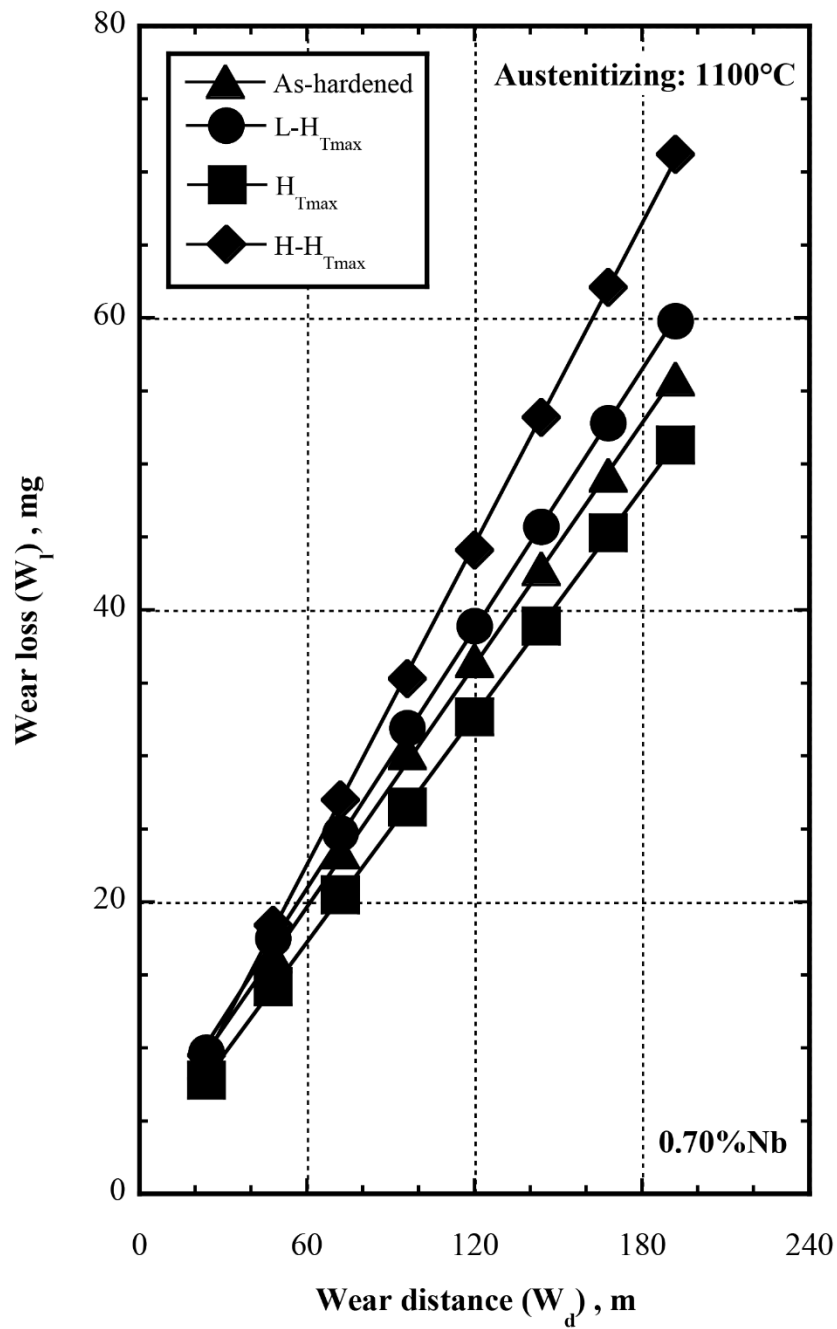
Specimen No.4 with 2.58%Nb, gave the wear results shown in Figure 4.43. The H_{Tmax} and As-hardened conditions produced comparatively low wear, the L- H_{Tmax} condition gave an intermediate value and the H- H_{Tmax} condition produced the highest total wear loss of the four. H_{Tmax} showed the lowest R_w at 0.2077 mg/m, only slightly lower than the As-hardened value of 0.2129 mg/m. L- H_{Tmax} followed at 0.2536 mg/m and H- H_{Tmax} reached the highest R_w of 0.2798 mg/m. Compared with specimen No.3, the R_w values of specimen No.4 were lower at every heat treatment condition. The grouping of the four conditions also differed from that of specimen No.3, where H_{Tmax} , L- H_{Tmax} and As-hardened had clustered together with only H- H_{Tmax} standing apart. In specimen No.4, only the As-hardened and H_{Tmax} conditions were closely paired, while L- H_{Tmax} remained clearly separated as an intermediate level. Set against the specimen hardened from 1050°C, the difference between the As-hardened and H_{Tmax} conditions narrowed further after hardening from 1100°C. Whereas the As-hardened condition had given the lowest R_w at 1050°C, this role passed to H_{Tmax} once the specimen was hardened from 1100°C.



As-hardened	▲	$W_l = 0.2130 W_d + 4.68$	$R^2 = 0.9951$
L- $H_{T_{max}}$	●	$W_l = 0.2404 W_d + 1.99$	$R^2 = 0.9997$
$H_{T_{max}}$	■	$W_l = 0.2275 W_d + 3.56$	$R^2 = 0.9989$
H- $H_{T_{max}}$	◆	$W_l = 0.2694 W_d + 1.09$	$R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

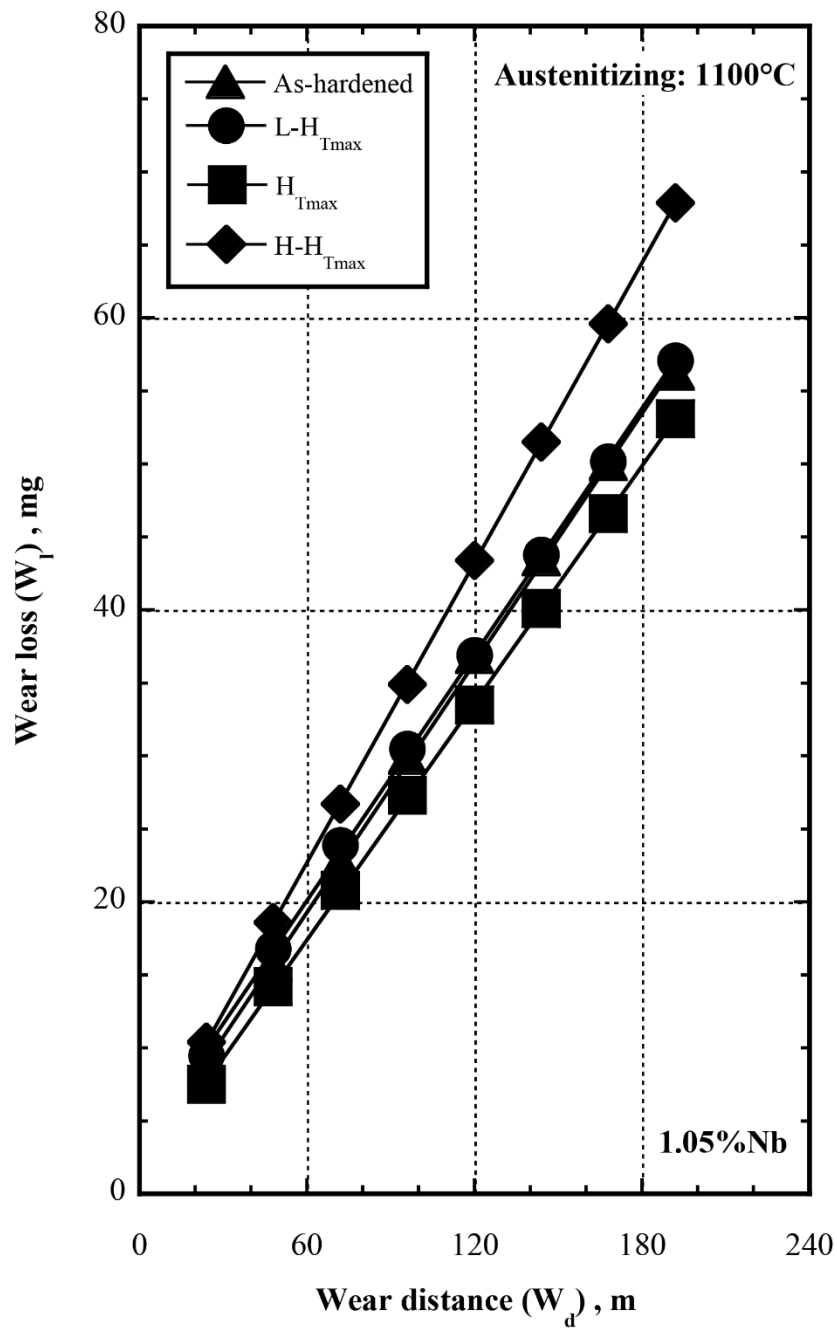
Figure 4.40 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.1 (Nb-free). Suga abrasive wear test. Load: 9.8 N (1 kgf).



As-hardened	▲	$W_l = 0.2761 W_d + 3.08$ $R^2 = 0.9993$
L- $H_{T_{max}}$	●	$W_l = 0.2962 W_d + 3.13$ $R^2 = 0.9997$
H- T_{max}	■	$W_l = 0.2586 W_d + 1.72$ $R^2 = 0.9999$
H- $H_{T_{max}}$	◆	$W_l = 0.3660 W_d + 0.58$ $R^2 = 0.9998$

W_l : Wear loss, W_d : Wear distance

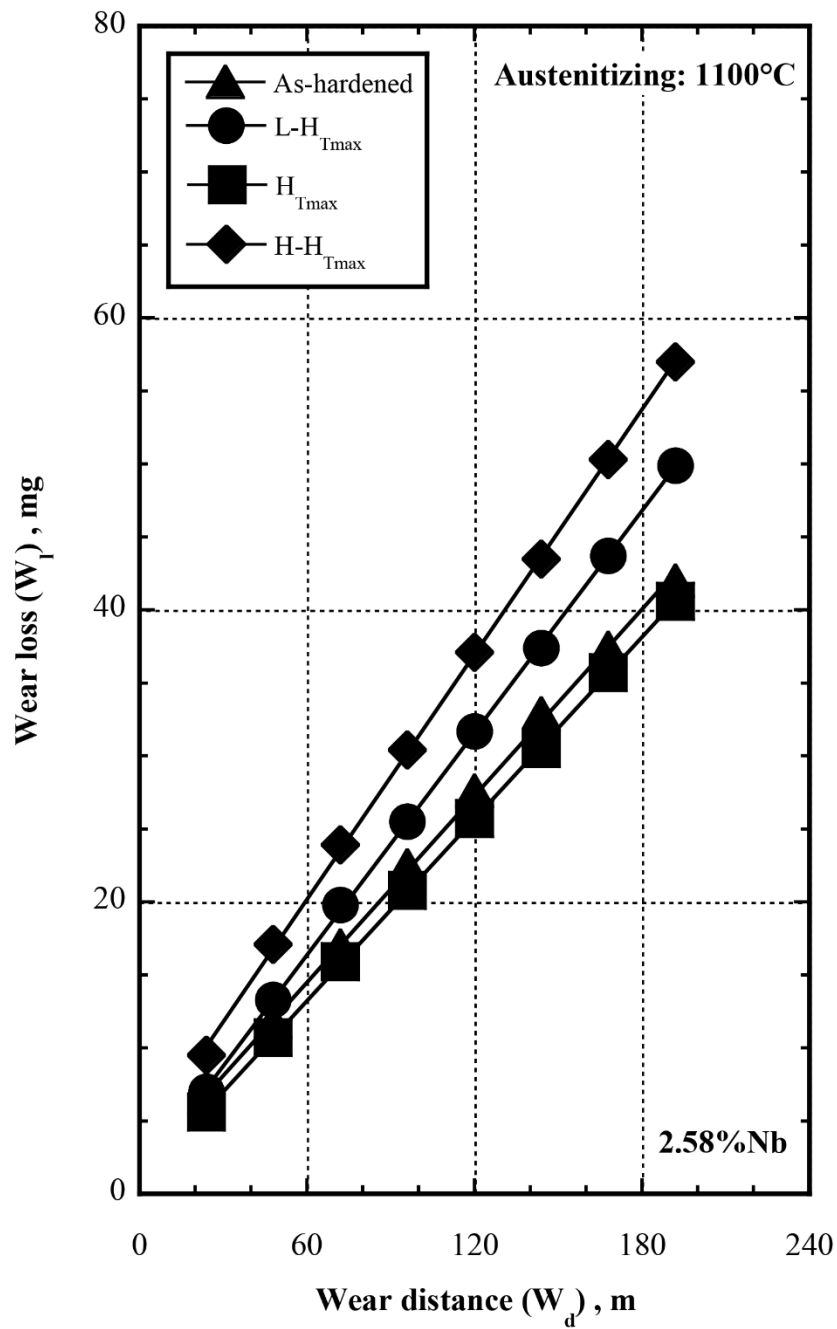
Figure 4.41 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.2 (0.70%Nb). Suga abrasive wear test. Load: 9.8 N (1 kgf).



As-hardened	▲	$W_l = 0.2847 W_d + 2.19$ $R^2 = 0.9994$
L-H _{Tmax}	●	$W_l = 0.2809 W_d + 3.23$ $R^2 = 0.9997$
H _{Tmax}	■	$W_l = 0.2705 W_d + 1.18$ $R^2 = 0.9999$
H-H _{Tmax}	◆	$W_l = 0.3425 W_d + 2.14$ $R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

Figure 4.42 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.3 (1.05%Nb). Suga abrasive wear test. Load: 9.8 N (1 kgf).



As-hardened	▲	$W_l = 0.2129 W_d + 1.71$	$R^2 = 0.9991$
L- $H_{T_{max}}$	●	$W_l = 0.2536 W_d + 1.15$	$R^2 = 0.9998$
$H_{T_{max}}$	■	$W_l = 0.2077 W_d + 0.76$	$R^2 = 0.9999$
H- $H_{T_{max}}$	◆	$W_l = 0.2798 W_d + 3.39$	$R^2 = 0.9996$

W_l : Wear loss, W_d : Wear distance

Figure 4.43 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.4 (2.58%Nb). Suga abrasive wear test. Load: 9.8 N (1 kgf).

Table 4.12 Wear rate (R_w) of heat-treated specimens. Suga abrasive wear test under a load of 9.8 N (1 kgf). Austenitizing: 1100°C.

Specimen	Wear rate (mg/m)			
	As-hardened	L-H _{Tmax}	H _{Tmax}	H-H _{Tmax}
No.1 (Nb-free)	0.2130	0.2404	0.2275	0.2694
No.2 (0.70%Nb)	0.2761	0.2962	0.2586	0.3660
No.3 (1.05%Nb)	0.2847	0.2809	0.2705	0.3425
No.4 (2.58%Nb)	0.2129	0.2536	0.2077	0.2798

4.3.2 Rubber wheel abrasive wear test (three-body type)

The test results for the specimens hardened from 1050°C and 1100°C are presented in Figure 4.44 to 4.47 and Figure 4.48 to 4.51, with the corresponding R_w values summarized in Table 4.13 and 4.14. A linear relationship between W_l and W_d was obtained for every specimen and heat treatment conditions and R_w was taken as the slope of this relationship. Overall, the R_w values obtained from the rubber wheel test were lower than those from the Suga test, reflecting the difference between three-body and two-body abrasive wear mechanisms.

4.3.2.1 Wear test results of specimens hardened from 1050°C

The wear behavior of specimen No.1 (Nb-free) is presented in Figure 4.44 and produced a linear relationship between W_l and W_d consistent with the linear trend observed for every specimen and heat treatment condition. The trend lines separated with the As-hardened specimen clearly apart at the lower end, the H_{Tmax} and L-H_{Tmax} specimens forming a closely spaced middle pair and the H-H_{Tmax} specimen at the upper end. Consistent with this arrangement, the As-hardened specimen obtained the lowest total wear loss, while the H-H_{Tmax} specimen obtained the highest. The corresponding R_w values were 0.0851 mg/m for the As-hardened specimen, 0.1101 mg/m and 0.1142 mg/m for the H_{Tmax} and L-H_{Tmax} specimens, and the highest value at 0.1257 mg/m for H-H_{Tmax} specimen. This grouping also differed in shape from that of the same specimen evaluated by the Suga test. In the Suga test, the four conditions separated into two closely spaced pairs, As-hardened with H_{Tmax} and L-H_{Tmax} with H-H_{Tmax}, with only a small overall difference in R_w between the two groups. In the rubber wheel test, by contrast, the As-hardened and H-H_{Tmax} specimens separated from the other two conditions, giving a much wider spread between the

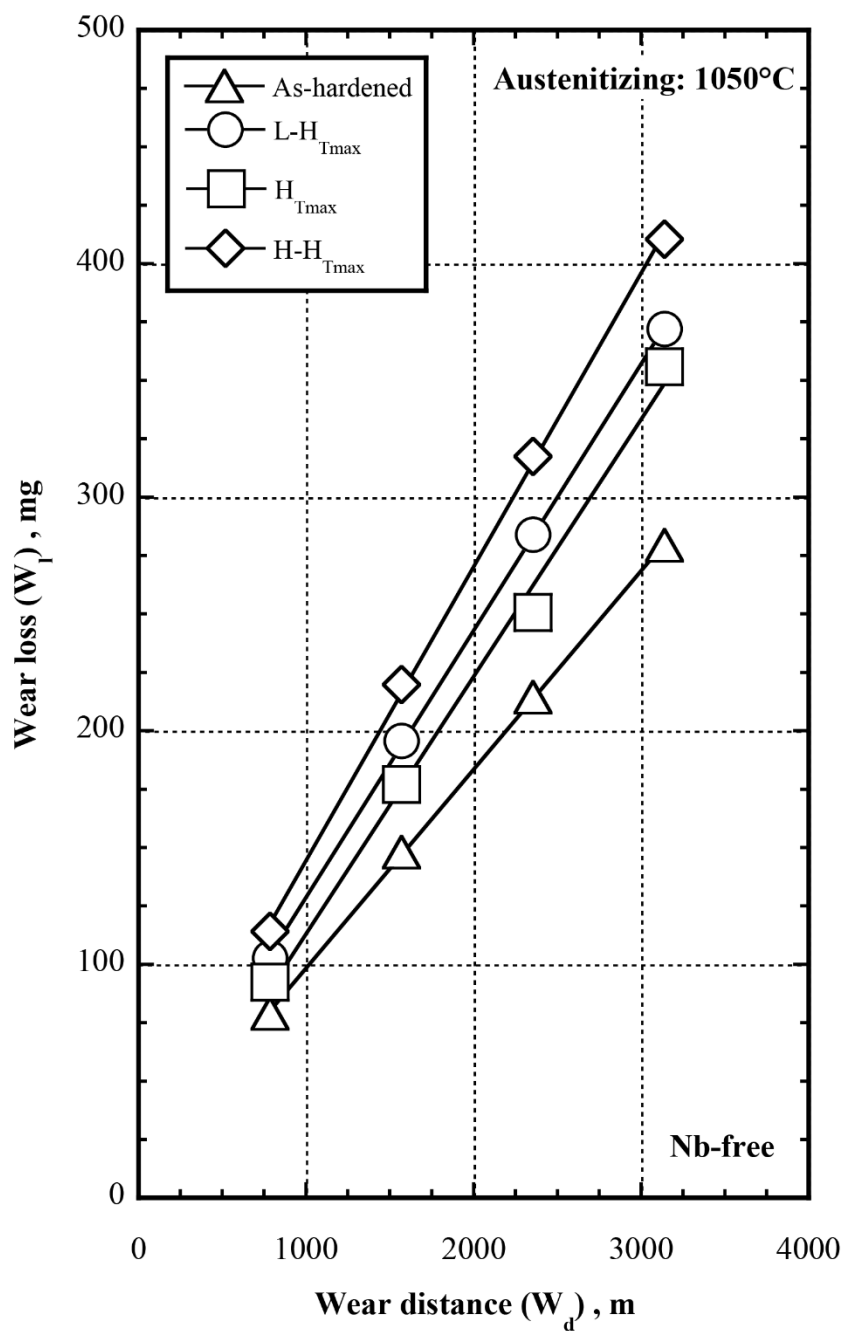
lowest and highest R_w values and indicating that the wear behavior of the Nb-free specimen was more sensitive to heat treatment condition under three-body abrasion than under two-body abrasion.

As for specimen No.2 with 0.70%Nb, the wear results are presented in Figure 4.45. The wear behavior is divided into two distinct groups, with the As-hardened and $H_{T_{max}}$ specimens located in the lower total wear loss group and the L- $H_{T_{max}}$ and H- $H_{T_{max}}$ specimens forming the higher total wear loss group. The As-hardened specimen showed the lowest total wear loss and the H- $H_{T_{max}}$ specimen the highest, consistent with this grouping. Within the lower group, the As-hardened and $H_{T_{max}}$ specimens reached R_w values of 0.0784 mg/m and 0.0890 mg/m. The higher group followed at 0.1043 mg/m for L- $H_{T_{max}}$ and 0.1170 mg/m for H- $H_{T_{max}}$. Whereas specimen No.1 had the As-hardened condition separated alone at the bottom and the $H_{T_{max}}$ and L- $H_{T_{max}}$ conditions forming the closely spaced middle pair, specimen No.2 produced a clean two-group split and its R_w values were lower than those of specimen No.1 at every heat treatment condition.

In the case of specimen No.3 with 1.05%Nb, the wear results are shown in Figure 4.46. The As-hardened, $H_{T_{max}}$ and L- $H_{T_{max}}$ specimens are located in the lower total wear loss region, while H- $H_{T_{max}}$ specimen is located in the higher total wear loss region. Total wear loss followed the same order, lowest in the As-hardened and highest in the H- $H_{T_{max}}$ specimen. The As-hardened specimen reached the minimum R_w value of 0.0724 mg/m followed by $H_{T_{max}}$, L- $H_{T_{max}}$ and the H- $H_{T_{max}}$ specimens with R_w values of 0.0883, 0.0943 and 0.1133 mg/m, respectively. Compared with specimen No.3 evaluated by the Suga test, where the As-hardened and $H_{T_{max}}$ specimens had formed one group and the L- $H_{T_{max}}$ and H- $H_{T_{max}}$ specimens another, the rubber wheel test again produced a different grouping shape.

For specimen No.4 with 2.58%Nb, the wear results are illustrated in Figure 4.47. Unlike the other three specimens, the four heat treatment conditions did not separate into distinct groups. Instead, R_w increased progressively from the As-hardened condition through $H_{T_{max}}$, and L- $H_{T_{max}}$ to H- $H_{T_{max}}$, in relatively even steps. Total wear loss followed the same increasing order, lowest in the As-hardened specimen and highest in the H- $H_{T_{max}}$ specimen. The R_w values were 0.0835 mg/m for As-hardened, 0.1025 mg/m for $H_{T_{max}}$, 0.1162 mg/m for L- $H_{T_{max}}$ and 0.1348 mg/m for H- $H_{T_{max}}$, the highest of the four. Whereas specimen No.3 had shown the As-hardened,

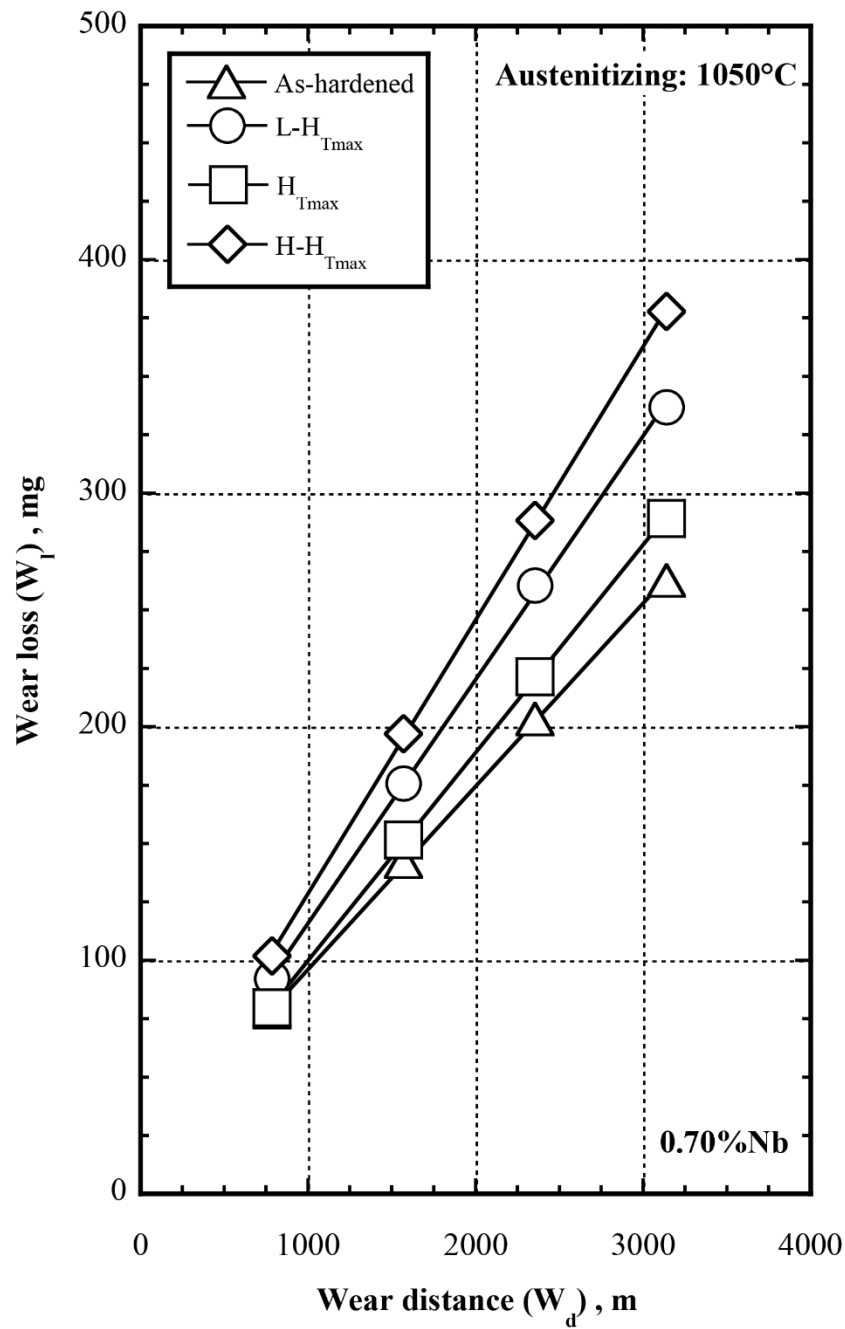
$H_{T_{max}}$ and $L-H_{T_{max}}$ specimens clustered together with only $H-H_{T_{max}}$ standing apart, specimen No.4 showed no such clustering and its R_w values were higher than those of specimen No.3 at every heat treatment condition. Compared with specimen No.4 evaluated by the Suga test, where the As-hardened condition located distinctly apart while the $L-H_{T_{max}}$ and $H-H_{T_{max}}$ conditions ran close together, the rubber wheel test again produced a different pattern, with all four conditions spread in a progressive sequence rather than forming distinct groups.



As-H	△	$W_l = 0.0851 W_d + 13.25$	$R^2 = 0.9998$
L- $H_{T_{max}}$	○	$W_l = 0.1142 W_d + 14.55$	$R^2 = 0.9998$
H- T_{max}	□	$W_l = 0.1101 W_d + 2.95$	$R^2 = 0.9948$
H- $H_{T_{max}}$	◇	$W_l = 0.1257 W_d + 18.65$	$R^2 = 0.9992$

W_l : Wear loss, W_d : Wear distance

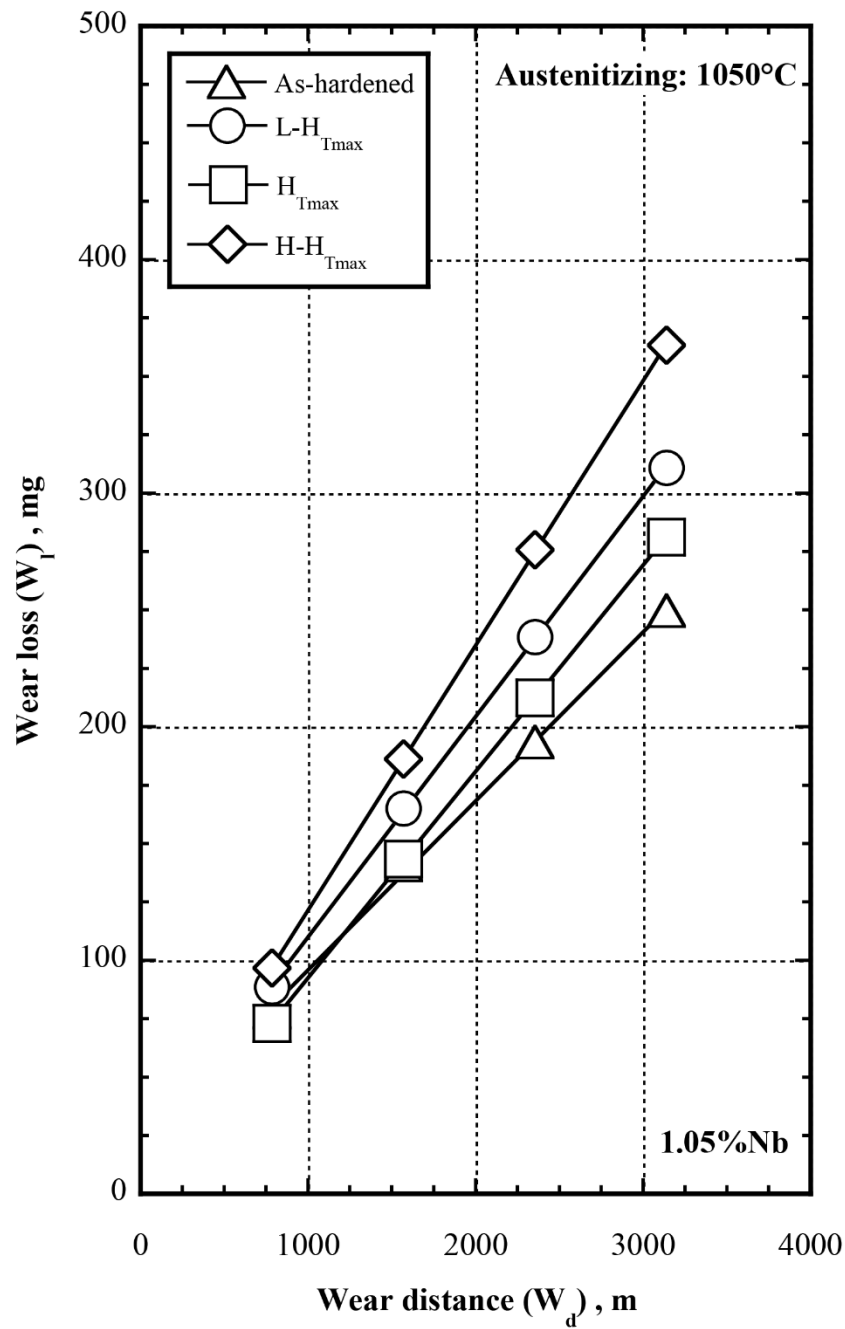
Figure 4.44 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.1 (Nb-free). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).



As-H	△	$W_l = 0.0784 W_d + 17.65$	$R^2 = 0.9996$
L- H_{Tmax}	○	$W_l = 0.1043 W_d + 11.55$	$R^2 = 0.9995$
H- T_{max}	□	$W_l = 0.0890 W_d + 10.90$	$R^2 = 0.9998$
H- H_{Tmax}	◇	$W_l = 0.1170 W_d + 11.60$	$R^2 = 0.9998$

W_l : Wear loss, W_d : Wear distance

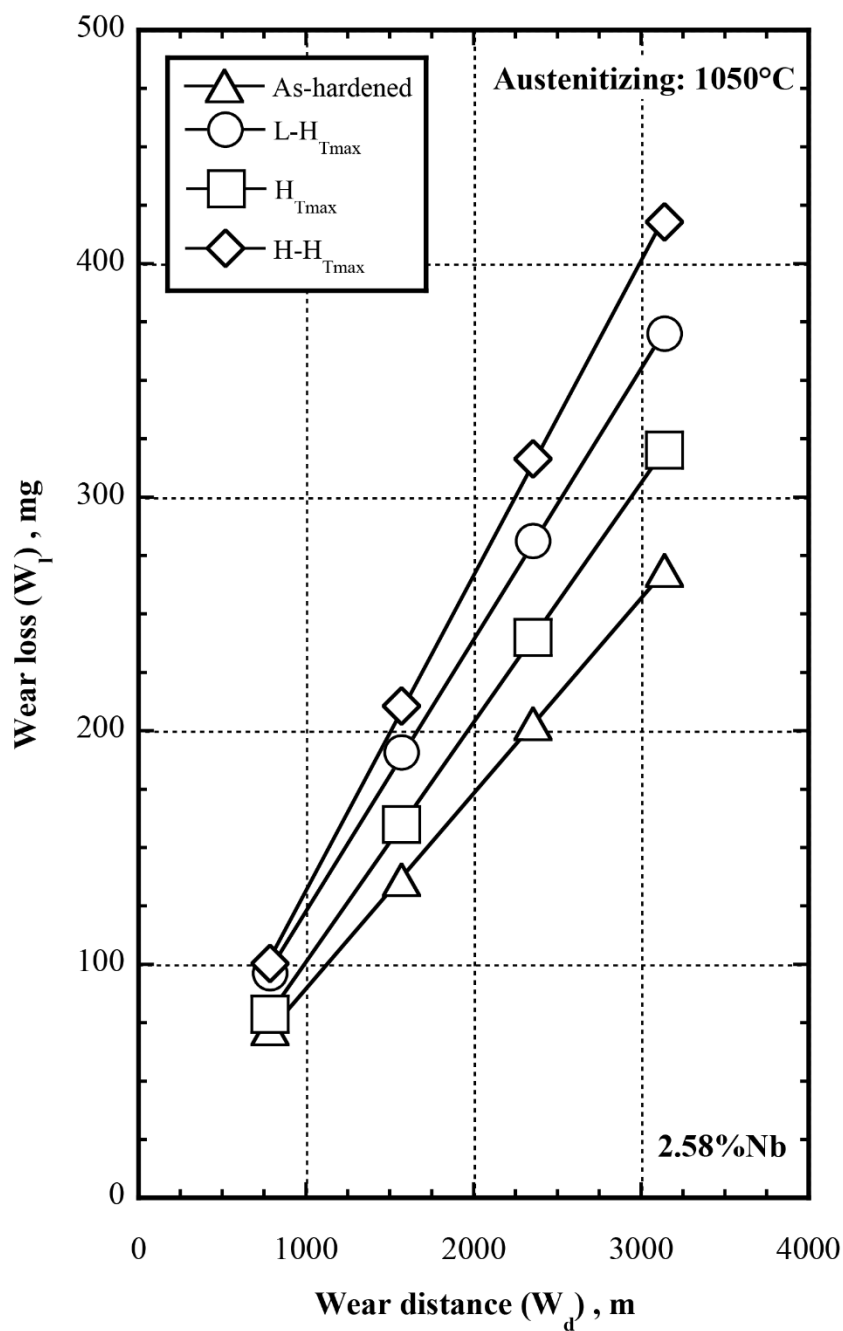
Figure 4.45 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.2 (0.70%Nb). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).



As-hardened	△	$W_l = 0.0724 W_d + 23.45$	$R^2 = 0.9987$
L- $H_{T_{max}}$	○	$W_l = 0.0943 W_d + 15.70$	$R^2 = 0.9998$
$H_{T_{max}}$	□	$W_l = 0.0883 W_d + 4.10$	$R^2 = 0.9999$
H- $H_{T_{max}}$	◇	$W_l = 0.1133 W_d + 8.15$	$R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

Figure 4.46 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.3 (1.05%Nb). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).



As-hardened	△	$W_l = 0.0835 W_d + 6.10$	$R^2 = 0.9999$
L- $H_{T_{max}}$	○	$W_l = 0.1162 W_d + 6.60$	$R^2 = 0.9998$
H- T_{max}	□	$W_l = 0.1025 W_d - 1.55$	$R^2 = 0.9999$
H- $H_{T_{max}}$	◇	$W_l = 0.1348 W_d - 3.30$	$R^2 = 0.9997$

W_l : Wear loss, W_d : Wear distance

Figure 4.47 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.4 (2.58%Nb). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).

Table 4.13 Wear rate (R_w) of heat-treated specimens. Rubber wheel abrasive wear test under a load of 85.3 N (8.7 kgf). Austenitizing: 1050°C.

Specimen	Wear rate (mg/m)			
	As-hardened	L- $H_{T_{max}}$	$H_{T_{max}}$	H- $H_{T_{max}}$
No.1 (Nb-free)	0.0851	0.1142	0.1101	0.1257
No.2 (0.70%Nb)	0.0784	0.1043	0.0890	0.1170
No.3 (1.05%Nb)	0.0724	0.0943	0.0883	0.1133
No.4 (2.58%Nb)	0.0835	0.1162	0.1025	0.1348

4.3.2.2 Wear test results of specimens hardened from 1100°C

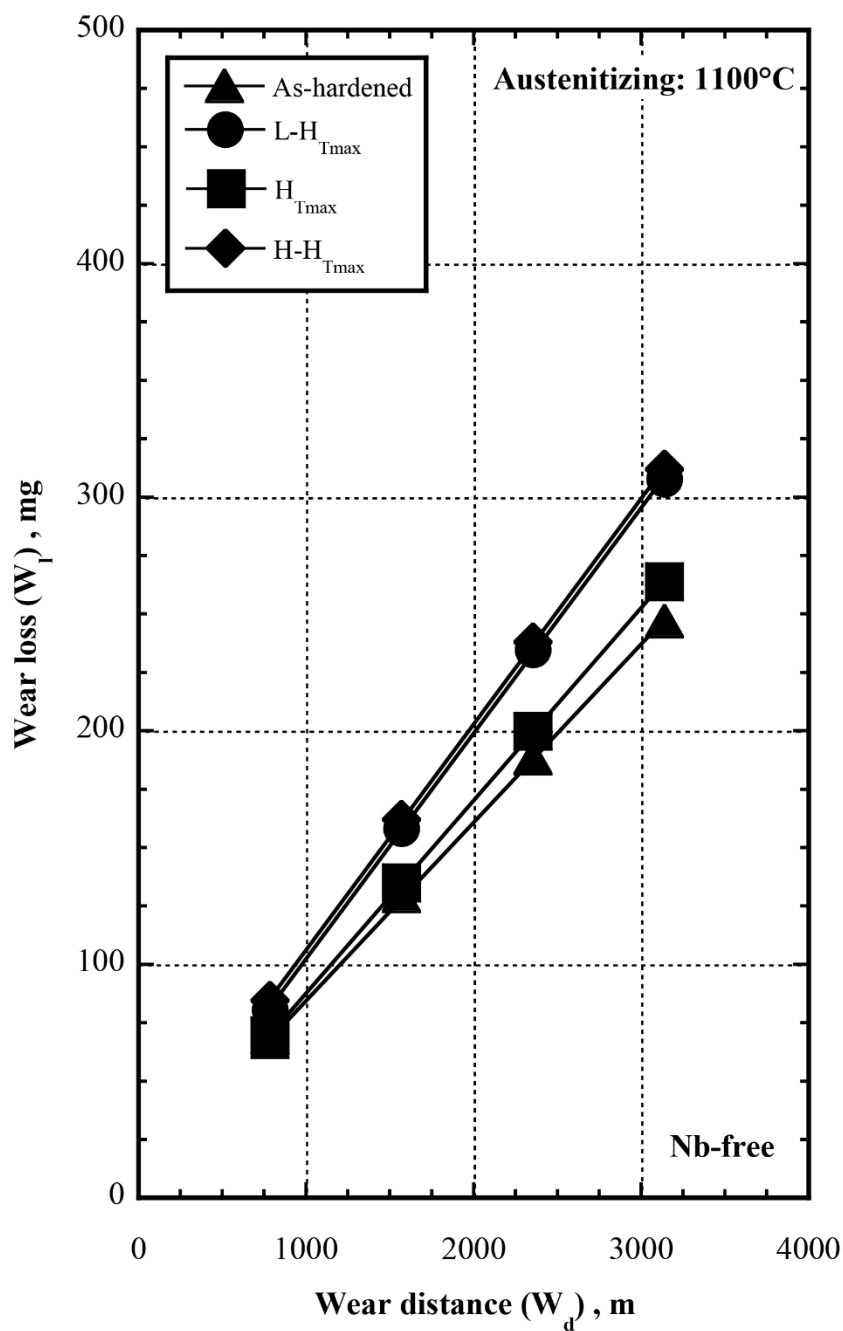
Figure 4.48 presents the wear results of specimen No.1 with Nb-free and a linear relationship between W_l and W_d was obtained for all specimens. The As-hardened and $H_{T_{max}}$ specimens belonged to the lower total wear loss group, while L- $H_{T_{max}}$ and H- $H_{T_{max}}$ specimens represented the higher group, with the latter two conditions reaching R_w values that were almost the same. Total wear loss lowest in the As-hardened specimen and highest is in the H- $H_{T_{max}}$ specimen. The corresponding R_w values were 0.0764 mg/m for As-hardened, 0.0825 mg/m for $H_{T_{max}}$, 0.0965 mg/m for L- $H_{T_{max}}$ and 0.0967 mg/m for H- $H_{T_{max}}$. In comparison with the same specimen hardened from 1050°C, where the As-hardened condition was separated at the bottom, while $H_{T_{max}}$ and L- $H_{T_{max}}$ clustered closely together at the higher position and H- $H_{T_{max}}$ reached the highest value, hardening from 1100°C produced a cleaner two-group split and the R_w was lower at every heat treatment condition.

In the case of specimen No.2 with 0.70%Nb, shown in Figure 4.49, the As-hardened and $H_{T_{max}}$ specimens were located in the lower wear rate region, while the L- $H_{T_{max}}$ and H- $H_{T_{max}}$ specimens occupied the higher region. The $H_{T_{max}}$ specimen exhibited the lowest total wear loss overall, while the As-hardened specimen showed the lowest R_w value, unlike specimen No.1, where the As-hardened condition had singly obtained both the lowest total wear loss and the lowest R_w . The H- $H_{T_{max}}$ specimen produced the highest total wear loss. The R_w reached 0.0721 mg/m for As-hardened and 0.0731 mg/m for $H_{T_{max}}$ in the lower group, compared with 0.0874 mg/m for L- $H_{T_{max}}$ and 0.0932 mg/m for H- $H_{T_{max}}$ in the higher group. This grouping is similar to that of specimen No.1. The same two-group pattern had already appeared

in specimen No.2 hardened from 1050°C and hardened from 1100°C led to a further reduction in R_w across all heat treatment conditions.

Specimen No.3 with 1.05%Nb, illustrated in Figure 4.50, showed the As-hardened and H_{Tmax} specimens occupying the lower wear rate region, with comparable R_w values of 0.0703 mg/m and 0.0720 mg/m. Although the H_{Tmax} specimen exhibited the lowest total wear loss, the lowest R_w value was obtained in the As-hardened specimen, reproducing the same discrepancy seen in specimen No.2. In contrast to specimen No.2, where the four conditions separated into two clear groups, specimen No.3 formed only one such group, As-hardened and H_{Tmax} . The L- H_{Tmax} and H- H_{Tmax} specimens instead separated from each other, with R_w increasing progressively away from H_{Tmax} to reach 0.0819 mg/m in L- H_{Tmax} and 0.0921 mg/m, the highest value in H- H_{Tmax} specimen which also gave the highest total wear loss. The R_w values of specimen No.3 were lower than those of specimen No.2 at every heat treatment condition. At 1050°C, this specimen had shown a different pattern, with the As-hardened condition standing apart on its own rather than pairing with H_{Tmax} , yet its wear resistance still improved after hardening from 1100°C, with R_w decreasing at every heat treatment condition.

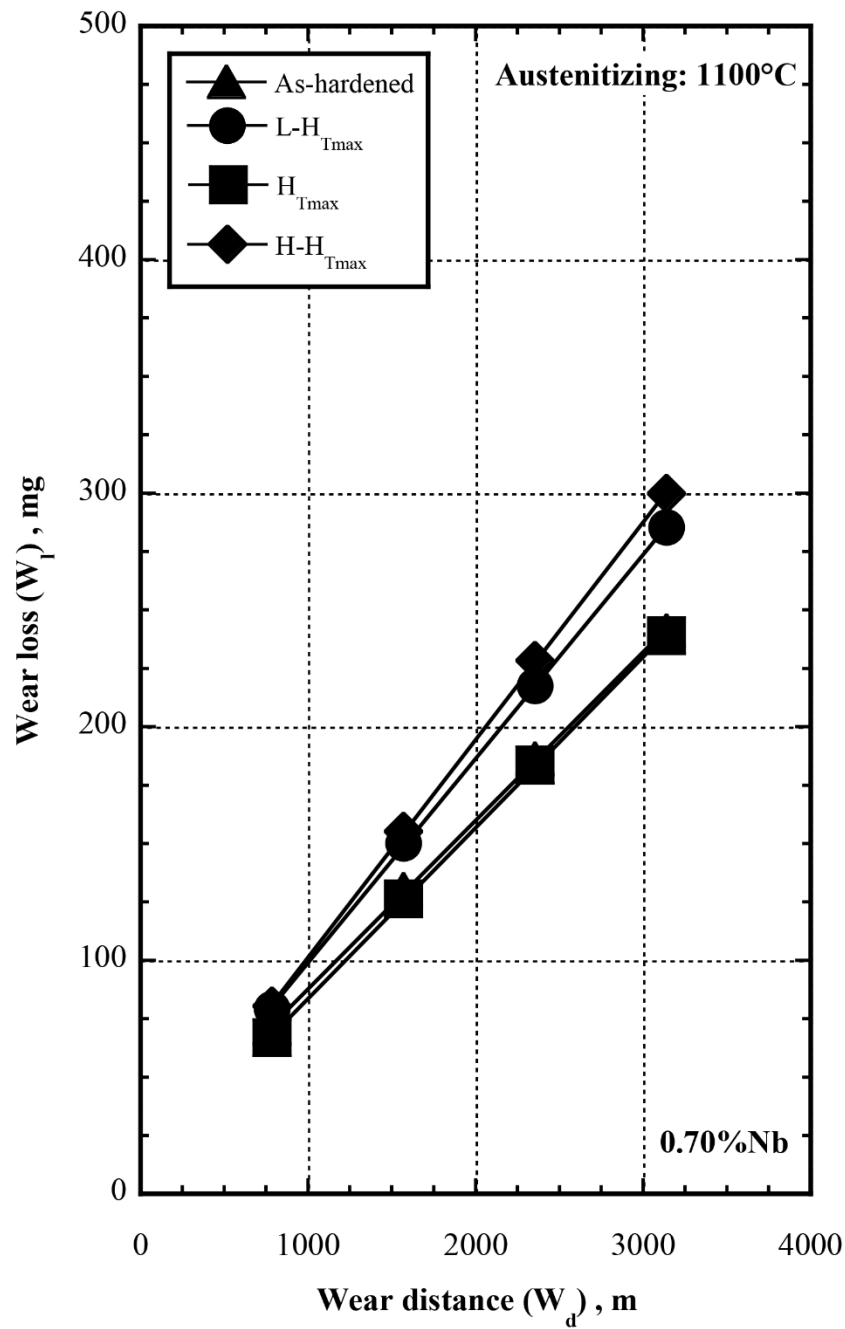
The wear results of specimen No.4 with 2.58%Nb are presented in Figure 4.51. As in specimen No.3, the As-hardened and H_{Tmax} specimens occupied the lower wear rate region, with R_w values of 0.0834 mg/m and 0.0850 mg/m. The H_{Tmax} specimen showed the smallest total wear loss, whereas the As-hardened specimen had the lower R_w , continuing the same difference found in specimens No.2 and No.3. Different from any of the other three specimens, it was the L- H_{Tmax} rather than the H- H_{Tmax} condition that reached both the largest total wear loss and the highest R_w , at 0.1110 mg/m, with H- H_{Tmax} following closely behind at 0.1066 mg/m. At every heat treatment condition, R_w was higher for specimen No.4 than for specimen No.3. The same specimen hardened from 1050°C had shown R_w rising in a gradual increase from As-hardened through H_{Tmax} and L- H_{Tmax} to H- H_{Tmax} , with no clear grouping at all. After hardening from 1100°C, the As-hardened and H_{Tmax} conditions paired closely together as in the other three specimens, while L- H_{Tmax} and H- H_{Tmax} remained close to each other but reversed which one gave the higher value.



As-hardened	▲	$W_l = 0.0764 W_d + 7.95$	$R^2 = 0.9998$
L- $H_{T_{max}}$	●	$W_l = 0.0965 W_d + 5.80$	$R^2 = 0.9998$
$H_{T_{max}}$	■	$W_l = 0.0825 W_d + 5.05$	$R^2 = 0.9999$
H- $H_{T_{max}}$	◆	$W_l = 0.0967 W_d + 9.40$	$R^2 = 0.9999$

W_l : Wear loss, W_d : Wear distance

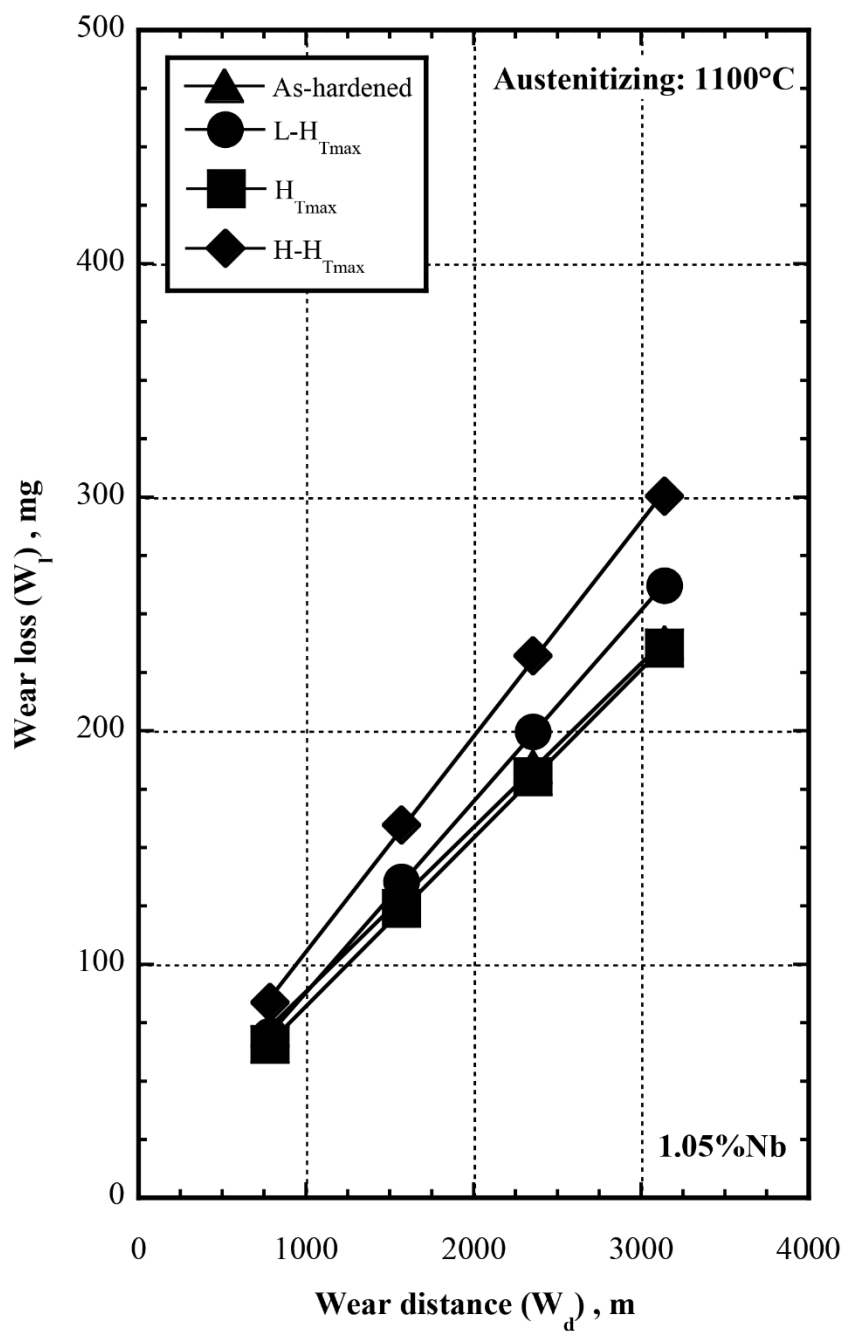
Figure 4.48 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.1 (Nb-free). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).



As-hardened	▲	$W_l = 0.0721 W_d + 15.65$	$R^2 = 0.9997$
L- H_{Tmax}	●	$W_l = 0.0874 W_d + 11.55$	$R^2 = 0.9999$
H- T_{max}	■	$W_l = 0.0731 W_d + 10.45$	$R^2 = 0.9997$
H-H- T_{max}	◆	$W_l = 0.0932 W_d + 8.10$	$R^2 = 0.9992$

W_l : Wear loss, W_d : Wear distance

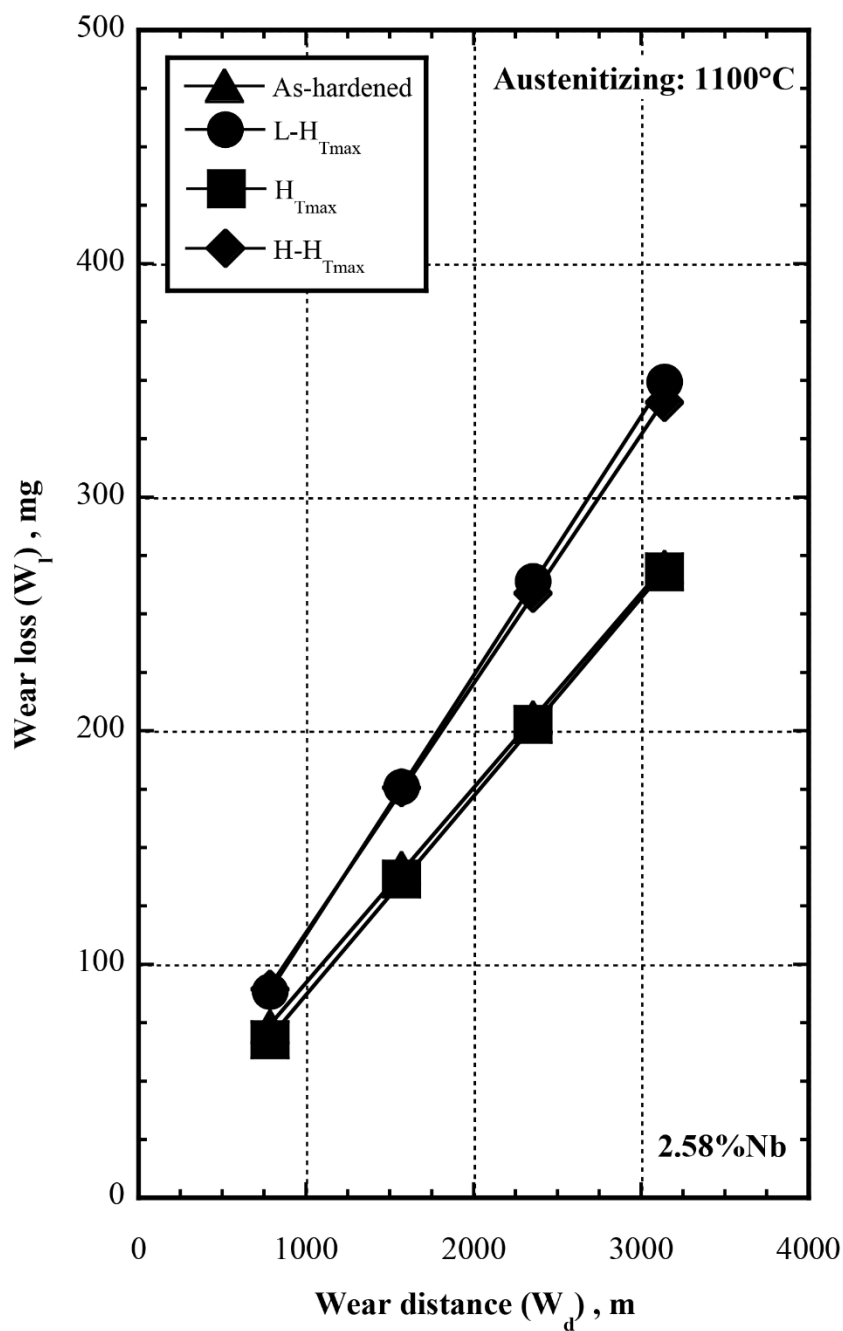
Figure 4.49 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.2 (0.70%Nb). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).



As-hardened	▲	$W_l = 0.0703 W_d + 18.10$	$R^2 = 0.9996$
L-H _{Tmax}	●	$W_l = 0.0819 W_d + 5.95$	$R^2 = 0.9999$
H _{Tmax}	■	$W_l = 0.0720 W_d + 10.00$	$R^2 = 0.9998$
H-H _{Tmax}	◆	$W_l = 0.0921 W_d + 13.30$	$R^2 = 0.9995$

W_l : Wear loss, W_d : Wear distance

Figure 4.50 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.3 (1.05%Nb). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).



As-hardened	▲	$W_l = 0.0834 W_d + 8.75$	$R^2 = 0.9998$
L- $H_{T_{max}}$	●	$W_l = 0.1110 W_d + 1.65$	$R^2 = 0.9999$
H- T_{max}	■	$W_l = 0.0850 W_d + 1.95$	$R^2 = 0.9999$
H-H- T_{max}	◆	$W_l = 0.1066 W_d + 6.80$	$R^2 = 0.9998$

W_l : Wear loss, W_d : Wear distance

Figure 4.51 Relationship between wear loss (W_l) and wear distance (W_d) of specimen No.4 (2.58%Nb). Rubber wheel abrasive wear test. Load: 85.3 N (8.7 kgf).

Table 4.14 Wear rate (R_w) of heat-treated specimens. Rubber wheel abrasive wear test under a load of 85.3 N (8.7 kgf). Austenitizing: 1100°C.

Specimen	Wear rate (mg/m)			
	As-hardened	L-H _{Tmax}	H _{Tmax}	H-H _{Tmax}
No.1 (Nb-free)	0.0764	0.0965	0.0825	0.0967
No.2 (0.70%Nb)	0.0721	0.0874	0.0731	0.0932
No.3 (1.05%Nb)	0.0703	0.0819	0.0720	0.0921
No.4 (2.58%Nb)	0.0834	0.1110	0.0850	0.1066

4.3.3 Comparison of abrasive wear severity between Suga and rubber wheel abrasive wear tests.

In order to assess the relative severity between the two wear conditions, the Suga and rubber wheel abrasive wear tests were conducted under the conditions described in Sections 3.6.2 and 3.6.3, respectively, with the SiC abrasive paper configuration and the rubber wheel wear mechanism illustrated in Figure 4.52 and Figure 4.53. Since the total worn area differed between the two tests (2.304 m² for Suga and 47.12 m² for rubber wheel), a normalized parameter was required to directly compare the severity of the two abrasive wear conditions.

To compare the severity of abrasive wear between the Suga and rubber wheel abrasive wear test, the 2.58%Nb specimen was selected for analysis. Since the two tests were conducted under different loading conditions and worn areas, the total wear loss alone may not directly reflect the severity of each wear condition. Therefore, the specific wear rate (S_{WR}) was introduced as a normalized parameter. The S_{WR} was calculated by dividing the total wear loss by the applied load and the worn area, as expressed in equation 4.1.

$$S_{WR} = \frac{\text{Total wear loss}}{(\text{Load} \times \text{Total worn area})} \quad (4.1)$$

The calculated S_{WR} values and the corresponding severity ratios between the Suga and rubber wheel abrasive wear test are summarized in Table 4.15. The results clearly show that the Suga abrasive wear test produced a much higher S_{WR} than the rubber wheel abrasive wear test for all heat treatment conditions. This indicates that the Suga test imposed a more severe abrasive wear condition. Based on the average severity

ratio, the Suga abrasive wear condition was approximately 27.42 times more severe than the rubber wheel abrasive wear condition.

The markedly higher severity of the Suga test relative to the rubber wheel test can be attributed to the constrained nature of the abrasive, its hardness relative to the eutectic carbides and the resulting difference in contact stress. In the Suga test, SiC particles (2500 – 2800 HV) are rigidly bonded to the paper and pressed against the specimen over a narrow 12 mm contact width. Being unable to roll or redistribute the load, each particle tip concentrates the applied stress locally, producing continuous two-body cutting and ploughing of both the carbides and the matrix (Glaeser, 1986). In the rubber wheel test, the silica sand particles (1100 HV) are entrained loosely between the compliant rubber wheel and the specimen. The rubber's elasticity allows the particles to partially embed and roll rather than slide, which distributes the contact load over a larger number of particles and lowers the local stress concentration at any single contact point. According to Lighvan (2021), the degree of constraint, rather than the magnitude of contact stress alone, is the dominant factor governing this difference, since constrained particles spend most of their travel sliding and cutting while loose particles spend most of it rolling. In addition, the silica sand hardness is insufficient to cut the eutectic carbide (1500 – 2500 HV), so wear under the rubber wheel condition is confined mainly to the matrix. This is in good agreement with Opapaiboon et al., who reported a comparable severity ratio (approximately 31) between the two tests for a compositionally similar multi-alloyed white cast iron, attributed to the same combination of constraint, hardness and contact stress effects (Opapaiboon et al., 2025b).

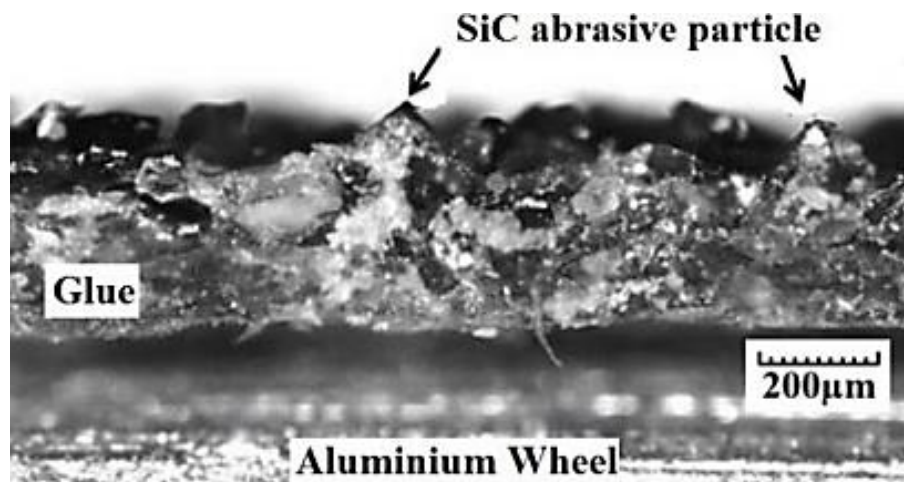


Figure 4.52 Cross-sectional structure of the SiC abrasive paper fixed on the aluminum wheel used in the Suga abrasive wear test (Aungsupaitoon, 2008).

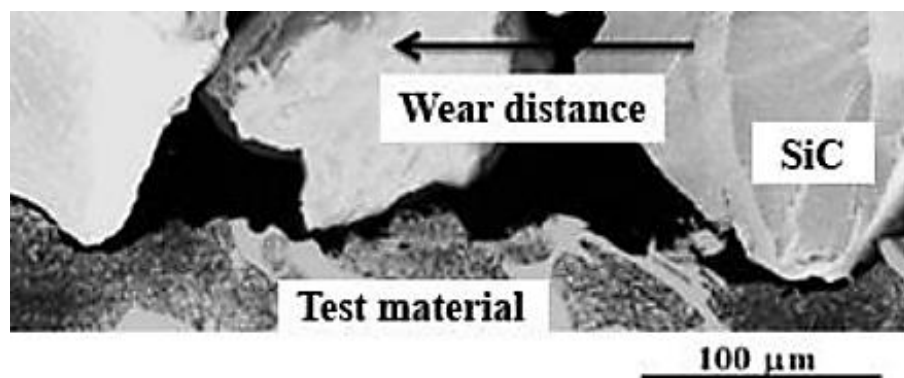


Figure 4.53 Schematic illustration of the wear mechanism in the rubber wheel abrasive wear test. (Aungsupaitoon, 2008).

Table 4.15 Specific wear rate of the 2.58%Nb specimen under different heat treatment conditions after austenitizing at 1100°C.

Heat treatment condition	Specific wear rate, (mg/N·m ²)		Ratio of set (Suga/Rubber wheel)
	Suga	Rubber wheel	
As-hardened	1.86	0.067	27.76
L-H _{Tmax}	2.21	0.087	25.40
H _{Tmax}	1.80	0.067	26.87
H-H _{Tmax}	2.52	0.085	29.65
Average			27.42

4.3.4 Effect of Nb addition on wear rate

The wear resistance of multi-alloyed white cast iron is governed by several factors, including the type and amount of eutectic carbides as well as the characteristics of the matrix structure. The matrix plays an essential role in supporting both eutectic and secondary carbides during abrasive wear. However, when the matrix strength is insufficient, abrasive particles can preferentially remove the matrix phase (Pourasiabi & Gates, 2022b). This weakens the support for the carbides and may lead to the detachment or spalling of both eutectic and secondary carbides. Therefore, the matrix strength, which is closely associated with matrix hardness, is an important factor controlling the wear resistance of the material. The results presented in Sections 4.1 and 4.2 indicated that Nb content affects the microstructure, hardness and volume fraction of retained austenite which influence the R_w of multi-alloyed white cast iron. Therefore, the relationship between R_w and Nb content was examined. Figure 4.54 shows the relationship between R_w and Nb content for the specimen subjected to the Suga abrasive wear test, while Figure 4.55 presents the corresponding results for the rubber wheel abrasive wear test.

In the Suga abrasive wear test, the R_w values of the As-hardened and H_{Tmax} specimens increased with increasing Nb content up to 1.05%Nb, followed by a decrease in specimen No.4 with 2.58%Nb, regardless of the austenitizing temperature. The increase in R_w from the Nb-free specimen to 0.70%Nb, despite the concurrent increase in eutectic MC carbide fraction, was mainly attributed to a decrease in overall hardness after heat treatment, even though the eutectic MC carbide fraction established during solidification remained unchanged. Although the eutectic MC carbide increased with Nb addition, it consumed more carbon during solidification and reduced the carbon content remaining in the matrix. Owing to limited carbon availability during austenitizing, the precipitation of secondary carbides was suppressed and more carbide-forming elements remained dissolved in the austenitic matrix. This promoted austenite stabilization and increased the V_γ compared with the Nb-free specimen. These factors weakened the matrix support and reduced the resistance to two-body abrasive wear, leading to an increase in R_w .

When the Nb content increased to 1.05%, the R_w values reached the highest values in both the As-hardened and H_{Tmax} conditions, mainly due to the lowest hardness among the specimens. Although primary MC carbide (MC_p) crystallization

occurred in this specimen, it exhibited the lowest amount of carbides formed during solidification. In addition, after heat treatment, this specimen showed the lowest area fraction of secondary carbides. In addition, the further increase in the V_γ contributed to additional softening of the matrix. As a result, the 1.05%Nb specimen had the lowest hardness which reduced its resistance to abrasive wear and led to the highest R_w values in the As-hardened and H_{Tmax} conditions.

When the carbon and Nb contents were increased to 2.40% and 2.58%, respectively, the lowest R_w values in both the As-hardened and H_{Tmax} conditions were obtained. This was mainly due to the resulting increase in hardness. Compared with the 1.05%Nb specimen, the concurrent increase in carbon content from 2.00% to 2.40% and Nb content from 1.05% to 2.58% promoted the crystallization of MC_p carbide and increased the amount of eutectic carbide. Moreover, the higher carbon availability allowed sufficient carbon to remain in the matrix for secondary carbide precipitation during heat treatment. This is evidenced by the marked increase in secondary carbide area fraction at both austenitizing temperatures, exceeding even the value obtained in the Nb-free specimen. In contrast, the V_γ increased only slightly at 1100°C, a markedly smaller increment than observed at lower Nb contents, although the increase at 1050°C was comparatively more moderate. Although this specimen exhibited the highest V_γ , the beneficial effects of increased carbide formation and higher matrix hardness appeared to be more dominant than the softening effect of retained austenite. Consequently, the matrix became stronger and provided better support to the carbides, resulting in improved resistance to abrasive wear and lowest R_w values among the specimens. Compared with the Nb-free specimen, this reduction in R_w corresponded to an improvement in wear resistance of approximately 5.5% and 8.7% in the H_{Tmax} specimens austenitized at 1050°C and 1100°C, respectively.

In the Suga abrasive wear test, the effect of austenitizing temperature on R_w showed a similar trend at both austenitizing temperatures. Nevertheless, the R_w values of the specimens hardened from 1100°C were lower than those of the specimens hardened from 1050°C. This result is attributed to the higher austenitizing temperature, which increased the volume fraction of retained austenite in the as-hardened condition and enhanced the secondary hardening response after tempering. Consequently, the higher hardness obtained after hardening from 1100°C improved the resistance to abrasive wear and reduced the R_w values.

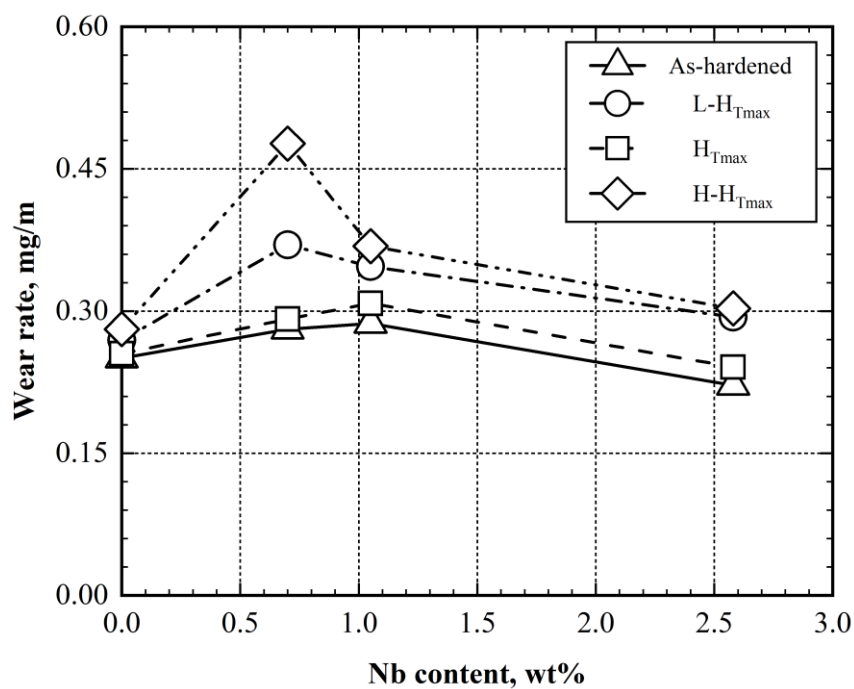
Regarding the effect of heat treatment condition on wear behavior, the As-hardened and $H_{T_{max}}$ specimens exhibited better wear resistance than the $L-H_{T_{max}}$ and $H-H_{T_{max}}$ specimens in the Suga abrasive wear tests. This behavior is considered to be associated with differences in the matrix structure of the specimens. The phases present in the matrix differed between the As-hardened and tempered conditions. In the As-hardened specimens, the matrix consisted of secondary carbides precipitated during austenitizing, martensite containing high carbon and dissolved carbide-forming elements formed during quenching and retained austenite that remained untransformed. In the tempered state, the matrix of the $H_{T_{max}}$ specimens was composed of very fine tempered martensite, secondary carbides and a small amount of retained austenite. In the $H-H_{T_{max}}$ specimens, excessive tempering promoted the coarsening and agglomeration of secondary carbides, resulting in a matrix containing coarse carbides and only a very small amount of retained austenite. In contrast, the $L-H_{T_{max}}$ specimens contain tempered martensite, secondary carbides and a higher amount of retained austenite than the $H-H_{T_{max}}$ specimens.

As shown in Figure 4.55, the wear behavior in the rubber wheel abrasive wear test, which represents three-body abrasion, was clearly different from that observed in the Suga abrasive wear test. The R_w values of all specimens slightly decreased with increasing Nb content up to 1.05%Nb and then increased in the specimen with higher carbon and Nb contents. Compared with the Nb-free specimen, this reduction in R_w at 1.05%Nb corresponded to an improvement in wear resistance of approximately 19.8% and 12.7% in the $H_{T_{max}}$ specimens austenitized at 1050°C and 1100°C, respectively. Regardless of austenitizing temperature, the decrease in R_w up to 1.05%Nb can be attributed to the morphology and distribution of carbides crystallized during solidification. As described in Section 4.1, the MC_p carbides in specimen No.3 showed a granular morphology and were smaller than those in specimen No.4. The smaller and more uniformly distributed MC_p carbides were considered to be more effective in improving the resistance to three-body abrasion.

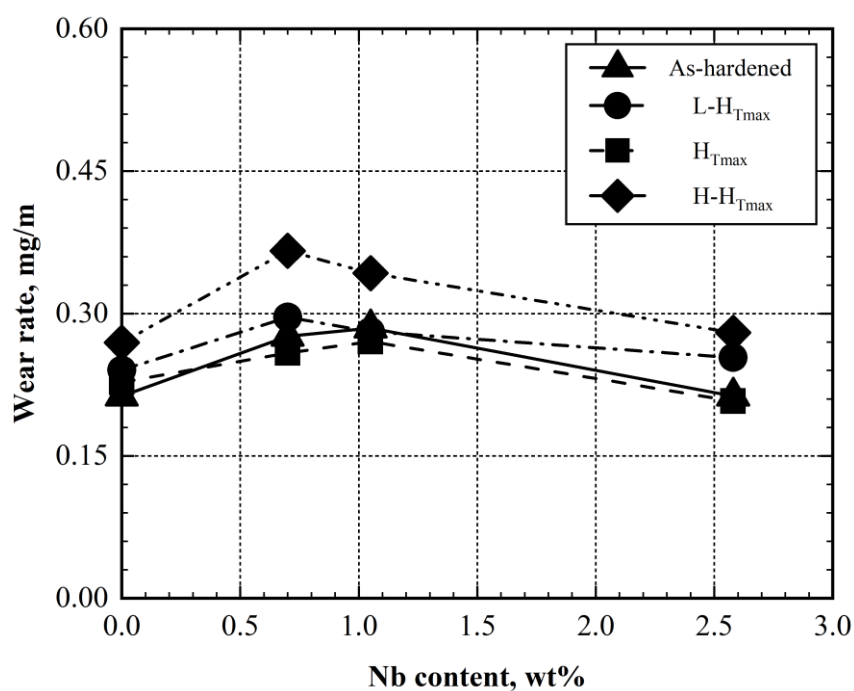
In contrast, specimen No.4 contained a larger amount of coarser MC_p carbides. Although this specimen exhibited higher matrix micro-hardness than specimen No.3, its R_w value in the rubber wheel test was higher. This indicates that the wear resistance in the rubber wheel abrasive wear test was not governed solely by matrix hardness. The intrinsic hardness advantage of the eutectic MC carbide only

translates into improved wear resistance when the surrounding matrix can retain the carbide in place and coarser carbides require proportionally greater matrix support to resist detachment. In three-body abrasion, SiO_2 particles move freely between the specimen and the rubber wheel and the matrix is removed. The carbides lose mechanical support and may become susceptible to cracking, detachment or possible carbide pull-out (de Mello & Polycarpou, 2010). A similar dependence of three-body abrasion resistance on carbide size and distribution rather than on hardness alone has been reported by Pourasiabi and Gates for NbC-bearing high-chromium white cast iron under high-stress abrasion, showing that coarser or excessive carbide volume fraction can offset the wear resistance benefit expected from higher hardness (Pourasiabi & Gates, 2022a). Therefore, the lower R_w value of the 1.05%Nb specimen compared with the harder 2.58%Nb specimen is likely due to a better balance between carbide size, carbide distribution and matrix support under three-body abrasive wear. The same trend in austenitizing temperature and heat treatment condition described for the Suga test were also observed in the rubber wheel abrasive wear test, with hardening from 1100°C and the As-hardened and $H_{T_{\max}}$ conditions consistently giving lower R_w values.

When the Suga and rubber wheel abrasive wear tests are compared, the R_w values obtained in the rubber wheel test are evidently much lower across all specimens and heat treatment conditions, consistent with the substantially higher S_{WR} measured for the Suga test in Section 4.3.3. This difference reflects the fundamentally different abrasive contact mechanism between the two tests, particularly the degree of constraint of the abrasive particles and the resulting contact stress, as discussed in detail in Section 4.3.3.

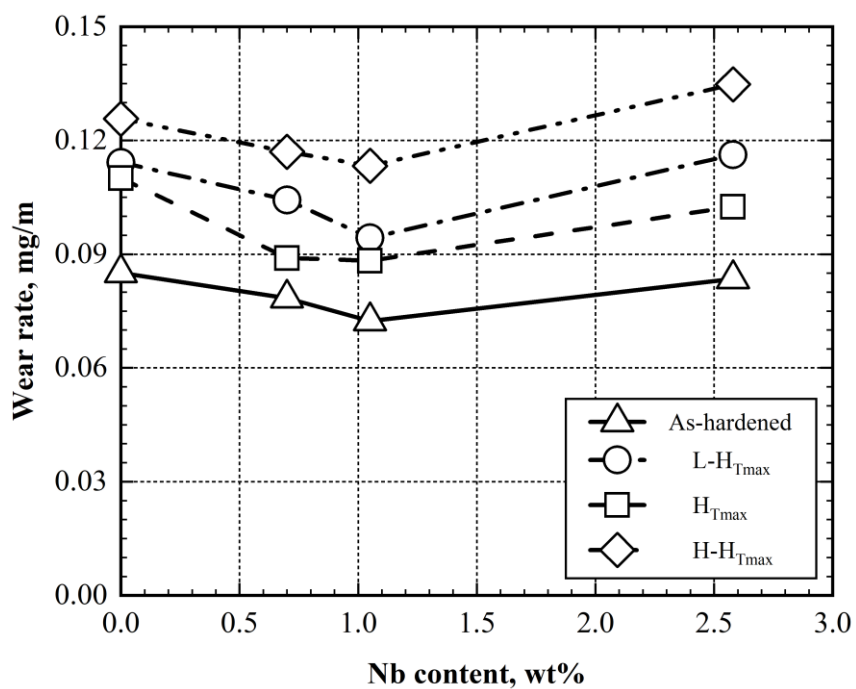


(5) 1050°C Austenitizing.

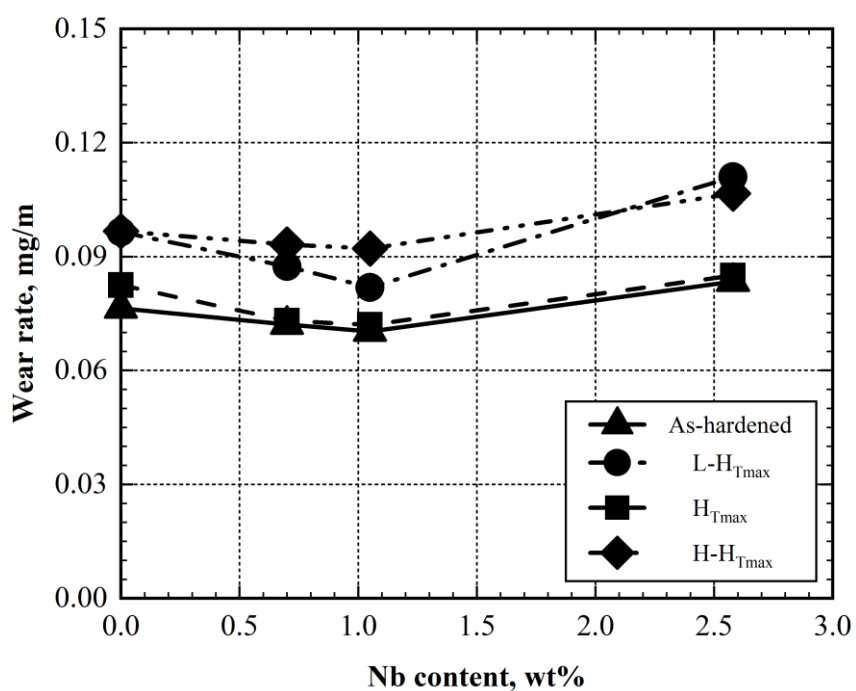


(b) 1100°C Austenitizing.

Figure 4.54 Relationship between wear rate (R_w) and Nb content of specimens evaluated by the Suga abrasive wear test under a load of 9.8 N (1 kgf); (a) 1050°C Austenitizing and (b) 1100°C Austenitizing.



(5) 1050°C Austenitizing.



(b) 1100°C Austenitizing.

Figure 4.55 Relationship between wear rate (R_w) and Nb content of specimens evaluated by the rubber wheel abrasive wear test under a load of 85.3 N (8.7 kgf); (a) 1050°C Austenitizing and (b) 1100°C Austenitizing.

4.3.5 Effect of hardness on wear rate

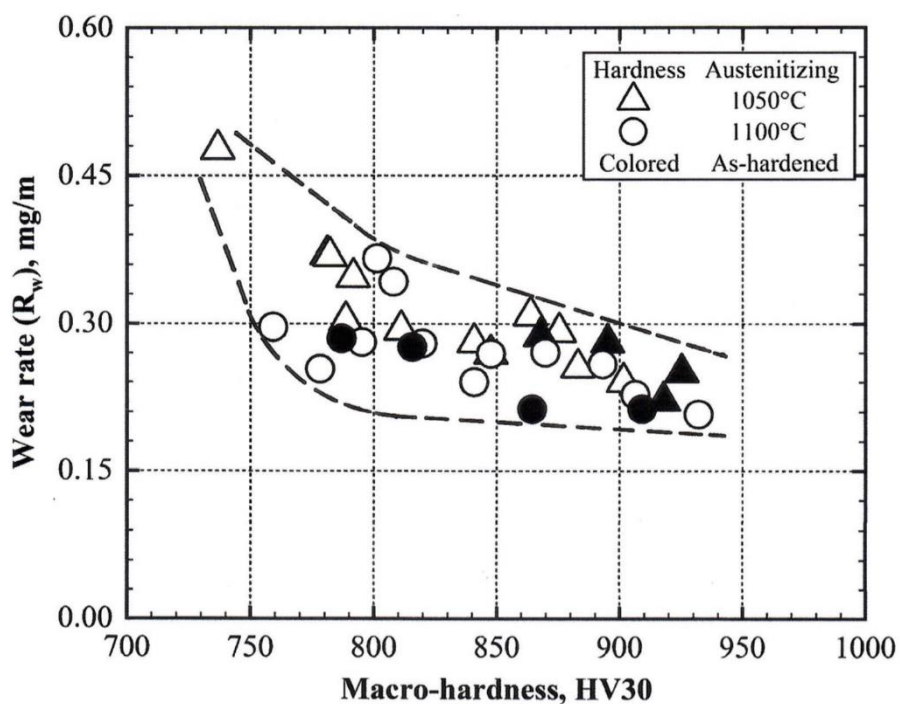
In general, the hardness of specimens is one of the main factors controlling wear resistance. Specimens with higher hardness usually exhibit better wear resistance than those with lower hardness. However, microstructure is also an important factor. The type, morphology and size of carbides crystallized during solidification, as well as the matrix structure that supports these carbides all play important roles in determining the hardness and wear resistance of the material. Figure 4.56(a) and Figure 4.56(b) show the relationship between R_w and macro-hardness and micro-hardness, respectively, for the Suga abrasive wear test. The corresponding relationships for the rubber wheel abrasive wear test are presented in Figure 4.57(a) and Figure 4.57(b). The results indicate that higher hardness values were generally obtained in the As-hardened and H_{Tmax} specimens, whereas lower hardness values were mostly observed in the L- H_{Tmax} and H- H_{Tmax} specimens.

When the relationship between hardness and R_w in both abrasive wear tests is considered, similar trends can be observed. The observed trend can be separated into two distinct regions. The first region corresponds to hardness ranges of approximately 730 to 800 HV30 and 700 to 750 HV0.1, in which R_w decreased rapidly with increasing hardness. Most of the specimens within this hardness region correspond to the L- H_{Tmax} and H- H_{Tmax} conditions, representing tempering states below and above the optimum tempering temperature, respectively. Nevertheless, as shown in Figure 4.56 and 4.57, specimens within this region do not necessarily share the same matrix condition or the same R_w . In the over-tempered condition, the retained austenite has already been consumed, but the secondary carbides have coarsened and become poorly distributed, weakening the matrix. In the under-tempered condition, the secondary carbides remain fine, but a substantial fraction of retained austenite has not yet transformed, leaving the matrix comparatively soft. These represent two distinct microstructural states that arrive at similar overall hardness through different routes, one governed by carbide coarsening and the other by incomplete austenite transformation. Between these two heat treatment conditions, Figure 4.56 and 4.57 both show that the under-tempered condition consistently gave a lower R_w than the over-tempered condition in both Suga and rubber wheel abrasive wear test, indicating that the loss of carbide dispersion through coarsening is more detrimental to wear resistance than the presence of untransformed austenite within this hardness region.

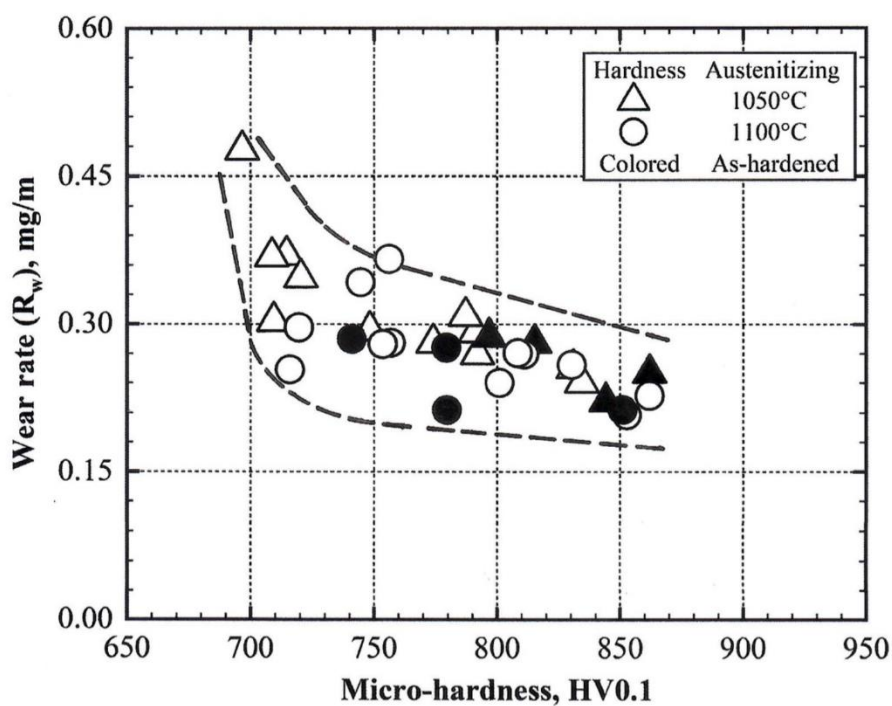
The second region corresponds to hardness values higher than approximately 800 HV30 and 750 HV0.1. In this region, R_w continued to decrease with increasing hardness, but the reduction became more gradual and this tendency was observed regardless of the abrasive wear test method, suggesting that hardness alone may not be the dominant factor controlling wear behavior in the high-hardness range. In this range, the data can be interpreted by considering two heat treatment conditions of matrix structure. The H_{Tmax} specimens, combining high hardness with a low V_γ , have matrix mainly composed of martensite and nearly pure secondary carbides. Although this structure provides high resistance to abrasive penetration, the low V_γ may reduce the toughness and crack-blunting ability of the matrix. Consequently, microcracks generated during abrasion can propagate more easily through the brittle martensitic matrix and lead to matrix pitting, so the improvement in wear resistance becomes limited despite the increased hardness (Sare, 1979). In contrast, the As-hardened specimens combine high hardness with a relatively high V_γ . Retained austenite can improve wear resistance by increasing matrix toughness and undergoing strain-induced transformation during abrasion. However, because the retained austenite in the As-hardened specimen is highly stabilized by carbon and alloying elements, its ability to transform into martensite during abrasion becomes limited, so its contribution to strain-induced hardening is reduced. In addition, the transformation of retained austenite to martensite during abrasive wear may induce local volume expansion and internal stress, which can promote matrix spalling. As a result, R_w does not decrease markedly in the As-hardened specimens even though the hardness remains high (Opapailboon et al., 2019b). Taken together, the H_{Tmax} and As-hardened conditions represent two different routes to high hardness within this region. The H_{Tmax} condition is limited by the reduced toughness of a matrix containing insufficient retained austenite, while the As-hardened condition is limited by an excessive amount of retained austenite in the matrix. This explains the more gradual decrease in R_w observed in this region.

Considering the austenitizing temperature, the specimens hardened from the higher austenitizing temperature exhibited lower R_w values than those hardened from the lower austenitizing temperature. This can be attributed to the greater dissolution of carbon and alloying elements into the matrix during austenitizing. These dissolved elements promote secondary carbide precipitation during subsequent tempering, leading to secondary hardening and improved mechanical support for

eutectic carbides. As a result, the resistance to abrasive wear is enhanced. Nevertheless, the effect of retained austenite should also be considered, since it is also increased by higher austenitizing temperature.

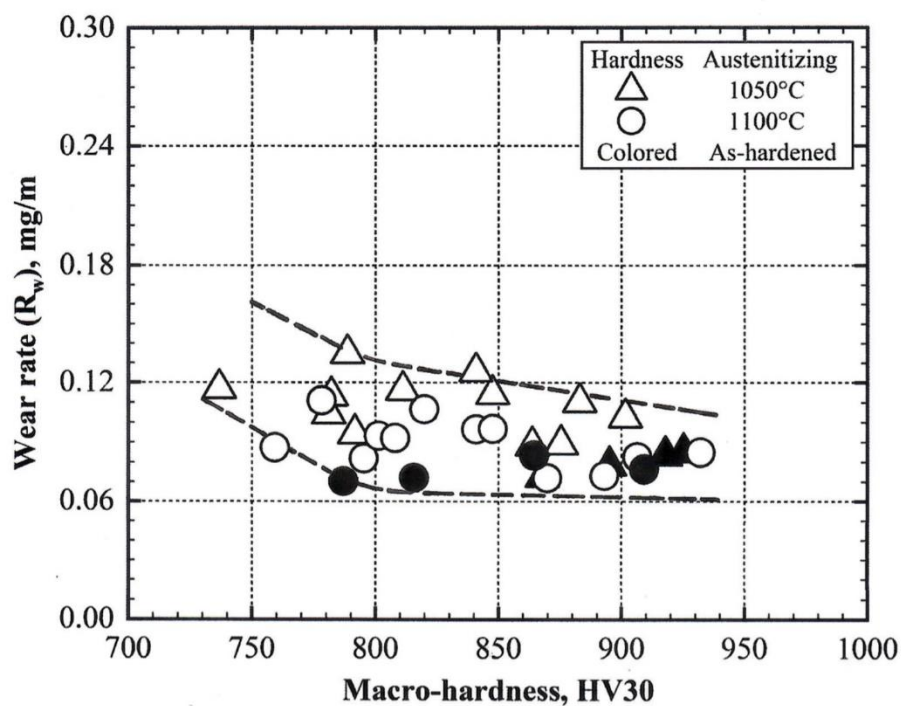


(5) Macro-hardness

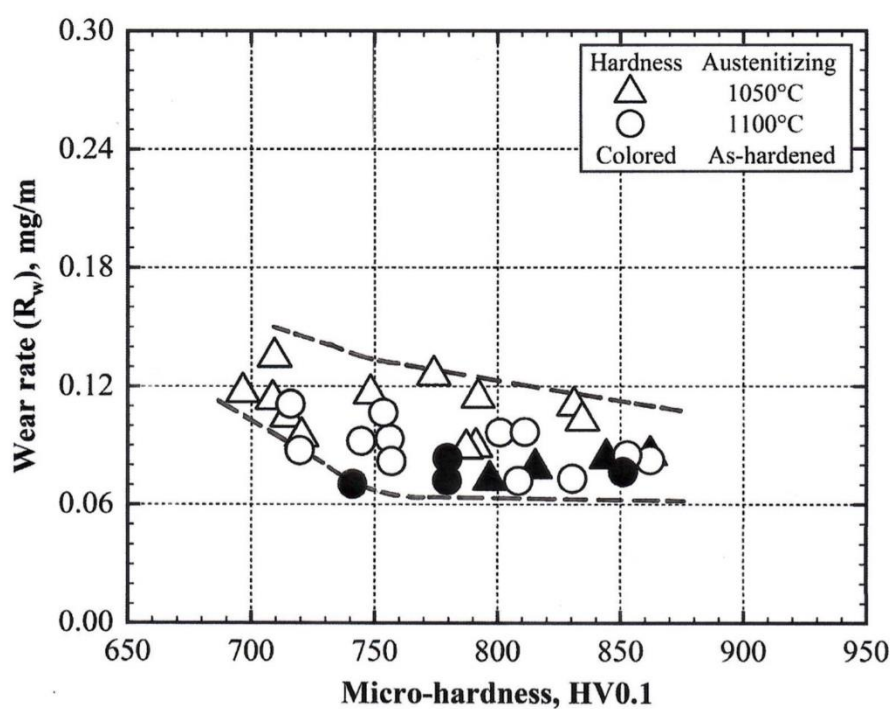


(b) Micro-hardness

Figure 4.56 Relationship between wear rate (R_w) and hardness of specimens evaluated by the Suga abrasive wear test under a load of 9.8 N (1 kgf); (a) Macro-hardness and (b) Micro-hardness.



(5) Macro-hardness



(b) Micro-hardness

Figure 4.57 Relationship between wear rate (R_w) and hardness of specimens evaluated by the rubber wheel abrasive wear test under a load of 85.3 N (8.7 kgf); (a) Macro-hardness and (b) Micro-hardness.

4.3.6 Relationship between wear rate and volume fraction of retained austenite (V_γ) in heat-treated state

As discussed in Section 4.3.5, the wear resistance of the specimens was not governed only by hardness. The characteristics of the matrix structure also played an important role in determining wear behavior. Since austenite is a phase with high toughness and can undergo strain-induced hardening during abrasion, the amount of retained austenite remaining in the matrix is also expected to influence the R_w . Therefore, the effect of retained austenite on the wear rate of the specimens is discussed in this section. The correlation between R_w and the V_γ is shown in Figure 4.58 for the Suga abrasive wear test and Figure 4.59 for the rubber wheel abrasive wear test.

For the Suga and rubber wheel abrasive wear test, the relationship between R_w and V_γ can be divided into two regions at approximately 8% V_γ , as shown in Figure 4.58 and 4.59, respectively. In the first region, where V_γ was lower than about 8%, the R_w values were widely scattered against V_γ , indicating that retained austenite was not the dominant factor controlling R_w in this range. This region was composed predominantly of H-H_{Tmax} specimens, together with most H_{Tmax} specimens. To clarify the factor actually governing R_w in this low V_γ region, the same specimens were re-plotted against micro-hardness in Figure 4.60 (Suga test) and Figure 4.61 (rubber wheel test). Unlike the scattered relationships with V_γ , R_w correlated much more clearly with micro-hardness in both figures, confirming that when V_γ is low, wear resistance is governed primarily by matrix hardness and the ability of the matrix to support hard carbides, consistent with the micro-hardness relationship shown in Figure 4.56(b) and Figure 4.57(b). This interpretation is in good agreement with previous reports that abrasive wear resistance is mainly controlled by matrix hardness when retained austenite is low.

In the second region, where V_γ exceeded approximately 8%, the R_w values tended to decrease with increasing V_γ , but the rate of decrease slowed progressively and approached a steady state at high V_γ levels. This tendency suggests that retained austenite contributed to improving wear resistance to some extent. During abrasion, retained austenite can absorb deformation energy and may transform into martensite by strain-induced transformation. This transformation increases the local hardness of the worn surface and helps reduce material removal. Doğan et al.

also reported that austenitic matrices can improve abrasion resistance because of their ductility, strain-hardening ability and metastability during wear (Doğan et al., 1997). However, this beneficial effect is expected only when retained austenite is present in an appropriate amount and has sufficient metastability. When the amount of retained austenite is excessive or when it is highly stabilized by carbon and alloying elements, particularly in the As-hardened specimens, it may not transform sufficiently into martensite during abrasion. In this case, the matrix may remain relatively soft and the mechanical support for eutectic carbides becomes less effective, resulting in easier matrix removal and carbide detachment. In addition, local transformation of retained austenite into martensite during abrasion may be accompanied by volume expansion and internal stress, which often results in spalling on the worn surface, while microcracking and pitting tend to occur within the surrounding martensite. Opapairoon et al. (2025b) reported that in the Suga wear test, S_{WR} decreased up to intermediate retained austenite level but increased again when retained austenite became excessive, whereas in the rubber wheel test the effect of retained austenite was less significant because of the lower wear stress. Although the present results showed a steady rather than a clear increase in R_w at the highest V_γ levels, a similar saturation of the beneficial effect of retained austenite may explain this tendency toward a steady value. This implies that retained austenite has both beneficial and detrimental effects and its effect must be balanced with matrix hardness, carbide distribution and the stability of retained austenite.

Taken together, both regions show that the rubber wheel test is consistently less sensitive to matrix-related factors than the Suga test, as shown by comparing Figure 4.58 and Figure 4.59. This includes hardness in the first region and retained austenite in the second. In the low V_γ region, R_w correlated more weakly with hardness in the rubber wheel test than in the Suga test, a trend also evident between Figure 4.60 and Figure 4.61 (Inthidech et al., 2021a). In the high V_γ region, the beneficial effect of retained austenite through strain-induced transformation was likewise less pronounced, with the lowest R_w obtained in a specimen with only moderate V_γ rather than the highest. This consistent pattern reflects the lower effective contact stress in the rubber wheel test under which carbide size and distribution play a comparatively larger role than matrix hardness or retained austenite content in governing wear resistance.

To further clarify the role of retained austenite, R_w was analyzed separately against V_γ for the As-hardened and H_{Tmax} conditions (Figure 4.62-4.65). In the As-hardened condition, R_w decreased with increasing $V_{\gamma-AsH}$ at lower V_γ levels but became steady once $V_{\gamma-AsH}$ exceeded approximately 20%, in both the Suga and rubber wheel tests. This indicates that the beneficial effect of retained austenite on wear resistance was limited to the lower V_γ range. Once the matrix already contained a sufficiently large and highly stabilized volume fraction of untransformed austenite, further increases in V_γ had little additional effect on wear resistance. This is consistent with limited strain-induced transformation in the As-hardened matrix at high V_γ , where the retained austenite becomes too stable to respond further to abrasion, so that carbide characteristics and local matrix support account for the remaining variation in R_w once this steady level is reached.

Although the As-hardened specimens generally exhibited low R_w , their high retained austenite content makes this condition less suitable for practical service, since strain-induced transformation, when it does occur, can generate volume expansion and internal stress that promote microcracking, pitting, interfacial damage, spalling and dimensional instability. The H_{Tmax} condition is therefore preferable, offering a more stable matrix of tempered martensite, fine secondary carbides and a smaller, better-controlled amount of retained austenite. In the H_{Tmax} condition, R_w decreased clearly with increasing $V_{\gamma-HTmax}$ in the Suga test (Figure 4.64), reflecting a favorable combination of high hardness and improved toughness as retained austenite increased from a low starting level. In the rubber wheel test (Figure 4.65), R_w decreased to a minimum at an intermediate $V_{\gamma-HTmax}$ before increasing again at the highest V_γ , consistent with the coarser, less well-supported MC_p carbides in the highest Nb specimen. This confirms that retained austenite is beneficial only when supported by fine, well-distributed carbides. Overall, these results show that the influence of V_γ on R_w depends strongly on how much retained austenite is already present and on the heat treatment condition. In the As-hardened condition, the benefit of V_γ saturates beyond approximately 20%. In the H_{Tmax} condition, where V_γ remains low to moderate, retained austenite contributes more clearly to wear resistance, provided it is supported by an appropriate carbide structure. This explains why the As-hardened and H_{Tmax} specimens generally outperformed the L- H_{Tmax} and H- H_{Tmax} specimens, but also why

R_w within the As-hardened group itself could not be predicted from V_γ alone once V_γ exceeded 20%.

Overall, the abrasive wear resistance of the specimens was not governed by a single factor but by the combined effect of Nb addition, heat treatment condition, matrix hardness, retained austenite and abrasive wear mode. Nb addition improved wear resistance through the formation of hard MC carbide. However, excessive or coarse MC_p carbides could reduce matrix support and promote carbide cracking or pull-out. Matrix hardness was an important factor reducing R_w , especially in the low retained austenite region, but hardness alone could not fully explain the wear behavior. Retained austenite also contributed to wear resistance by improving matrix toughness and possibly promoting strain-induced hardening, provided that it remained in an appropriate amount and had suitable metastability. Among the heat-treated conditions, the As-hardened and H_{Tmax} specimens generally showed lower R_w than the L- H_{Tmax} and H- H_{Tmax} specimens. In particular, the H_{Tmax} condition provided a favorable matrix structure consisting of tempered martensite, fine secondary carbides and a moderate amount of retained austenite. Therefore, the best wear resistance was achieved by an optimum matrix-carbide structure rather than maximizing Nb content, hardness or retained austenite alone.

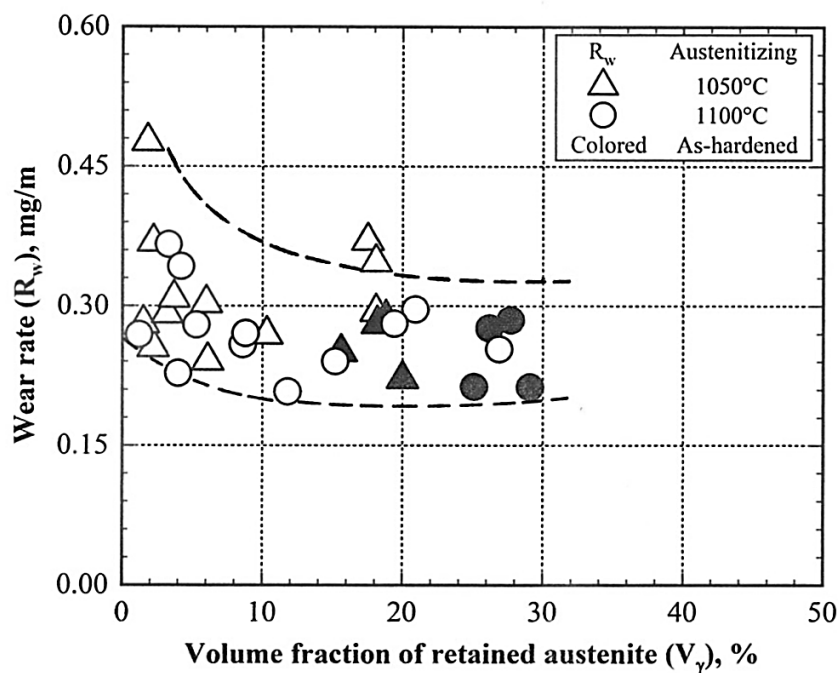


Figure 4.58 Relationship between wear rate (R_w) and volume fraction of retained austenite (V_γ) of all the specimens determined by the Suga abrasive wear test under a load of 9.8 N (1 kgf).

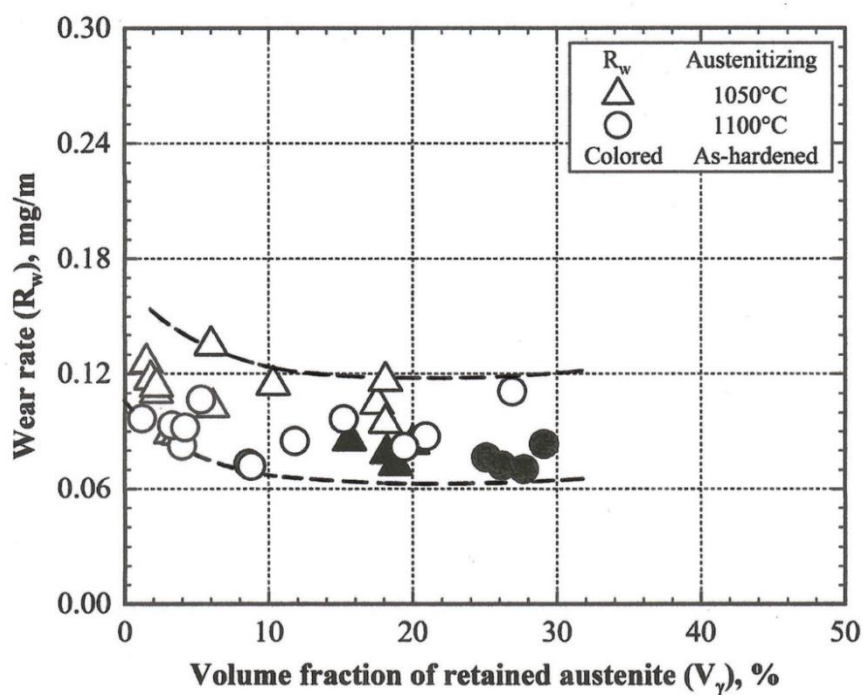


Figure 4.59 Relationship between wear rate (R_w) and volume fraction of retained austenite (V_γ) of all the specimens determined by the rubber wheel abrasive wear test under a load of 85.3 N (8.7 kgf).

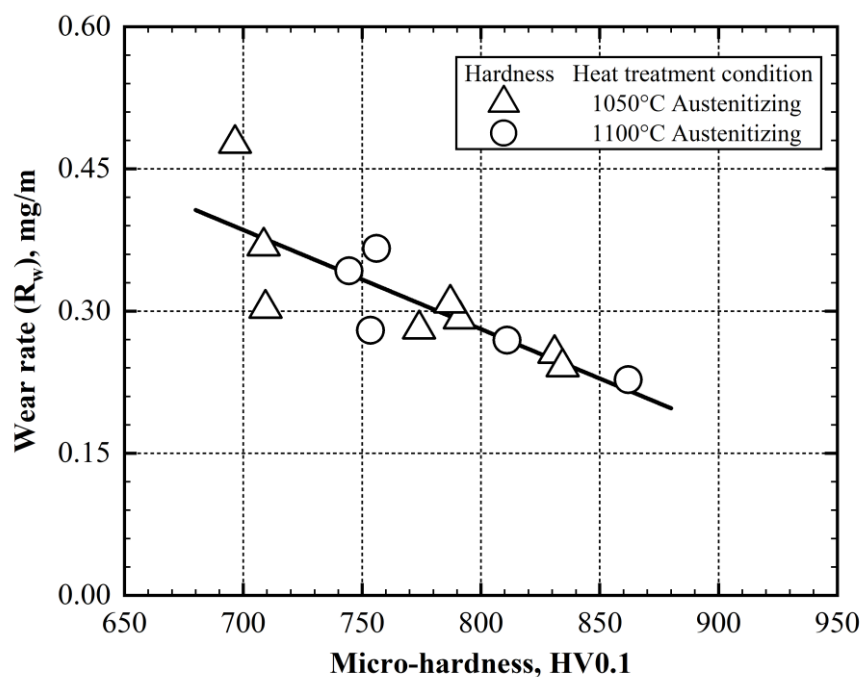


Figure 4.60 Relationship between wear rate (R_w) and micro-hardness for specimens with a volume fraction of retained austenite (V_γ) less than 8% (Suga abrasive wear test).

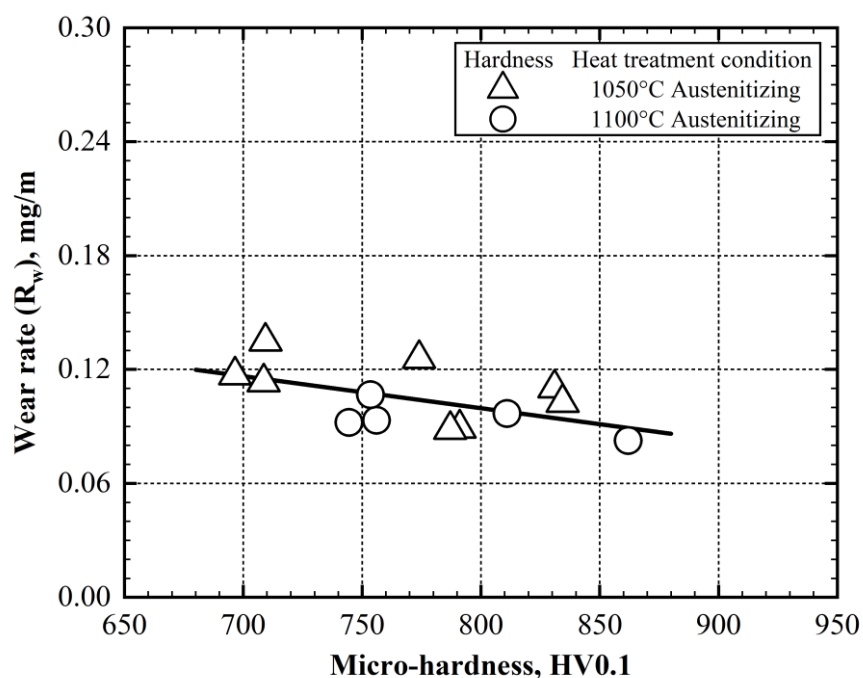


Figure 4.61 Relationship between wear rate (R_w) and micro-hardness for specimens with a volume fraction of retained austenite (V_γ) less than 8% (Rubber wheel abrasive wear test).

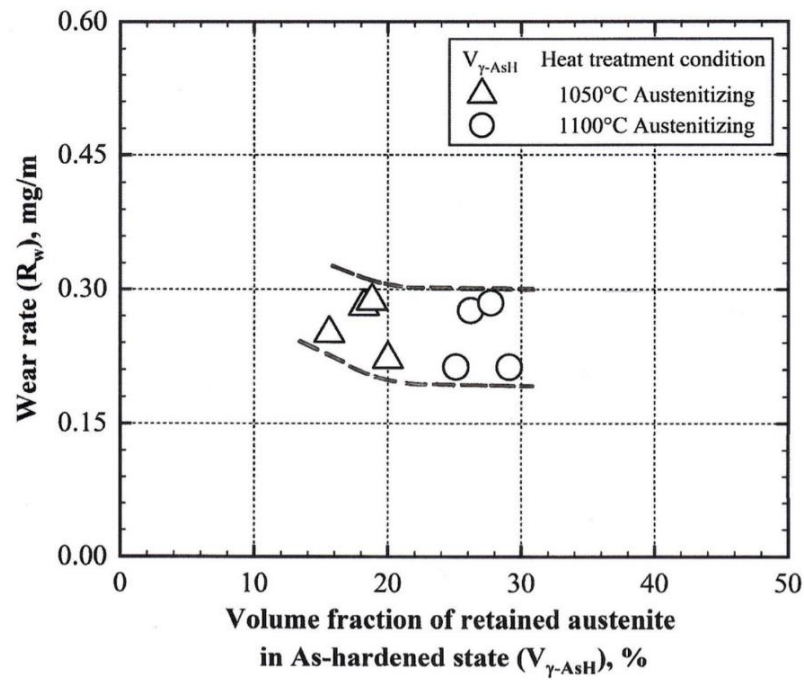


Figure 4.62 Relationship between wear rate (R_w) and volume fraction of retained austenite in as-hardened state ($V_{\gamma-AsH}$) determined by the Suga abrasive wear test.

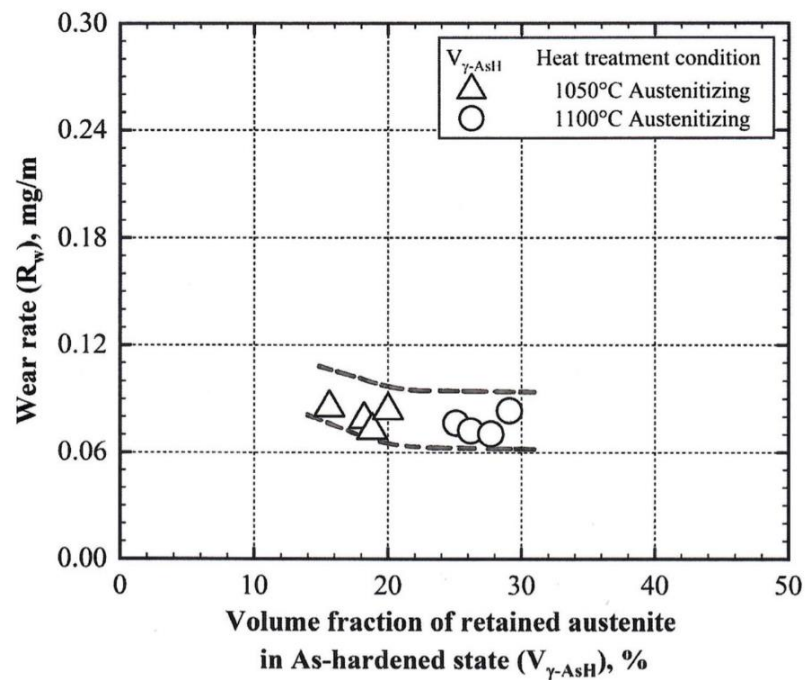


Figure 4.63 Relationship between wear rate (R_w) and volume fraction of retained austenite in as-hardened state ($V_{\gamma-AsH}$) determined by the rubber wheel abrasive wear test.

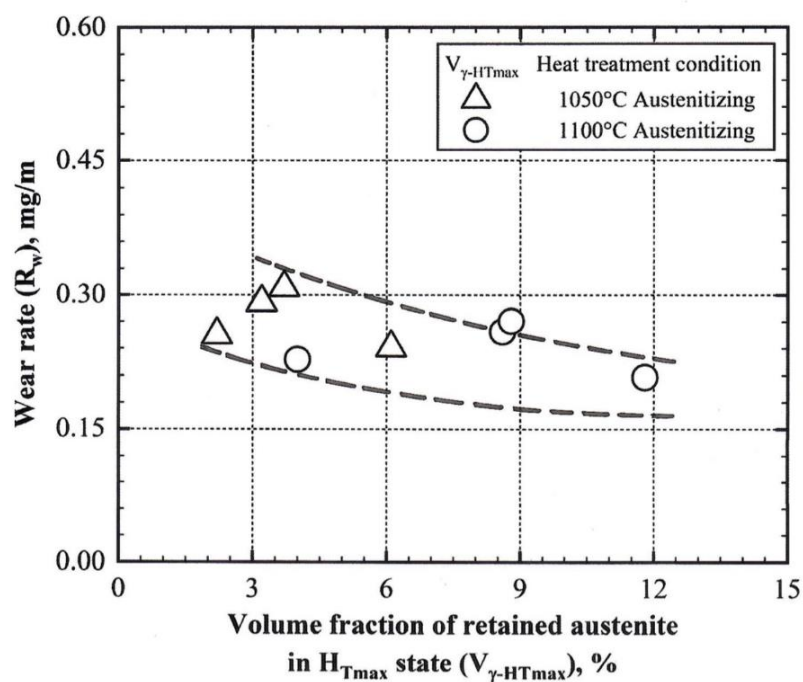


Figure 4.64 Relationship between wear rate (R_w) and volume fraction of retained austenite at H_{Tmax} ($V_{\gamma-HTmax}$) determined by the Suga abrasive wear test.

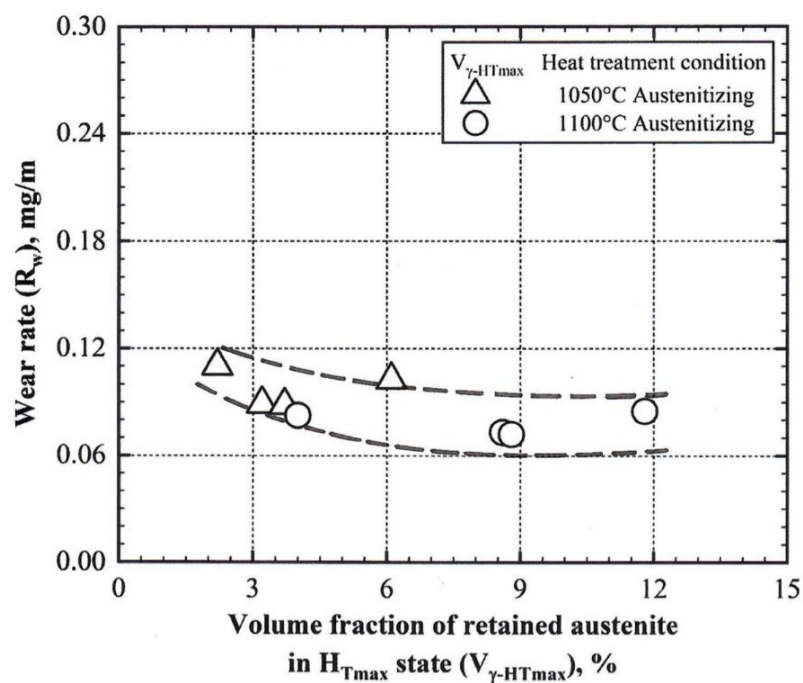


Figure 4.65 Relationship between wear rate (R_w) and volume fraction of retained austenite at H_{Tmax} ($V_{\gamma-HTmax}$) determined by the rubber wheel abrasive wear test.

4.3.7 Observation of worn surface

Scanning electron microscopy (SEM) examination of the worn surfaces of the 2.58%Nb and 1.05%Nb specimens hardened from 1100°C after the Suga abrasive wear test is presented in Figure 4.66 and 4.67, respectively. The corresponding worn surfaces of the 1.05%Nb and 2.58%Nb specimens after the rubber wheel abrasive wear test are presented in Figure 4.68 and 4.69, respectively. The predominant wear mechanisms occurring in both tests include scratching (S), grooving (G), pitting (P) and carbide pull-out (PO). Regarding the Suga abrasive wear test, scratching and grooving tracks were observed parallel to the direction of abrasion. Pitting was predominantly concentrated within the eutectic carbide region. Cracks may initiate and propagate within the eutectic carbides, driven by either localized wear-induced stress concentrations or the impact of abrasive particles that develop into pits, and where matrix support was most severely lost, into fully detached carbide fragments observed as pull-out (PO). Furthermore, the matrix phase exhibited significantly more severe wear than the carbide region due to its lower hardness. In the case of the rubber wheel abrasive wear test, the worn surface exhibits less pronounced wear features overall. Since the SiO₂ particles can move freely and their hardness is higher than that of the matrix phase but lower than the carbide phase, severe damage and wear mostly occurred within the matrix region, which could progressively erode the support and detach poorly supported carbides through pull-out.

Regarding the influence of Nb content on the worn surface, in the Suga abrasive wear test, the specimen exhibiting the lowest wear rate was the 2.58%Nb specimen hardened from 1100°C (as shown in Figure 4.66), compared to the 1.05%Nb specimen in Figure 4.67. This behavior can be attributed to the severity of the Suga test, which is two-body type abrasive wear mode utilizing highly hard silicon carbide particles. Because this condition is considerably more severe than the rubber wheel test, matrix hardness becomes the dominant factor governing wear resistance (Bedolla-Jacuinde, 2016). In contrast, during the rubber wheel abrasive wear test, the specimen with the lowest wear rate was the 1.05%Nb specimen, despite its lower overall hardness. This behavior reflects the three-body abrasive wear mechanism, in which the freely moving SiO₂ particles preferentially attack and erode the softer matrix phase (Inthidech et al., 2021a). The MC_p carbides in the 1.05%Nb specimen were relatively smaller than those in the 2.58%Nb specimen and were uniformly distributed

throughout the matrix, thereby providing more effective protection and resistance against three-body abrasive wear. Furthermore, the worn surface of the 2.58%Nb specimen after the rubber wheel test as shown in Figure 4.69 revealed severe pitting (P) wear within the MC_p carbide regions. This distinctive surface damage strongly correlates with its higher wear rate. As the surrounding matrix is preferentially removed, the large, brittle primary carbides lose essential mechanical support, which accelerates localized stress concentration, micro-fracture and subsequent pull-out (de Mello & Polycarpou, 2010). The cavities left behind by these detached carbide fragments are what appear as pitting (P) in the SEM micrographs of the 2.58%Nb specimen.

To evaluate the influence of heat treatment condition on the worn surface, the As-hardened and $H_{T_{max}}$ specimens which exhibited the least surface damage demonstrated higher wear resistance than the L- $H_{T_{max}}$ and H- $H_{T_{max}}$ specimens. From the microstructural perspective, the matrix of the As-hardened specimen was predominantly composed of as-quenched martensite, secondary carbides and a certain amount of retained austenite which enhanced its wear resistance. Among the tempered specimens, the worn surface roughness increased in the order of $H_{T_{max}}$, L- $H_{T_{max}}$ and H- $H_{T_{max}}$. The minimal damage observed in the $H_{T_{max}}$ specimens can be attributed to a microstructure consisting of tempered martensite and precipitated secondary carbides, together with a significant reduction in retained austenite. In contrast, over-tempering in the H- $H_{T_{max}}$ specimens caused matrix softening and led to severe damage via extensive grooving wear. These morphological observations correlate well with the wear test results presented in Sections 4.3.1 to 4.3.6.

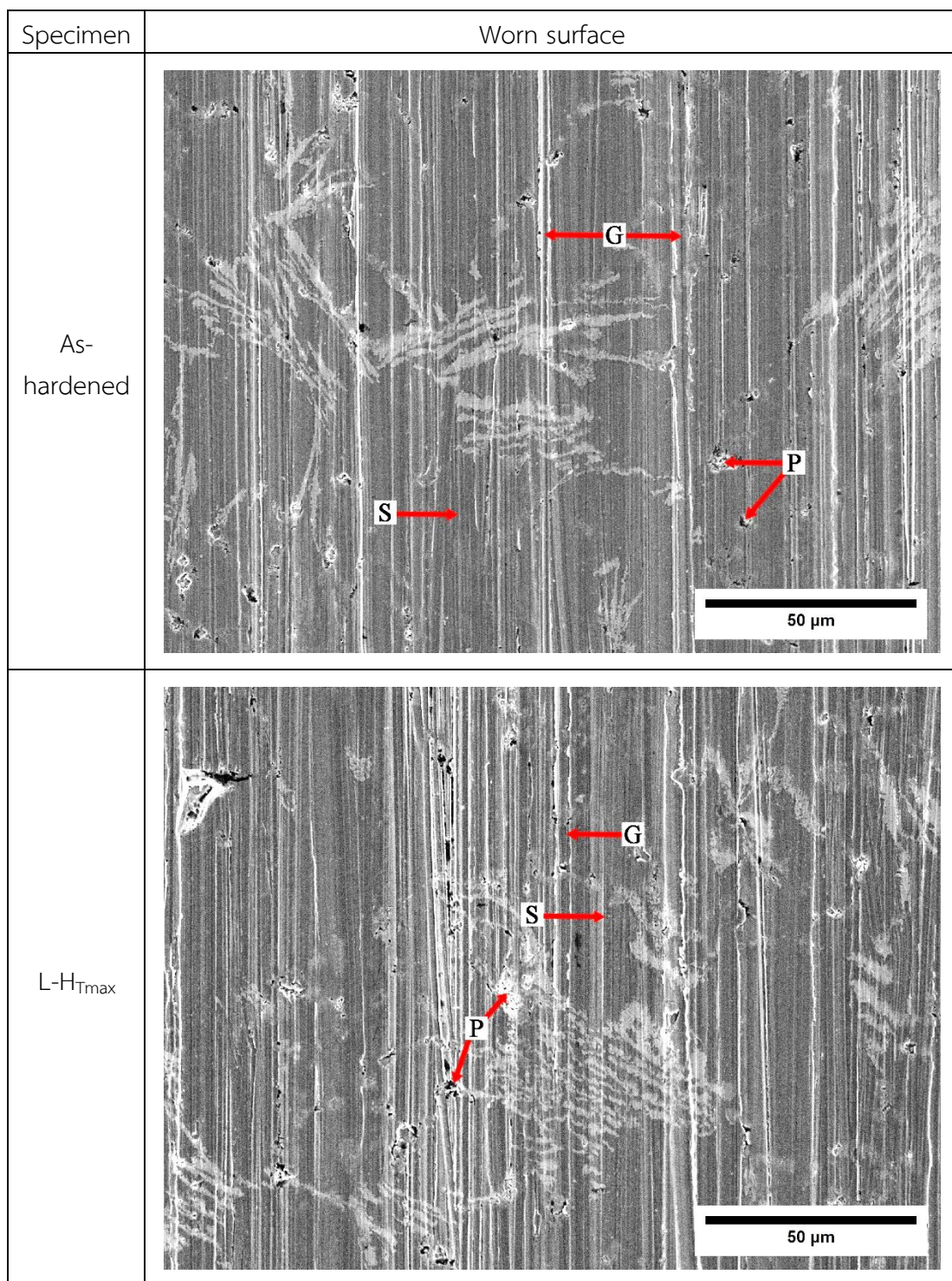


Figure 4.66 SEM microphotographs of worn surface after Suga abrasive wear test in 2.58%Nb specimen hardened from 1100°C austenitizing. (Lowest wear rate)

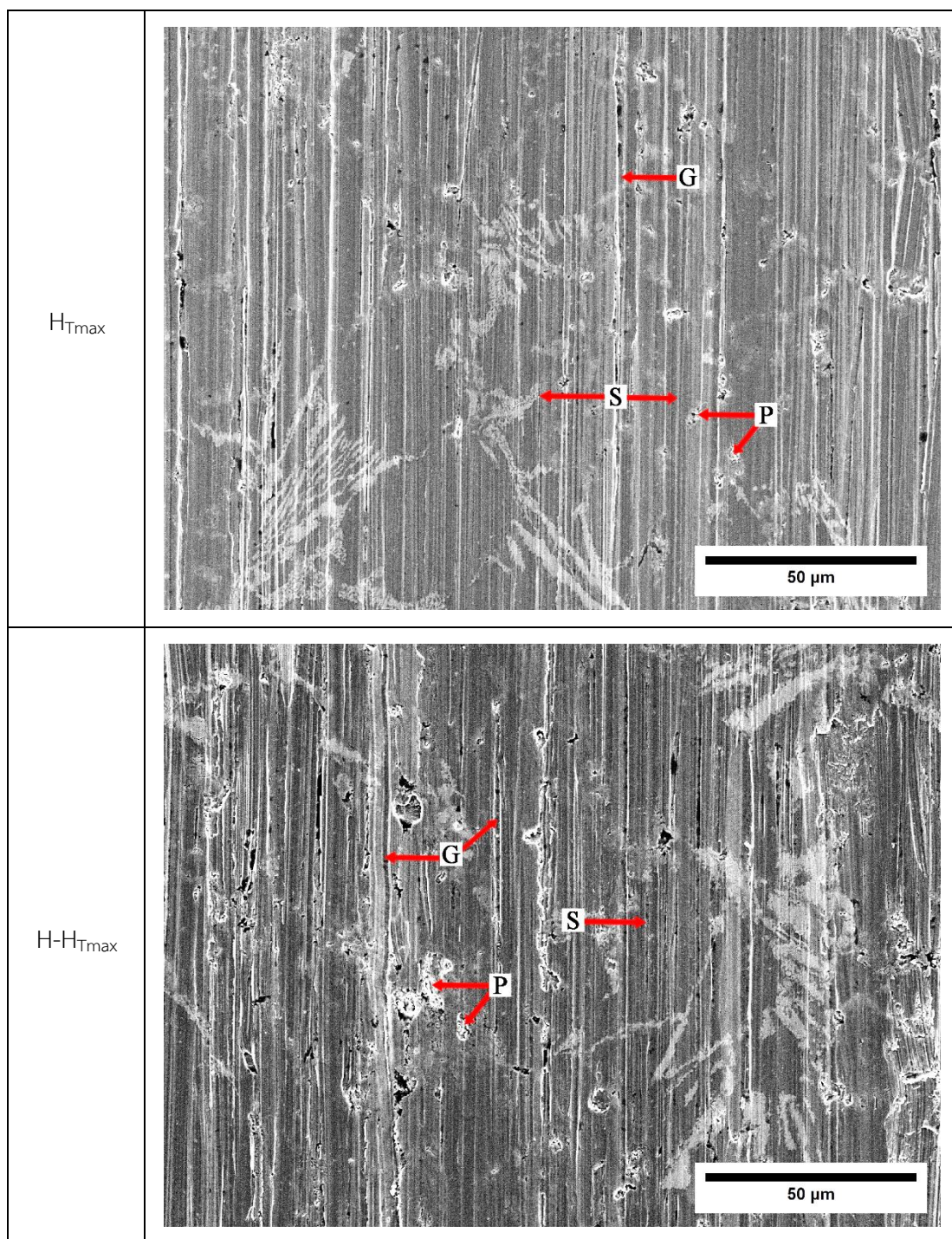


Figure 4.66 SEM microphotographs of worn surface after Suga abrasive wear test in 2.58%Nb specimen hardened from 1100°C austenitizing. (Lowest wear rate)

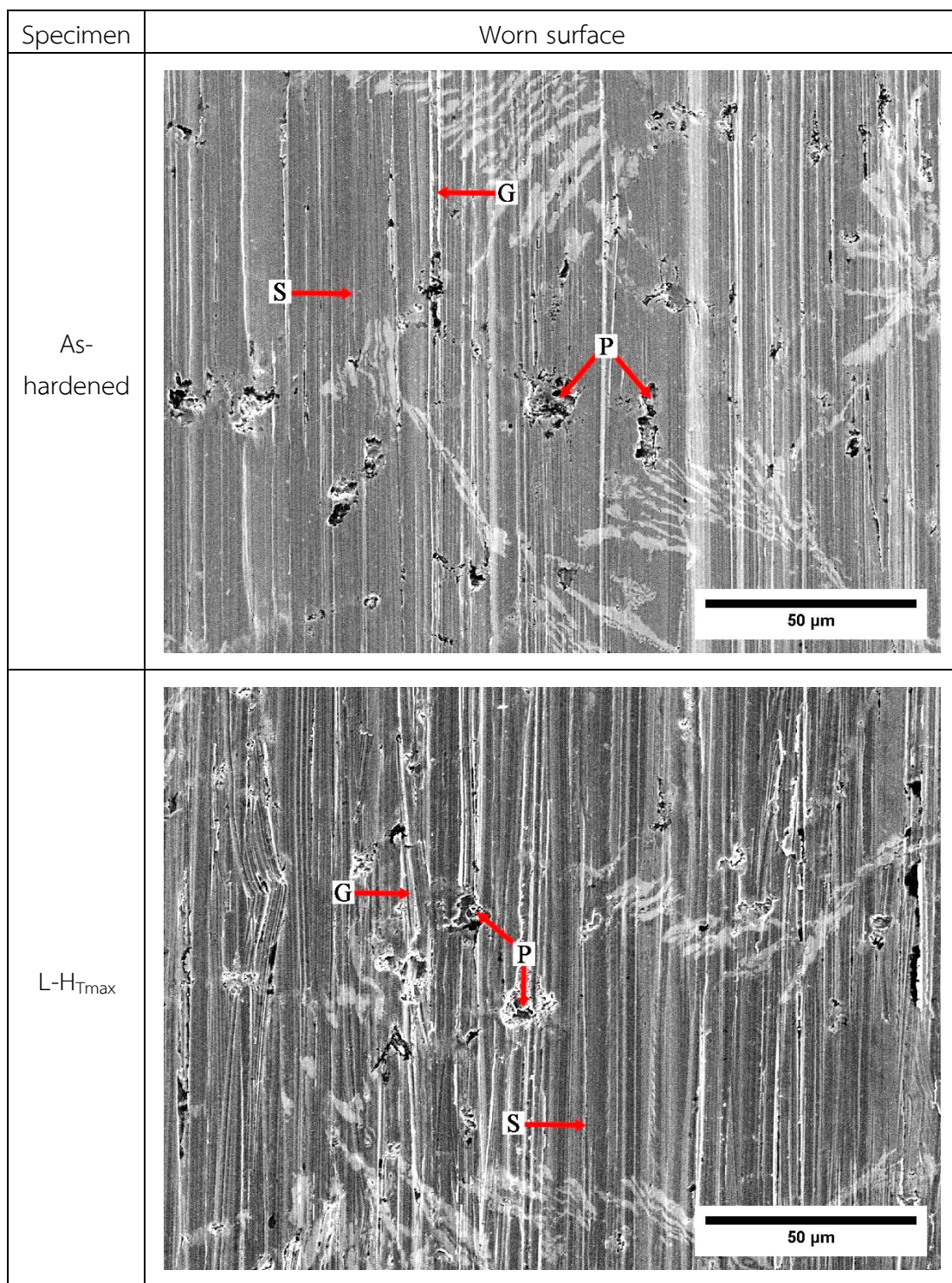


Figure 4.67 SEM microphotographs of worn surface after Suga abrasive wear test in 1.05%Nb specimen hardened from 1100°C austenitizing. (Highest wear rate)

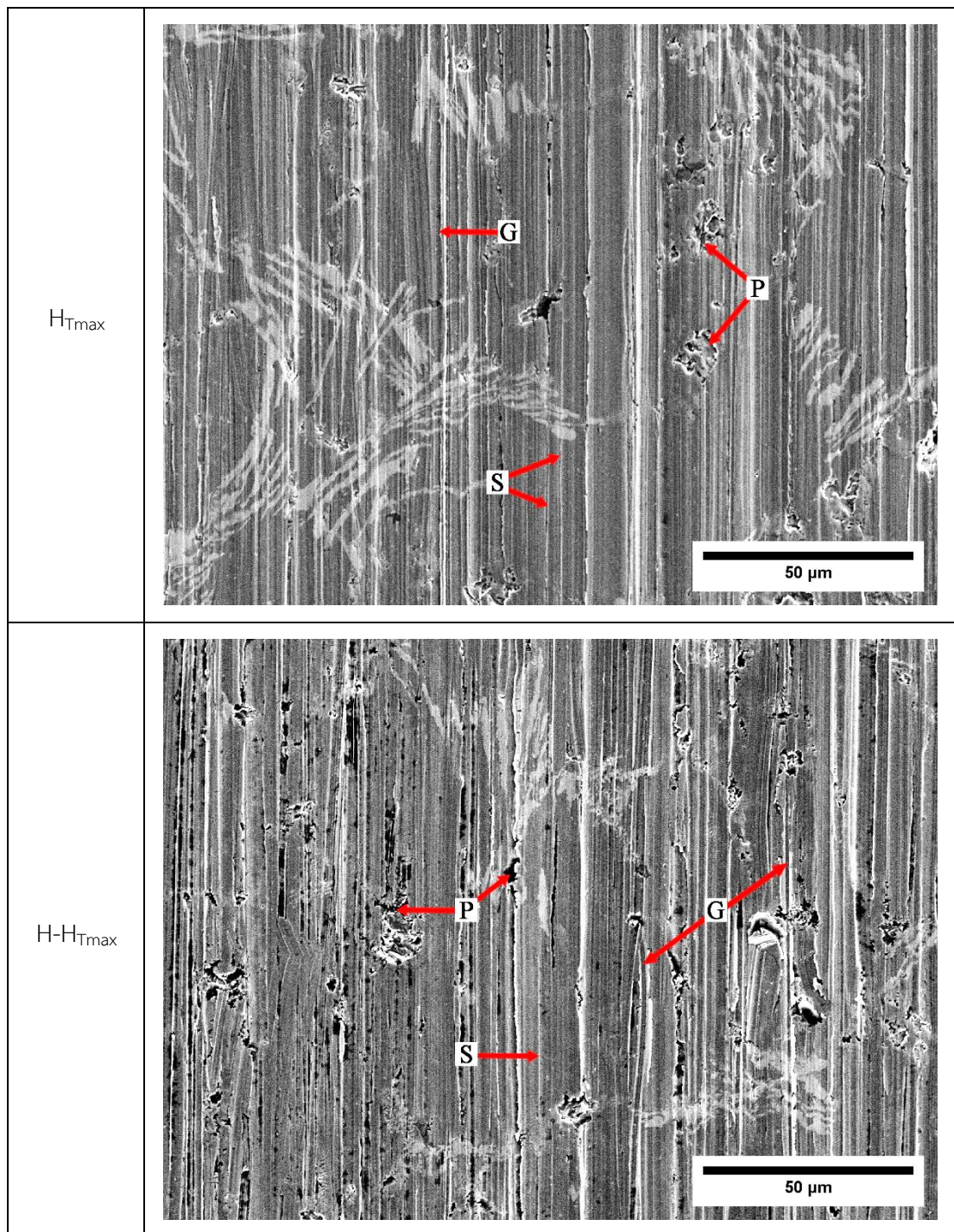


Figure 4.67 SEM microphotographs of worn surface after Suga abrasive wear test in 1.05%Nb specimen hardened from 1100°C austenitizing. (Highest wear rate)

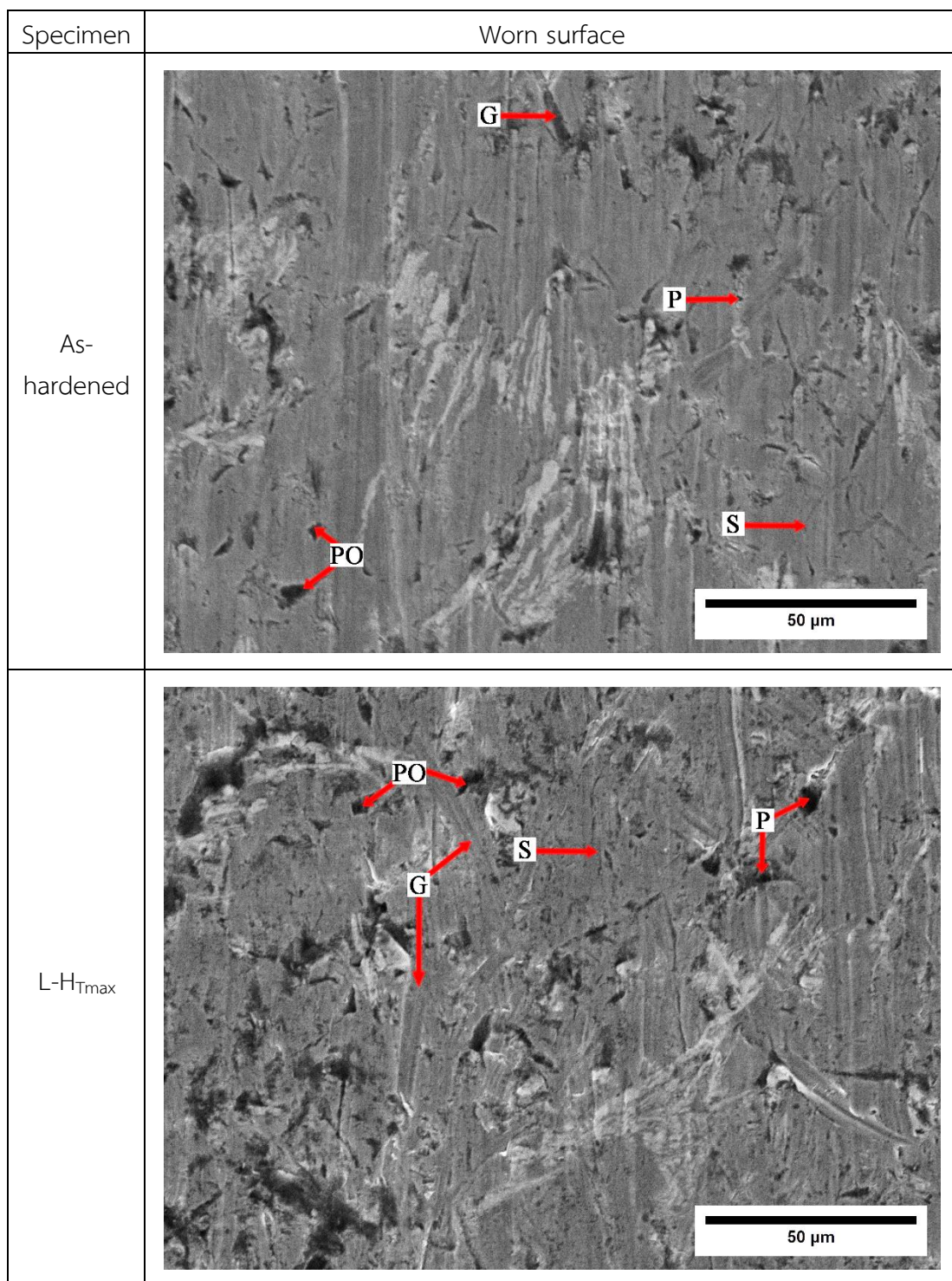


Figure 4.68 SEM microphotographs of worn surface after rubber wheel abrasive wear test in 1.05%Nb specimen hardened from 1100°C austenitizing. (Lowest wear rate)

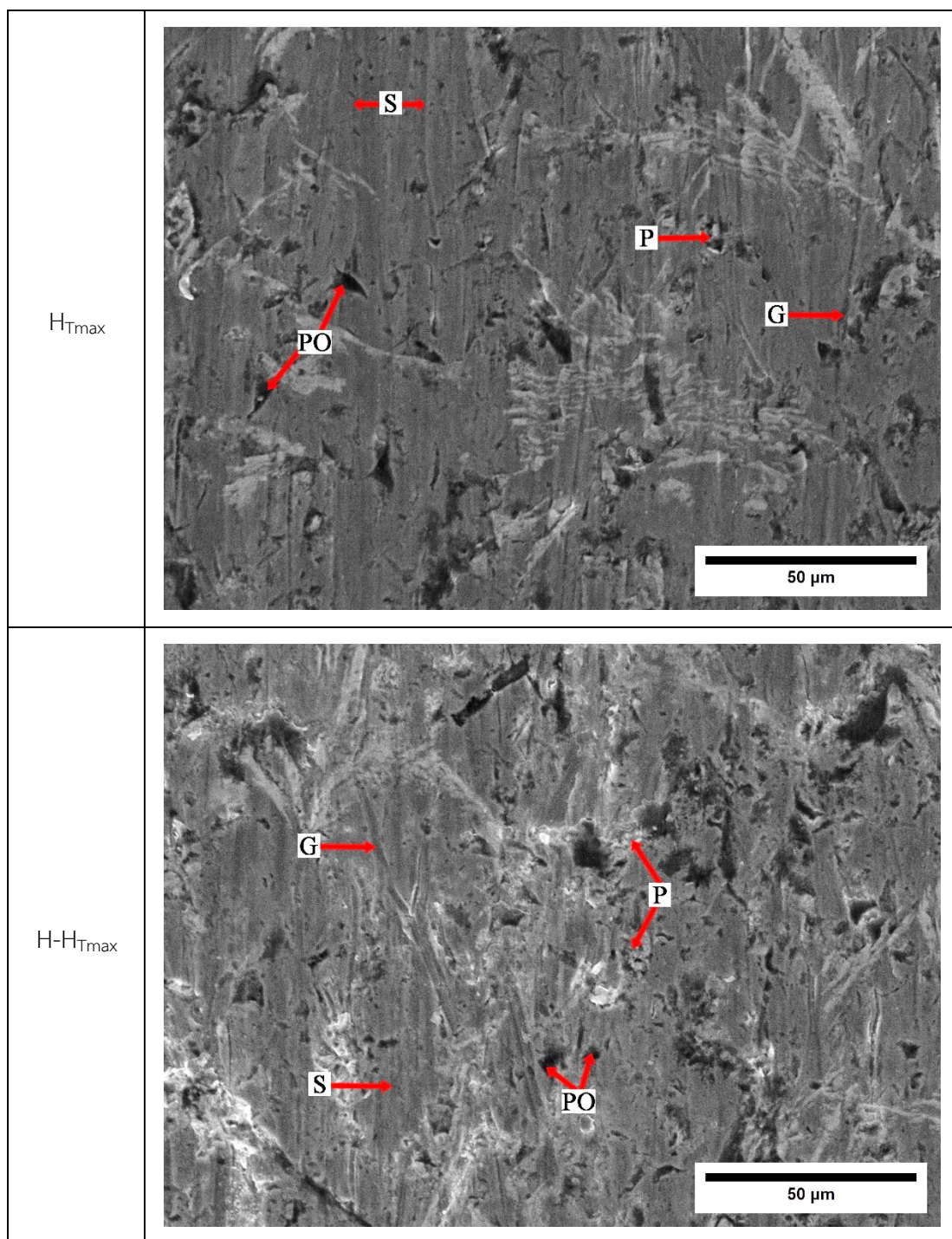


Figure 4.68 SEM microphotographs of worn surface after rubber wheel abrasive wear test in 1.05%Nb specimen hardened from 1100°C austenitizing. (Lowest wear rate)

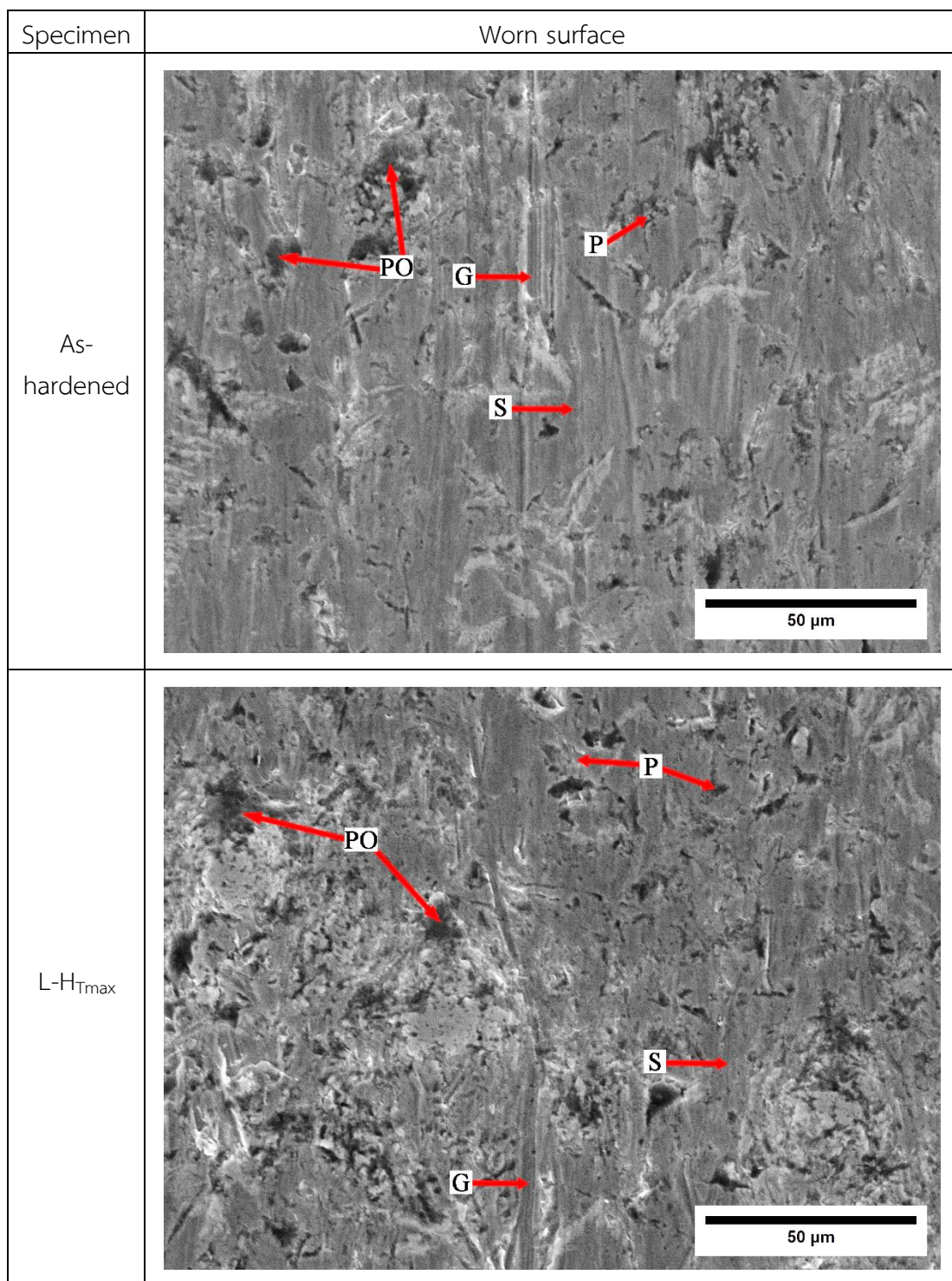


Figure 4.69 SEM microphotographs of worn surface after rubber wheel abrasive wear test in 2.58%Nb specimen hardened from 1100°C austenitizing. (Higher wear rate)

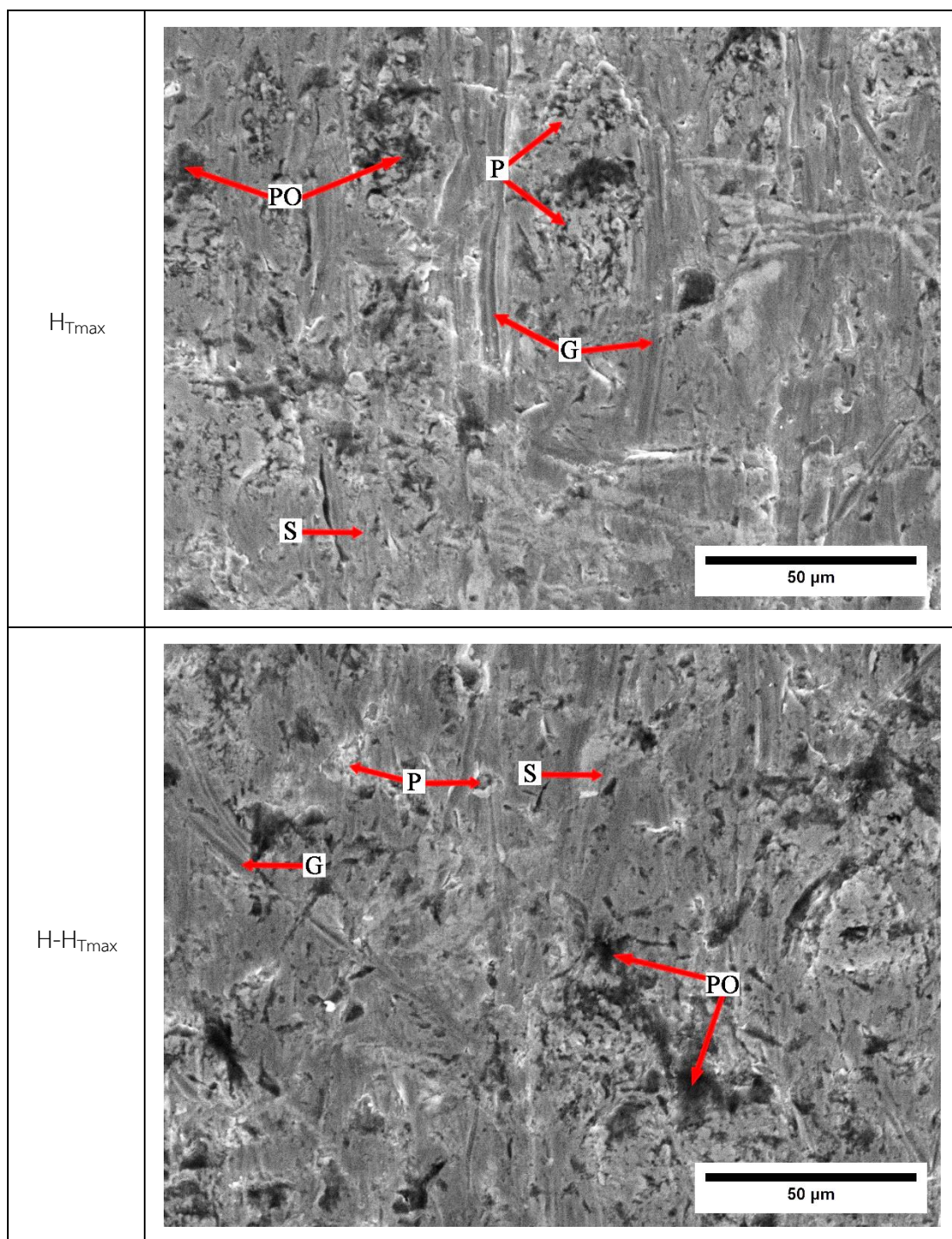


Figure 4.69 SEM microphotographs of worn surface after rubber wheel abrasive wear test in 2.58%Nb specimen hardened from 1100°C austenitizing. (Higher wear rate)