

Composition, Microstructure, Hardness, and Wear Properties of High-Speed Steel Rolls

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The effects of alloying elements on the microstructural factors, hardness, and wear properties of four high-speed steel (HSS) rolls fabricated by centrifugal casting were investigated. A hot-rolling simulation test was carried out using a high-temperature wear tester capable of controlling speed, load, and temperature. The test results revealed that the HSS roll containing a larger amount of vanadium showed the best wear resistance because it contained a number of hard MC-type carbides. However, it showed a very rough roll surface because of cracking along cell boundaries, the preferential removal of the matrix, and the sticking of the rolled material onto the roll surface during the wear process, thereby leading to an increase in the friction coefficient and rolling force. In order to improve wear resistance with consideration to surface roughness, it is suggested that a reduction in the vanadium content, an increase in solid-solution hardening by adding alloying elements, an increase in secondary hardening by precipitation of fine carbides in the matrix, and formation of refined prior austenite grains by preaustenitization treatment be employed to strengthen the matrix, which can hold hard carbides in it.

I. INTRODUCTION

ROLLING conditions are gradually becoming complicated and demanding in hot-rolling plants, because of the increasing volume of thin plates and high-strength steel plates. These conditions require rolls with high hardness and excellent wear properties such as a low friction coefficient and low surface roughness. Accordingly, high-speed steel (HSS) rolls, characterized by enhanced thermal fatigue properties and excellent resistance to wear and abrasion, are replacing conventional high-chromium (Hi-Cr) iron rolls.^[1,2]

The wear process of rolls is a very complex one and is integrally affected by various factors such as abrasive wear, oxidation wear, cracking by thermal fatigue and heat impact, fatigue wear by fall-off of carbides, and sticking of rolled material onto the roll.^[3,4] Wear involves microscopic and dynamic processes occurring in the interface between the roll and the rolled material and is almost impossible to directly observe. Interpretation is also difficult because kinetic, thermal, chemical, and material factors are intricately in interaction. Since the roll surface is repeatedly under stress and heat during hot rolling and in friction with the rolled material, the surface layer is often damaged. This poses serious problems in roll performance and considerably deteriorates the surface quality of rolled products because of the surface roughness of the roll.

In Hi-Cr rolls, a network of carbides formed along solidification cell boundaries work favorably for improving wear resistance. However, because the carbide networks often initiate thermal fatigue cracks and function as crack

propagation sites, they seriously harm thermal fatigue properties.^[5,6,7] These shortcomings are microstructurally alleviated in HSS rolls, where very hard carbides improve hardness and wear resistance while the tempered martensitic matrix, containing fine carbides, is responsible for fracture toughness.^[8] Under this concept, HSS rolls contain strong carbide-forming elements such as V, Mo, and W, which help form very hard carbides that significantly improve hardness and wear resistance.^[9,10,11] The properties of HSS rolls are determined by various microstructural factors such as (1) the type, morphology, volume fraction, and distribution of carbides; (2) the characteristics of the martensitic matrix; and (3) the solidification cell structure, composed of carbides. Carbides directly influence the wear resistance of rolls and the surface quality of rolled plates, since carbides are very hard. The matrix is related to the overall hardness, strength, and fracture toughness of the rolls and plays a sustaining role for hard carbides. The cell size also plays a critical role in determining the overall carbide distribution, directly affecting mechanical properties. Thus, these microstructural factors should be controlled in optimal combinations in order to manufacture HSS rolls with excellent properties.

Recently, it was reported that regions of carbide prominence are formed since the matrix, not as hard as the carbides, is selectively exposed to wear, causing a rough roll surface and increasing rolling force.^[12] Increased rolling force is associated with an increase in the friction coefficient between the roll and the rolled material. This poses problems by limiting the dimensions of rolled products, disturbing the precision control of products, and demanding more electric consumption. Thus, considering the adverse effects of very hard carbides on surface roughness, further studies should be made on the proper matrix hardness, carbide hardness, and the amount of carbides required to simultaneously improve wear resistance and surface roughness without forfeiting the advantage of enhanced hardness and wear resistance.

In the present study, the effects of alloying elements on

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Table I. Chemical Compositions of HSS Rolls and a High Chromium Cast Iron Roll Investigated (Weight Percent)

Roll	C	W	Mo	V	Cr	Si	Mn	Ni	Al	W_{eq}^*
A	2.0	1.9	1.3	2.8	5.4	0.5	0.3	0.5	0.02	4.5
B	2.0	1.9	1.2	4.5	5.8	0.6	0.3	1.0	0.02	4.3
C	2.0	1.8	1.1	6.1	5.6	0.6	0.3	1.0	0.02	4.0
D	2.0	1.8	2.5	4.0	5.5	0.6	0.3	0.5	0.02	6.8
Tested HSS**	2.0	6.5	4.1	4.3	5.6	0.2	0.9	1.4	—	14.7
Hi-Cr	2.8	—	1.3	—	18.0	0.8	0.5	1.0	—	2.6

* $W_{eq} = W + 2Mo$; tungsten equivalent.

**HSS roll tested in rolling stands.

the formation behavior of carbides in HSS rolls was analyzed. This analysis included the effects of the type, volume fraction, and distribution of carbides on wear properties and surface roughness. Four experimental HSS rolls were fabricated by adding V, W, and Mo, and high-temperature wear tests were conducted to experimentally simulate the wear and surface roughness experienced by the rolls during hot rolling. Based on test results, the mechanisms of wear and surface roughening are delineated, and the correlation between carbides and wear properties are integrally analyzed in comparison to the actual wear process.

II. EXPERIMENTAL

Materials used in this study are four HSS rolls fabricated by centrifugal casting and a commercial Hi-Cr roll, whose chemical compositions and tungsten equivalents ($W_{eq} = W + 2Mo$) are listed in Table 1. The basic chemical composition of the HSS rolls was 2.0C-4.5V-1.2Mo-1.9W-5.8Cr (wt pct, $W_{eq} = 4.3$); the roll with this particular composition was designated as roll B. The other rolls contained varying contents of V, W, and Mo in order to investigate the effects of vanadium content and a tungsten equivalent. Roll A contained less vanadium (2.8 pct V), whereas roll C had more (6.1 pct V). Roll D was fabricated to have a higher tungsten equivalent (2.5 pct Mo - 1.8 pct W, $W_{eq} = 6.8$). A small horizontal centrifugal caster was used for producing the experimental rolls. Molten metal was poured into a high-speed rotating mold to form the shell part. Immediately after solidification, the core part of nodular graphite cast iron was introduced into the cast rolls, 400 mm in diameter (shell thickness: 65 mm) and 600 mm in length. These rolls were austenitized at 1050 °C for 2 hours, air cooled, and double-tempered at 540 °C for 2 hours. By the first tempering, the large amount of retained austenite transforms to martensite. The second tempering increases the fracture toughness by transforming martensite to tempered martensite. The Hi-Cr roll was austenitized at 950 °C to 1050 °C, air cooled, and then double-tempered at 450 °C to 600 °C. Metallographic specimens were obtained from the shell part.

The specimens were polished and etched in 5 pct nital and Murakami etchant (3 g $K_3Fe(CN)_6 + 10$ g NaOH + 100 mL H_2O),^[13] which allows for selective etching depending on the type of carbide. The microstructure was observed using an optical microscope and a scanning electron microscope (SEM), and the volume fraction of carbides was quantitatively analyzed using an image analyzer. After the volume fraction of overall carbides was first obtained from the nital-etched specimens, those of M_7C_3 -type

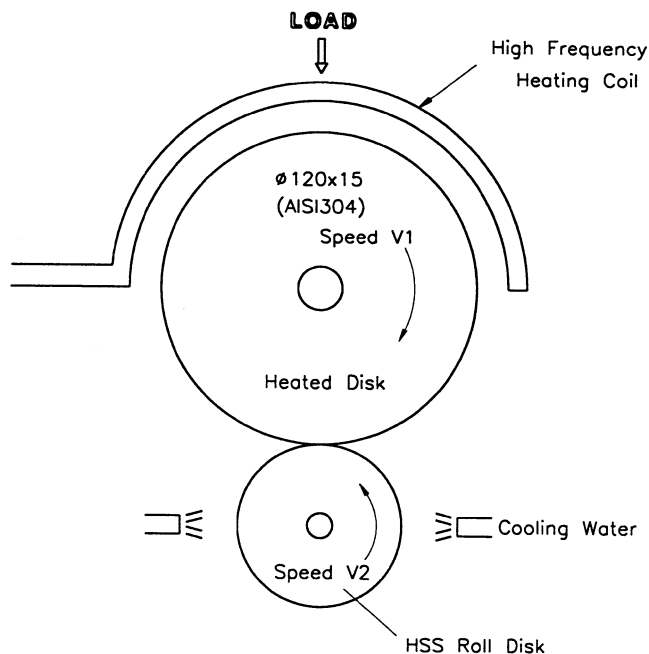


Fig. 1—Schematic diagram of the high-temperature wear tester used for hot rolling simulation.

and MC-type carbides were respectively obtained using the Murakami etching method. The overall bulk hardness and microhardness of the matrix were measured under a load of 30 kg and 0.5 kg, respectively, using a Vickers hardness tester.

Actual rolling is a process in which the rolled material is squeezed between two rotating rolls. In the present experiment, the HSS rolls and the rolled material (AISI 304 stainless steel) were machined into a disk shape, and then hot-rolling simulation tests were conducted using a rotating disk-type high-temperature wear tester, as shown in Figure 1. Figure 2 shows the shape and dimensions of the wear-test specimen. One disk is heated by an induction heating coil to function as steel slabs or strips, while the other serves as the roll. The AISI 304 stainless steel was used as the rolled material. During the experimental run, the revolving speed, load, and torque measurements were obtained using a computer/data logger. The high-temperature wear test was carried out under a stress of 600 MPa according to the Hertz calculation method,^[14] while the temperature of the 304 stainless steel disk was 850 °C. The backward-slip ratio was 34 pct, and the revolving speed of the roll disks was 0.3 m/s. The roll disks were water-cooled, as in the actual process. The weight of the roll disk was measured after predetermined revolutions to evaluate the

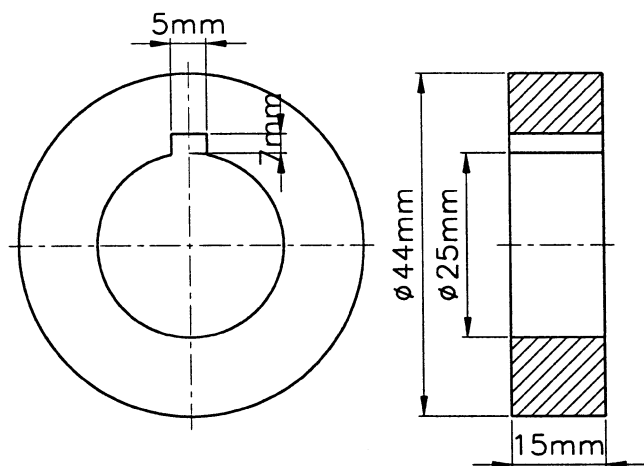


Fig. 2—Shape and dimensions of the wear test specimen (unit: mm).

wear loss. The friction coefficient (μ) was obtained by measuring the torque (T) applied to the rotating axis and vertical load (P) applied to the roll, using the following equation:^[12]

$$\mu = \frac{T}{P \cdot R}$$

where R is radius of the roll disk.

The surface and the cross section of the roll disk were characterized by an SEM after the wear test, and the average surface roughness was measured by a roughness gage. The surface of an HSS roll tested in the rolling stands was observed and compared to the wear process during the rolling simulation test. Its chemical composition is also shown in Table I.

III. RESULTS

A. Microstructure

Figures 3(a) through (e) provide optical micrographs of the four HSS rolls and the Hi-Cr roll. In the HSS rolls, M_7C_3 -type and M_2C -type carbides along cell boundaries and hard MC-type carbides inside cells are shown. Figures 3(a) through (c) indicate that the amount of MC-type carbides increases with an increasing vanadium content and that these carbides are relatively small and mainly distributed inside cells in a strawberry shape. In roll B, with a tungsten equivalent of 4.2, MC-type carbides are homogeneously distributed within cells (Figure 3(b)), whereas in roll D, with a tungsten equivalent of 6.8, they are mainly along cell boundaries (Figure 3(d)). All the rolls consist of lath-type tempered martensitic matrices, together with some mixed with plate-type tempered martensite. The ratios of these two martensites vary with the roll specimens, showing a tendency for more martensite plates to be present in roll D, which has a higher tungsten equivalent. Fine spheroidal $M_{23}C_6$ -type carbides, precipitated in the matrix by autotempering during slow cooling after casting, are also observed (Figure 4(a)).^[15] In the matrix of roll D, dark regions are found, as marked by arrows in Figure 3(d), and are confirmed to be bainitic structures (Figure 4(b)). The microstructure of the Hi-Cr roll consists of the chromium-containing carbides (white area) of M_7C_3 -type

and the tempered martensite matrix, as shown in Figure 3(e). Within the matrix, there are many areas with well-developed fine carbides.

Table II shows the quantitative analysis data of the carbide volume fraction and the stoichiometric carbon equivalent (SCE)^[17] required for carbide formation. Although M_2C -type carbides are formed in all the rolls by adding W and Mo, these are considered to be M_7C_3 -type carbides because of their meager amount, which is mostly mixed with M_7C_3 -type carbides. The total carbide volume fraction, excluding $M_{23}C_6$ -type carbides, is about 11 to 12 pct in all the four rolls. Comparing rolls A through C, the MC-type carbide fraction increases as the vanadium content increases, whereas the M_7C_3 -type carbide fraction is reduced. This indicates that the major effect of vanadium is to form MC-type carbides.^[18,19] The value, deducting the SCE from the carbon content, roughly equals the carbon content inside the matrix. This decreases, in order, for rolls A, B, and C, indicating that both the matrix hardness and the amount of tempered plate martensite inside the matrix are reduced. Particularly, roll C does not show any presence of plate martensite, as shown in Figure 3(c). Comparing rolls B and D, with differing Mo contents, the overall carbide volume fraction tends to increase with increasing Mo content. The total carbide volume fraction is much higher (26 pct) in the Hi-Cr roll.

B. Hardness

To investigate the cause of surface deterioration, the overall bulk hardness and matrix hardness were measured and are listed in Table III. Although the overall bulk hardness is slightly higher in roll B, it is roughly in the same range, with a Vicker's hardness number of 700. The matrix hardness of rolls A through C drops as the vanadium content increases. This is associated with the lower carbon content in the matrix, as vanadium forms MC-type carbides. This is also confirmed by the C pct-SCE values in Table II. The Hi-Cr roll has lower hardness values because the carbides of the Hi-Cr roll, mostly M_7C_3 -type, have the lower hardness.

C. Wear Properties

Figure 5 shows the weight loss of the roll disk vs the number of revolutions for rolls A through D. The wear amount almost linearly increases with an increasing number of revolutions. Roll C, with the highest vanadium content, shows the least weight loss among rolls A through C. This is because very hard MC-type carbides are increasingly formed with increasing vanadium content, as shown in Table II. Their even distribution inside cells might help improve wear resistance. Comparing rolls B and D, roll D, with a higher tungsten equivalent, has a slightly better wear property than roll B due to the larger amount of carbides, but the difference is much less than in the case of the vanadium content variation. These HSS rolls have superior wear resistance to the Hi-Cr roll—3 to 4 times better. This is because the latter is solely composed of M_7C_3 -type carbides, which have a relatively low hardness.^[12,20] These weight-loss results hardly show any correlation to the hardness results.

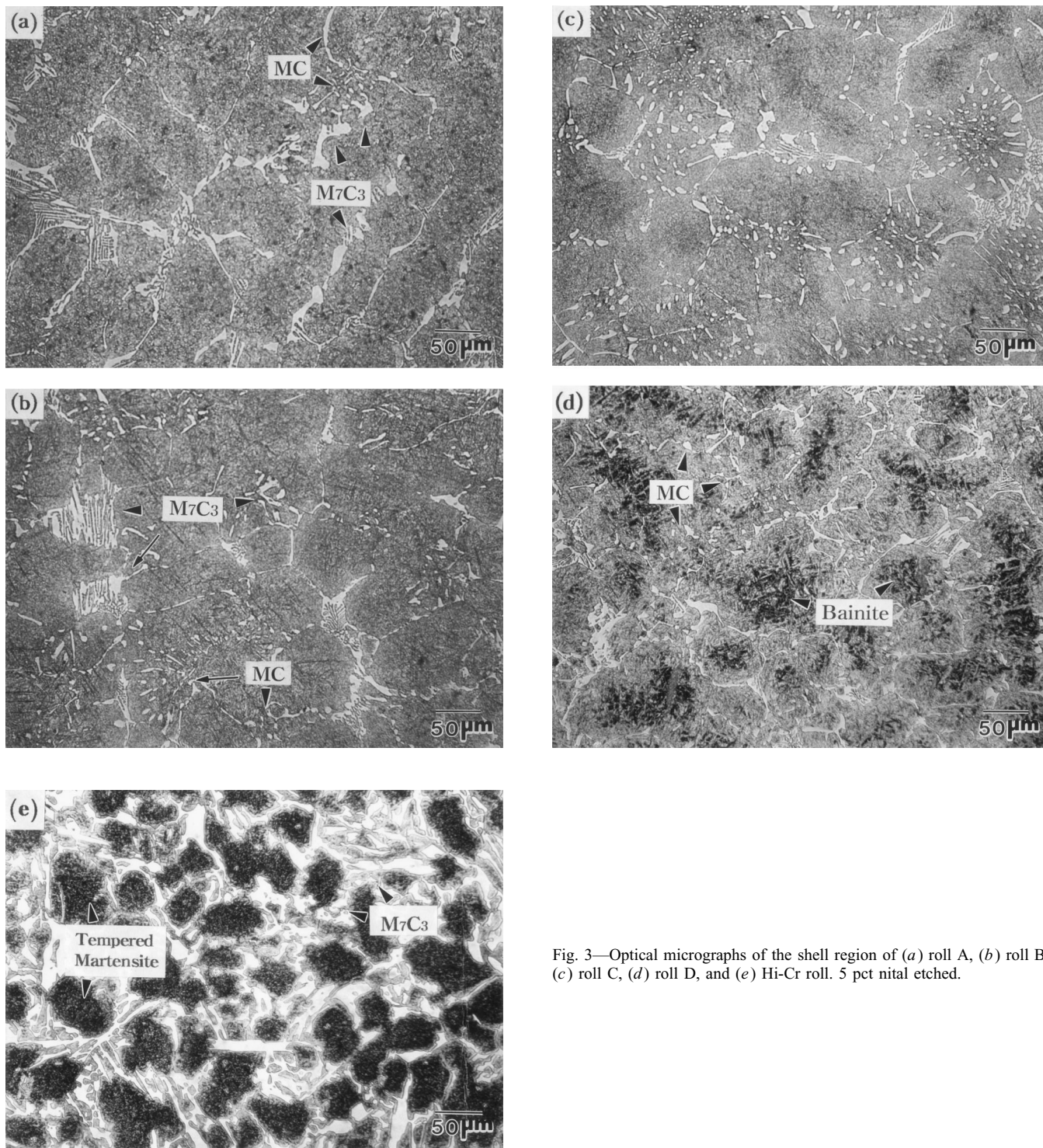


Fig. 3—Optical micrographs of the shell region of (a) roll A, (b) roll B, (c) roll C, (d) roll D, and (e) Hi-Cr roll. 5 pct nital etched.

D. Surface Roughness

Figures 6(a) through (d) are SEM micrographs of the surface of the roll disks after the wear test at 12,000 revolutions. Roll A shows comparatively even wear of the matrix and carbides (Figure 6(a)). However, it is observed in rolls B and C that cracks are formed at the interfaces between the cell-boundary carbides and the matrix. The matrix is also severely scratched or peeled off, and some portions near the carbides have fallen off (Figures 6(b) and (c)). Particularly, roll C shows the far-rougher roll surface

in the regions with a large amount of MC-type carbides. As in rolls B and C, roll D shows a similar wear pattern, with its matrix peeled off (Figure 6(d)).

Figure 7 shows the relationship of the carbide volume fraction vs the friction coefficient. The friction coefficient decreases as the total carbide fraction increases, because the increasing carbide fraction reduces the contact of the 304 steel with the matrix. However, when considering MC-type carbides apart from total carbides, the increase of the MC-type carbide fraction is closely related to the increase of

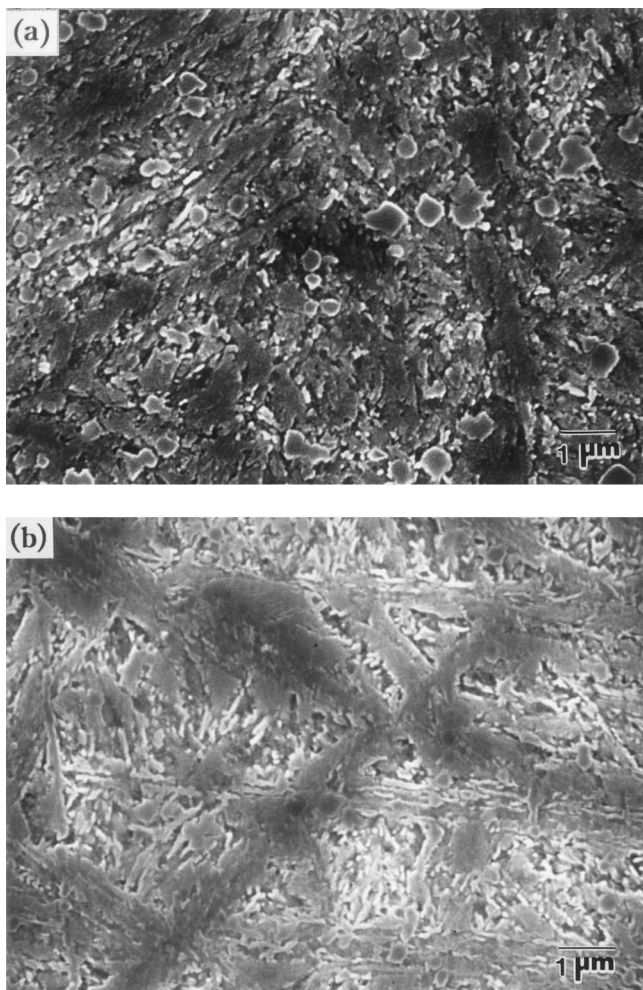


Fig. 4—SEM micrographs showing (a) the martensitic matrix containing fine spheroidal carbides in roll C and (b) bainitic structure in roll D. 5 pct nital etched.

Table II. Results of Quantitative Analysis of HSS Rolls and a High Chromium Cast Iron Roll

Roll	Volume Fraction of Eutectic Carbides (Pct) ^[16]			SCE**	C Pct SCE†
	MC	M ₇ C ₃ *	Total		
A	1.8	10.6	12.4	1.02	0.98
B	3.8	6.7	10.5	1.38	0.62
C	8.2	3.2	11.4	1.68	0.32
D	4.4	6.9	11.3	1.35	0.65
Tested HSS	4.9	10.6‡	15.5	1.66	0.34
Hi-Cr	—	26.0	—	—	—

*Includes small amounts of M₇C₃-type carbides.

**SCE = 0.06Cr + 0.033W + 0.06Mo + 0.2V (wt pct); stoichiometric carbon equivalent.

†This value estimates maximum carbon content contained in the matrix.

‡This value indicates the volume fraction of M₇C₃ carbides because the tested HSS roll contains many M₂C carbides instead of M₇C₃ carbides.

the friction coefficient, indicating that the presence of MC-type carbides is the main factor in increasing the friction coefficient.

Figure 8(a) and (b) are the surface roughness data vs vanadium content and tungsten equivalent. The increase in vanadium content results in a rougher surface. The surface

Table III. Matrix and Overall Bulk Hardness of HSS Rolls and a High Chromium Cast Iron Roll

Roll	Matrix Hardness* (VHN)	Overall Bulk Hardness** (VHN)
A	585	698
B	558	738
C	523	706
D	556	687
Hi-Cr	513	527

*Hardness value measured using 0.5 kg load.

**Hardness value measured using 30 kg load.

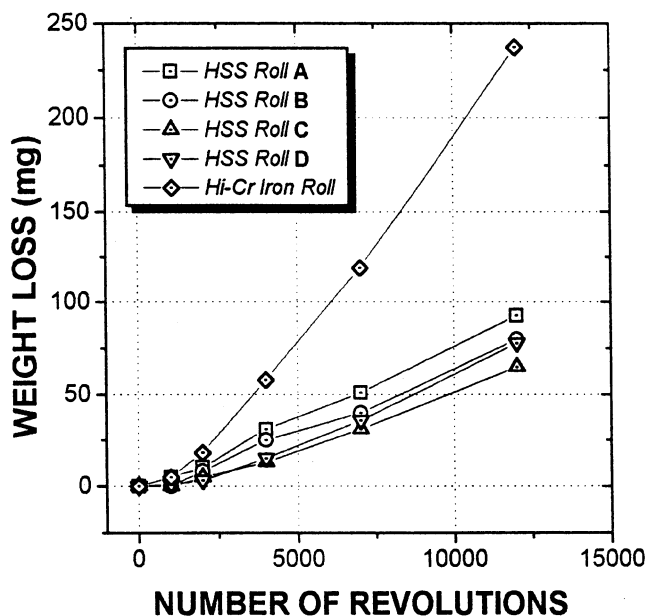


Fig. 5—Weight loss of the roll disk vs the number of revolutions for four HSS rolls and a high chromium cast iron roll.

deterioration becomes more severe (Figure 8(a)) according to the decrease of the matrix hardness, as can be seen in Table III. However, the tungsten equivalent does not lead to much of a difference in surface roughness (Figure 8(b)). The relation of surface roughness to the friction coefficient is shown in Figure 9, indicating a linear increase.

Figures 10(a) through (e) are SEM micrographs of the cross section of the surface layer sectioned in the diameter direction. Roll A keeps a nearly plane surface (Figure 10(a)), but severe prominence and depression have developed on the surface in roll C (Figure 10(b)). This is magnified, as in Figure 10(c), showing crack propagation along cell-boundary carbides. Further progress of cracking might induce fall-off of the matrix surrounded by cells. Because of continuous friction of this fallen-off matrix with the 304 steel, wear debris, which abrades in the gap between the roll disk and 304 steel disk, is formed, and continues scratching off the matrix, resulting in the rougher surface. In other surface regions, sticking of the 304 steel is observed between the severely roughened regions (Figure 10(d)). This surface region was sectioned and observed using an SEM. An approximately 10- μ m-thick layer is formed due to the repeated sticking of the 304 steel on the roll surface, inside which slip bands are developed parallel to the rolling direction (Figure 10(e)).

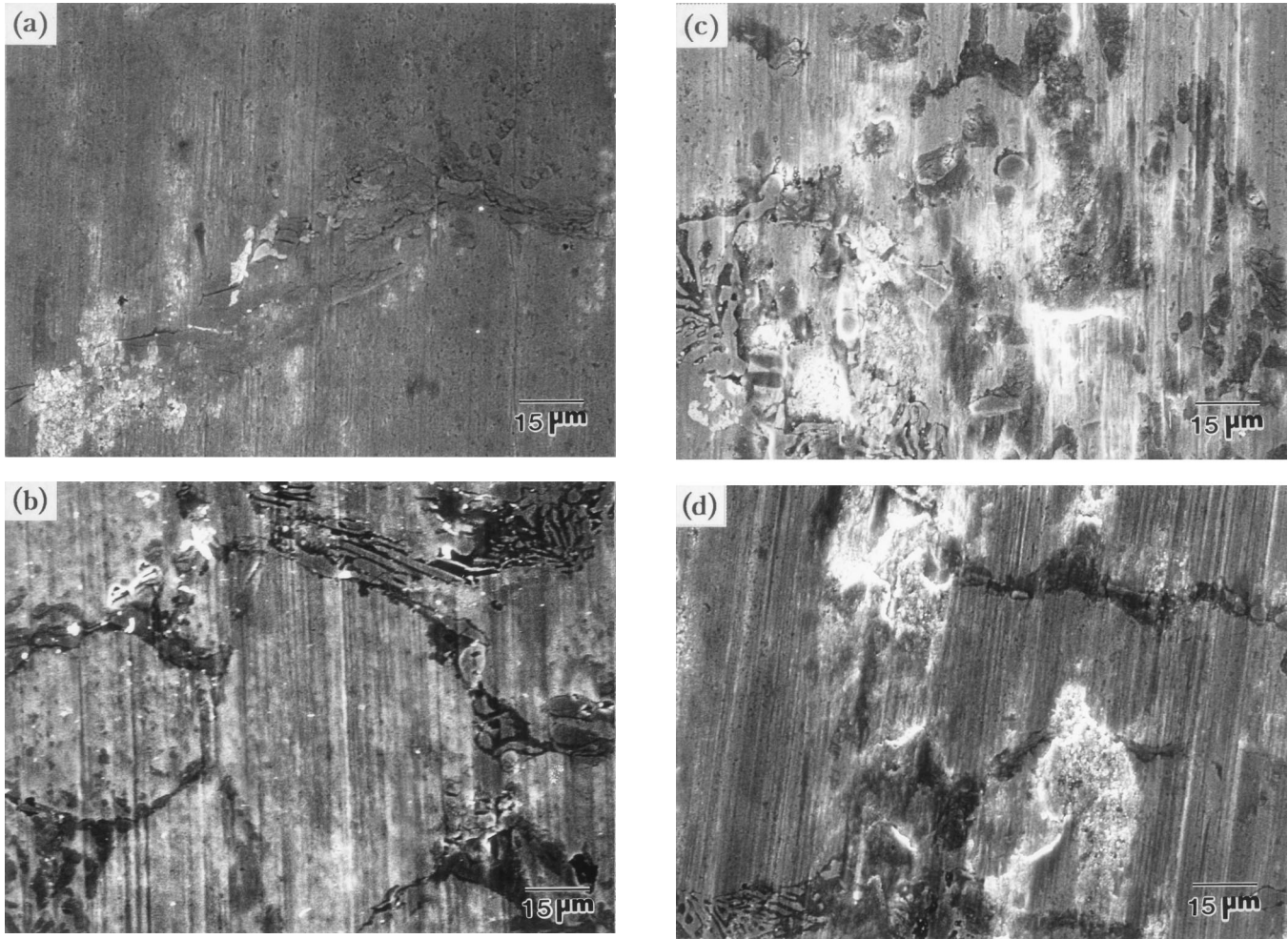


Fig. 6—SEM micrographs of the surface of the roll disks for (a) roll A, (b) roll B, (c) roll C, and (d) roll D after 12,000 revolutions.

E. Observation of Conventional HSS Roll Surface

Figure 11(a) is an SEM micrograph of the worn surface of an HSS roll tested at the Pohang Iron and Steel Co. (POSCO, Pohang, Korea) hot-strip mill. This roll contains much tungsten and molybdenum (Table I) and consists of many M_2C carbides formed mainly along the cell boundary, as shown in Figure 12. The black oxide layer formed during hot rolling is separated from the surface, showing a similarity to the wear surfaces of Figures 6(b) through (d) after the experimental wear test. The SEM observation of this roll, sectioned parallel to the rolling direction (Figure 11(b)), reveals a similar result to that of the wear test of Figure 10(b). Thermal cracks initiate and propagate along cell-boundary carbides, thereby removing parts of the roll surface.

IV. DISCUSSION

The microstructural factors affecting the properties of HSS rolls are (1) the type, morphology, volume fraction, and distribution of carbides; (2) the fine spheroidal carbides precipitated inside the matrix; and (3) the solidification cell structure. Due to the hardness of carbides themselves, they directly influence the overall bulk hardness, wear properties, and surface roughness. Furthermore, depending on

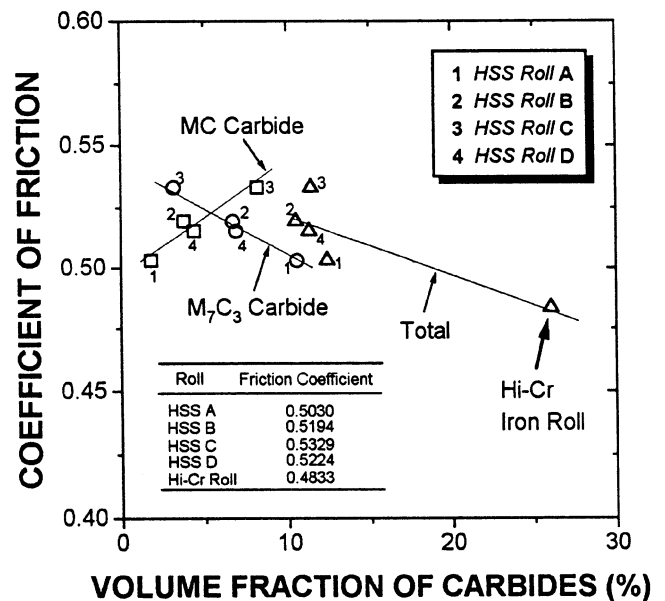
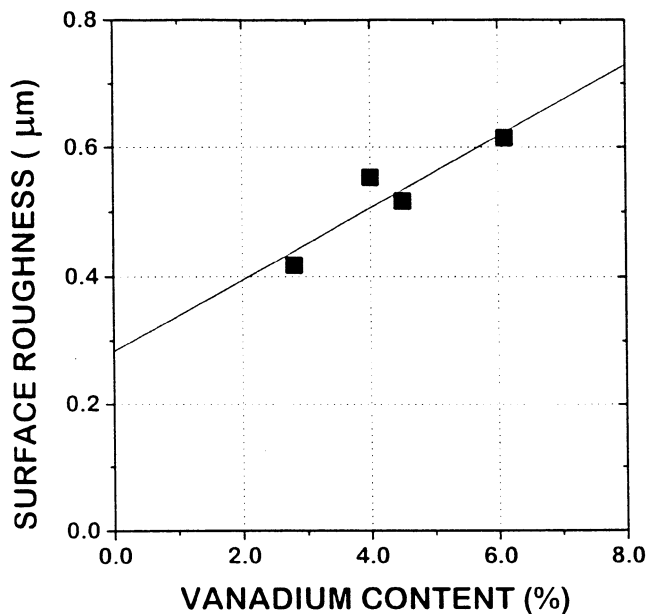
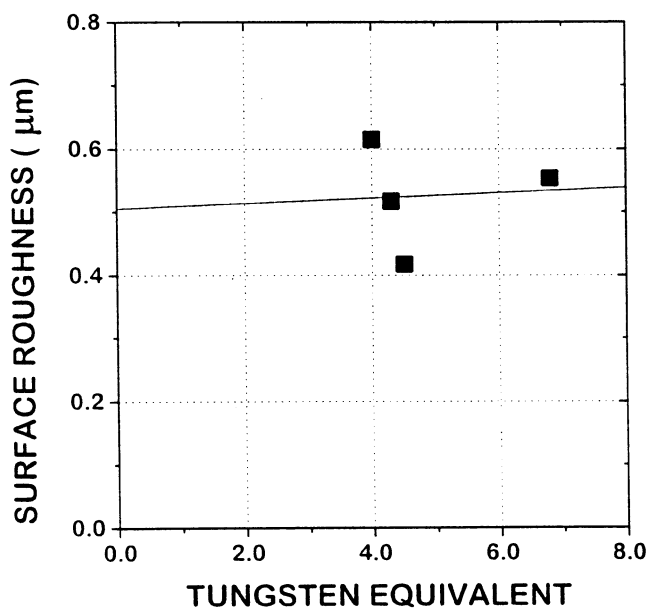


Fig. 7—Friction coefficient vs volume fractions of total carbides, MC-type carbides, and M_7C_3 -type carbides.

their morphology, volume fraction, and distribution, they also significantly affect thermal fatigue and fracture char-



(a)



(b)

Fig. 8—Surface roughness vs (a) vanadium content and (b) tungsten equivalent after 12,000 revolutions.

acteristics. The matrix is related to the overall bulk hardness, strength, and fracture toughness and holds hard carbides in it. The size of the cell structure plays a critical role in determining the overall distribution of carbides, thereby affecting thermal fatigue, fracture toughness, and the separation of black iron-oxide layers.^[21] Thus, these microstructural factors should well be incorporated and optimized in order to manufacture HSS rolls with excellent performance. To especially improve wear resistance, the most crucial factor is to form a large amount of discontinuous carbides homogeneously distributed inside cells or along cell boundaries.^[20] In this respect, roll C, containing a large amount of hard MC-type carbides, has the best wear properties, as can be confirmed by Table II and Figure 5.

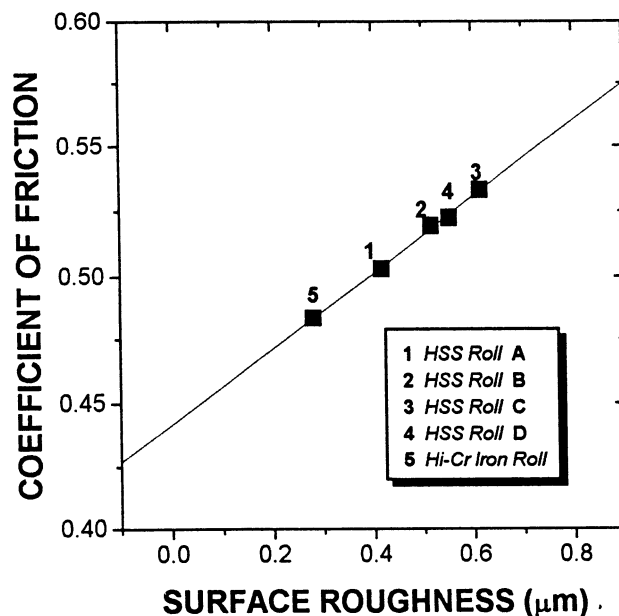


Fig. 9—Friction coefficient vs surface roughness for four HSS rolls and a high chromium cast iron roll.

However, considering surface roughness together with wear resistance, very hard carbides may work negatively, because they cause severe surface unevenness during the wear process due to the large difference in hardness between the carbide and the matrix. As shown in Figure 5, at the initial stage of wear testing up to revolution 1000, no weight loss is observed in roll C. This does not mean that there is no wearing under progress, but the result is caused by the simultaneous sticking of the 304 steel to the roll surface. In Figure 10(d) of actual wearing, severe unevenness and sticking of the 304 steel are observed, indicating that the saturation point has already been reached following the initial stage. It is reported that the sticking of the 304 steel debris gradually increases and reaches the saturation stage, after which the sticking stops.^[22,23] If the wear-test results of roll C include the weight of the stuck 304 steel, its wear resistance could be worse than that of roll A, which has a clean surface without having any sticking. Although the amount of the stuck 304 steel was not measured in this study, it can be inferred that roll A has better wear resistance than roll C when both the sticking and surface roughness are taken into consideration.

In roll C, ample formation of MC-type carbides also reduces matrix hardness (Table III). As the matrix is worn prior to other parts and cell-boundary carbides fall off during the wear process, severe prominence and depression take place on the surface. Figure 10(c) shows the process of crack initiation and propagation along cell boundaries and the final stage of fall-off of the matrix surrounded by the cell boundary. This surface roughness causes the friction coefficient to rise, as can be confirmed from the relation of MC-type carbides to the friction coefficient in Figure 7. On the other hand, the matrix hardness of roll A is higher than that of roll C because of the smaller amount of MC-type carbides formed in the matrix. The M_7C_3 -type carbides, having a lower hardness, are primarily in charge of wear resistance instead of MC-type carbides, and the hardness gap between the matrix and the M_7C_3 -type car-

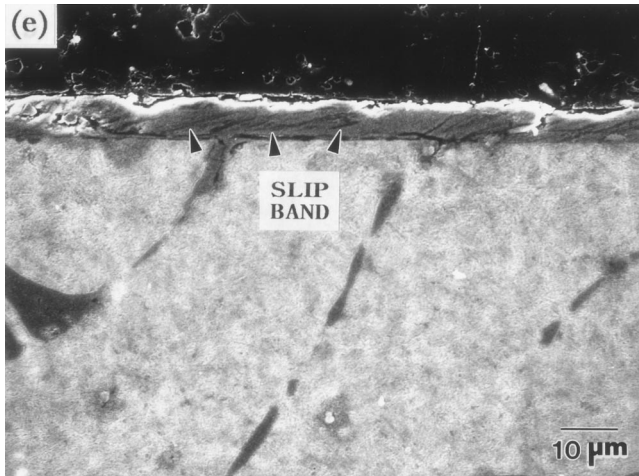
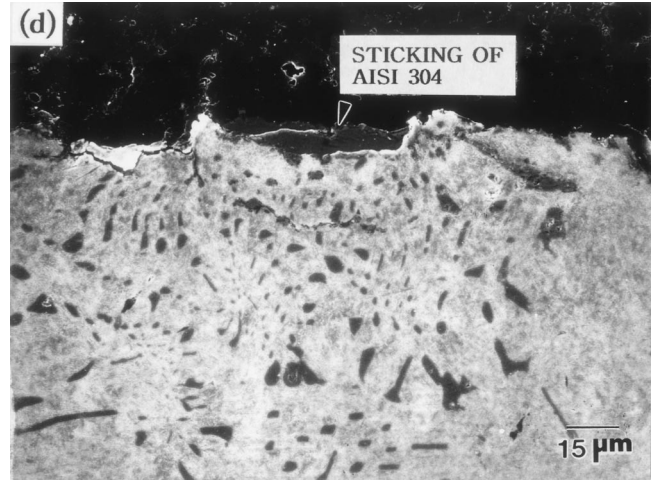
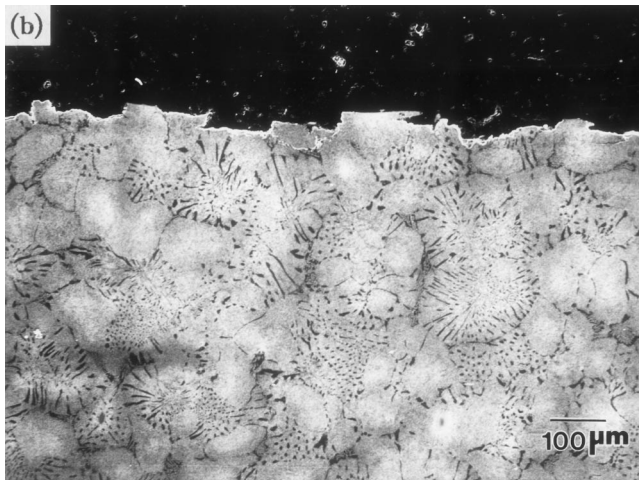
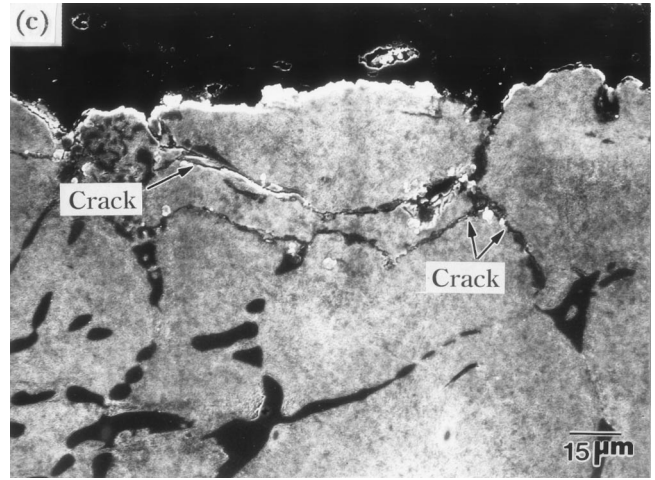
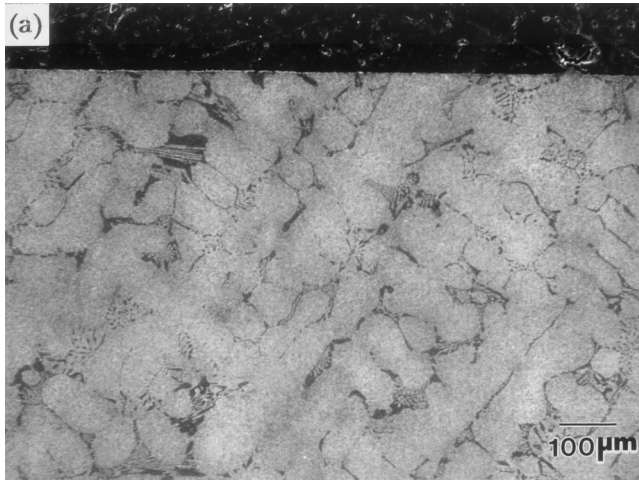


Fig. 10—SEM micrographs of the cross sections of (a) roll A and (b) through (e) roll C. Note the cracks propagated along cell boundaries and the sticking phenomenon of the 304 stainless steel in roll C.

bides is reduced due to a harder matrix. Thus, very few carbides fall off, thereby leading to a homogeneous wear of the matrix and carbides and to a far better surface roughness than roll C, as shown in Figure 10(a).

These wear-test results indicate that surface roughness should also be taken into account to improve the overall wear properties of the HSS rolls and that a different interpretation from the general wear pattern is required. Although an increase in MC-type carbides distributed evenly

inside cells may enhance wear resistance, it reduces matrix hardness and broadens the hardness gap from that of carbides, resulting in severe surface irregularity and in deteriorated surface roughness. The MC-type carbides also promote the sticking of the 304 steel and considerably increase rolling force by increasing the friction coefficient. The heavy formation of MC-type carbides, thus, has a negative effect on overall wear properties. Therefore, to improve wear properties with consideration to surface

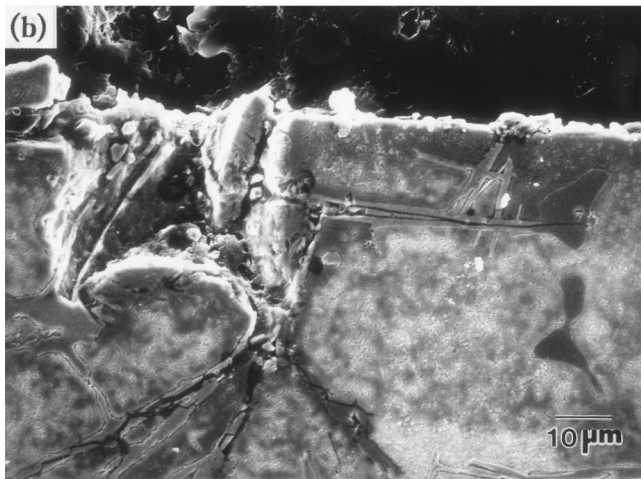


Fig. 11—SEM micrographs of (a) the surface and (b) the cross section of an HSS roll tested in rolling stands.

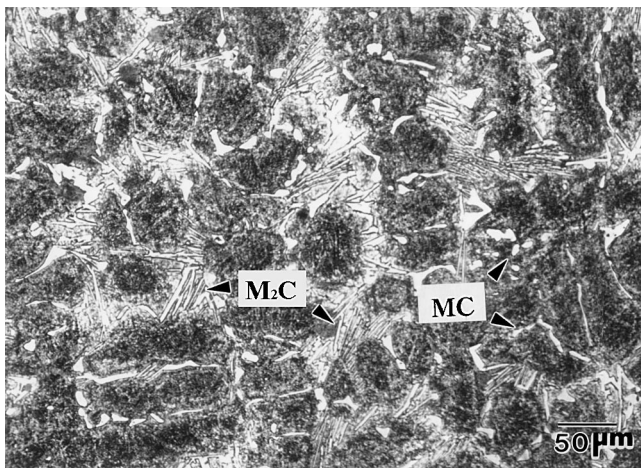


Fig. 12—Optical micrograph of the shell region of the HSS roll tested in rolling stands, showing many M_2C carbides formed mainly along the cell boundary.

roughness, it is necessary to strengthen the matrix to hold carbides.

Methods to strengthen the matrix of the HSS rolls include (1) reducing the vanadium content to increase the carbon content of the matrix; (2) adding more alloying el-

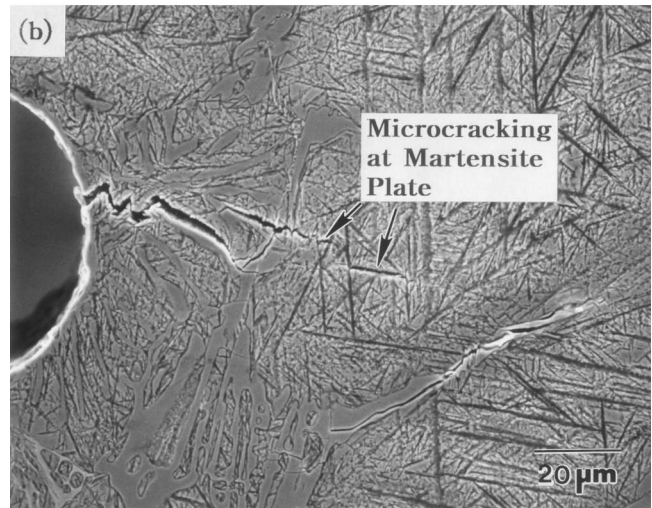
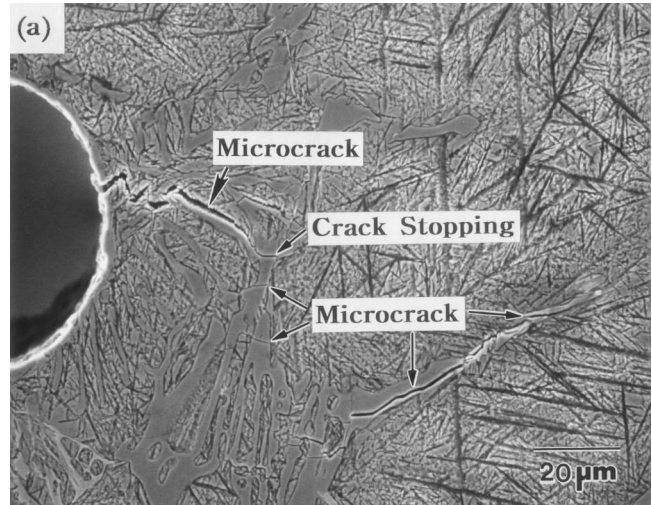


Fig. 13—A series of SEM micrographs near a notch tip of an HSS roll, showing (a) microcrack formation at carbide/matrix interfaces and at carbides themselves and (b) microcrack propagation through martensite plates of the matrix. This roll was austenitized at 1100 °C and double-tempered at 560 °C.

ements of W, Mo, and Cr to improve solid-solution hardening; (3) precipitating many fine carbides in the matrix by optimizing the austenitization and double-tempering treatments; and (4) promoting the formation of refined grains in the matrix. Under the heat-treatment conditions used in the present study, the appropriate compositions for wear properties are 2 to 3 pct of V, 3 to 4 pct of W and Mo, and 6 to 7 pct of Cr. The Cr content might be slightly raised over the basic composition of roll B, because it is effective in enhancing resistance to oxidation and in matrix hardening due to the precipitation of fine $M_{23}C_6$ -type carbides. Matrix hardening can also be achieved by varying austenitizing and tempering temperatures. A higher austenitizing temperature promotes greater formation of plate-type martensite, and the tempering temperature showing maximum hardness leads to the strengthened matrix. However, it is not desirable to raise the matrix carbon content or the austenitizing temperature to any extreme extent, since plate martensite may unduly harm fracture toughness. Figures 13(a) and (b) are SEM micrographs of an HSS roll austenitized at higher temperatures, showing microcracking in the

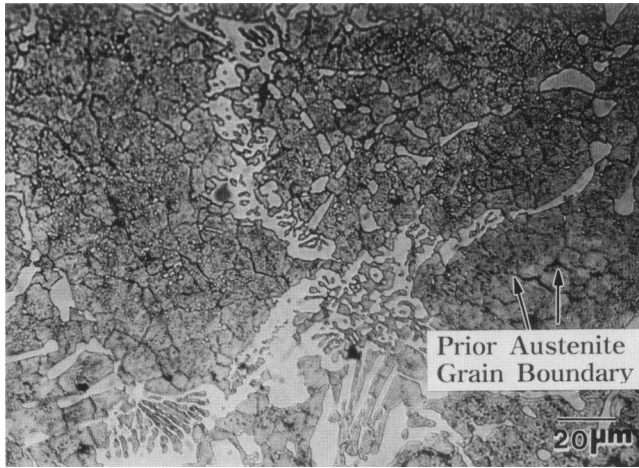


Fig. 14—Optical micrograph of roll A, showing small prior austenite grain boundaries formed after preaustenitization treatment.

plate-type martensitic matrix. In this case, cracks directly cut across the martensitic matrix rather than propagate along cell boundaries because of the brittle nature of the matrix, thus considerably lowering the fracture toughness. One method used to strengthen the matrix by heat treatment without deteriorating the fracture toughness is to form refined prior austenite grains in the matrix. Prior to austenitization, the temperature is raised to the austenite region of about 850 °C, cooled down to 500 °C to 600 °C, and raised again for austenitization. This way, small prior austenite grains are formed in the matrix. Figure 14 is an optical micrograph of roll A, showing an example of refined prior austenite grains formed after the preaustenitization treatment. In this case, as prior austenite grains become smaller, the martensite packet size might also be reduced, thus probably enhancing the fracture toughness as well as strengthening the matrix due to the presence of grain boundaries.

The wear mechanism of the HSS rolls occurring during hot rolling was investigated in the present study in relation to the wear-test results and microstructural factors. Based on the analysis data, suggestions are made to simultaneously improve wear properties as well as surface roughness by distributing carbides with a slightly lower hardness in the strengthened matrix. For further improvement of properties, future studies should focus on identifying the microstructural conditions, alloy design, and process parameters that can enhance the matrix. The test simulating the hot-rolling process is expected to prove effective in studies related to wear properties, as it helps clarify the mechanisms involved in wear and surface roughness.

V. CONCLUSIONS

The present study presents essential data for producing rolls with enhanced wear properties by examining the effect of alloying elements on the microstructural factors and wear properties of four HSS rolls.

1. As the vanadium content increases, the amount of MC-type carbides increases, and they are smaller in size and evenly distributed inside cells. The matrix consists primarily of lath-type tempered martensite, together with small amounts of plate-type martensite.

2. The HSS rolls displayed a wear resistance 3 to 4 times better than the Hi-Cr roll. The roll with highest vanadium content shows the best wear properties, due to the presence of many MC-type carbides. As the matrix is worn prior to carbides and falling-off of cell-boundary carbides occurs, severe unevenness takes place on the roll surface. Because of the low matrix hardness, the surface roughness seriously deteriorates and the friction coefficient increases, which causes the sticking of the 304 steel on the roll surface.
3. In the roll with a reduced amount of vanadium, a smaller amount of MC-type carbides is formed, with a higher matrix hardness. The wear resistance is primarily taken care of by M_7C_3 -type carbides with a lower hardness. As a result, since carbides do not fall off but are held inside the matrix, surface roughness is enhanced by more homogeneous wear of the matrix and carbides.
4. To improve wear properties with consideration to surface roughness, a strengthened matrix is required to hold hard carbides in it. For this purpose, suggestions are made to reduce the vanadium content, to induce solid-solution hardening by adding alloying elements, to promote secondary hardening by precipitation of fine carbides, and to form small prior austenite grains by applying a preaustenitization treatment.

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