

IMPROVING THE LIFE OF CONTINUOUS CASTING ROLLS THROUGH SUBMERGED ARC CLADDING WITH NITROGEN-ALLOYED MARTENSITIC STAINLESS STEEL

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ABSTRACT

Continuous casting rolls are subjected to extreme temperature fluctuations and harsh environmental conditions during service. High slab contact pressures, high surface temperatures, adhesive or abrasive wear by slag or oxides and thermal fatigue all contribute towards deterioration of the roll surface. Rolls may also suffer surface degradation as a result of corrosion. The slab is cooled using water spray, often containing hydrofluoric acid as product of the reaction between water and the mould flux powder used to keep the molten metal from adhering to the copper mould plate. This results in an aggressive low pH environment that may promote pitting or crevice corrosion and stress corrosion cracking failures. In order to prolong the life of continuous casting rolls in service, the rolls are surfaced using martensitic stainless steel weld cladding. After surfacing, the roll may be given a post-weld heat treatment to temper the martensite to the required hardness level, or to relieve thermally-induced stresses. This investigation examined the submerged arc cladding of continuous casting rolls using nitrogen-alloyed martensitic stainless steel, and focused on the effect of the weld thermal cycle and post-weld heat treatment on carbide precipitation in the clad layer. The precipitation of chromium-rich carbides during welding or post-weld heat treatment may lead to weld decay due to chromium depletion, rendering the roll surface susceptible to pit nucleation and crack initiation.

IW-Thesaurus keywords: *Cladding; Martensitic stainless steels; Submerged arc welding.*

1 BACKGROUND

Continuous caster rolls in steel mills operate in particularly harsh service environments. Within the caster, molten steel poured from a tundish flows into an oscillating copper mould where it forms a skin against the water-cooled copper plates. The partially solidified slab is then pulled out of the mould through a series of containment segments containing drive rolls. The slab is usually bent into the curved section of the caster, and unbent towards the run-out portion. Within the containment section of the caster, the slab is subjected to water spraying to promote solidification [1]. High

slab contact pressures, elevated temperatures, plastic deformation of roll surfaces on contact with the hot slab, severe abrasive wear by oxides and slag, cyclic thermal and mechanical stresses and corrosion, either individually or in combination, contribute towards the deterioration of caster rolls in service [2].

Although high-strength low-alloy steel forgings and castings were used in the 1970's and 1980's for caster roll applications, rolls used in continuous casting lines have since been found to exhibit increased life if surfaced, by welding, with a layer of more noble material resistant to contact with the hot slab and the harsh environment within the caster [2]. The introduction of weld cladding with modified 12 % chromium AISI 400 series martensitic stainless steels in the middle to late 1980's significantly improved product quality and the life expectancy of continuous casting rolls and these materials are now the industry norm in this application.

The introduction of submerged arc cladding with relatively inexpensive AISI 420 martensitic stainless steel tubular wires, containing around 0.2 % carbon and 12 % chromium, greatly improved the service life of caster rolls, and became standard on steel mills during the late 1980's and 1990's [3]. The 420 weld deposit,

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although a significant improvement on low-alloy forgings, still exhibited unsatisfactory properties in the weld metal and reheated heat-affected zone due to the high hardness and low ductility of the medium-carbon martensite phase that forms on cooling. Preheating at temperatures of 315 °C and higher, high interpass temperatures and slow cooling rates are required in these deposits to prevent brittle fracture and to minimize the likelihood of hydrogen-induced cracking [3]. Slow cooling rates facilitate the precipitation of chromium-rich $M_{23}C_6$ carbides at the prior austenite grain boundaries or martensite lath boundaries. During multiple pass welding, the formation of carbides is very difficult to prevent, especially in the reheated or inter-bead zones. Once welding is completed, the martensite should be tempered, which leads to additional carbide precipitation. The precipitation of chromium-rich carbides may lead to chromium depletion of the surrounding matrix, a phenomenon known as sensitization. The sensitized regions exhibit reduced corrosion resistance and increased susceptibility to pit nucleation and the initiation of stress corrosion cracks in service [2]. A derivative of AISI 414 is also used, but although characterized by superior performance, may exhibit unacceptable deterioration of the roll surface within three months of installation.

As an alternative to using carbon to increase hardness and form austenite at elevated temperatures within the 12 % chromium alloys, nitrogen can be added to martensitic stainless steels for cladding applications. Alloys containing nitrogen as a substitute for carbon display improved mechanical properties (including higher strength, ductility and toughness), and much improved temper resistance. Although nitrides may precipitate in these martensitic steels, the nitride precipitates form at considerably higher temperatures and are finer than carbides, inhibit grain growth and improve strength. Nitride precipitation does not occur preferentially at grain boundaries, but more uniformly throughout the microstructure [2].

This report describes the current philosophy behind the development of optimized alloys for roll refurbishing, with the objectives of increasing the service life of rolls well beyond 2 million tons of cast steel, reduce the number of maintenance stops, increase productivity, improve product quality (less roll cracking and bending), and reduce the amount of time spent on machining rolls subsequent to cladding. It also investigates the possibility of eliminating the need for post-weld heat treatment of refurbished rolls.

2 SURFACE DETERIORATION OF CASTER ROLLS

In order to optimize cladding alloys and increase the service life of continuous casting rolls, a thorough understanding of the possible mechanisms of failure of these rolls is required. Continuous casting rolls may deteriorate during service or suffer catastrophic fail-

ure through a number of possible failure mechanisms. In general, the risk of catastrophic failure (as a result of spalling, neck breakage or roll breakage) is lower with clad rolls, and surface deterioration is the dominant cause of roll failure. Possible surface deterioration mechanisms are considered in more detail below [1, 4, 5].

2.1 Abrasive/adhesive wear and oxidation

In the finishing stands and lower segments of the caster roll line, oxidation of the steel surface in the presence of steam produces oxides which aggravate abrasive wear. Oxidation results in an “elephant skin” surface appearance due to grain boundary oxidation of the outer clad layers, as well as cracking and spalling of the cladding. An example of the surface of a worn casting roll is shown in Figure 1. The risk of wear and oxidation failure can be reduced by ensuring that the clad layer has a high chromium content (greater than approximately 12 %), which facilitates the formation of a continuous stable oxide layer that has the ability to regenerate spontaneously. Alloying with nitrogen is beneficial due to the formation of a stable nitrogen-modified $FeO.Cr_2O_3$ spinel phase on the roll surface, which ensures a lower coefficient of friction and superior resistance to gouging and adhesive wear [5, 6].

2.2 Thermal and mechanical fatigue

Fluctuating stresses within the surface layers of casting rolls may give rise to fatigue failures during service. The fluctuating stresses may be mechanical, caused by cyclic loading per revolution due to the passage of the slab or by bending of the rolls [1, 4, 6]. Fluctuating thermal stresses develop due to changes in the temperature of the roll surface as it contacts the hot slab (leading to peak temperatures in the region of 600 °C), followed by water spraying as the roll revolves. A network of fine thermal fatigue cracks, known as “fire-cracking”, may develop, and is recognized as the major

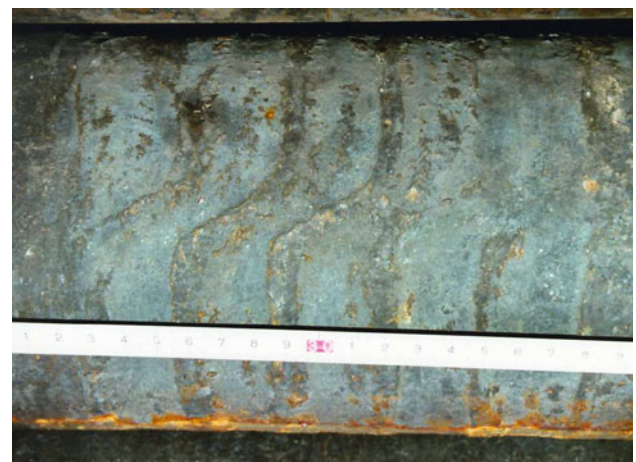
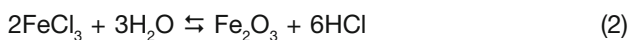
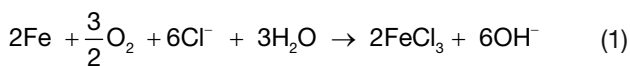


Figure 1 – Wear of a casting roll
(Photograph courtesy of Welding Alloys Ltd, UK) [6]

cause of material loss in roughing stands. Thermal fatigue cracking can be minimized by ensuring that the roll surface exhibits a low coefficient of thermal expansion and high thermal conductivity.

2.3 Localized corrosion

Localized pitting and crevice corrosion develops as a result of exposure to aggressive acidic environments within the caster. Anodic dissolution of the steel may be accelerated within surface discontinuities, promoted by progressive acidification of the pit or crevice environment (as illustrated by Equations (1) and (2)). Extensive pitting corrosion of the surface of a caster roll is shown in Figure 2. High chromium, molybdenum and nitrogen contents increase resistance to localized corrosion [6].



2.4 Premature inter-bead corrosion

Multiple pass welding and slow cooling after cladding promote the precipitation of chromium-rich M_{23}C_6 carbides at the prior austenite grain boundaries or martensite lath boundaries. The formation of these carbides is very difficult to prevent, especially in the reheated or inter-bead zones between runs. Resulting chromium depletion of the surrounding matrix may give rise to sensitization and preferential corrosive attack of the chromium depleted overlap regions. A reduction in the carbon content, as well as higher levels of chromium and nitrogen, are beneficial in preventing attack [2, 6]. An example of inter-bead corrosion caused by sensitization is shown in Figure 3.

2.5 Stress corrosion cracking

A major cause of failure of casting rolls is crack initiation as a result of the presence of corrosion pits and

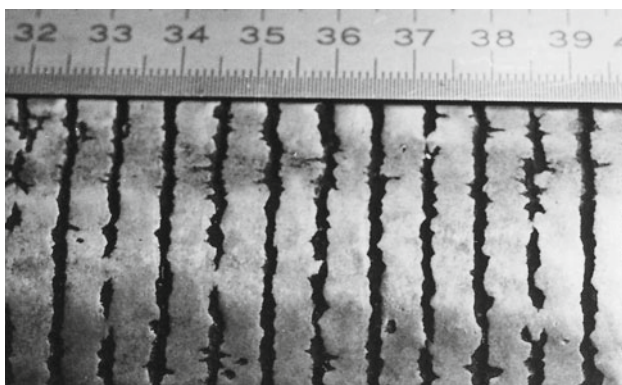


Figure 2 – Pitting corrosion of a continuous casting roll surface

(Photograph courtesy of Welding Alloys Ltd, UK) [6]



Figure 3 – Inter-bead corrosion of a continuous casting roll surface

(Photograph courtesy of Welding Alloys Ltd, UK) [6]

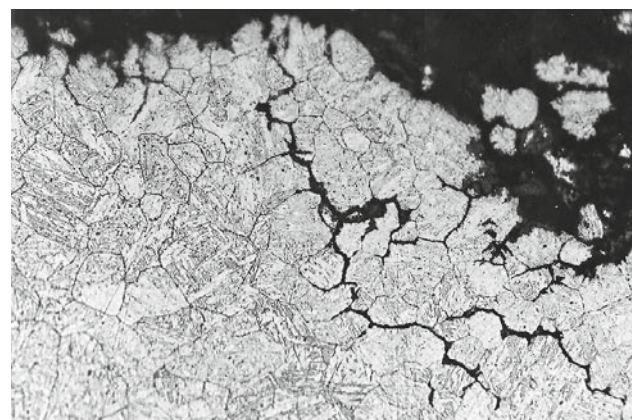


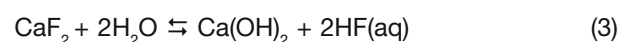
Figure 4 – Intergranular stress corrosion cracking of a continuous casting roll surface

(Photograph courtesy of Welding Alloys Ltd, UK) [6]

crevices, particularly in the reheated or overlap zones of weld beads. These areas may be sensitized by the weld thermal cycle and are often prone to chromium depletion and intergranular attack. Once initiated, these cracks can propagate under the influence of applied tensile stresses which damage the passive film at the crack tip [6]. A micrograph displaying stress corrosion cracking along the prior austenite grain boundaries of a caster roll clad layer is shown in Figure 4.

2.6 Mould flux-induced corrosion

Cooling water used in the upper segments of the caster contains hydrofluoric acid as a result of a reaction with the mould flux powder used to keep the molten metal from adhering to the copper mould plates [6]. These reactions are illustrated by Equations (3) and (4).



The formation of hydrofluoric acid lowers the pH of the cooling water and creates a severely corrosive environment in contact with the roll surface. An example of mould flux-induced corrosion is shown in Figure 5.

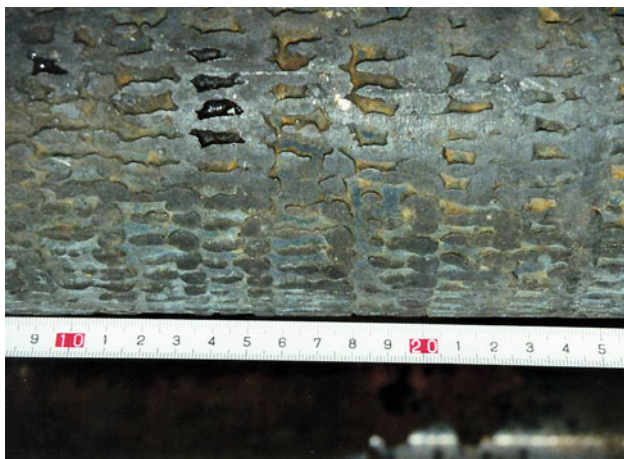


Figure 5 – Mould flux-induced corrosion of a continuous casting roll surface
(Photograph courtesy of Welding Alloys Ltd, UK) [6]

3 REQUIREMENTS FOR CASTER ROLL REFURBISHING

In order to optimize roll reclamation and service life, cladding consumables have to be designed with the aim of preventing the failure mechanisms described above. The most important requirements for continuous caster roll clad layers are listed below [6, 7]:

- Elevated temperature oxidation resistance.
- Resistance to localized pitting and crevice corrosion.
- Maximum resistance to stress corrosion cracking and corrosion fatigue.
- Maximum tempering resistance to prevent softening during service.
- Resistance to thermal and thermomechanical fatigue.
- High thermal conductivity.
- Low coefficient of thermal expansion.
- High hardness and resistance to abrasive and adhesive wear.
- High strength and toughness.
- Good weldability.
- Reasonably inexpensive.

In order to obtain the required hardness, strength, wear resistance and corrosion resistance required in this application, stainless steels with martensitic microstructures are preferred. A martensitic structure ensures high hardness, strength and wear resistance, as well as a low coefficient of thermal expansion (and hence improved thermal fatigue resistance). A minimum of 12 % chromium provides adequate corrosion and elevated temperature oxidation resistance. The composition should be adjusted to guarantee minimal retained austenite for improved thermal fatigue resistance (austenite and martensite exhibit a large coefficient of thermal expansion/contraction mismatch), and a δ -ferrite

content of less than 9 %. The presence of ferrite is believed to reduce the hardness and strength of the deposit, but small amounts of δ -ferrite may improve machinability after cladding. The addition of controlled amounts of nickel and molybdenum improves the high temperature properties. Nickel also raises the toughness of the deposit, whereas molybdenum improves resistance to localized corrosion.

Further improvement can be obtained by substituting part of the carbon normally added to martensitic stainless steels with nitrogen. Nitrogen-enhanced 12 % chromium martensitic stainless steels exhibit faster repassivation kinetics, a fine homogeneous distribution of nitride precipitates which inhibit grain growth, and an increase in the stability and passivity range of the oxide film [2]. This results in improved oxidation and corrosion resistance, higher strength, improved impact toughness, enhanced temper resistance, and higher resistance to thermal and thermomechanical fatigue. These improved properties result in considerably less material loss during service, as shown in Figure 6 for unmodified 420 and 414, compared to nitrogen-enhanced 414 (labelled 414N) martensitic stainless steel.

4 CURRENT CLADDING PRACTICE AT ARCELOR-MITTAL, VANDERBIJLPARK

Current production procedures for roll refurbishing consist of preparing the roll body, followed by weld overlaying using submerged arc cladding. Prior to welding, all existing defects are machined away, often necessitating deep machining into the base metal. The forged base of a caster roll is typically a low-alloy steel containing sufficient Cr, Ni, Mo and V to prevent catastrophic failure and a high enough yield stress and Young's modulus to minimize deflection and flexing during service.

In order to refurbish the roll, build-up of a buffer material to a certain diameter is often required prior to the

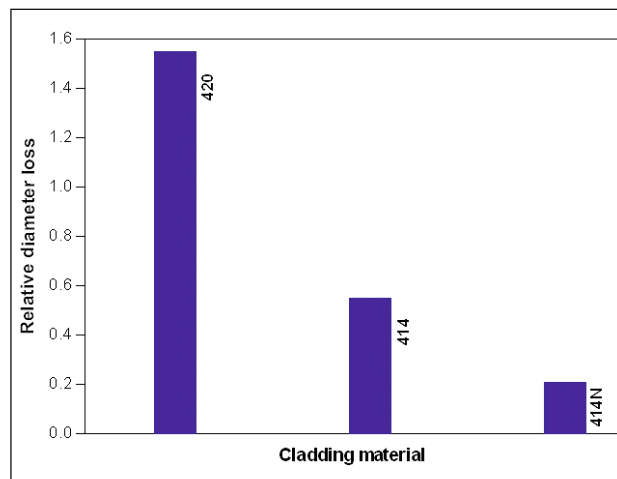


Figure 6 – Comparison between the relative diameter loss of continuous casting rolls surfaced with 420, 414 and nitrogen-enhanced 414N cladding material [7]

application of the final hardfacing layers. A low-carbon buffer layer with a bainitic structure, ensuring excellent impact toughness and high strength, is used to build up the rolls using submerged arc welding (supplied by Welding Alloys South Africa (Pty) Ltd (WASA) under the trade name of BUF-S). Welding of the buffer layer is followed by the application of one layer of a 17 % chromium 430-type ferritic stainless steel. This layer minimizes the dilution effect of the low-chromium buffer layer and ensures that the chromium level of the martensitic cladding rapidly approaches the required level of between 13 and 14 %.

The rolls are then surfaced with a 414N-S wire, a nitrogen-bearing, low-carbon tubular submerged arc welding wire that produces a martensitic deposit with a composition optimized for corrosion resistance. This wire is custom produced by Welding Alloys South Africa (Pty) Ltd (WASA) for Arcelor-Mittal under the trade name of Chromecore 414N-S. The typical all-weld metal chemical composition of this consumable is shown in Table 1. The amount of cladding deposited is determined by the need to resurface the rolls to the required diameter of 230 mm. The chromium content of the fifth layer is required to be between 13 and 14 %, and the hardness of a three-layer deposit a minimum of 38 HRC (hardness on the Rockwell C scale). The typical mechanical properties of an all-weld metal 414N-S deposit are shown in Table 2.

Single-wire submerged arc welding utilizing neutral flux and a wide weave (25 to 30 mm) to facilitate machining and provide a more uniform deposit (resulting in a bead width of 30 to 40 mm) is used (see Figure 7). A preheat temperature of 230 °C and a maximum interpass temperature of 350 °C are specified. Cladding is usually performed above the M_s (martensite start) temperature to avoid brittle fracture and to minimize the risk of hydrogen-induced cracking.

Although it is widely accepted that this technology dispenses with post-weld heat treatment, the client preferred that the rolls be stress relieved prior to being placed in service. An investigation was therefore launched to study the effect of various combinations of post-weld heat treatment temperatures and times on the deposit hardness and the extent of carbide precipitation. The results of this investigation are considered below.

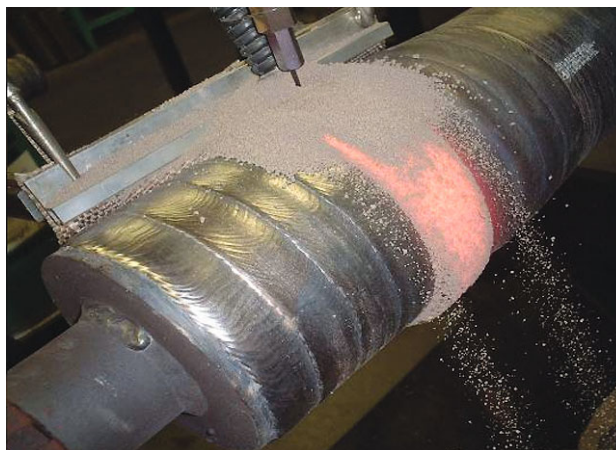


Figure 7 – Submerged arc cladding of continuous caster rolls using Chromecore 414N-S wire

(Photograph courtesy of Welding Alloys Ltd, UK) [7]

5 THE INFLUENCE OF POST-WELD HEAT TREATMENT ON THE 414N-S DEPOSIT HARDNESS AND EXTENT OF CARBIDE PRECIPITATION

In order to examine the effect of post-weld heat treatment on carbide precipitation and the hardness of the weld deposit, a low-alloy steel test plate was surfaced with 414N-S. A four-layer clad deposit was submerged arc welded onto the base plate using Chromecore 414N-S tubular wire and welding parameters similar to those used in production. Preheating of 230 °C was applied, and the measured interpass temperature ranged between 290 °C and 355 °C. Samples of this test plate were exposed to two standard post-weld heat treatments applied in production, i.e. 480 °C for 5 h, and 580 °C for 8 h. A third post-weld heat treatment, designed on the basis of a Hollomon-Jaffe temper parameter similar to that of the second production post-weld heat treatment (at 580 °C for 8 h), was applied at 620 °C for 1 h.

The measured hardness values of the four layers in each weld deposit in the as-welded condition and after post-weld heat treatment are given in Table 3. Each hardness value represents the average of at least three measurements.

Table 1 – Typical all-weld metal chemical composition of 414N-S (weight %, balance Fe)

C	Mn	Si	Cr	Ni	Mo	N
0.08	1.00	0.60	13.5	4.30	0.50	0.10

Table 2 – All-weld metal mechanical properties of a 414N-S deposit (room temperature)

Yield strength	Tensile strength	Reduction in area	Elongation	Hardness	Impact toughness
1 020 MPa	1 110 MPa	43 %	18 %	42-45 HRC	19 J (As-welded) 35 J (Heat treated)*

* Heat treated at 580 °C for 8 h.

Table 3 – Average hardness of four layers of 414N-S weld deposit, measured in the as-welded condition, and after three different post-weld heat treatments

Sample condition	Hardness (HRC)			
	First layer	Second layer	Third layer	Last layer
As-welded	44.1	46.2	45.9	45.0
PWHT @ 480 °C for 5 h	45.3	47.9	47.0	47.6
PWHT @ 580 °C for 8 h	40.0	40.2	40.3	42.3
PWHT @ 620 °C for 1 h	30.1	31.2	29.8	30.9

HRC: hardness on the Rockwell C scale.

Metallographic specimens were prepared by sectioning samples from the clad layer after welding, and after each of the three post-weld heat treatments specified above. The samples were mounted in resin and polished to a mirror finish. In order to identify any carbide precipitates within the microstructure, the samples were electrolytically etched in a saturated oxalic acid solution (according to the procedure detailed in Practice W of ASTM 763-93) [8]. Optical micrographs of the last two layers of the clad specimens are shown in Figures 8 to 11. The first two layers were not evaluated since the absence of a 17 % chromium buffer layer is likely to result in extensive chromium dilution in the first runs.

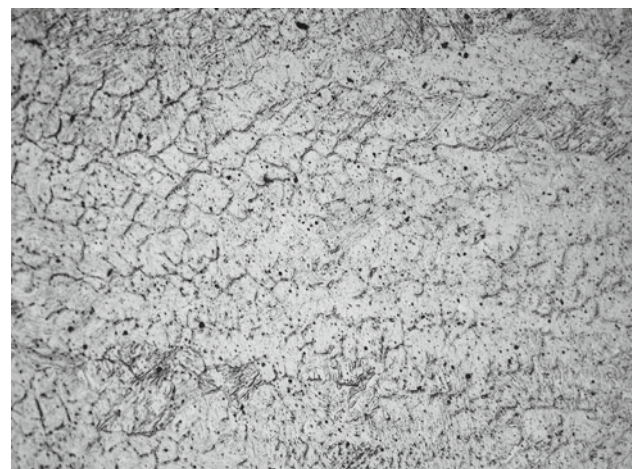
Examination of the 414N-S deposit in the as-welded condition [Figures 8 a) and b)] revealed a martensitic microstructure with some δ -ferrite located at the cores of the original dendrites that formed on solidification. The observed δ -ferrite content was within acceptable limits, and no retained austenite was evident, resulting in acceptable hardness levels (Table 3). Isolated carbide particles were observed, however, the absence of continuous films of grain boundary carbide indicates that the deposit was not sensitized by the weld thermal cycle. Examination at a higher magnification (Figure 12) revealed that the observed carbide particles were not specifically associated with the δ -ferrite phase.

The optical micrographs shown in Figures 9 a) and b) suggest that stress relieving at 480 °C for 5 h (current practice) did not result in extensive grain boundary carbide precipitation, and the deposit microstructure appears very similar to that of the as-welded specimen. Table 3 also indicates that this heat treatment did not reduce the deposit hardness to any significant extent. Although the amount of stress relieving obtained at 480 °C is likely to be limited, this heat treatment is unlikely to cause any detrimental changes to the mechanical properties or corrosion resistance of the deposit.

As shown in Table 3, heat treating the 414N-S deposit at 580 °C for 8 h resulted in a reduction in hardness, although the measured hardness was still within the specified range. Figures 10 a) and b) indicate that this standard heat treatment resulted in extensive carbide precipitation. Within the last layer, the carbides appeared to be associated with the δ -ferrite phase (Figure 13), whereas in the third layer the carbides seemed to outline the prior austenite grain boundaries.



a) The third layer

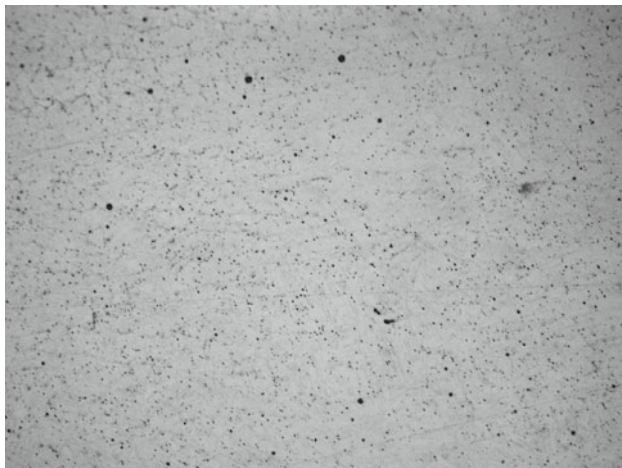


b) The last layer

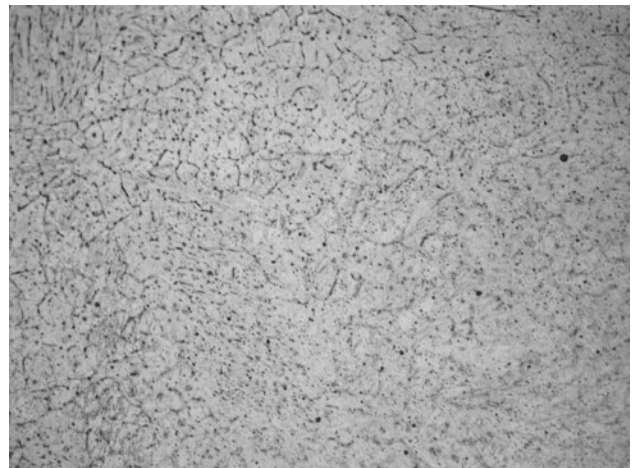
(Magnification: 450x)

Figure 8 – Optical micrographs of a 414N-S deposit in the as-welded condition

Table 3 indicates that post-weld heat treatment at 620 °C for 1 h caused an unacceptable reduction in deposit hardness below the specified hardness range for this application. This heat treatment also resulted in extensive carbide precipitation at the prior austenite grain boundaries and the martensite lath boundaries, as shown in Figures 11 a) and b). Although the extent of chromium depletion caused by the observed carbide precipitation was not evaluated during the course of this investigation, the structure is likely to be sensitized, and may exhibit increased susceptibility to localized corrosion and the initiation of stress



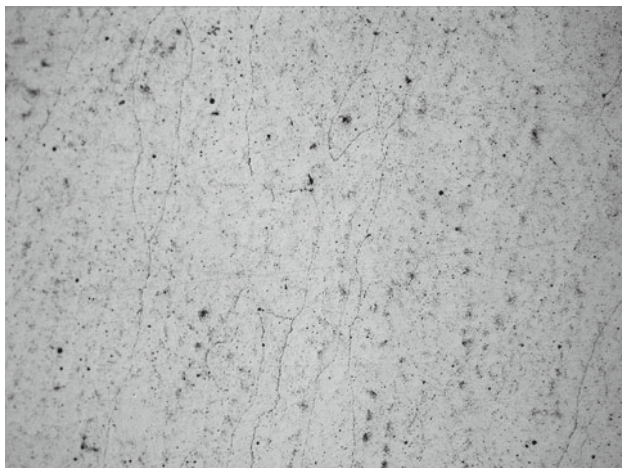
a) The third layer



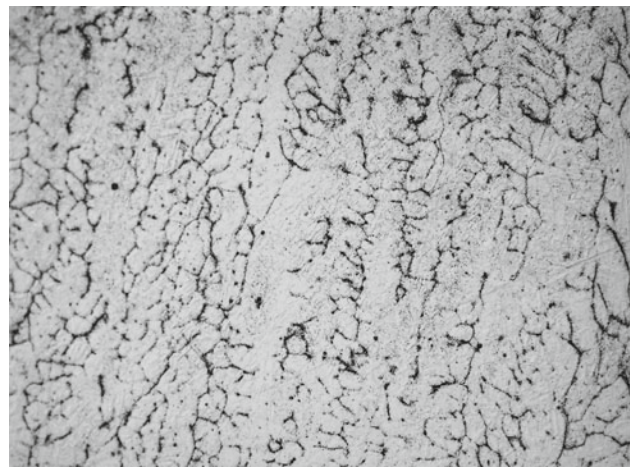
b) The last layer

(Magnification: 450x)

Figure 9 – Optical micrographs of a 414N-S deposit after post-weld heat treatment at 480 °C for 5 h



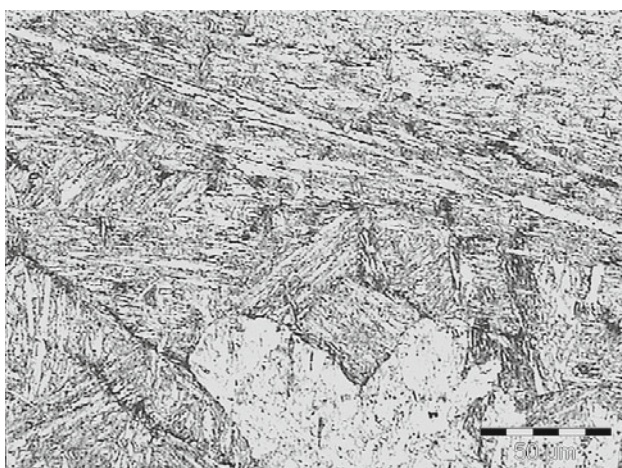
a) The third layer



b) The last layer

(Magnification: 450x)

Figure 10 – Optical micrographs of a 414N-S deposit after post-weld heat treatment at 580 °C for 8 h



a) The third layer



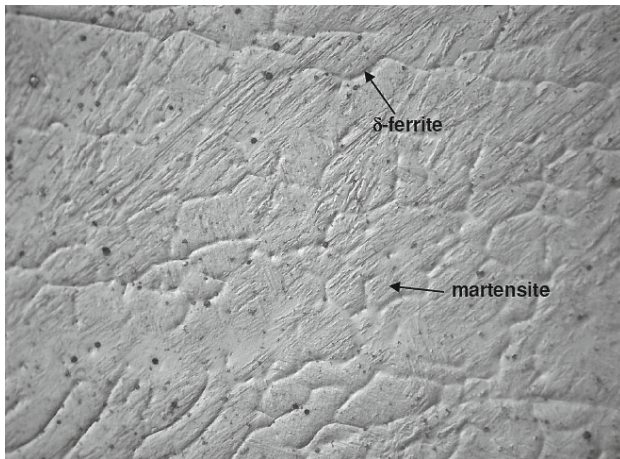
b) The last layer

(Magnification: 450x)

Figure 11 – Optical micrographs of a 414N-S deposit after post-weld heat treatment at 620 °C for 1 h

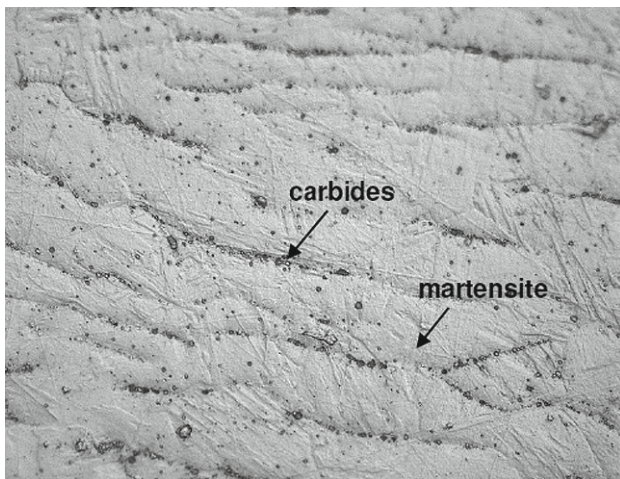
corrosion cracks during service. This heat treatment, although likely to result in higher levels of stress relief

during post-weld heat treatment, is therefore not recommended.



(Magnification: 900x)

Figure 12 – Optical micrograph of the last layer of a 414N-S deposit in the as-welded condition



(Magnification: 900x)

Figure 13 – Optical micrograph of the last layer of a 414N-S deposit after post-weld heat treatment at 580 °C for 8 h

6 CONCLUSIONS

- Nitrogen-alloyed 12 % chromium martensitic stainless steels offer a unique combination of high hardness, wear resistance, resistance to high temperature oxidation, corrosion resistance and resistance to thermal and mechanical fatigue. This combination of properties renders these materials the alloys of choice for refurbishing continuous caster rolls.
- Partially substituting carbon with nitrogen reduces the susceptibility of 12 % chromium martensitic stainless steels to inter-bead corrosion caused by sensitization. The lower-carbon martensite that forms in these

deposits and the reduced tendency towards chromium-rich carbide precipitation during cooling negate the need for post-weld heat treatment after cladding.

- The properties of a nitrogen-enhanced 414N-S deposit in the as-welded condition satisfy the requirements of minimal retained austenite and δ -ferrite, and limited $M_{23}C_6$ carbide precipitation. The hardness in the as-welded condition is well above the specified minimum value for this application. If a post-weld stress relief heat treatment is required, it should be performed at a temperature no higher than 500 °C to prevent softening and carbide precipitation.

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