

Scale Formation at High Temperature



Scale (or iron oxide) forms on the surface of steel during continuous casting, slab reheating and subsequent rolling process. For heat-treated products, especially for quench-and-tempered products, scale can reduce heat transfer efficiency during quenching. Therefore, scale is deemed to be undesirable for the quenching process and should be blasted off prior to austenitization. Although scale formation at high temperature has been studied extensively, it is still not completely clear how the steel chemistry and slab reheating atmosphere impact scale formation and its adherence to the steel. This article discusses scale formation on four steels in a simulated slab reheating condition with different excess oxygen levels.

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Introduction

Oxidation is a natural event during the reheating of steel slabs. It is important to understand how steel oxidizes at high temperature since the end results from this process could affect the surface quality of final products. There are several phenomena associated with high-temperature oxidation, such as (1) surface cracking (or hot shortness) due to the liquid Cu penetration along the grain boundaries of austenite, (2) the formation of adhesive scale and (3) the formation of red scale. All these can increase the difficulties for subsequent processing and affect the surface quality of final products. Cu hot shortness is the most severe defect on the surface of a hot-rolled steel product and has been studied extensively.^{1–4} For steels made from the scrap-based electric arc furnace (EAF) process, the probability of having Cu hot shortness is high if residual elements, such as Sn, Sb and As, are not controlled properly. All these elements can exacerbate Cu hot shortness by enhancing the formation of liquid Cu phase at the scale/steel interface and accelerating the Cu penetration into the steel along the austenite grain boundaries, weakening the grain boundaries. Over the last several decades, one of the solutions to mitigating Cu hot

shortness in steels is by adding proper amounts of Si and Ni. Si promotes the occlusion of Cu into scale, while Ni increases the solubility of Cu in austenite. The presence of these elements reduces Cu segregation to the scale/steel interface and prevents Cu from forming a liquid phase at the scale/steel interface. However, the addition of Si introduces another issue, i.e., increasing the propensity of forming a Si-rich phase (or fayalite) at the scale/steel interface, which increases the bonding force between scale and steel, forming a sticky scale. This will increase the difficulties of scale removal for the subsequent processes.^{5–9} The third phenomenon associated with high-temperature oxidation is the formation of red scale, i.e., red-colored scale on the surfaces of hot-rolled steel products.^{10, 11} Compared to the studies conducted to analyze the effect of steel chemistry, fewer studies were carried out to analyze the effect of heating conditions on the formation of these defects. In a recent study conducted by Haapakangas et al., the effects of furnace atmosphere with different levels of water vapor on the oxidation rate, formation of adhesive scale, and porosity in scale were investigated.¹² In another study carried out by Gaiser et al., the oxidation phenomenon at an ESP (Endless

Table 1

Chemical Compositions of Steels, wt. %

Steel	C	Mn	Si	Al	Cu	Ni	Cr	Mo	Sn	Cu/Ni
A	0.161	1.20	0.06	0.030	0.29	0.14	0.15	0.03	0.010	2.07
B	0.175	1.23	0.18	0.029	0.25	0.14	0.08	0.03	0.010	1.79
C	0.057	1.31	0.21	0.034	0.32	0.17	0.16	0.04	0.009	1.88
D	0.073	1.26	0.33	0.029	0.34	0.32	0.61	0.03	0.010	1.06

Strip Production) mill was investigated.¹³ The effects of furnace atmosphere with different levels of water vapor on oxidation were also analyzed. More studies are needed to better understand how hydrogen and oxygen affect scale formation. To shed more light on this topic, simulated slab reheating of four steels under four different conditions (1%, 3%, 5% and 10% oxygen) was studied and the results are presented in this article.

Materials and Experimental Setup

Materials and Experimental Setup

The steels used in this study are four low-carbon and medium-carbon steels with various of Si, Cr, Cu and Ni additions. The detailed compositions of the steels are shown in Table 1. Among the four steels, Steel A has the lowest Si addition; the Cr content in this steel is medium compared to the other three steels. Steels B and C have similar additions of Si, but the Cr content in Steel C is higher than that in Steel B. The Si, Cu, Ni and Cr contents in Steel D are the highest among the four steels used in this study. The high levels of these elements were aimed for since this steel was made to the specification of weathering steel. The Cu contents of these steels vary in a narrow range. The Ni contents in Steels A, B and C are similar. However, the Ni content in Steel D is much higher, resulting in a low Cu/Ni ratio (1.06) for Steel D. A low Cu/Ni ratio is desirable to prevent steel from Cu hot shortness. Although the steels were made with a scrap-based EAF steelmaking process, the contents of Sn in these steels are relatively low. Since Sn can exacerbate Cu hot shortness, a low level of Sn is also desirable. All the samples used for the heating experiments were cut from 1.25-inch-thick hot-rolled steel plates. One side of the samples (top surface) was kept in the as-received

condition, while the other side (bottom surface) was machined to investigate how the initial surface condition affects oxidation during the simulated slab reheating. The experimental setup is shown by the schematic drawing in Fig. 1. The samples were inserted into a FeCrAl horizontal tube furnace, and the system was sealed and flushed with argon prior to heating. After purging, the furnace was programmed and heated to the desired exposure temperature (1,210°C). It took about 60 minutes to reach 1,000°C, followed by an additional ~30 minutes to reach and stabilize at the aim temperature of 1,210°C. Temperature stabilization was confirmed using thermocouples positioned near the samples. Once the furnace reached the intended exposure temperature, the samples were held (soaked) for 30 minutes. At the end of the exposure period, the samples were removed from the furnace and cooled in air. In total, the heating and soaking sequence lasted approximately 120 minutes. All exposures were held under flowing gas of 250 mL/minute. The furnace atmosphere is a mixture of O₂ and N₂. Four different combinations of O₂ and N₂, 1% O₂ + 99% N₂, 3% O₂ + 97% N₂, 5% O₂ + 95% N₂, and 10% O₂ + 90% N₂, were used for the study. The heating parameters and furnace atmosphere are shown in Table 2.

Figure 1

Experimental setup.

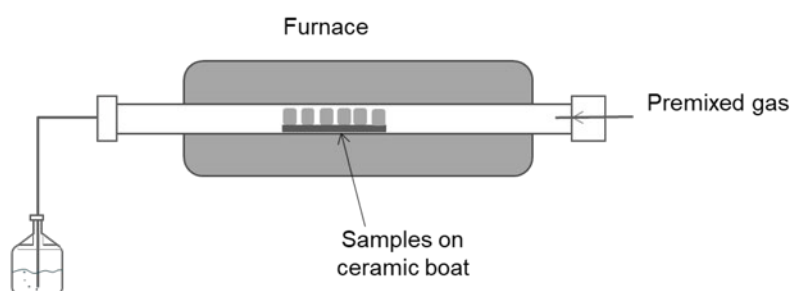


Table 2

Heating Parameters and Furnace Atmosphere

Temperature, °C	Heating time, min.	Soaking time, min.	Furnace atmosphere
1,210	90	30	1% O ₂ , balance N ₂
			3% O ₂ , balance N ₂
			5% O ₂ , balance N ₂
			10% O ₂ , balance N ₂

Sample Preparation for SEM Examination

After the heating experiment, all of the samples were handled with care and cold-mounted in epoxy to preserve the oxide layers on the sample surfaces for examination. The samples were ground with SiC papers down to

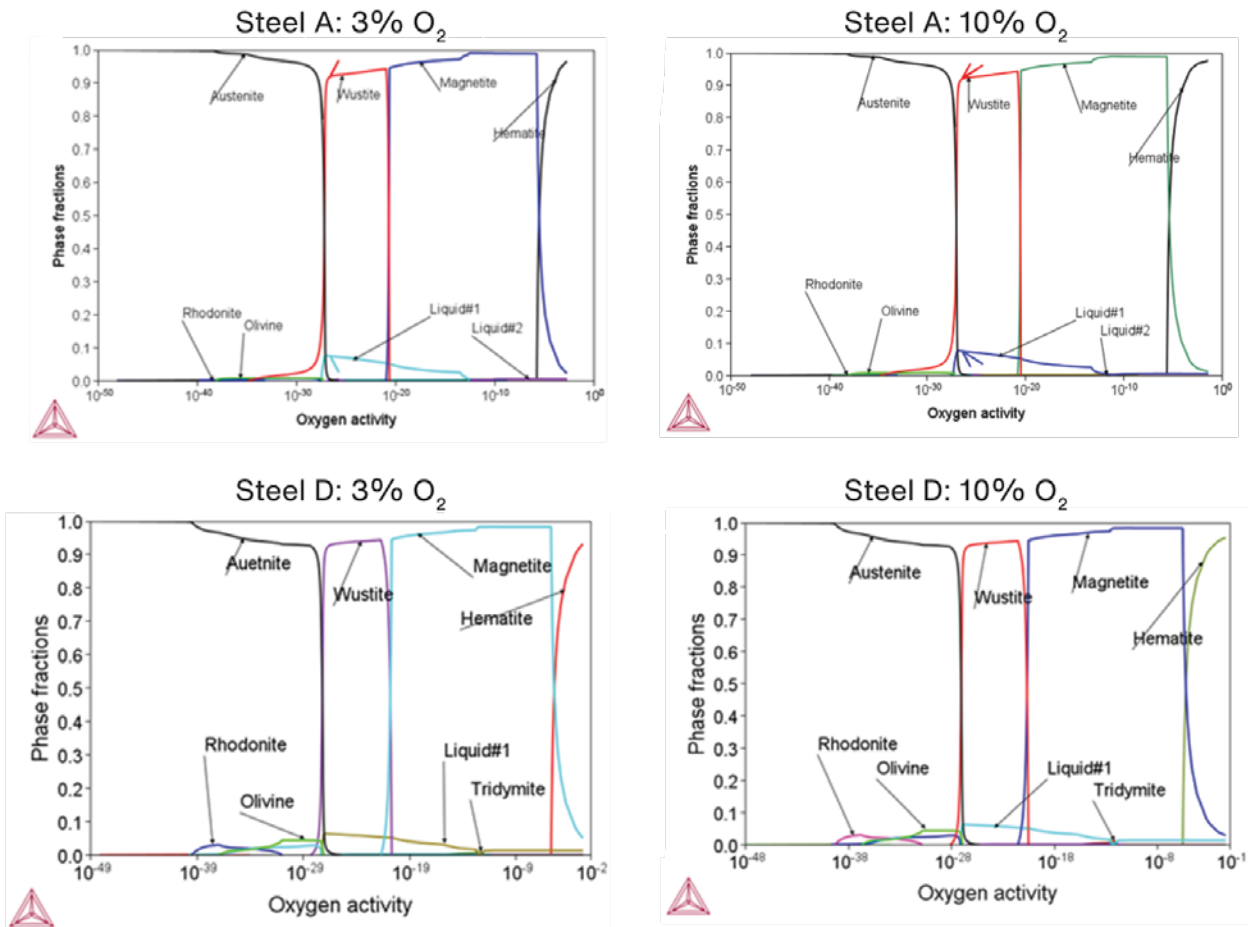
P2400, then polished with 0.25 μm diamond suspension in the final step. The samples were subsequently cleaned in acetone and ethanol in an ultrasonic bath. The samples were then coated using gold sputtering to ensure sufficient surface conductivity during imaging. Cross-sectional (CS) analysis was carried out to examine the oxide/metal interfaces using a Zeiss Sigma 300VP scanning electron microscope (SEM). Both secondary electron (SE) and backscattered electron (BSE) detectors were employed, in combination with an energy-dispersive x-ray (EDX) detector. Elemental distribution was mapped using energy-dispersive spectroscopy (EDS) at an acceleration voltage of 15 kV. Under these conditions, the spatial resolution of the technique is approximately 1 μm.

Thermodynamic Calculations

Thermodynamic calculations were performed for Steels A and D using the Thermo-Calc software TCFE steels/

Figure 2

Thermodynamic calculation on phase stability of Steels A and D.



Fe-alloys Database version 13. These calculations were performed in the equilibrium condition at 1,210°C, and the phase stability as a function of oxygen activity is plotted in Fig. 2. Oxides forming at lower oxygen activity are expected to be close to the metal surface, while those forming at higher oxygen activity are expected to be most likely at the surface exposed to the furnace atmosphere. As an example, the results obtained for Steels A and D under the conditions of 3% O₂ and 10% O₂ are presented here. It can be seen from Fig. 2 that both steels are expected to form same phases at equilibrium. Major phases, such as hematite (Fe₂O₃), magnetite (Fe₃O₄), and wüstite (FeO), are expected to form at high oxygen activity. Minor phases, such as rhodonite (MnSiO₃) and olivine (fayalite, Fe₂SiO₄), are expected to form at low oxygen activity while two liquid phases are expected to form at intermediate oxygen concentration. Steel D demonstrates a slightly more complex phase evolution where

one additional phase, tridymite (a polymorph of SiO₂) is expected to form. Comparing Steel A to Steel D at the same oxygen concentration, a difference in the phase fraction is observed in the low oxygen activity phases, e.g., olivine (fayalite) and rhodonite. The phases in Steel D seem to have a higher volume fraction compared to Steel A. This is suspected to be due to the high Si content in Steel D, which is the highest among the steels examined in this study.

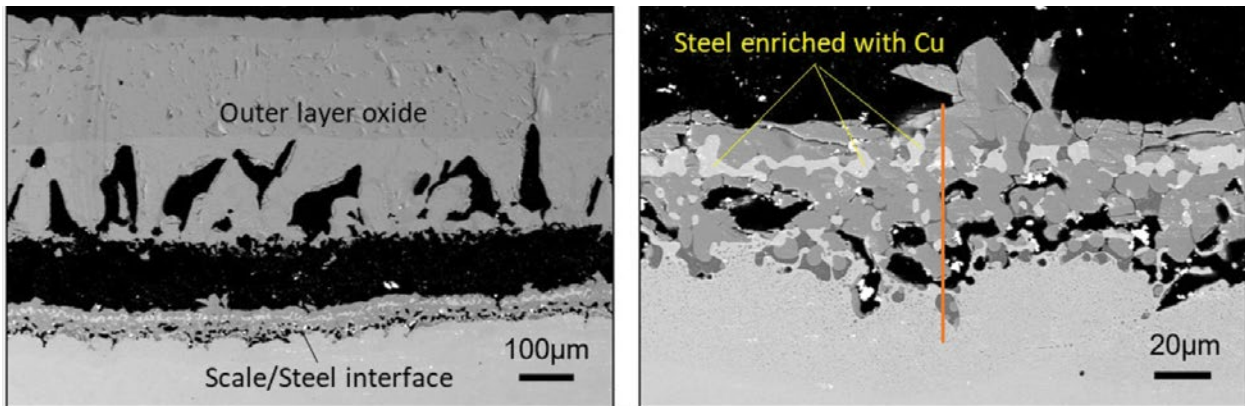
Experimental Results

Characterization and Measurement of Scale

Since the bottom surface of the samples (the machined side) was not in free contact with the furnace atmosphere, the oxidation process may have been affected. Therefore, only the top surface of the simulated samples was examined. A cross-section of a simulated sample (Steel B) is

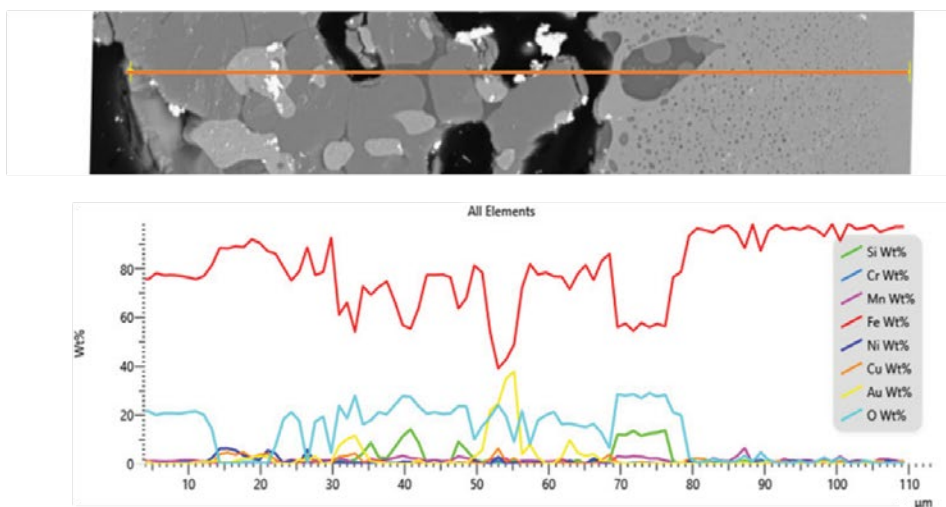
Figure 3

A cross-section of Steel B showing the through-thickness scale structure.



(a) Full thickness of scale

(b) Scale/steel interface



(c) EDS line scan

Table 3

Measurement of Scale Thickness

Steel	1% O ₂		3% O ₂		5% O ₂		10% O ₂	
	Internal, μm	Outer, μm	Internal, μm	Outer, μm	Internal, μm	Outer, μm	Internal, μm	Outer, μm
A	45–55	250–300	40–55	290–310	25–45	280–320	45–55	200–240
B	45–55	160–200	10–40	290–310	45–65	180–220	50	240–320
C	45–65	150–250	45–55	280–320	40–60	170–230	35–55	380–420
D	45–55	300–330	15–25	145–185	15–60	380–400	10–30	350–550

shown in Fig. 3. The full thickness of scale on the top surface (as-received surface) is shown in Fig. 3a. The scale consists of a thick outer layer and a thin layer at the scale/steel interface. Based on the oxygen concentration, it was determined that the very outer surface is hematite (Fe_2O_3), followed by magnetite (Fe_3O_4), and then wüstite (FeO) near the scale/steel interface. It can be seen that porosity or pores formed in the middle layer of the scale, which separated the outer layer from the steel substrate. Although the simulated samples were carefully handled, a majority of the outer layer has broken off the samples, making the measurement of scale thickness difficult. Nevertheless, an attempt was made to estimate the scale thickness of the simulated samples, and the results are shown in Table 3. It can be seen that Steel D (high Si), which was heated with 10% O_2 , has the thickest scale layer, while Steel A (low Si), which was heated with 1% O_2 , has the thinnest scale layer. For the same heating condition, the scale layer on Steel D is thicker than that on Steel A, except for the 3% O_2 condition for which scale on Steel D partially broke off, resulting in a lower measurement. Internal oxidation was observed in all the simulated samples with varying thickness. In some cases, some steel islands enriched with Cu were seen in the scale layer near the scale/steel interface, as shown in Figs. 3b and 3c. Oxides enriched with Si are also seen

in the scale near the scale/steel interface. Since Cu and Si enrichment at the scale/steel interface plays an important role in Cu hot shortness and the formation of adhesive scale, the scale layer near or at the scale/steel interface was examined closely.

Heating With 1% O₂

SEM images of the cross-sections prepared from the samples heated with 1% O_2 are shown in Fig. 4. The characterization focused on the scale/steel interface since this is the location where Si- and Cu-rich phases could form during the simulated heating treatment. It can be

Figure 4

Samples heated with 1% O_2 .

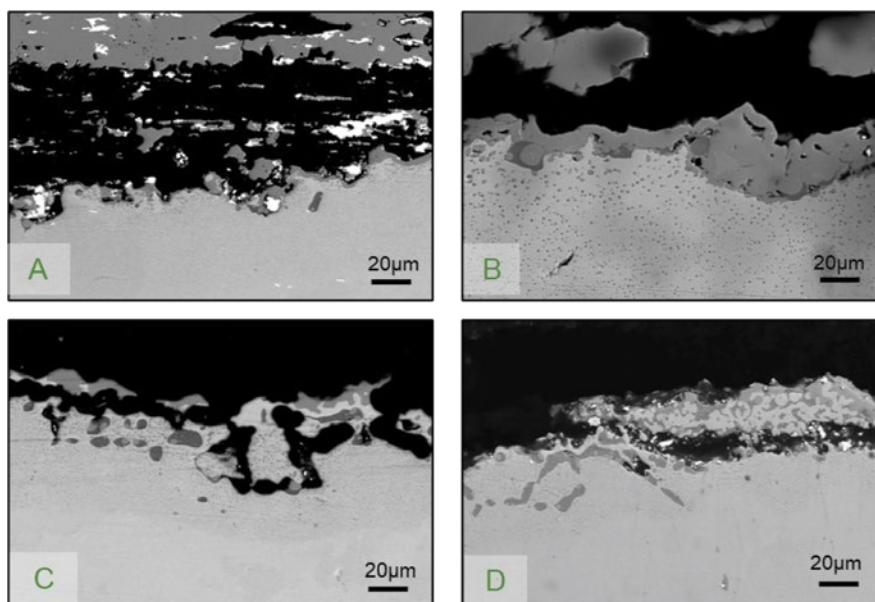
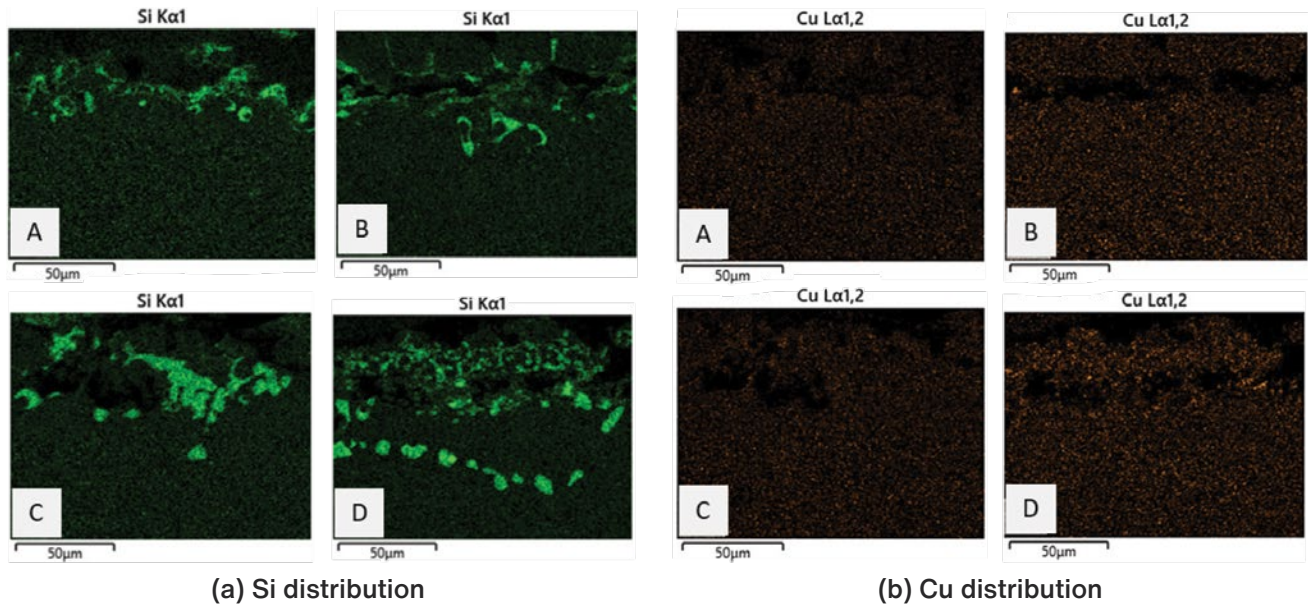


Figure 5

Si and Cu distributions at the scale/steel interface of samples heated with 1% O₂.



seen from the SEM images that the roughness of the interface is different among the four samples. Under this heating condition (1% O₂), the scale/steel interface of Steel A is the smoothest among the four steels examined, while that of Steels C and D are the roughest. Compared to Steel A, Steels C and D have higher Si and Cr, both can cause a rough scale/steel interface. From the EDS elemental mappings shown in Fig. 5, it can be seen that Si-rich phase segregated at the scale/steel interface. It appears that the volume fraction of Si-rich phase is higher in Steels C and D than Steels A and B. Although the Si content in Steel B is similar to that of Steel C, the lower Cr in Steel B may have reduced the propensity of Si segregating to the scale/steel interface, resulting in a lower volume fraction of Si-rich phase in this steel. It is interesting to note that in Steel D, some finger-like oxides (or anchors) extended from the scale/steel interface into the steel substrate. This kind of scale is difficult to remove by descaling during hot rolling. From the Cu distribution maps shown in Fig. 5b, it appears

that Cu segregation is relatively light in these steels under this heating condition. Only a few small spots of Cu-rich phase were observed in Steel D. No mass liquid Cu phase or Cu films were observed either at the scale/steel interface or along the austenite grain boundaries below the scale/steel interface.

Figure 6

Samples heated with 3% O₂.

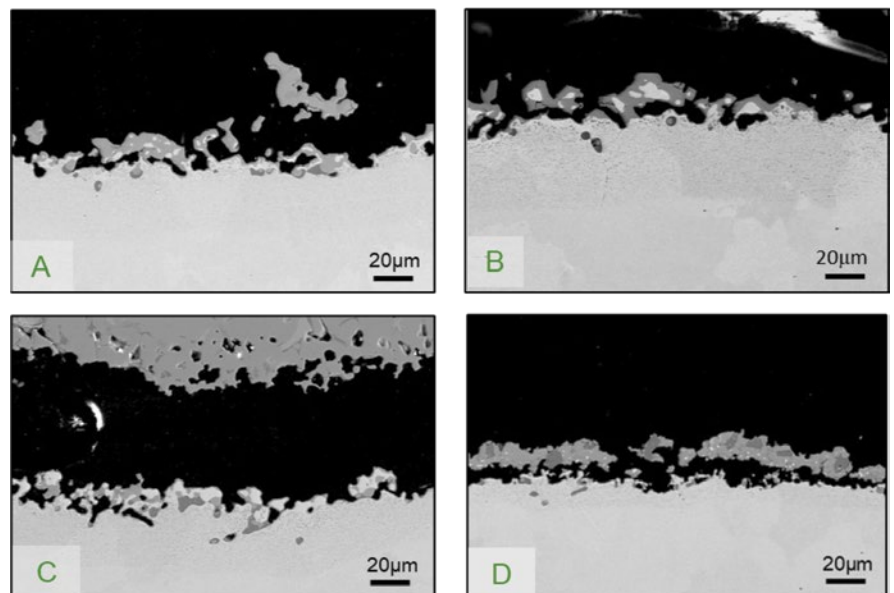
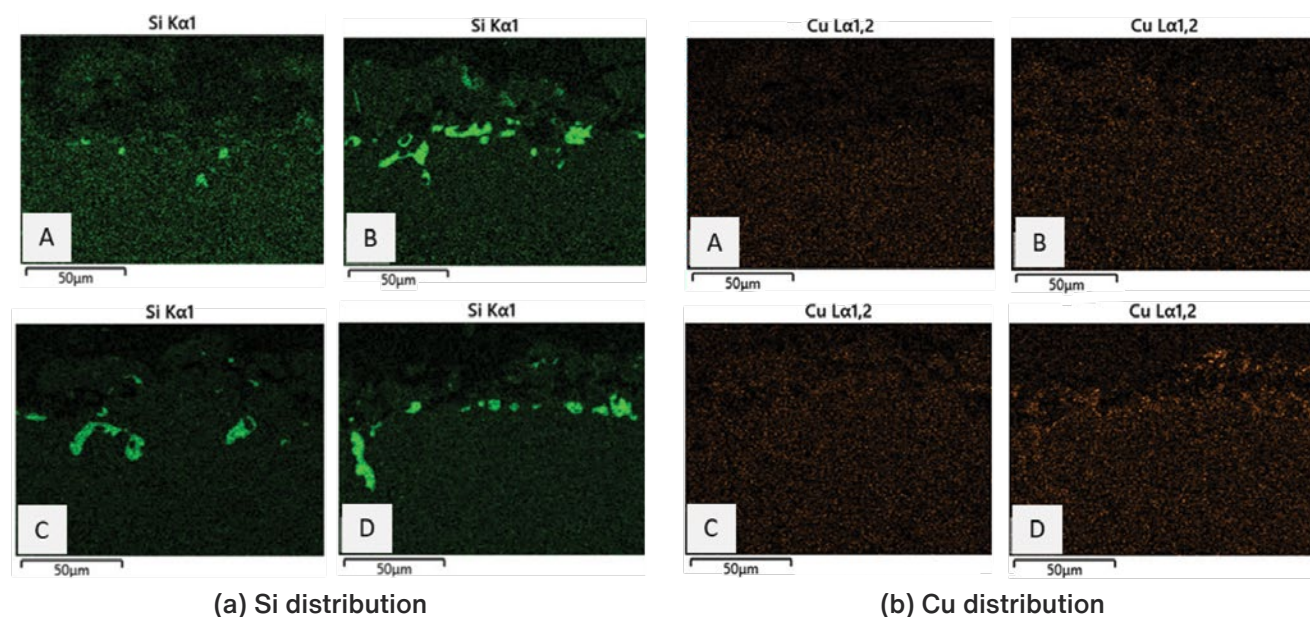


Figure 7

Si and Cu distributions in scale and at the scale/steel interface of samples heated with 3% O₂.



Heating With 3% O₂

SEM images of the cross-sections prepared from the samples heated with 3% O₂ are shown in Fig. 6. Similarly, the scale/steel interface was examined carefully to determine if any Si- and Cu-rich phases formed during the simulated heating treatment. It can be seen from the SEM images that the roughness of the interface is different among the four steels examined. Under this heating condition (3% O₂), the roughness of the scale/steel interface of these steels is similar. Compared to the samples heated with 1% O₂, the interface of these samples is smoother. From the EDS elemental mappings shown in Fig. 7, it can be seen that Si segregation at the scale/steel interface in these samples is lighter than that in the samples heated with 1% O₂. Si segregation in Steel A is the lightest among the four steels examined, while Si segregation in the other three steels is similar. From the Cu distribution maps shown in Fig. 7b, it appears that Cu segregation is only seen in Steel D. No mass liquid Cu phase or Cu

films were observed either at the scale/steel interface or along the austenite grain boundaries below the scale/steel interface in these samples.

Heating with 5% O₂

SEM images of the cross-sections prepared from the samples heated with 5% O₂ are shown in Fig. 8. It can

Figure 8

Samples heated with 5% O₂.

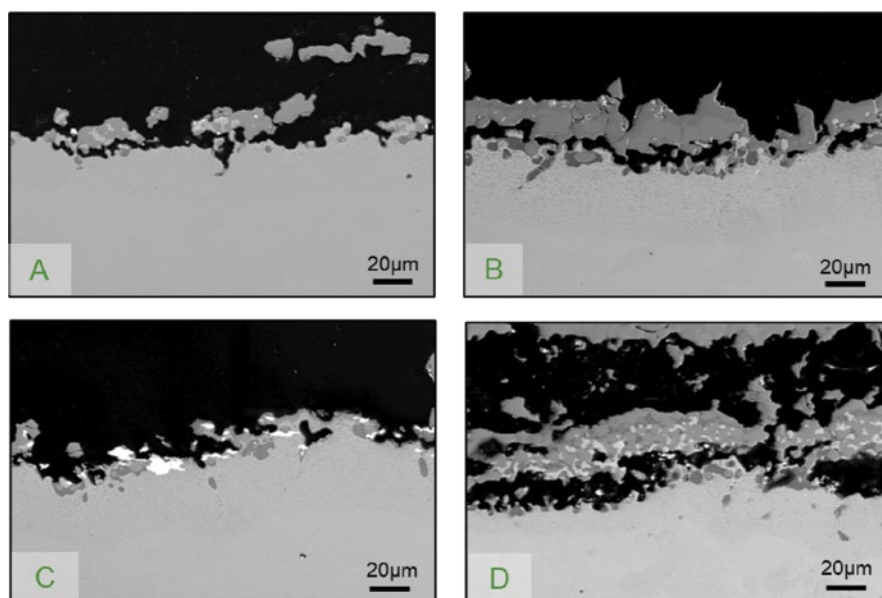
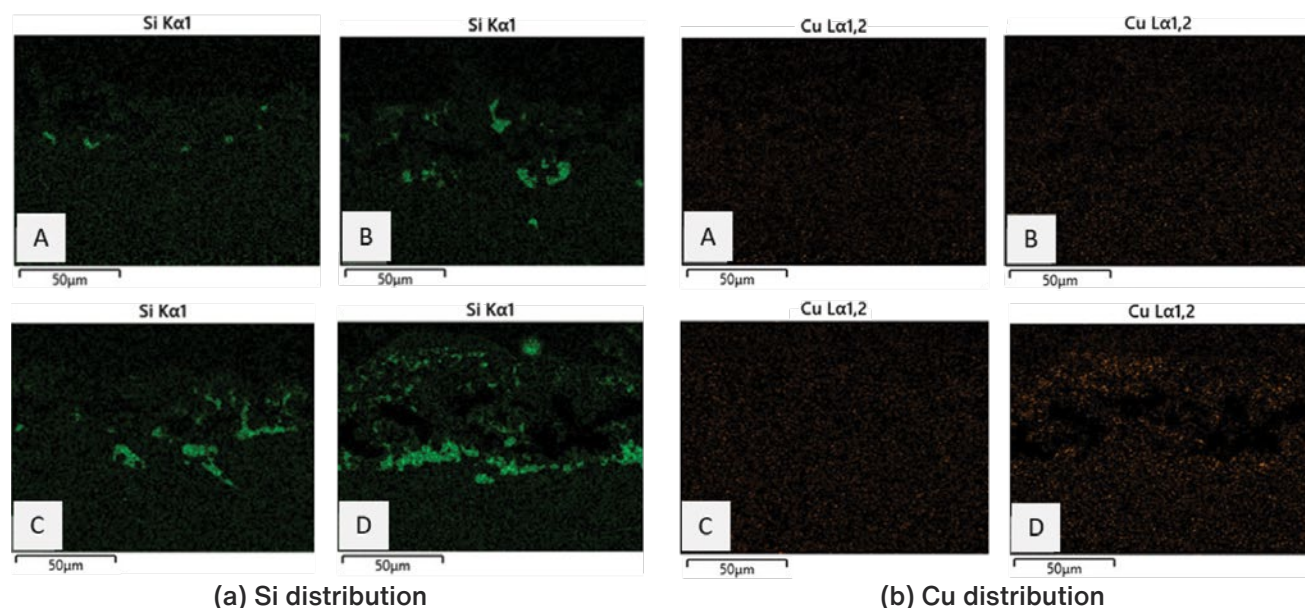


Figure 9

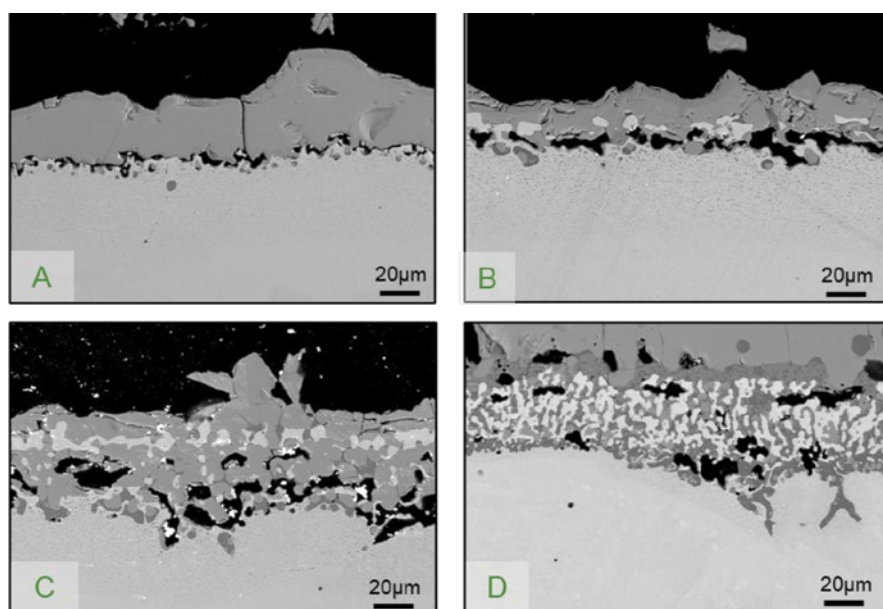
Si and Cu distributions in scale and at the scale/steel interface of samples heated with 5% O₂.



be seen from the SEM images that the roughness of the interface of the samples heated with 5% O₂ is similar to that of the samples heated with 3% O₂ shown in Fig. 6. The main difference among these four samples is that it appears in Steel D, more steel was occluded into the scale near the scale/steel interface (the light-colored phase in the scale). From the EDS elemental mappings shown in Fig. 9, it can be seen that Si segregation at the scale/steel interface in these samples is also similar to that in the samples heated with 3% O₂. Among the four steels, Si segregation in Steel A is the lightest, while Si segregation in Steel D is the heaviest. From the Cu distribution maps shown in Fig. 9b, it appears that Cu segregation is only noticeable in Steel D. No mass liquid Cu phase or Cu films were observed either at the scale/steel interface or along the austenite grain boundaries below the scale/steel interface in these samples.

Figure 10

Samples heated with 10% O₂.

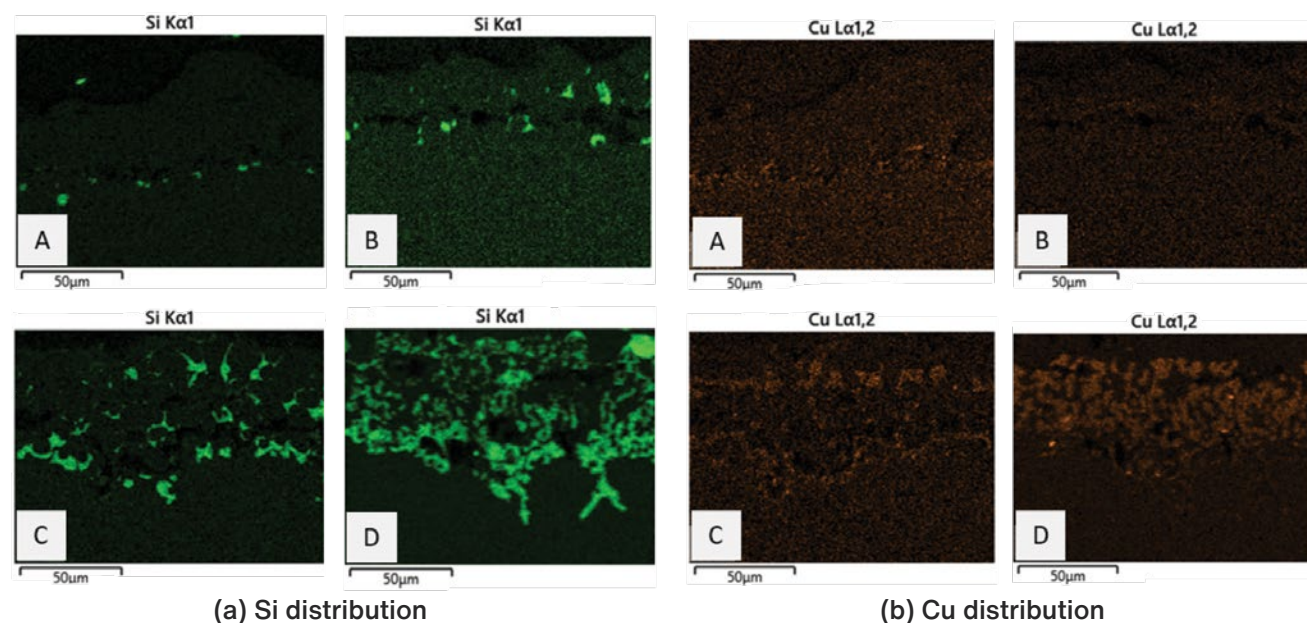


Heating With 10% O₂

SEM images of the cross-sections prepared from the samples heated with 10% O₂ are shown in Fig. 10. It can be seen from these SEM images that the scale/steel interface of Steels A and B is smoother than that of Steels C and D. It also appears that not much steel was occluded into the

Figure 11

Si and Cu distributions in scale and at the scale/steel interface of samples heated with 10% O₂.



scale near the scale/steel interface in Steel A. In the other three steels, Fe-rich phase (the light-colored phase) can be seen in the scale near the scale/steel interface. Especially in Steel D, a large volume fraction of steel was occluded into the scale. This is the most severe case among all the heating simulated samples. In this sample, finger-like Si-rich oxides extended into the steel substrate below the scale/steel interface, anchoring the scale layer into the steel substrate. This makes the scale more adhesive and difficult to be removed by descaling during hot rolling. From the EDS elemental mappings shown in Fig. 11, it can be seen that Si segregation at the scale/steel interface in Steel A is the lightest, while Si segregation in Steel D is the heaviest. From the Cu distribution maps shown in Fig. 11b, it appears that Cu segregation is noticeable in Steels C and D. However, except for a few small spots of Cu-rich phase in Steel D, no mass liquid Cu phase or Cu films were observed either at the scale/steel interface or along the austenite grain boundaries below the scale/steel interface in these samples.

Discussion

Effect of Steel Chemistry

During slab reheating, depending on steel chemistry, Si-rich oxide and liquid Cu phase could form at the scale/steel interface. The former can form an adhesive scale layer at the scale/steel interface, making descaling difficult. The latter can penetrate the steel substrate along the austenite grain boundaries, causing Cu hot shortness (or cracking). It is well known that Si promotes occlusion of

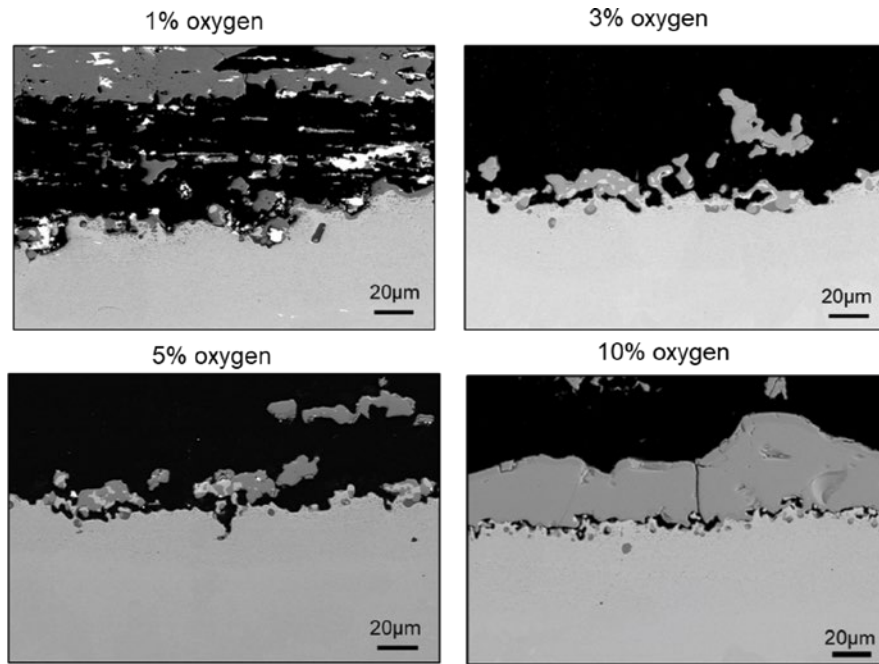
Cu-rich phase into scale, thereby reducing Cu concentration at the scale/steel interface. Ni is known to increase the Cu solubility in austenite, thereby slowing down the formation of liquid Cu phase at the scale/steel interface. For the four steels examined in this study, Steel A has the lowest Si content, thereby it has the lowest propensity of forming Si-rich oxides at the scale/steel interface. As shown in Figs. 4–11, it is noticeable that segregation of Si-rich oxides at the scale/steel interface in this steel is indeed the lightest among the four steels examined. By contrast, Steel D has the highest Si level. The Cr content in this steel is also the highest among the four steels examined. The results shown in Figs. 4–11 indicate that segregation of Si-rich oxides at the scale/steel interface in this steel is the heaviest. This is not only due to the high Si level in this steel but also the high content of Cr. It is well known that Cr promotes the formation of stable Si-rich oxides, such as SiO₂ and Fe₂SiO₄ (fayalite), in steel at high temperature. The combination of Cr and Si produces more stable and adherent oxide scale compared to steels containing only Si. This is further demonstrated by comparing Steel B and Steel C. Steels B and C have similar Si levels. However, the Cr content in Steel C is higher than that in Steel B, resulting in heavier segregation of Si-rich phase at the scale/steel interface in Steel C. In terms of Cu hot shortness, Ni plays a very important role. As shown by the Cu equivalent developed by Wolf and Schwabe in Eq. 1,¹⁴ Ni significantly reduces the impact of copper on hot shortness. This suggests that the Cu limit set for avoiding hot shortness could be higher if Ni is added and the other tramp elements are kept low. For the four steels examined in this study, the Cu contents are in

the range of 0.25 to 0.34 wt. %. Although the Cu content in Steel D is the highest, the higher Ni addition in this steel effectively reduced the equivalent Cu to a very low level (<0.15). This significantly reduced the propensity of

Cu hot shortness for this steel. In fact, all these steel plates did not have any cracks on their surfaces. It can be seen from the SEM images and elemental mappings shown in Figs. 4–11 that no mass liquid Cu phase was observed at the scale/steel interface of these steels.

Figure 12

Scale/Steel interface of Steel A heated with 1%, 3%, 5% and 10% O₂.



$$Cu_{eq} = Cu + 10Sn - Ni$$

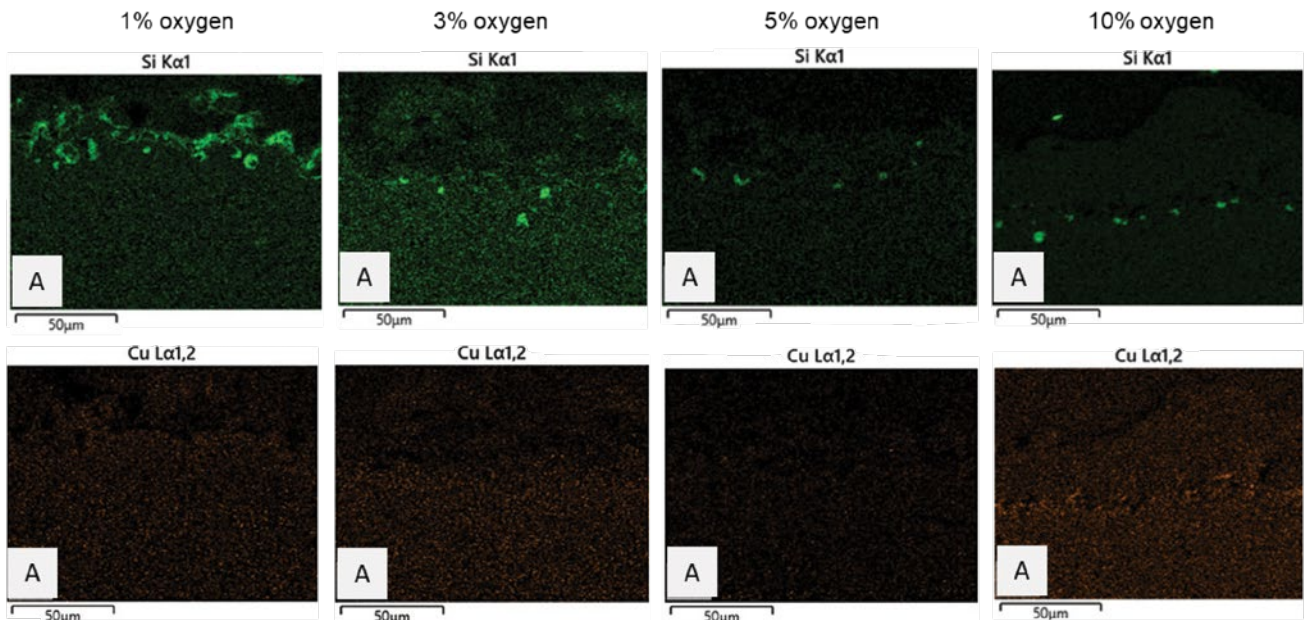
(Eq. 1)

Effect of Furnace Atmosphere (O₂ Level)

In addition to steel chemistry, the atmosphere, more specifically, the oxygen level, in the reheating furnace can also affect the formation of scale during slab reheating. In the present study, steel samples were heated with a furnace atmosphere at four different oxygen levels, 1%, 3%, 5% and 10% O₂. As two extreme cases, the results obtained from Steels A and D are shown in Figs. 12–15 and discussed here. It can be seen from the SEM images shown in Fig. 12

Figure 13

Element distributions at scale/steel interface of Steel A heated with 1%, 3%, 5% and 10% O₂.



that the roughness of the scale/steel interface produced for this steel (Steel A) is similar for the four heating conditions. Among the four steels, Steel A has the lowest Si content, which reduced the propensity of the formation

of Si-rich oxides at the scale/steel interface. As shown in Fig. 13, no mass Si-rich oxides were observed at the scale/steel interface. Cu segregation in this steel is also very light for the four heating conditions used. Except for

a few small spots under the 10% O₂ heating condition, no apparent liquid Cu phase was observed in these samples. The scenario for Steel D, however, is different. This steel has higher Si and Cr, both of which enhance the formation of adhesive scale and produce a rough scale/steel interface. As shown in Fig. 14, the scale/steel interface of samples heated with 1% and 10% O₂ appears to be rougher compared to that of the samples heated with 3% and 5% O₂. In addition to the rougher interface, finger-like Si-rich oxides anchored the scale layer into the steel substrate. This is further shown by the elemental mappings in Fig. 15. It can also be seen that it appears more steel was occluded into scale for the samples heated with 1% and 10% O₂. The mechanism for the 1% O₂ heating condition

Figure 14

Scale/Steel interface of Steel D heated with 1%, 3%, 5% and 10% O₂.

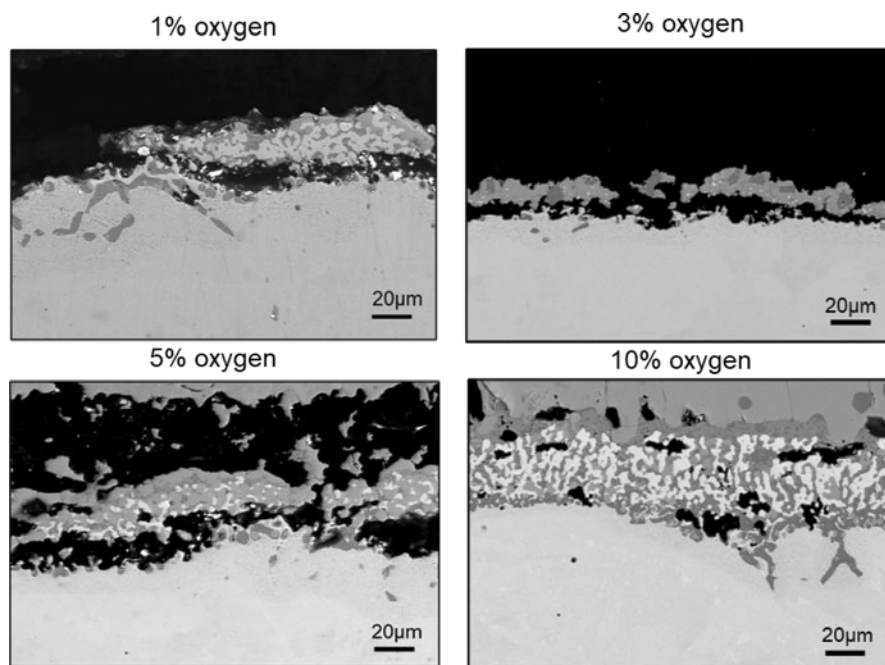
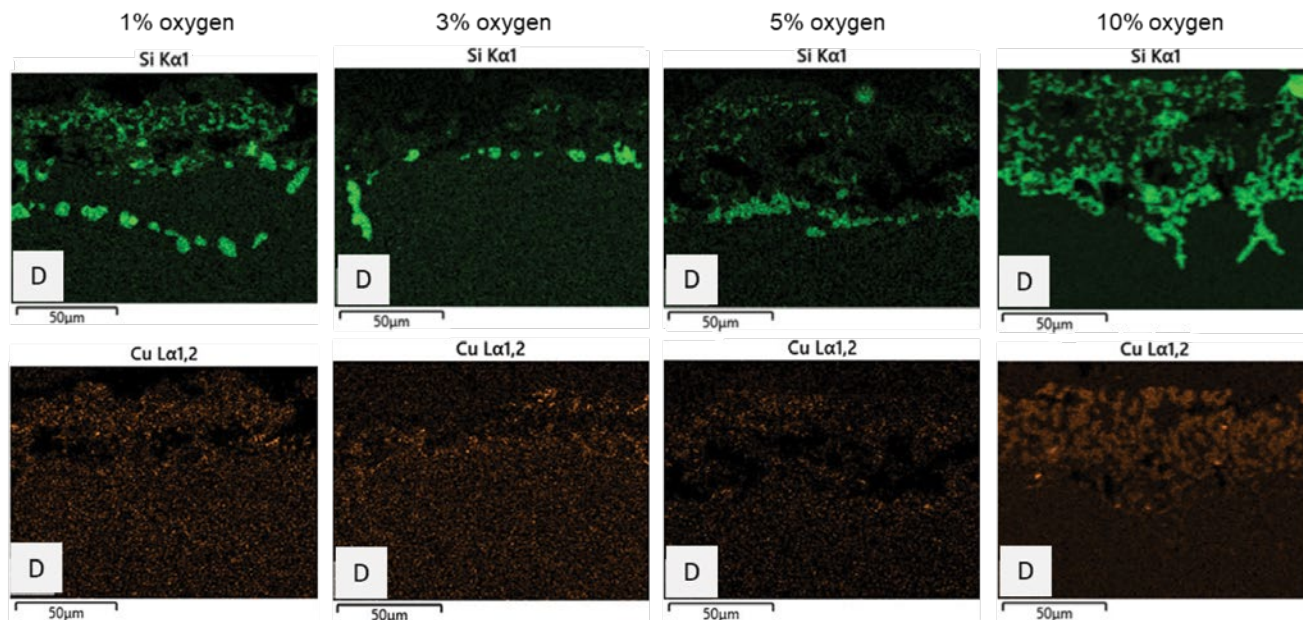


Figure 15

Element distributions at scale/steel interface of Steel D heated with 1%, 3%, 5% and 10% O₂.



is not clear. For the 10% O₂ heating condition, it is possible that O₂ attacks the austenite grain boundaries quicker than oxidizing the steel, resulting in some unoxidized steel islands, which were occluded into scale. Although Cu segregation was seen in the samples heated with the four conditions, no mass liquid Cu phase was observed at the scale/steel interface. Among the four conditions, heating with 10% O₂ produced heavier Cu segregation. Overall, it appears that heating with 3% and 5% O₂ produced smoother scale/steel interface and lighter Si and Cu segregation. A furnace atmosphere with about 3–5% excess oxygen is desirable to minimize the Si-rich oxide and Cu segregation at the scale/steel interface, thereby improving the surface quality of final products.

Summary

The effects of chemical composition and furnace atmosphere on high-temperature oxidation were studied. The results and observations obtained from this study are summarized as follows:

- Overall, it appears that the scale/steel interface of the samples heated with 1% and 10% oxygen is rougher than that of the samples heated with 3% and 5% oxygen. For the samples heated with 1% and 10% oxygen, Si-rich oxide appears to be anchored into the steel below the scale/steel interface (the finger-like feature). This kind of scale is difficult to remove by descaling.
- From the SEM/EDS elemental mapping results, it appears that Si-rich oxide heavily segregated to the scale/steel interface in the samples heated with 1% and 10% oxygen, while less Si-rich oxide segregated to the scale/steel interface in the samples heated with 3% and 5% oxygen. The results also showed that Cu segregation near the scale/steel

interface is the strongest in the sample heated with 10% oxygen, while the Cu segregation in the samples heated in the other three conditions is low. In all cases, no mass liquid Cu phase was observed either at the scale/steel interface or along the austenite grain boundaries.

- Among the four steels studied, the low-Si steel (Steel A) performed the best in terms of the scale/steel interface condition, while the steel with high Si and Cr (Steel D) performed the worst. It appears that Cr enhanced the formation of the Si-rich phase at the scale/steel interface. The scale formed on Steel D would likely be the most difficult to remove due to a rough scale/steel interface formed in the steel.
- Overall, it appears that a furnace atmosphere with about 3–5% excess oxygen is desirable to minimize the Si-rich oxide and Cu segregation at the scale/steel interface, thereby improving the surface quality of final products. A furnace atmosphere of less than 3% oxygen appears to be not favorable due to the formation of a more tenacious scale enriched in Si in the case of silicon-containing steel. A furnace atmosphere of more than 5% oxygen is also not favorable due to enrichment of Cu and other noble elements at the steel/scale interface.

Acknowledgments

The authors would like to thank the rolling mill personnel at SSAB Americas for their great support in collecting samples for this study. The authors are also grateful to Ben Kowing and Sunday Abraham for their critical review of the manuscript, and the SSAB Americas senior management for their permission to publish this work.

This article is available online at AIST.org for 30 days following publication.

References

1. S.J. Seo, K. Asakura and K. Shibata, *ISIJ International*, Vol. 37, 1997, pp. 240–249.
2. R.Y. Chen and W.Y.D. Yuen, *ISIJ International*, Vol. 45, 2005, pp. 807–816.
3. S.W. Kim and H.G. Lee, *Steel Research International*, Vol. 80, 2009, pp. 121–129.
4. G. Gaiser, P. Presoly, C. Bernhard and S. Ilie, *High Temperature Corrosion of Materials*, 102:8, 2025, pp. 1–24.
5. T. Kizu, Y. Nagataki, T. Inazumi and Y. Hosoya, *ISIJ International*, Vol. 41, 2001, pp. 1494–1501.
6. M. manna, K. Bhave and A.S. Pathak, *ISIJ International*, Vol. 61, 2021, pp. 1623–1632.
7. R. Osei, S. Lekakh, R. O'Malley, L. Peterson and O. Sasso, *AISTech 2021 Conference Proceedings*, 2021, pp. 596–607.
8. M. Zhou, G. Xu, H. Hu, Q. Yuan and J. Tian, *Metals*, Vol. 7, No. 8, 2017.
9. C. Wang, H. Wu and Y. Zhang, *Journal of Materials Research and Technology*, Vol. 30, 2024, pp. 840–853.
10. T. Fukagawa, H. Okada and Y. Maehara, *ISIJ International*, Vol. 34, 1994, pp. 906–911.
11. H. Okada, T. Fukagawa, H. Ishihara, A. Okamoto, M. Azuma and Y. Matsuda, *ISIJ International*, Vol. 35, 1995, pp. 886–891.
12. J. Haapakangas, S. Riikonen, S. Airaksinen, E.P. Heikkinen and T. Fabritius, *Metals*, Vol. 14, No. 189, 2024.
13. G. Gaiser, P. Presoly, C. Bernhard, K. Baumgartner and S. Grosseiber, *ISIJ International*, Vol. 64, 2024, pp. 1439–1449.
14. M.M. Wolf and H. Schwabe, *Proceedings of the 2nd European Electric Steel Congress – Part II*, AIM, Florenz, 1986. ♦