Role of Microstructural Constituents on Surface Crack Formation During Hot Rolling of Standard and Low Nickel Austenitic Stainless Steels

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[Manuscript received 3 November 2012, in revised form 15 December 2012]

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The effect of alloy segregation and delta (δ) ferrite contents on surface cracking of three standard (*i.e.* AISI 304L, AISI 310S and AISI 321) and two low nickel (*i.e.* LNi-1 and LNi-0.3) austenitic stainless steels (ASS) during hot rolling was investigated using optical microscopy (OM), automatic image analyzer, scanning electron microscopy (SEM), energy dispersive X-ray spectroscopy (EDX) and electron probe micro analyzer (EPMA). It was observed that the amount of δ -ferrite varied among different grades and also distributed heterogeneously across the width of the steel plates. In general, low nickel ASS showed higher amount of δ -ferrite compared to the standard ASS grades. The tendency to surface cracking during hot rolling gradually increased with increasing δ -ferrite content. Interestingly, carbon and nitrogen exerted maximum effect on δ -ferrite formation. The higher carbon and nitrogen content in the steel decreased δ -ferrite content. In addition, the segregation of Cu and Mn plays significant role in low nickel ASS and Ni-Cr in case of standard ASS has profound effect on surface cracking of the steel plates. A possible cause of surface crack formation/origination in steel plates during hot rolling was discussed.

KEY WORDS: Austenitic stainless steel; Hot rolling; Surface cracking; Alloy segregation; δ ferrite

1. Introduction

Hot-rolling is commonly used in manufacturing steel plates. Because of high temperature stability, austenitic stainless steels (ASS) usually exhibit substantially higher hardness during hot-rolling compared to ferritic or mild steels. To avoid excessive mill loading, hot working are often carried out at high rolling temperatures. Furthermore, sometimes inherently low hot-ductility in ASS may lead to edge cracking and other defects^[1]. There are many factors that can affect the hot ductility of steels, such as material flow properties, temperature, strain rate, composition, grain size, precipitates, non-metallic inclusions, previous thermal and mechanical treatments

and hot deformation parameters^[2]. Hence, a true understanding of hot rolling variables and their interactions with the material are needed to find out suitable hot rolling conditions in order to avoid cracking and other defects. In other words, consistently good quality hot rolled steel sheet could be produced only through strict control of hot working conditions.

Several extensive reviews on steel hot ductility and/or cracking in relation to the hot working (*i.e.* hot rolling) process has been published^[3–5]. For the formation of microstructure and its control during hot working, one of the most important mechanism is the dynamic recrystallization (DRX). The DRX process mainly influences the final microstructure and mechanical properties of the deformed material and thus, the hot formability of the material^[6]. DRX usually takes place in materials with low to medium stacking fault energy^[7,8] such as austenitic stainless steel. Further, as DRX is a flow softening mechanism, the

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load required for deformation is considerably reduced if the material is worked under those conditions suitable for DRX to operate. It is well known that DRX is initiated at elevated temperatures when the imposed deformation level reaches critical strain (*i.e.* required strain to start dynamic recrystallization)^[9]. The characteristics of DRX, such as dynamic grains evolved and the effect of initial microstructure etc., have been fairly clarified under hot working. The grain size in DRX is critically dependent on temperature, strain rate and flow stress^[10,11]. Such grain size formed during DRX leads to the formation of finer grain structure. However, the rate of recrystallization may be reduced^[12] when carbides and nitrides are taken into solution at soaking temperatures during the multi-pass rolling operation and strengthen the matrix due to precipitation on dislocations and in grain boundaries. On the other hand, the second phase particles (*i.e.* δ -ferrite), in general, enhance recrystallization^[8]. Also, the non-metallic inclusions particularly the non-deformable inclusions have a significant detrimental effect on the hot-working behaviour^[13]. However, presence of non-equilibrium δ -ferrite in austenitic stainless steels impairs the hot ductility at hot-rolling temperatures, leading to the initiation of cracks at the austenite/ferrite interface^[14]. It has been observed that, with increasing ferrite content in the standard 304 ASS grade, susceptibility to longer edge cracks increases leading to deleterious surface of the rolled steel^[15]. However, controversy still exists about the amount of ferrite with respect to cracking. Previous work on both as-cast and wrought structures has shown that, depending upon temperature and ferrite morphology, ductility though exists at ferrite contents in the range of 15%–30%, although there are findings indicating that, in practice, smaller ferrite contents can produce defective material^[14,16]. Furthermore, Czerwinski *et al.*^[15] reported that, carbon and nitrogen, among the alloying elements present in 304 ASS grade, have the critical effect on δ -ferrite content and edge crack formation. But, there is still little available report on the effect of alloying elements on the formation of δ -ferrite and surface crack in different ASS alloy systems.

Beside the standard ASS grades, low nickel austenitic stainless steels (low Ni ASS) have a unique combination of mechanical, chemical and physical properties, such as high strength and toughness, good corrosion resistance and low magnetic susceptibility^[17–19]. Thus, they are widely used for the production of heavy generator retaining rings, transport, electric power generation, cryogenic vessels and many other engineering applications. Like standard ASS grades low Ni ASS plates are also susceptible to surface cracking during hot rolling. Reported works regarding the hot ductility/workability aspect of these steels are however very meagre. Recently, Chen *et al.*^[20] and Cui *et al.*^[21] determined the hot ductility of 18Mn18Cr0.5N steel under dif-

ferent deformation conditions using numerical simulation. The effect of nitride precipitation characteristics on hot ductility have also been reported and found detrimental effect on hot ductility in high nitrogen CrMn austenitic stainless steels^[22,23]. Furthermore, the activation energy for hot working (*i.e.* required work done to start DRX to refine the microstructure) of low nickel Fe-Mn steels is a function of the Mn content^[24,25]. High Mn content thus responsible for increasing activation energy in low nickel Fe-Mn steels should hinder the DRX. However, Farahat *et al.*^[25] recently reported that the Mn alone is not sufficient to increase the activation energy of DRX in low nickel ASS grades. Manganese together with other alloying elements present in stainless steels is responsible for increasing the activation energy. It is to be mentioned here that the delay in DRX will facilitate to form the twins and dislocations within the grains which thereafter act as a precursor to δ -ferrite formation during hot working. Apart from chemical composition of ASS, Wang *et al.*^[26] showed that the hot workability of 18Mn18Cr0.5N steel gradually decreased with increasing preheating temperature between 1100 and 1200 °C, and quickly deteriorated up to 1250 °C. They have further demonstrated that, above 1200 °C, δ -ferrite particles appeared in 18Mn18Cr0.5N steel promoted cavity coalescence on grain boundary and accelerated surface crack formation during the hot working process. However, difference in opinion exists about the presence of δ -ferrite on surface cracking during hot working. Pickering^[27] concluded that a small amount of δ -ferrite is beneficial to the hot workability, because it can purify the grain boundary and restrain the intergranular fracture caused by impurities. On the other hand, Tarboton *et al.*^[28] confirmed that δ -ferrite can lower the hot workability of high nitrogen CrMn austenitic stainless steel. Dubey *et al.*^[29] however suggested that a low δ -ferrite level is harmless, but excess amount is harmful. As per Shi *et al.*^[22] cracking in ferrite is attributed due to lower strength of δ -ferrite (quarter of the austenite) and more severe plastic deformation in ferrite than in austenite. Thus, there is currently no unified viewpoint on the formation of δ -ferrite as well as their presence on susceptibility to cracking quantitatively during hot working in low Ni and/or high nitrogen CrMn austenitic stainless steels.

In the present study, the effect of microstructural constituents on surface cracking of three standard (*i.e.* AISI 304L, AISI 310S and AISI 321) and two low nickel (*i.e.* LNi-1 and LNi-0.3) austenitic stainless steels (ASS) during hot rolling was investigated to establish suitable hot working conditions of ASS.

2. Experimental

In this study, five plates of austenitic stainless steel (ASS) after hot rolling, containing surface cracks of different intensity, were used as research materials.

Among those materials three of them were standard ASS grades (*i.e.* AISI 304L, AISI 310S and AISI 321) and the rest two were low nickel ASS grades (*i.e.* LNi-1 and LNi-0.3). The average chemical compositions of hot rolled plates are shown in Table 1. The plates were first preheated at 1240°C for 5 h and then rolled

up to eight passes to get the desired dimension. The rolling temperature was maintained within the range of 950°C to 1150°C. After the rolling passes macro picture of the experimented specimens were captured by a digital camera to observe different surface qualities as shown in Fig. 1.

Table 1 Chemical composition (in wt.%) of standard and low Ni ASS specimens

Elements	304L	321	310S	LNi-1	LNi-0.3
C	0.017	0.020	0.050	0.10	0.10–0.12
Si	0.45	0.54	1.26	0.60	0.40–0.60
Mn	1.46	1.66	1.58	9.0–9.2	9.2–9.3
S	0.006	0.004	0.007	0.015	0.015
P	0.035	0.034	0.040	0.070	0.07
Cr	18.46	17.30	24.90	15.00–15.50	13.50–14.0
Ni	8.10	9.06	19.18	1.0–1.05	0.30–0.35
Cu	0.23	0.20	0.050	1.60–1.70	1.55–1.60
Al	0.010	0.032	0.013	–	–
Ti	0.016	0.26	0.010	–	–
N	0.00570	0.00735	0.00455	0.014–0.016	0.016–0.018
Nb	–	–	0.010	–	–
B	–	–	0.0058	–	–
Mo	–	–	0.02	–	–
C+N	0.023	0.027	0.054	0.115	0.127

Fig. 1 Macrographs of specimen, showing surface cracks: (a) 304L; (b) 321; (c) 310S (d) LNi-1 and (e) LNi-0.3

The specimens for metallography were sectioned, prepared and polished using different grades of emery papers followed by final polishing in the disc polishing machine using diamond compound ($2\ \mu\text{m}$ particle size). The specimens were then etched with Kalling's No.2 reagent (*i.e.* 5 mL alcohol+5 mL HCl+0.5 g CuCl_2) for 10 – 15 s to reveal δ -ferrite in the austenite matrix. Microstructural examination was carried out using a light optical microscope (Maker: Carl ZEISS India Pvt. Ltd.; Model: Imager.A1m). A quantitative analysis of the ferrite content and morphology was conducted using an image analyzing computer AxioVision/AxioCam Version 4.8.

In order to examine the details of surface crack morphologies and the alloy composition around the surface cracks in the ASS hot rolled plates, a scanning electron microscopy (SEM) (Maker: JEOL Ltd., Japan; Model: LSM- 6360) with energy dispersive X-ray spectroscopy (EDX) (Maker: Oxford Instruments; Model: Inca-mics, Sl. No.: 01736) was carried out for each specimen.

After SEM and EDX analysis, two specimens (*i.e.* 304L and LNi-1) were mirror polished using auto polisher machine first then they were examined under electron probe micro analyzer (EPMA) (Maker: JEOL Ltd., Japan, Model: JXA-8230) machine in unetched condition to reveal the distribution of constituent elements.

3. Results and Discussion

3.1 Phase composition and distribution

To understand the variation in phase composition and distribution in ASS plates after hot rolling, it is necessary to get a preliminary idea about the phenomenon occurred during the solidification. In industrial conditions the solidification of any ASS is not an equilibrium process and a two-phase microstructure of austenite and δ -ferrites is obtained. Although the solidification can start with the crystallisation of δ -ferrite or austenite depending on the chemical composition of the melt and amount of δ -ferrite can vary in the cast ASS, formation of such δ -ferrite content however may decrease due to homogenisation. Be-

cause of the microstructural similarity, one specimen from standard ASS grades (304L) and one specimen from low Ni ASS grades (LNi-1) were chosen. Fig. 2 shows the typical as-cast microstructure for two cast ASS grades of 304L and LNi-1 before hot rolling revealing large austenite grains along with δ -ferrite. In the as-cast micrograph of 304L specimen (Fig. 2(a)) the distribution of both constituents (austenite and δ -ferrite) in the microstructure indicates that δ -ferrite has formed at the end of solidification from the enriched melt, while in the second case *i.e.* the as-cast micrograph of LNi-1 (Fig. 2b), the solidification started with the two phase region, *i.e.* formation of δ -ferrite + austenite. In addition, the twins like appearances are visible in the micrograph of LNi-1 along with δ -ferrite. It is also clear, by comparing the microstructures in Fig. 2, that with the primary solidification in austenite greater segregations are formed and, as a consequence, the hot ductility of the steel with the as-cast microstructure is diminished. From this point of view, depending upon the microstructure, two main assumptions can be made. All low Ni ASS grades must have higher percent of δ -ferrite in the as cast steel plates than the standard grades and all the grades, *i.e.* standard and low Ni ASS, should have variable δ -ferrite content in there microstructure after hot rolling. Moreover, in the continuously cast slab stresses and deformation are induced by the temperature gradient and the ferrostatic pressure. After solidification, the grains become coarser and their boundaries are frequently oriented orthogonally to the slab surface and to the metal flow by deformation. This and the presence of impurities and segregations may grow to unacceptable surface defects in slab after hot rolling^[30].

Stainless steels of types 304L, 321, 310S, LNi-1 and LNi-0.3 after hot-rolling, examined at room temperature, were essentially austenitic. It is assumed, however, that traces of δ -ferrite might be present in their microstructure along with the austenite. To verify the presence of δ -ferrite in all the ASS plates after hot rolling, the microscopical analyses by optical microscope were carried out. After chemical etching using a selected solution, the micrograph of both standard and low Ni ASS grades (Fig. 3 and Fig. 4) reveals δ -ferrite in the austenite matrix. The δ -ferrite became

Fig. 2 Cast microstructures of specimens 304L (a) and LNi-1 (b)

a dark phase, while the austenitic matrix remained bright. The ferrite morphology observed in the rolling plane of three standard ASS specimens is shown in Fig. 3a-c. In general, it has the appearance of islands forming a chain-like morphology elongated in the rolling direction. Similar type of morphological appearance of δ -ferrite is examined in the case of low Ni ASS grades (Fig. 4). Although, the micrographs show higher amount of δ -ferrite present in low Ni ASS grades comparing to the standard ASS grades. A simple visual assessment of Fig. 3 and Fig. 4 demonstrates that the individual steel plates contained essentially different volume fractions of δ -ferrite. Moreover, the distribution of δ -ferrite is non-uniform in each plate. In order to provide the quantitative description of the phases, the volume fraction of δ -ferrite was measured using automatic image analyzer.

The ferrite content in selected test specimens, as measured on the plate surface, is tabulated in Table 2. In addition to quantitative measurement of the δ -ferrite present in the surface of standard ASS grades (*i.e.* 304L, 321 and 310S), δ -ferrite is predicted from the chemical composition of the respective steel plates (Table 1) using the Schaeffler diagram^[31]. The contents of δ -ferrite, estimated from that diagram using chromium and nickel equivalents, calculated according to the formulas presented in Ref. ^[31], are shown in Table 2. It is worth mentioning here that Schaeffler diagram is not properly applicable for the low Ni ASS grades (*i.e.* LNi-1 and LNi-0.3) due to very low Ni content and high Mn content. These variations were not included properly in the diagram. For the plates examined, the predicted content of δ -ferrite varies between 6% and 12% for standard grades. It

Fig. 3 Micrographs of 304L (a), 321 (b) and 310S (c), revealing the δ -ferrite formation along with micro and macro cracks in austenite matrix

Fig. 4 Micrographs of LNi-1 (a) and LNi-0.3 (b), revealing the δ -ferrite formation in austenite matrix (white)

Table 2 Predicted and measured δ -ferrite contents of as-cast and cracked SS specimens

Sample	Schaeffler	Diagram	Predicted	Measured
	Ni _{eq} [*]	Cr _{eq} [†]	δ -ferrite content/ vol.%	δ -ferrite content/ vol.%
304L ^a	9.511	19.16	12	12.32
LNi-1 ^a	—	—	—	29.35
304L ^b	9.511	19.16	12	19.62
321 ^b	10.53	18.14	10	18.33
310S ^b	21.61	26.83	6	14.82
LNi-1 ^b	—	—	—	39.24
LNi-0.3 ^b	—	—	—	31.82

Note: ^{*}Ni_{eq} = Ni+30C+30N+0.5Mn; [†]Cr_{eq} = Cr+Mo+1.5Si+0.5Nb, a: as-cast condition before hot rolling; b: after hot rolling at 1200 °C

is observed from the Table 2 that the predicted δ -ferrite content (12%) is highest in 304L ASS followed by 321 (10%) and 310S (6%). However, the magnitude of measured δ -ferrite content in each specimen is much higher than the bulk and/or cast condition of the plates, although the order remains same. The measured values of δ -ferrite in 304L is 19.62% followed by the 321 and 310S (Table 2). Among the two low Ni ASS grades the highest δ -ferrite content (39.24%) is observed in LNi-1 specimen. However, the measured values of δ -ferrite in LNi-1 and LNi-0.3 specimens are quite higher as compared to their respective cast conditions and the standard grades as well. The increase in δ -ferrite content after hot rolling can be explained by the thermal condition experienced during reheating before rolling. As the plates were reheated before rolling at 1240 °C for 5 h, which is close to the δ + γ region, δ transformation from γ has occurred resulting in more δ -ferrite at the expense of austenite. In addition, there is a dependency between δ -ferrite and the chemical composition, as expressed by a sum of C and N (Table 1). The plates with the higher δ -ferrite content have the lowest sum of C and N. This is true for both standard and low Ni ASS grades. For example, 304L and LNi-1 steels have lowest sum of C and N and higher δ -ferrite content, thus have highest surface crack concentration (Fig. 1) among their types. Since both C and N enhance the formation of austenite, this finding seems to support the observation concerning the detrimental influence of δ -ferrite on surface crack generation. At this point, it should be mentioned that some other results reported in the literature claim that unstabilized nitrogen degrades hot ductility and workability in nickel based alloys and 18-8 stainless steel^[32]. Furthermore, the higher δ -ferrite content of low Ni ASS grades compared to the standard grades are likely due to their metastable chemical composition at high temperature. The activation energy for hot working (*i.e.* required work done to start dynamic recrystallization to refine the microstructure), of low Ni ASS grades, was found to be a function of the Mn content. The Mn was found to delay the dynamic recrystallization, because the corresponding activation energy increases with increasing Mn content (*i.e.* required activation energy to start dynamic recrystallization). Again, the critical strain (*i.e.* required strain to start dynamic recrystallization) was also found to increase with increasing Mn content^[24]. This delayed DRX and higher critical strain, of low Ni ASS grades, also increases the formation of twins and dislocation pile-ups in the microstructure (Fig. 4), which act as a precursor to δ -ferrite formation^[25] during hot rolling. It is to be mentioned here that the distribution of δ -ferrite in steel plates is inhomogeneous. The highest ferrite concentration is detected in the plate-edge region and it decreased while moving towards the middle of the plate. The presence of this non-equilibrium δ -ferrite at hot rolling temperature impaired the hot ductility^[15] because ferrite and

austenite present different softening mechanisms at high deformation temperatures; while ferrite recovers austenite re-crystallises leading to the initiation of cracks at austenite/ferrite interface (Fig. 3 and Fig. 4).

3.2 Analysis of surface cracks formation in steel plates

The microscopic analysis of the steel microstructure in the regions adjacent to the cracks was performed in order to assess the location of the cracks with respect to δ -ferrite islands and the network of the grain boundaries of austenite. Since the surfaces of the large and open cracks were heavily oxidized at hot-rolling temperatures, they were not useful for fractographic analysis. Therefore our effort was focused on tiny branches of microcracks which surround the major cracks and penetrate deeper into the steel plate than it was anticipated from the macroscopical observations of the plate surface (Fig. 1). Although there is a general agreement in the literature that cracks may develop at the ferrite/austenite interface during hot rolling, the influence of δ -ferrite on the hot workability of different ASS is considered to be more complex. According to Kane^[32], the workability of 310 stainless steel ingots can be improved by certain amounts of δ -ferrite. It is believed that δ -ferrite retards the grain boundary migration of austenite during solidification which, in turn, leads to corrugated grain boundaries of austenite considered to be resistant to crack propagation. Moreover, the presence of the ferrite/austenite interfaces acting as nucleation sites increases the recrystallization rate. In experiments by Kane^[32], cracks developed at the ferrite/austenite interface only in the cases when ferrite phase was continuous and/or temperature was low, however, they did not propagate deeply into the steel. On the other hand, there are data indicating that even small amounts of δ -ferrite are critical for the hot workability of steel and may generate surface defects like cracks^[16]. In this study, Fig. 3(a) and Fig. 3(b) show the microstructure of the crack region. The location of the crack with respect to austenite grain boundaries indicates its interfacial character. In general, the crack path does not change its direction at the austenite grain boundaries. Moreover there is an evident dependence in alignment between the crack path and the δ -ferrite islands (Fig. 3(c)). The cracks are situated at ferrite/austenite interfaces and then connect the neighbouring islands of ferrite through bridges of austenite. Similar behaviour of the crack characteristics is also observed for low Ni ASS grades (Fig. 4).

In order to go further inside into the surface cracks of the ASS specimens, detail SEM and EDX study was carried out for all the specimens (Fig. 5–Fig. 9). SEM micrographs were taken from the cracked specimens and revealed both long macro-crack and several micro-cracks. SEM clearly reveals the interfacial

Fig. 5 SEM and EDX analysis of 304L at crack and no-crack regions

Fig. 6 SEM and EDX analysis of 321 at crack and no-crack regions

behaviour of the cracks in all the tested steels. There are two clearly visible regions, *i.e.* crack and no crack regions, present in the micrographs. EDX analysis at or near the cracks depicts the segregation of ferrite stabilizing elements such as Cr, Al, Si *etc.* and the segregation of austenite stabilizing elements such as

Ni, Cu, C *etc.* in no-crack regions. The higher segregation of ferrite stabilizing elements in the localized area is mainly responsible for the structural heterogeneity of the individual ASS plates, depending upon the different chemical compositions (Table 1), leading to higher band/thin film like δ -ferrite formation

Fig. 7 SEM and EDX analysis of 310S at crack and no-crack regions

Fig. 8 SEM and EDX analysis of LNi-1 showing segregation of alloying elements around the crack

(Fig. 3 and Fig. 4). Therefore, it is obvious that there should be a variation in δ -ferrite content among the steel plates after hot rolling (Table 2).

To clarify the segregation of alloying elements

more clearly EPMA analysis was carried out for 304L ASS (Fig. 10 and Fig. 11) and LNi-1 (Fig. 12 and Fig. 13). These two specific steel plates were chosen because they have highest δ -ferrite content (Table 2)

Fig. 9 SEM and EDX analysis of LNi-0.3 showing alloy segregation around the crack

Fig. 10 EPMA analysis of 304L cracked reveals alloy segregation around a crack

and surface crack concentration (Fig. 1) among their classes. EPMA analysis of 304L specimen confirms that the crack region is enriched with Cr and the surrounding regions of the crack, *i.e.* no-crack regions, are enriched with Ni as shown in Fig. 10. These results ultimately suggest that the δ -ferrite has taken a major role in crack formation. It has been well established that higher fraction of δ -ferrite is very detrimental to the hot workability of ASS grades^[16,33]. However, the hot workability could be improved in ASS by certain amount (about 4%) of δ -ferrite, as

δ -ferrite retards the grain boundary migration of austenite during solidification which, in turn, leads to corrugated grain boundaries of austenite considered to be resistant to crack propagation. On the other hand, it has been reported that the hot workability is also diminished at a content of δ -ferrite above 8%^[30]. Another interesting observation was evolved from the EPMA point analysis (Fig. 11) that some regions in the 304L specimen show Cu segregation. Although Cu may produce an eutectic melt, wetting the grain boundaries, lowering their resistance and

leading to crack formation by the axial flow of the metal^[34], sufficient solubility of copper in austenite and the presence of nickel prevent the formation of eutectic composition in standard ASS grade^[30].

Fig. 11 EPMA point analysis of different positions in specimen 304L reveals higher Cu segregation (in wt.%) near the crack

Fig. 12 EPMA analysis of LNi-1 cracked specimen reveals different alloy segregation around a crack

Fig. 13 EPMA Line analysis across the white patch, evident from backscattered image of LNi-1 specimen, revealing the Cu segregation

4. Conclusions

(1) Segregation of alloying elements, *i.e.* ferrite stabilizing elements and austenite stabilizing ele-

Like standard ASS grade, the EPMA analysis of LNi-1 (Fig. 12) also reveals the alloy segregations particularly segregation of ferrite stabilizing elements at the crack locations. Whereas, no-crack regions around the crack show segregation of austenite stabilizing elements. This result again suggests the formation of δ -ferrite in low Ni grades and the δ -ferrite acted as a crack precursor as mentioned earlier. However, in addition to δ -ferrite content, EMPA analysis of LNi-1 specimen also demonstrates that the alloy consists of several Cu segregations as a continuous white patch throughout the specimens (Fig. 13). Furthermore, the segregated amount of Cu in low Ni grades is much higher than that of standard ASS grades. Detrimental effect of Cu may induce due to lower Ni content which is unable to prevent the formation of eutectic composition. The higher amount of δ -ferrite and segregation of Cu is more likely responsible for more frequent surface cracking of LNi-1 (low Ni grade) (Fig. 1) than standard ASS grade.

ments, are solely responsible for higher δ -ferrite formation and variation among the plates. Presence of higher δ -ferrite in standard ASS grades (*i.e.* 304L, 321, 310S) and low Ni grades (*i.e.* LNi-1 and LNi-

0.3) are mainly responsible for surface cracks. Surface cracks are mainly formed at the δ -ferrite and austenite interfaces during hot rolling.

(2) Copper segregation in low Ni grades (*i.e.* LNi-1 and LNi-0.3) compared to standard ASS grades (*i.e.* 304L, 321, 310S) is an additional factor to surface crack generation.

(3) In order to avoid the surface cracking during hot rolling, alloy composition and reheating conditions should be adjusted for minimum amount of δ -ferrite.

Acknowledgements

The authors would like to thank BRG Iron and Steel Co. Pvt. Ltd., Dhenkanal, Orissa, India, for providing the experimental materials.

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