Hot Cracking of Low Alloy Steels in Simulated Continuous

Casting-Direct Rolling Process*

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Synopsis
Surface cracking during direct rolling of continuously cast low alloy
steel slabs has been investigated by means of hot tensile tests simulating
the process. The hot ductility of directly cooled specimens is not reduced
by high strain rate deformation corresponding to that in rolling. While, in the specimens predeformed at temperatures ranging from r/a duplex phase to low temperature r region at slow strain rate corresponding to that in bending and/or straightening operations of CC slabs, ductile intergranular fracture of austenite is easily induced by the subsequent deformation even at high strain rate, resulting in poor ductility. This ductility loss can be explained in terms of dyn2mic precipitation of carbide, nitride, or carbo-nitride such as A1N and/or NbC (or NbC(N)) within the matrix as well as on the r grain boundaries during the prior slow strain rate deformation. The ductility is also influenced by the final deformation conditions because o f the morphology change of the precipitates in the process, although the embrittlement mechanism is essentially the same as in the CC process.

Key words: continuous casting; low alloy steel; hot ductility; hot crack-ing; dynamic precipitation; hot tensile test; intergranular fracture.

I. Introduction

It has been well established that a ductility of low manganese steels in austenite temperature region after remelting or high temperature solution treatment is largely reduced by high strain rate deformation ac-companied by an intergranular fracture due to dy-namic precipitation of Fe-rich (Fe, Mn)S and/or grain boundary segregation of S atoms.'-4) The same mechanism is apt to be applied to a surface cracking of continuously cast low alloy steel slabs during hot rolling without reheating process (referred to as CC-DR process), since the deformation conditions are very similar to each other. However, there remains some difficulties to adopt. That is, (1) the hot embrittle-ment due to S and/or sulfide precipitation occurs by the deformation after relatively rapid cooling followed by high temperature solution treatment, and (2) the dynamic precipitation of sulfide is much reduced by such slow cooling and/or by the prior slow strain rate deformation corresponding to bending and/or straight-

ening operations in the CC process because of easy coarsening of the sulfide particles, resulting in high ductility in the subsequent high strain rate deforma-tion.2)
The slow strain rate deformation in the tempera-ture region from r/a duplex to low temperature Y causes dynamic precipitation of carbide, nitride, or carbonitride such as A1N and/or NbC (or NbC(N)), resulting in hot embrittlement accompanied by an intergranular fracture of austenite."4-13) This is be-lieved to be the main mechanism of transverse surface

crackings of CC slabs. However, in the case that the

precipitation does not lead to the transverse cracks in the CC process, it is not clear how the precipitates affect on the ductility in the following direct rolling

process.

In the present article, a mechanism of surface cracking in the direct rolling process has been studied by means of hot tensile tests simulating the process with particular attention to the dynamic precipitation of A1N and/or NbC (or NbC(N)) during slow strain rate deformation prior to high strain rate deformation.

II. Experimental Procedure

Chemical compositions of the steels used are given in Table 1. Steels X 1 and X3 are Al-killed, X2 and X4 are Nb containing steels, and X3, X4 are steels with a small amount of Ti addition to Xl and X2, respectively. The ingots of 50 kg in weight prepared by an induction furnace were hot-forged and rolled into 12 mm thick plates. From these plates, tensile specimens, whose gage length, gage diameter and fillet radius were 20, 8 and 5 mm, respectively, were machined in the direction parallel to that of rolling. These were set to an Instron-type tensile machine, and heated at 1 300°C for 10 min in an infrared image furnace attached to the tensile machine. After vari-

ous thermo-mechanical treatments during cooling at a rate of 2°C/s from the solution treatment tempera-

Table 1. Chemical compositions of the steels used. (wt%)

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Originally published in Tetsu-to-Hagane, 73 (1987), 876, in Japanese. English version received on March 28, the final form on July 8, 1988. CQ 1988 ISIJ
Iron and Steel Research Laboratories, Sumitomo Metal Industries, Ltd., Nishinagasu-hondori, Amagasaki 660. Kashima Steel Works, Oaza-hikari, Kashima-cho, Kashima-gun, Ibaraki Pref. 314.

1988; accepted in

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ture, they were finally deformed up to failure at tem-peratures ranging from 650 to 1 100°C at initial strain rates from 10-4 to 10-1 The base thermal history was simple temperature homogenizing for 1 min at final deformation temperatures after directly cooling from the solution treatment temperature. In addi-tion, the specimens were predeformed up to 10 % at a strain rate of 4 X 10-4 s-1 at given temperatures and kept for 10 min at final deformation temperatures after taking the load off immediately at a strain rate of 10-1 s-1. At this stage, any crack initiation was not recognized in the specimens examined. The ten-sile properties in the latter case were calculated by using cross section area of the specimens before the final deformation, assuming that the specimens were uniformly deformed by the predeformation.
On the fractured specimens, fractographic and metallographic examinations were conducted by means of scanning electron (SEM), optical and trans-mission electron microscopy (TEM). In order to identify the precipitate particles, X-ray analysis in STEM-EDX mode was also used in addition to se-lected area electron diffraction in TEM.

III. Experimental Results

1. Tensile Properties

Figure 1 shows variation of tensile properties with strain rate in the specimens deformed at 800°C after direct cooling from the solution treatment tempera-ture. The ductility evaluated by reduction in area

(RA) decreases with decreasing strain rate in each steel. The ductility loss is enhanced by Nb addition, and significantly reduced by small amount of Ti

addition in both Nb-free and Nb added steels. The

peak stress in stress-strain curve increases with lower-ing temperature, and strengthening due to Nb and/or Ti addition is also observed especially in lower tem-perature region. The ductility at high strain rate is not reduced at all temperatures examined, except that small ductility trough is observed at N 750°C which seems to be just below the Ar3 point, as shown in Fig.
2. Figure 3 shows tensile properties of Steels Xl to X3 fractured at 800°C at a strain rate of 0.83 x 10-1

Fig. 1. Variation of tensile properties with strain rate; Steels Xl to X4 specimens, deformed at 800°C after
directly cooled from the solution treatment tem-
perature of 1 300°C.

Fig. 2. Variation of
temperature;
10-1 s-1, after
temperatures.

tensile properties with deformation
the specimens, deformed at t = 0.83 x
directly cooled to the deformation
(& : strain rate)

Fig. 3. Effect of thermo-mechanical history on tensile properties of various steels fractured at 800°C at
=0.83x 10-1 s-i.
The prior treatments were (A) none, (B) isotherm-
ally held at 800°C for 20 min, and (C) prede-
formed to 10 % at 800°C at t=4 X 10-4 s-1 and
held for 10 min at 800°C.

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s-1 after various thermo-mechanical treatments. In

each steel, the ductility was significantly reduced by 20 min holding before the final deformation (compare B with A). Furthermore, the ductility was much lower in the case of C in which the specimens were

predeformed to 10 % at 800°C at a strain rate of 4 x 10-4 s-i. It should be noted that the strength of case C was much larger than that of A.
The thermo-mechanical treatment C in Fig. 3 simulates bending and/or straightening operations and the subsequent direct rolling process, and the effects of both temperatures on the ductility in the final high strain rate deformation were examined. Figure 4 shows variation of tensile properties with final de-formation temperature of the specimens which were predeformed to 5 or 10 ° o at 800°C at a strain rate of 4x 10-4 s-1. Although the ductility trough of Al-killed Steel X1 ranges from 700 to 950°C, the ductility loss occurs only around 760°C in Ti added Steel X3. The ductility trough is largely widened and deepened by small amount of Nb addition (Steel X2), and the ductility loss due to Nb addition is also reduced by Ti addition (Steel X4). It should be noted that the ductility trough of Steel X4 is separated into two parts, i.e., one is around 750°C and the other is around 900°C. In each steel, the ductility trough was wid-ened and deepened by increasing the strain in the slow strain rate predeformation, as shown in the case of Steel X2. Figure 5 shows effect of strain in the

predeformation on the tensile properties of the final deformation. The both deformation temperatures were 800°C. With increasing the prestrain, the duc-tility decreased and the strength increased. Figure 6 shows the results of the same experiment at 760°C as in Fig. 5. Increase of prestrain did not give a large ductility decrease in this case. Figure 7 summarizes the effects of pre- and final deformation temperatures

on the ductility in the final deformation of Steels Xl to X3. The areas surrounded by the contour lines of 50 % RA show conditions of relatively high surface cracking tendency during the direct rolling. In Al-killed Steel Xl, the area is long, and the direct rolling at temperatures around 760°C has a high surface cracking frequency. The area of Nb Steel X2 is very wide, and largely depends on the predeformation tem-

perature as well as on the final deformation tempera-ture. While, in Steel X3 with small amount of Ti addition to X l, the area was very small at both tem-

peratures of 760°C. Similar effect of Ti addition was also recognized in the case of Nb steel, and the area in Steel X4 was almost the same as in Steel X3

in Fig. 7.

Fig. 4. Variation of tensile properties with final deforma-tion temperature; the specimens, predeformed to
10% •, A) or 5 % (0) at s=4x 10-4 s-i,
held for 10 min at various temperatures and re-strained up to failure at =0.83x 10-1 s-1.

Fig. 5. Effect of strain at predeformation on tensile defor-mation; the specimens, predeformed at 800°C at
E=4x 10-4 s-1, held for 10 min at 800°C, and then restrained up to failure at =0.83x 10-1 s-1.

Fig. 6. Effect of strain at predeformation on tensile prop-erties; the specimens, predeformed at 800°C at
=4x 10-4 s-1, held for 10 min at 760°C, and then
restrained up to failure at 760°C at -0.83 x 10-1
s-1.

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2. Fractography

Figure 8 shows fracture surfaces of Steels Xl to X3, which were deformed at a strain rate of 4 x 10-4 s-1 at 800°C after direct cooling from the solution

treatment temperature of 1 300°C. The fracture mode of Steels X 1 and X2 shows brittle mannar con-

sisting of an intergranular fracture of austenite (Figs. 8(a) and 8(b), respectively). The detailed examina-tion revealed that they were ductile intergranular fracture in which only austenite grain boundary layers were plastically deformed. While in Steel X3, duc-tile transgranular mode was mixed within the inter-

granular fracture (Fig. 8(c)). Figure 9 shows effect of thermo-mechanical history on the fracture surface of Steel X2 specimens fractured at 800°C at a strain rate of 0.83 x 10-i s-i. Typical of ductile transgranu-lar fracture mode was observed in the specimen frac-tured after direct cooling from the solution treatment temperature, as in Fig. 9(a). Whereas, an inter-granular fracture was mixed within the ductile dimple pattern in Fig. 9(b) in which the specimen was held for 20 min at 800°C prior to the deformation, and the major fracture mode was an intergranular fracture in Fig. 9(c) where the specimen was predeformed to 10 % at a slow strain rate at 800°C. Figure 10 shows

Fig. 7. Effects of pre- and final deformation temperatures on the ductility in the final deformation.
Reduction of area in the region surrounded by con-
tour lines was less than 50 % for each steel.
The specimens were predeformed to 10 % at E =
4>< 10-4 s-1, held for 10 min at the final deforma-
tion temperatures, and then restrained up to failure
at =0.83x 10-1 s-1.

Fig. 8. Fracture surfaces; the specimens, directly cooled to the deformation temperature of 800°C, and de-
formed up to failure at =4x 10-4 s-1.

Fig. 9. Effect of thermo-mechanical history on the fracture surface of Steel X2 specimens ruptured at 800°C at
t = 0.83 x 10-1 s-1.

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a fracture surface and a microstructure on the longi-tudinal cross section in the vicinity of the fracture surface of Steel X 1 specimen, which was predeformed to 10 % at a strain rate of 4 X 10-4 s-i at 760°C, held for 10 min, and then fractured by high strain rate deformation at this temperature. The cracks prop-agate along austenite grain boundary allotriomorphs of ferrite, and the dimples on the intergranular facets are deeper than those of Fig. 8(a), indicating that this fracture was caused by the deformation at tem-perature below Ar3 point.

3. Metallography
Figure 11 illustrates carbon extraction replica im-ages of fractured Steel X2 specimens. In the speci-men fractured by slow strain rate deformation at 800°C, fine precipitate particles of ~nm dispersed within the matrix in addition to relatively coarse grain boundary precipitates, as in Fig. 11(a). These particles were identified as NbC or NbC(N) by a selected area diffraction in TEM and X-ray analyses in STEM-EDX mode. Similar results were obtained

in the specimen, which was predeformed at a slow strain rate in low temperature r region and fractured by high strain rate deformation, as shown in Fig. 11(b). Whereas, in the specimen reheated to the final deformation temperature of 1 100°C for 10 min followed by the slow strain rate deformation in a similar fashion to that in Fig. 11(a), the precipitates became very coarse to 500 nm, as in Fig. 11(c). Such morphology change was also observed in A1N precipitation in Steel X1. In Steels X3 and X4 which contained small amount of Ti, very coarse TiN particles were formed, as shown in Fig. 12(a). In

addition, lined-up fine NbC particles were observed within the matrix of Steel X4 specimens deformed at slow strain rate in low temperature r region, as shown in Fig. 12(b).

Iv. Discussion

Ductility of low alloy steels with coarser structure such as in as-cast steels is largely reduced by slow strain rate deformation at temperatures ranging from r Ja to low temperature r, accompanied by intergranu-lar fracture of austenite.4-13) This is caused by dy-namic precipitation of carbide, nitride, or carbonitride such as NbC (or NbC(N)) or AlN.9"°"2) That is, the strain concentration within soft r grain boundary layers such as precipitation free zones or the grain boundary allotriomorphs of ferrite leads to microvoids due to decohesion of the grain boundary precipitates from the matrix. Final fracture occurs by their micro-, void coalescence, resulting in the ductile intergranular

Fig. 10. Fracture surface and microstructure on the longi-tudinal cross section of Steel X3; the specimen, predeformed to 10 % at 760°C at s = 4 x 10-4 s-1,
held for 10 min at 760°C, and then restrained up
to failure at =O.83 X 10-1 s-1.

Fig. 11. Extraction replica images of Steel X2 specimens.

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fracture of austenite. Transverse cracking of CC slabs, mainly formed in the bending and/or straightening operations, can be explained by this mechanism.
The morphology of the precipitates hardly changes by the long time holding in the low temperature r region (Figs. 11(a) and 11(b)). Even if the micro-voids do not initiate by the slow strain rate deforma-tion, the specimen is easily fractured by the sub-sequent deformation even when the strain rate is fast. Of course, the ductility loss will be enhanced by de-creasing the strain rate in the final deformation, since the newly formed precipitates are accumulated.14) From the above results, it can be concluded that the main cause of the surface cracking in the CC-DR process is in the dynamic precipitation of carbide, nitride or carbonitride during bending and/or straight-ening operations. The mechanism is almost the same as in the transverse cracking of CC slabs. The strain rate of 10-1 s-1 for the simulation of DR process is considered to be reasonable for the rough rolling of slabs in the present study, although it is much smaller than that of plate rolling or hot strip rolling.
The fine precipitates formed dynamically can grow into coarse particles by heating at temperatures above 1 000°C, resulting in high ductility in the sub-sequent high strain rate deformation, in keeping with the previous results.15) Also lowering the N content or trapping of N as coarse TiN particles by small

amount of Ti addition is effective for suppression of the surface cracking, since A1N precipitation is largely reduced. Such effect of Ti was recognized in Nb steel, and the shallow ductility trough was separated into two parts (Fig. 4). The ductility trough around 900°C corresponds to precipitation behavior of NbC)s) The ductility improvement due to Ti addition can be caused by the suppression of NbC (N), in addition to the fact that the ductility loss due to A1N precipitation is cancelled.

It is quite difficult to avoid ductility loss at tem-

peratures around 750°C even by small amount of Ti addition (Fig. 4). The effect of prestrain on the duc-tility in the final deformation was very small, com-

pared to that in low temperature r region (Figs. 5 and 6). These results indicate that grain boundary ferrite layers have important role for the ductility loss through strain concentration. There may be enough initiation sites for grain boundary microvoid forma-tion in non-pure commercial base steels.

V. Conclusion

Surface cracking during direct rolling of continu-ously cast low alloy (0.15%C-0.4%Si-1.3%Mn) steel slabs has been investigated by means of hot tensile tests simulating the CC-DR process. The results ob-tained are as follows :

(1) The hot ductility of the specimens cooled di-rectly from high solution treatment temperature is not reduced by high strain rate deformation correspond-ing to that in rolling at temperatures ranging from 600 to 1 100°C. While, in the specimens predeformed at temperatures ranging from r/a duplex phase to low temperature r region at slow strain rate corresponding to that in bending and/or straightening operations of CC slabs, ductile intergranular fracture of austenite is easily induced by the subsequent deformation even at high strain rate, resulting in poor ductility.

(2) The ductility trough temperature ranges from 750°C, just below the Ar3 point, to low temperature

r region, and is widened to higher temperature of 1 050°C by a small amount of Nb addition.

(3) This ductility loss can be explained in terms of dynamic precipitation of carbide, nitride, or car-bonitrides such as NbC (or NbC(N)) and/or A1N within the matrix as well as on the r grain boundaries during the prior slow strain rate deformation. Al-though the embrittlement mechanism is essentially the same as in the CC slabs the ductility is influenced by the subsequent deformation conditions because of the morphology change of the precipitates.

(4) The ductility loss in the DR process can be largely suppressed by small amount Ti addition to trap N as coarse TiN particles in Nb steels as well as in Al-killed steels.

(5) It is very difficult to avoid hot embrittlement
at temperatures just below the Ar3 points even by small amount Ti addition, indicating the importance of strain concentration within soft grain boundary ferrite layers for the embrittlement.

Fig. 12. TEM micrographs showing
coarse TiN particle and (b)
particles around coarse TiN
matrices of Ti added steels.

precipitations of (a)
finely lined up NbC
particles within the

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