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Prevention of Slab Surface Transverse Cracking by Microstructure Control

Toru KATO, Yoshiki ITO, Masayuki KAWAMOTO, Akihiro YAMANAKA and Tadao WATANABE

Sumitomo Metal Industries, Ltd., Corporate Research and Development Laboratories, Sunayama, Hasaki, Kashima, Ibaraki, 314-0255 Japan.

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Slab surface transverse cracking is well known to be induced by strain concentration at film-like primary ferrite, *i.e.* allotriomorphs of ferrite formed along the austenite grain boundaries. In the present study, a new concept for the prevention of transverse cracking by means of microstructure control at continuous casting strand is examined. Three kinds of examinations in charge of each objective were conducted; (a) ingot cooling tests for microstructure control with secondary cooling; (b) hot tensile tests for hot ductility with the microstructure; and (c) continuous casting tests for cracking susceptibility on continuously cast slab. Results obtained are concluded as follows.

(1) Slab surface microstructure could be controlled by secondary cooling condition. Surface structure control (SSC) cooling, providing intensive cooling until less than A_3 transformation temperature just below mold and subsequently reheated up to 1 250 K in secondary cooling, brings film-like ferrite free structure.

(2) Hot tensile tests subsequent to *in-situ* remelting and solidification prove that hot ductility is much improved and ductility trough almost disappeared with that microstructure control. The results also confirm that *in-situ* remelting of specimen is indispensable on the hot tensile test to evaluate the effect of microstructure on susceptibility to transverse cracking.

(3) Continuous casting test confirms that susceptibility to transverse cracking could be alleviated with this microstructure control.

(4) The prevention of transverse cracking and microstructure control is a result of uniform fine precipitates dispersion, such as (Ti, Nb)(C, N), according to SSC cooling.

KEY WORDS: transverse cracking; hot ductility; hot tensile test; microstructure; continuous casting; secondary cooling; precipitation.

1. Introduction

Slab surface transverse cracking of continuously cast slab is sometimes noted in some low alloy steel. Cracking occurs along prior austenite grain boundaries associated with allotriomorphs of ferrite, which grows along grain boundaries and seems film-like. It takes place during bending or straightening operation of casting and is revealed to be related with ductility trough in the vicinity of austenite–ferrite transformation temperature. Much research has been performed on hot ductility and revealed the mechanism of the embrittlement.[1,2] The cracking is induced by strain concentration at film-like primary ferrite formed along the austenite grain boundaries. In the conventional continuous casting process, making slab surface temperature avoid the range of the trough during bending or straightening operation is not enough to prevent cracking. Nevertheless, uniform thermal profile across the slab width is difficult especially on Ni bearing low alloy steel because of unique subscale formation,[3] and a consequent change in spray cooling property.[4]

It is obvious that slab surface microstrucure correlates with the cracking. A coarser austenite grain is confirmed to reduce hot ductility and encourage the cracking.[5–7] An increase of the cracking in peritectic grade of steel is thought to attribute to uneven solidification, and hence grain coarsening.[5] Refining of austenite grain size of as-cast slabs, however, is difficult because grain size at slab surface is first settled in the mold. Utilization of re-crystallization by means of thermal cycle is considerable. Hot tensile tests also prove effects of thermal cycle on the hot ductility, despite some of them show negative affects.[8–11] Simple intensification of secondary cooling, however, brings slab temperature drop and sometimes involves harmful thermal stress in the slab. Few attempts are conducted on slab surface microstructure control for crack prevention.

In the present study, active microstructure control of slab surface region in the continuous casting strand was examined. Ingot cooling tests show slab surface microstructure could be controlled by secondary cooling condition. *In-situ* solidified hot tensile tests proved this microstructure control improve hot ductility of steel. And continuous casting tests as comprehensive verification demonstrate that transverse cracking could be prevented by this slab surface microstructure control. Microstructure is metallographically analyzed and mechanism of the phenomena is also discussed.

2. Experimental

2.1. Fundamental Conditions

Ingot cooling tests, hot tensile tests subsequent to *in-situ* melting and solidification, and continuous casting tests were conducted. Standard composition of the steels used are shown in **Table 1**. Common grade of Ni, Cu, Ti and Nb bearing low alloy steel was used throughout the experiments. As steels are melted by air induction furnace in ingot cooling tests and continuous casting tests, nitorogen content is higher than that of standard commercial products by blast furnace-BOF process. Nondescript thermal history of mild cooling; that is gradual cooling after solidification, corresponding to conventional secondary cooling of slab in the continuous casting strand, and surface structure control cooling (SSC) were examined in each experiment. SSC cooling is aimed at slab surface microstructure control, providing thermal cycle at the initial stage of secondary cooling; *i.e.* intensive cooling until less than A_3 transformation temperature and subsequent reheating up to 1 250 K. The following stage after reheating is similar to mild cooling. Metallographic examination by means of optical microscopy was conducted with the specimens obtained by these experiments after nital etching. Fine precipitates were also examined by means of transmission electron microscopy.

2.2. Ingot Cooling Test

Microstructure control at slab surface region was first investigated by ingot cooling tests. Experimental apparatus is shown in previous study.[12] 200 kg of molten steel was poured into a rectangular cast iron mold with 180 mm in thickness simulating the heat capacity of conventional slab. Then they were withdrawn downward before completing solidification and cooled with air-water mist spray immediately with a various programmed spray pattern. The thermal history of the ingot was measured by thermocouples placed at 5 mm inside to the surface before casting. Although those thermocouples were inserted in protection tube, the tips were uncovered in order for sensitivity. Specimens for metallographic examination were cut from a section perpendicular to the surface.

2.3. Hot Tensile Test

Hot tensile test is conducted in order to verify the effect of microstructure control on the hot ductility. Usual hot tensile tests with reheated specimens have some disadvantage that microstructure of steel slab, *in-situ* solidified, cannot be reproduced in the specimen. Although some *in-situ* solidified hot tensile tests simulating thermal history of continuously cast slabs had conducted in which specimen was placed in coaxial silica tube,[8,9] some problem such as microvoid growth or friction between tube and specimen still remains.

New hot tensile tests *in-situ* remelting were conducted to confirm the effect of microstructure control on the cracking susceptibility. A cold crucible type induction heater was employed to hold the molten specimen and install a hydraulic tensile testing machine, shown in **Fig. 1**. As there is no surrounding tube around the specimen, harmful influence could be neglected. Temperature is measured just above crucible and controlled according to calibration curve measured in advance. Temperature accuracy, stability and distribution were also inspected beforehand.[13] These apparatus were placed in a chamber and experiments were carried out in argon atmosphere.

The 25 kg of steels were melted in a vacuum induction furnace and cast into ingots. The ingots were hot forged and tensile specimens of 10 mm diameter were machined from them.

During the experiment, the middle part of the specimen was *in-situ* remelted about 30 K above the liquidus. More than 30 mm of melting zone, enough for fracture, could be obtained. After 240 s holding, the specimens were solidified at $5\,K\cdot s^{-1}$ to 1 523 K. Then cooling rate is changed to $0.4\,K\cdot s^{-1}$ until test temperature in mild cooling, while it is changed to $20\,K\cdot s^{-1}$ to 870 K, keeping it constant for 120 s subsequently reheated up to 1 430 K at $3.0\,K\cdot s^{-1}$ then cooled again at $0.4\,K\cdot s^{-1}$ until test temperature in SSC cooling. After reaching the test temperature and a further holding time of 120 s all specimens were deformed until fracture at strain rate of $3.3\times10^{-4}\cdot s^{-1}$ comparable to that during straightening of continuous casting. Reduction of area at fracture and microstructure of specimen are investigated. For the purpose of austenite grain size examination, some specimens were kept cooling without deformation. Usual tensile tests without *in-situ* remelting were also investigated with this apparatus to make a comparison with conventional hot tensile tests. Specimens were reheated up to 1 623 K and held for 240 s at which the specimens were never remelted. Other conditions were similar to *in-situ* melted tests.

Table 1. Standard chemical composition of steels used (mass%).

C	Si	Mn	Cu	Ni	Nb	Al	Ti	N
0.07	0.2	1.5	0.3	0.7	0.02	0.02	0.01	0.007

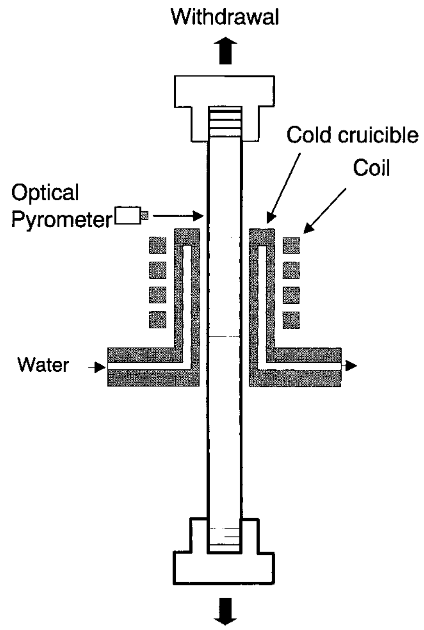

Fig. 1. Schematic illustration of experimental apparatus for *in-situ* solidified hot tensile test.

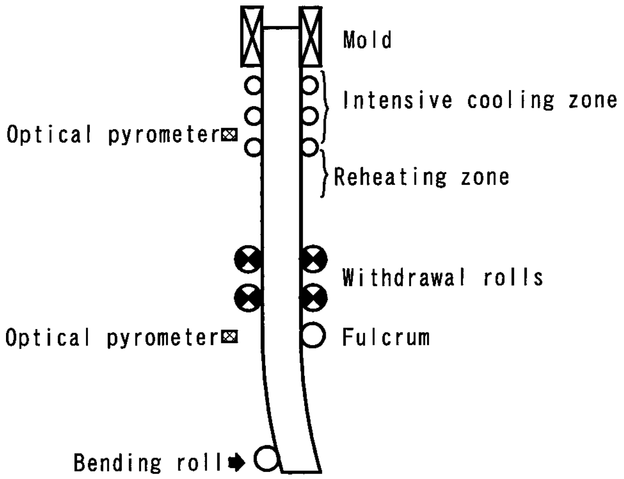

Fig. 2. Schematic illustration of slab bending test by a pilot continuous caster.

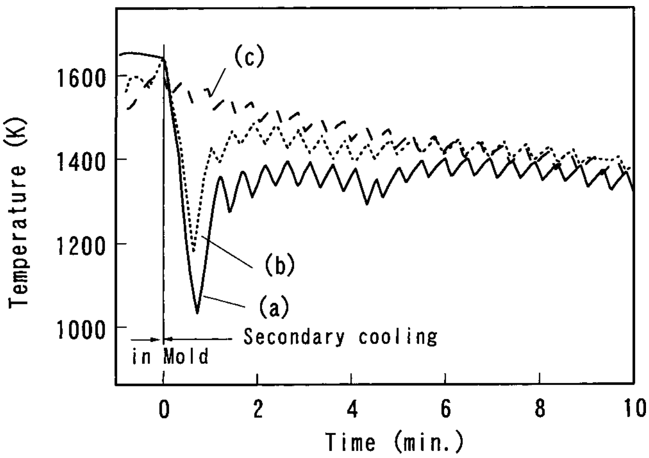

Fig. 3. Temperature profile for ingot cooling test measured at 5 mm inside to the surface.

2.4. Continuous Casting Test

Effect of the microstructure on the cracking susceptibility is actually assessed by slab bending tests. **Figure 2** shows the schematic illustration of the pilot caster used. The caster, which is essentially a vertical, slab-bending device is installed below the pinch rolls. The slabs 150 mm thick and 600 mm wide were cast at 1.0 m/min, standing by in the strand to adjust the surface temperature, then bent at the prescribed surface temperature. All processes were completed within 1 200 s after casting. Water density of intensive cooling zone is altered to control the slab surface microstructure. Slab surface temperature is measured by optical pyrometers set at the bottom of intensive cooling zone and bending zone, which is accompanied by air purge to get rid of water due to mist cooling. Bending radius was 10 m and average bending strain and strain rate at the slab surface were estimated as 0.8% and 2×10^{-4} 1/s respectively, corresponding to those of conventional caster. Surface cracks were detected by using the paint test on a machined surface.

3. Results

3.1. Ingot Cooling Test

Figure 3 shows some examples of thermal histories measured from 5 mm inside to the ingot surface. Temperature oscillation in each profile is due to intermittent cooling, which controls the cooling rate. In one cooling, the ingot is gradually cooled after withdrawal under mild cooling just as in a conventional continuous casting operation. In other coolings, the ingot is rapidly cooled after withdrawal and subsequently reheated up to 1 250 K by its own heat capacity. The minimum temperature during the rapid cooling is variously altered and Fig. 3 shows examples of (a) 1 050 K and (b) 1 170 K respectively,

Figure 4 shows microstructures at 5 mm inside to the ingot surface. Metallographic examination revealed that the features of microstructures are rather different, though either fundamental microstructure is ferrite–pearlite. A characteristic structure is formed under pattern (a), where substantial volume of ferrite is idiomorph, which seems granular morphology, whereas grain boundary allotriomorph of ferrite is formed under pattern (b) and (c), as conventionally produced slabs. In pattern (a), austenite grain boundaries are obscure because there is no ferrite allotriomorphs association along the grain. Similar structure is obtained when the minimum temperature during intensive cooling are between 870 K and 1 050 K. As strain concentration at ferrite allotriomorphs cause transverse cracking, reduction of them is expected to prevent cracking. On the other hand, it is conspicuous along grain boundaries under pattern (b) and (c), showing minimum temperature of 1 170 K during intensive cooling, pattern (b), is inadequate for microstructure control. Thus grain boundaries can be easily traced with them, as conventionally produced slabs. In this paper, secondary cooling pattern for microstructure control, which provides momentary intensive cooling until less than 1 050 K and subsequent reheating is defined as SSC cooling.

3.2. Hot Tensile Test

Hot ductility *in-situ* solidification was plotted in **Fig. 5** for both cooling patterns. Representative ductility curve as being made up of three regions is obtained under mild cooling. SSC cooling, however, gave a significant improvement on the ductility, thus tough embrittlement almost disappeared. For specimen having low ductility, *i.e.* RA of $<$40%, crack lies within the grain boundary allotriomorphs of ferrite and the fracture mode is intergranular brittle. It is, however, transgranular ductile under SSC cooling, which possesses higher ductility.

Microstructure of specimen, which passed through prescribed thermal history, however, without strain, is shown in **Fig. 6**. Prior austenite grain boundary is associated with ferrite allotriomorphs under mild cooling, whereas such ferrite grain could not be seen at austenite grain boundary under SSC cooling corresponding to ingot cooling tests. These results suggest that microstructure control possibly reduces susceptibility to transverse cracking. Further, similar grain size is obtained in the specimen between those coolings.

Hot ductility without remelting is plotted in **Fig. 7**. Hot tensile tests have been generally carried out under reheating process, and clarified that transverse cracking is related

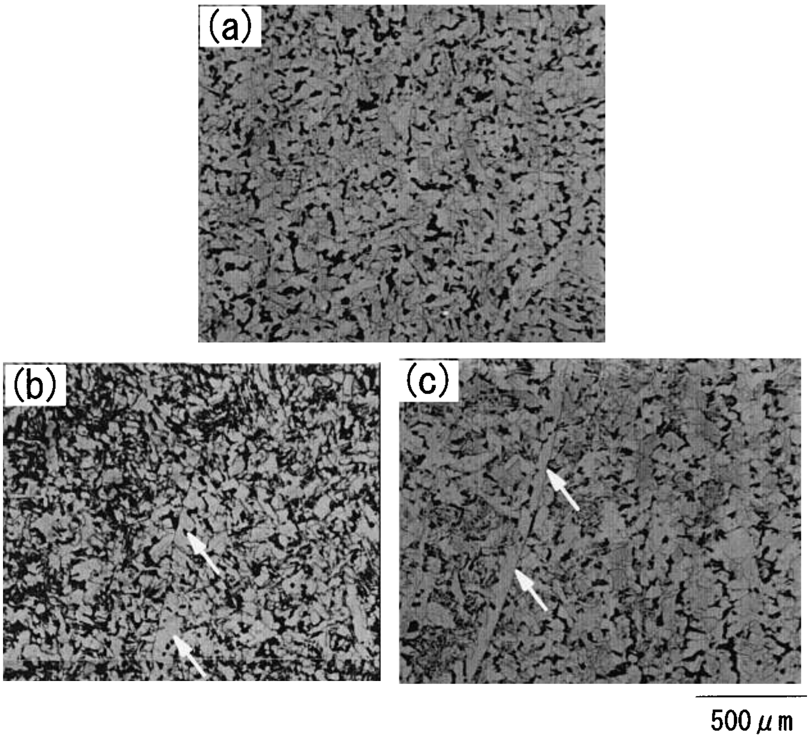

Fig. 4. Microstructure of the ingots from 5 mm inside to the surface under respective cooling patterns shown in Fig. 3. Arrows in (b) and (c) indicate grain boundary allotriomorph of ferrite.

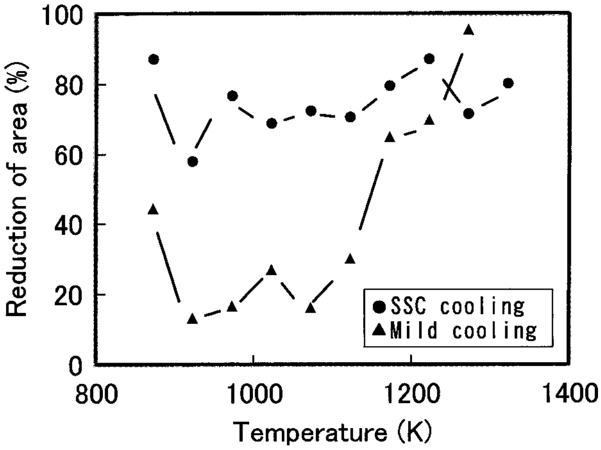

Fig. 5. Effect of microstructure control on reduction of area at hot tensile test.

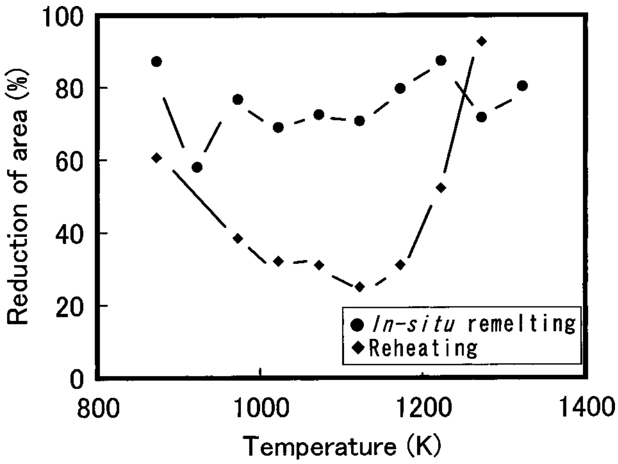

Fig. 7. Effect of *in-situ* remelting and solidification on reduction of area under SSC cooling.

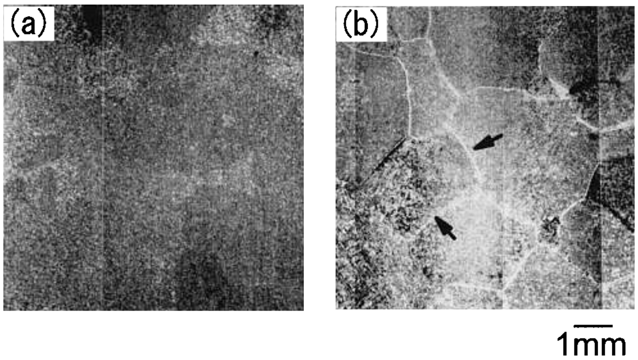

Fig. 6. Microstructure of the hot ductile specimen without deformation under (a) SSC cooling and (b) mild cooling. Arrows in (b) indicate grain boundary allotriomorphs of ferrite.

with hot ductility around γ–α transformation temperature.[1–5] Although precipitates are considered to be resolved during homogenizing treatment during reheating process, present study revealed that the treatment was insufficient to simulate as-cast properties. Remelting prior to deformation is indispensable to evaluate the effect of microstructure on susceptibility to transverse cracking.

3.3. Continuous Casting Test

Slab surface temperature measured at intensive cooling zone and bending zone is shown in **Fig. 8**. Although it is difficult to measure them throughout the experiment due to the water drops of mist cooling, those at intensive cooling zone are about 1 000 K under SSC cooling and 1 300 K under mild cooling, respectively. Those at bending were

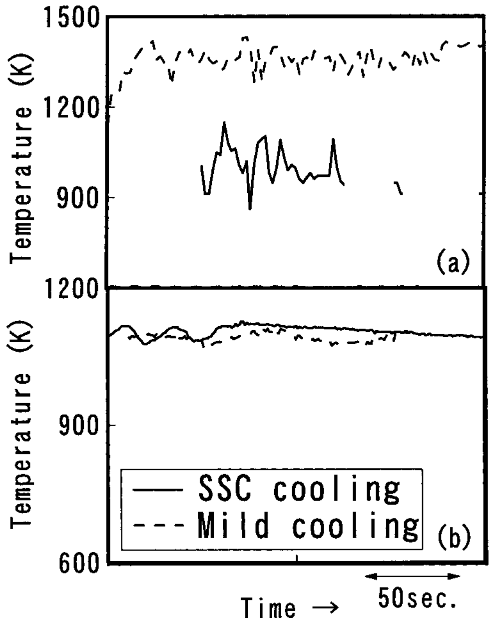

Fig. 8. Surface temperature of continuously cast slabs measured at (a) bottom of intensive cooling zone and (b) bending zone.

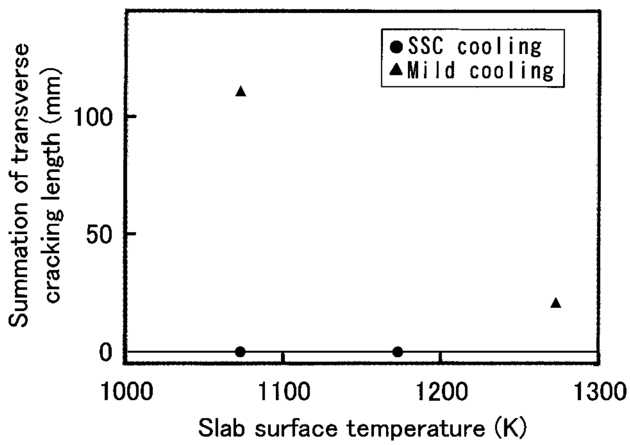

Fig. 9. Influence of thermal history on the slab surface transverse cracking investigated by a pilot continuous caster.

about 1 070 K in both cooling conditions. Thus, the aimed thermal history as examined in ingot-cooling tests and hot tensile tests could be reproduced in continuous casting tests.

Summation of transverse cracking length measured on the surface of a half width with 400 mm in length is shown in **Fig. 9**. The transverse cracks take place along the grain boundary under mild cooling, just as in commercial products. A number of cracks are found on the surface under mild cooling bent at 1 070 K. Some cracks still remain even at 1 270 K despite leaving ductility trough. In contrast, no crack is found under SSC cooling both at 1 070 K and 1 170 K bending. Although thermal stress is likely to increase according to intensive cooling, the transverse cracks are obviously alleviated by microstructure control, *i.e.* applying the SSC cooling. Other surface cracking or sub-surface cracking is not found under SSC cooling. Moreover, the cracking could be reproduced experimentally despite rather different dimension with a conventional caster.

Surface microstructure of these continuously cast slabs were shown in **Fig. 10**. Idiomorph of ferrite structure is formed under SSC cooling, whereas ferrite allotriomorphs structure associated with austenite grain boundary is formed under mild cooling, corresponding to ingot cooling tests and hot tensile tests. The results proved that susceptibility to transverse cracking could be reduced by appropriate surface microstructure control by means of secondary cooling in continuous casting strand.

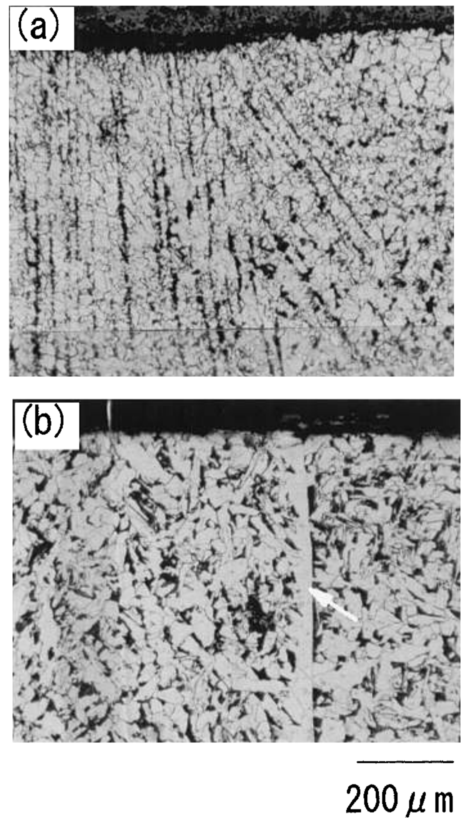

Fig. 10. Slab surface microstructure of continuously cast slabs under (a) SSC cooling and (b) mild cooling. An arrow in (b) indicates grain boundary allotriomorphs of ferrite.

3.4. Austenite Grain Size

As fine austenite grain size is well known to alleviate cracking susceptibility,[5–7] those sizes in each cooling conditions are examined. However, the microstructure without ferrite allotriomorph association at grain boundary under SSC cooling makes difficult to measure the size. Therefore another ingot cooling tests were conducted, in which the thermal history in the early stage were the same and the cooling rate in the latter half were increased. The operation makes microstructure from ferrite–pearlite to bainite on which austenite grain boundaries are easily observed. Similar experiments under SSC cooling and mild cooling were carried out, and then austenite grain size was examined on cross sections. Austenite grain structure at surface region is shown in **Fig. 11**. The austenite grain size within 5 mm from surface under SSC cooling is equivalent with that under mild cooling. Therefore, the improvement of hot ductility and prevention of transverse cracking are not a result of austenite grain refinement due to recrystalization.

3.5. Carbide and/or Nitride Precipitation

Micrographs of carbon extraction replicas from continu-

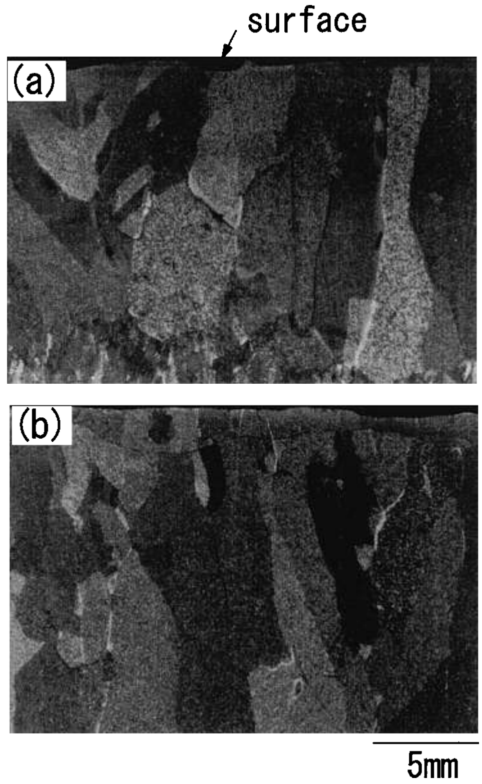

Fig. 11. Austenite grain size of the ingots under (a) SSC cooling and (b) mild cooling.

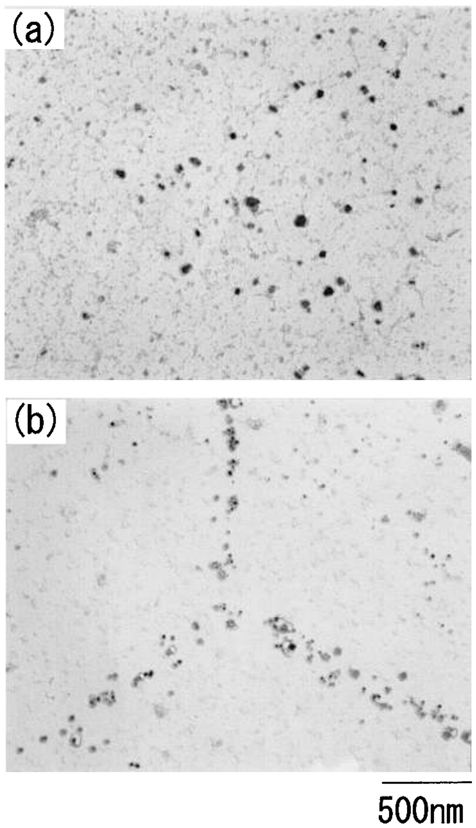

Fig. 12. Extraction reprica image of fine precipitates at austenite grain boundary and matrix under (a) SSC cooling and (b) mild cooling.

ously cast slab surface region, bent at 1 070 K after SSC cooling and mild cooling, are shown in **Fig. 12**. In the specimen after SSC cooling, free from cracking, very fine precipitates, less than 20 nm in diameter, are dispersed within the grains, as shown in Fig. 12(a). It should be noted that grain boundary precipitation was not observed in this specimen. In the meanwhile, precipitates line up along prior austenite grain boundary under mild cooling, providing cracking, as shown in Fig. 12(b). The mean precipitate diameter is −100 nm, larger than precipitates dispersed in the grain. The fine precipitates dispersed in the matrix are much fewer than that in Fig. 12(a). Similar dispersion of precipitates is obtained in the specimen after hot tensile test under both cooling conditions.[13]

The precipitates extracted from continuously cast slab surface region were analyzed. EDS and diffraction pattern analysis identified the precipitates along the prior austenite grain boundary after mild cooling as (Ti, Nb)(C, N) as shown in **Fig. 13**. As the fine precipitates dispersed in the matrix shown in Fig. 12(a) were too fine for similar analysis, dark field image and diffraction pattern were undertaken as shown in **Fig. 14**. Some brightness could be seen as relatively large precipitates identified as (Ti, Nb)(C, N). Interplanar spacing measured from diffraction pattern is consistent with them in ASTM card, as shown in **Table 2**. Therefore, very fine precipitates transgranularly dispersed must be also (Ti, Nb)(C, N).

4. Discussion

4.1. Alleviation of Cracking Susceptibility

Experimental results prove that slab surface microstructure could be controlled by secondary cooling condition. Microstructure free from grain boundary ferrite allotriomorphs, where strain concentrated during unbending operation of continuous casting strand, could be obtained as shown in Fig. 4(a) by SSC cooling, which provides thermal cycle in the early stage of secondary cooling.

Hot tensile tests *in-situ* solidification revealed that hot ductility is much improved by SSC cooling as shown in Fig. 5, and consistent with the experimental results of slab bending test. However, similar thermal cycle but without remelting deteriorate hot ductility. Although cracking susceptibility is well known to correlate with conventional tensile test without remelting,[1,2,5,6] this result obviously indicates that even the homogenizing treatment at very high temperature prior to hot tensile test is insufficient to evaluate the cracking susceptibility of continuously cast slab considering microstructure.

Continuously cast slab bending tests demonstrate that transverse cracking could be prevented by the microstructure control as shown in Fig. 9. Uniform dispersion of fine precipitates as shown in Fig. 12(a) under SSC cooling causes ductility improvement, and consequently alleviates the susceptibility to transverse cracking.

4.2. Microstructure Control

Although the early stages of the cooling are varied between SSC cooling and mild cooling, the last stages of them are similar. Therefore thermal cycle leaves a trace as unknown substance in the matrix during reheating opera-

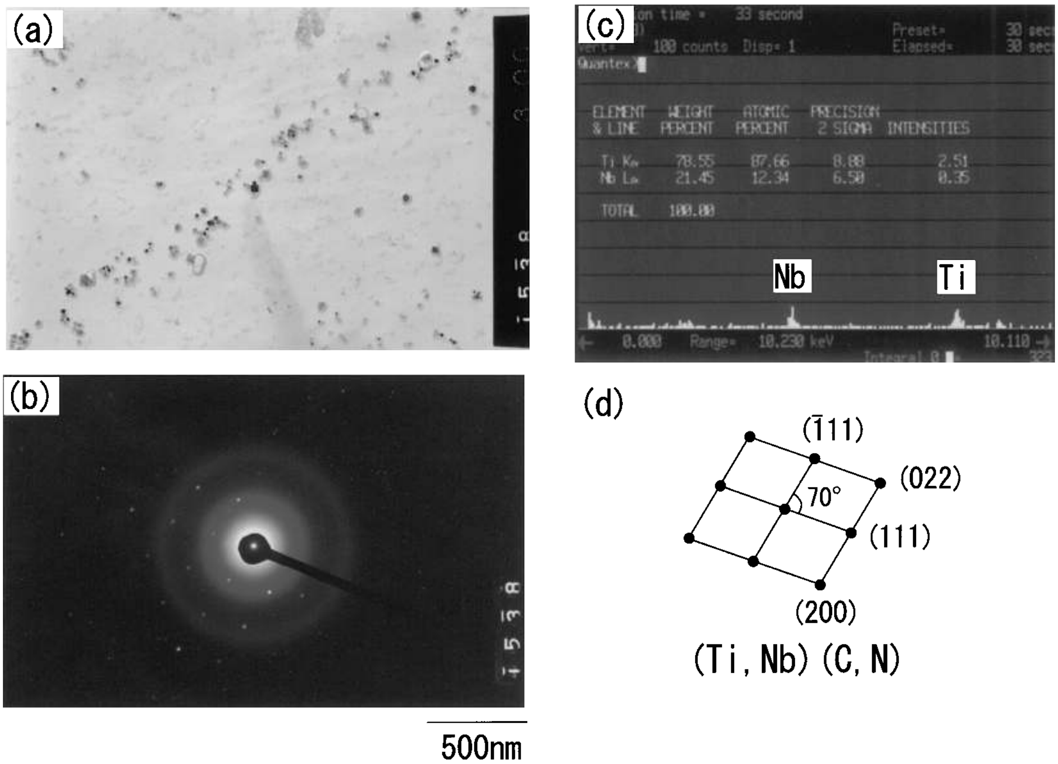

Fig. 13. Analysis of grain boundary precipitates under mild cooling by electron diffraction pattern and EDS (a) bright field image, (b) dark field image, (c) EDS spectra, and (d) schematic representation.

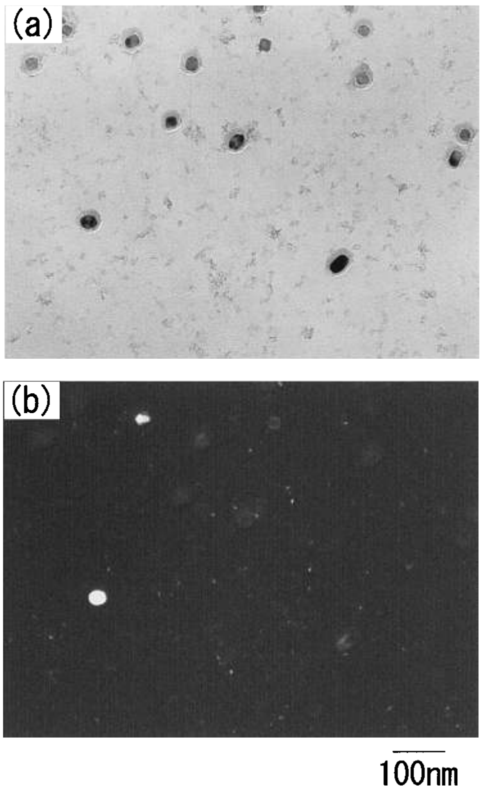

Fig. 14. Analysis of fine precipitates under SSC cooling (a) bright field image and (b) dark field image.

tion under SSC cooling.

According to Andrew's equation,[14] A_3 transformation temperature is estimated as 1 090 K. Cracking insensitive microstructure is obtained when minimum temperature dur-

Table 2. Comparison of inter planar spacing.

Measured	(Ti,Nb)(C,N) by ASTM card	
spacing	spacing	hkl
2.51	2.508	111
2.16	2.179	200
1.53	1.535	220

ing initial intensive cooling is below this temperature, whereas it is not obtained when minimum temperature does not reach that temperature, as shown in Fig. 4. These results indicate that phase transformation begins during initial intensive cooling under SSC cooling.

In general, the austenite grain size is refined by recrystallization, as commonly applied for mechanical property control of steel. It has been well known that fine austenite grain size bring improvement of hot ductility.[5–7] The austenite grain size close to the slab surface under SSC cooling, however, is equivalent with that under mild cooling as shown in Fig. 11. Although measured temperature should involve some error due to dull response and accuracy, these results show that phase transformation has never completed during thermal cycle. That is, the change in the microstructure according to secondary cooling does not originate in the refinement of the austenite grain size by recrystallization.

As the temperature reaches less than A_3 temperature during the thermal cycle, ferrite grain must once begin to precipitate. Such precipitates as (Ti, Nb)(C, N) must precipitate simultaneously. Solubility product of TiN, which is most liable to precipitate among carbide or nitride, in austenite and ferrite phase is expressed as Eqs. (1) and (2), respectively.[15,16] And they are shown in **Fig. 15**.

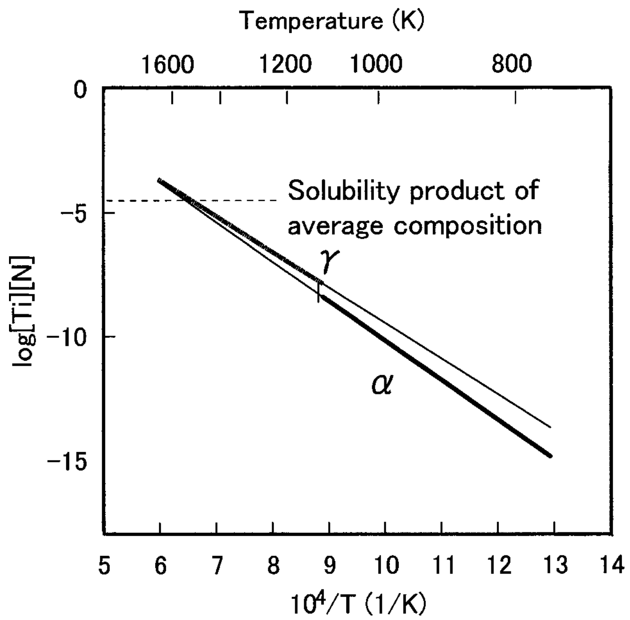

Fig. 15. Solubility product of TiN in ferrite and austenite calculated by Eqs. (1) and (2).

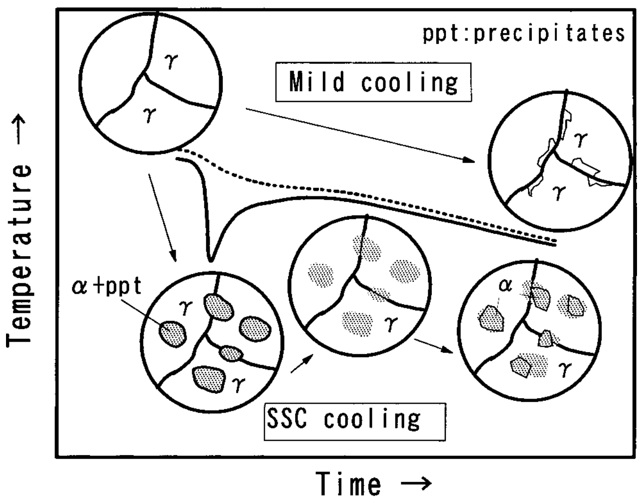

Fig. 16. Schematic illustration of mechanism forming microstructure.

$$\log[\mathrm{Ti}\%][\mathrm{N}\%]_{\gamma} = -14\,400/T + 4.94 \quad \text{.........(1)}^{15)}$$

$$\log[\mathrm{Ti}\%][\mathrm{N}\%]_{\alpha} = -15\,960/T + 5.79 \quad \text{.........(2)}^{16)}$$

Substituting the actual content and assuming equilibrium, the initiative of TiN precipitation is estimated at 1 583 K in austenite phase. Note that solubility product in ferrite phase is one digit smaller than in austenite below A_3 temperature. Slab surface transverse cracking is well known to be affected by carbide and/or nitride precipitation. Such precipitation at austenite grain boundary or matrix causes strain concentration and low ductility during unbending operation.

Behaviour of microstructure and precipitates during SSC cooling and mild cooling is schematically illustrated as shown in **Fig. 16**. In SSC cooling, ferrite grain must first precipitate during the thermal cycle. When cooling rate is fast enough and sufficient supercooling is achieved, the ferrite grains precipitate not only at austenite grain boundary but also inside the grain. According to the low solubility in ferrite phase compared with the austenite, fine (Ti, Nb)(C, N) precipitates simultaneously with ferrite precipitation. During the reheating, all ferrite must transform to austenite again. However, fine precipitates must be undissolved within the re-transformed austenite because of inadequate solubility product shown as Fig. 15. At the following stage of cooling, the ferrite grain for the second time must originate from these fine precipitates. On the other hand, low cooling rate due to gradual cooling cause ferrite allotriomorph precipitation at the prior austenite grain boundary in mild cooling.

Improvement of hot ductility and alleviation of cracking susceptibility consequent upon uniform dispersion of fine precipitates is consistent with these behaviors. A significant increase in hot ductility seems to be attributed to restraining of grain boundary allotriomorphs in γ–α duplex phase region, and uniform dispersion of fine precipitates in low temperature austenite region. In the SSC cooling process, slab surface temperature must reach below the A_3 transformation temperature during the intensive cooling in order to control the microstructure. As the precipitation need some time and supercooling, adequate cooling must be indispensable during the thermal cycle. Some research concluded that thermal cycle prior to hot tensile tests deteriorates hot ductility.[9] According to their quenched specimen, transgranular precipitation of ferrite never takes place during the thermal cycle. Inadequate temperature drop must cause hot ductility decrease consequent upon ferrite allotriomorphs and ranging fine precipitate formation along austenite grain boundary, as they mentioned. The austenite grain size at surface region is equivalent under both SSC and mild cooling, as shown in Fig. 11. Thus, phase transformation does not need to be completed during the thermal cycle under SSC cooling.

5. Conclusions

(1) Ingot cooling tests proved that the surface microstructure of continuously cast slab could be changed by secondary cooling condition. SSC cooling providing initial intensive cooling less than A_3 transformation temperature and subsequent reheating up to 1 250 K in secondary cooling, bring reduction of grain boundary allotriomorph of ferrite.

(2) A new hot tensile test after *in-situ* remelting and solidification proved that the microstructure control significantly improve the hot ductility itself. As for the hot tensile test, remelting of specimen preceding the deformation is indispensable to evaluate the effect of microstructure on susceptibility to transverse cracking.

(3) Slab surface transverse cracking could be reproduced experimentally by a pilot caster. Susceptibility to transverse cracking could be alleviated with this microstructure control.

(4) Replica examination proves that this microstructure change is accomplished by utilization of carbo-nitride precipitation. Very fine precipitates of (Ti, Nb)(C, N), less than 20 nm in diameter, are dispersed transgranularly under SSC cooling.

(5) This slab surface microstructure control can be adapted to the continuous casting process with mere im-

provement of conventional secondary cooling system.

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