1. Introduction

It is well known that continuously cast steel exhibits low ductility in the solidification range. This behavior has been recognized to be responsible for the problem of internal cracking, and sometimes, surface cracking if the internal crack propagates to the surface in continuous casting. In the study on hot cracking, several experimental methods such as ‘Submerged Split-Chill Tensile’ and ‘in-situ melt bending test’ and ‘in-situ solidified test’ etc., have been used to measure critical strain for hot cracking during continuous casting. However, since the measured data varies widely depending on the experimental conditions, reliable trend in cracking susceptibility with chemical compositions according to alloying elements has not been obtained by direct measurements. Only an evaluation equation for critical strain, which was obtained through curve fitting using reported data from many studies, has been reported. In general, hot cracking is believed to occur in alloys with a large freezing range, to be precise, a large brittle temperature range between zero-strength-temperature (ZST) and zero-ductility-temperature (ZDT). However, while the Fe–Ni alloys have a narrow solidification interval, they show very high cracking susceptibility during continuous casting.

The purposes of this study were to examine the hot cracking behavior of Fe–Ni alloys by using a high-speed camera to observe hot cracking in samples that were temperature-gradient heated and tensile deformed. The susceptibility to hot cracking was reliably evaluated based on critical strain and crack propagation length as functions of Ni content and cool side temperature. Grain size and shape in the brittle temperature region depended on liquid boundary migration due to solute diffusion within the liquid boundary under temperature gradient. High susceptibility to hot cracking of Fe–Ni alloys could be explained by the high sulfur segregation near the grain boundary and the large grain size within the solid/liquid coexisting region due to the rapid migration of the liquid boundary.

2. Experimental Procedures

The chemical compositions of the steel used in this study were represented in Table 1. A range of Ni content steel samples were used (0.07%C–0.021–9.46%Ni). The center in one side of each specimen, 150 mm in length, 20 mm in width, 10 mm in thickness, was heated in Ar atmosphere at 6°C/s to about 1 350°C, heated at 0.5°C/s to about liquidus, and held for 500 sec for homogenization. At this time, cool side temperature was about 1 250°C, cooled by Ar or He gas at −2.5°C/s to the target temperature (950–1 250°C), and held for 120 sec to obtain a steady temperature gradient profile across the entire width. The specimen was then subjected to mechanical tests in tension. Temperature was measured with a hot thermocouple located outside the melted region of the hot side, and compensated by the temperature difference between the center region of the hot side and the periphery of the melted region before melting occurred. Tensile tests were carried out at a stroke rate of 2 mm/s for 1 sec. During hot deformation, hot cracking near the hot side was monitored and recorded by using a high-speed camera. After deformation, the hot side of the temperature-gradient specimens were cooled at −1°C/s to cool side temperature to prevent existing cracks from propagating further, and then both sides were air cooled to room temperature. Another cooling procedure used after deformation was to immediately spray water on the hot side to investigate the in-situ microstructure developed during the hot deformation. The temperature-gradient tensile test combined with high-speed camera is schematically represented in Fig. 1. All the test specimens were sectioned parallel to the width direction that corresponds to the temperature-gradi-
ent direction, and examined metallographically. Samples with hot cracks were dipped in liquid nitrogen and fractured by impact to reveal the crack surface. Auger analysis was performed in a Physical Electronics Phi 600 Auger microprobe by slicing off the measured surfaces at a rate of 0.217 nm/s to investigate the microsegregation near the crack surface.

3. Experimental Results

3.1. Evaluation of Crack Susceptibility

Figure 2 shows hot crack occurrence followed by melt flowing into the crack near the hot surface during tensile deformation. Only a portion of the specimen’s width is shown due to the cover that prevents heat loss. In order to evaluate critical strain for hot cracking, it is necessary to know the length of the uniform deformation zone, the exact time deformation started and the time the crack occurred. Figure 3 shows the measured displacements of various positions within the deformation zone using a program that allows tracking a movement in any position during hot deformation. From this result and the fact that only the upper jig in the tester moves up during the tensile test, the length of the uniform deformation zone could be defined by roughly

| Sample | C  | Si  | Mn  | P  | S  | Cr  | Ni  | Al  | Cu  |
|--------|----|-----|-----|----|----|-----|-----|-----|-----|
| A1     | 0.087 | 0.266 | 0.63 | 0.012 | 0.012 | 0.032 | 0.021 | 0.002 | 0.031 |
| A2     | 0.07 | 0.249 | 0.635 | 0.012 | 0.013 | 0.031 | 1.47 | 0.017 | 0.031 |
| A3     | 0.071 | 0.242 | 0.64 | 0.011 | 0.012 | 0.029 | 3.35 | 0.03 | 0.03 |
| A4     | 0.071 | 0.245 | 0.643 | 0.008 | 0.013 | 0.026 | 8.51 | 0.03 | 0.03 |
| A5     | 0.069 | 0.252 | 0.643 | 0.006 | 0.012 | 0.025 | 9.46 | 0.027 | 0.029 |

Fig. 1. The apparatus of thermal-gradient tensile and high speed camera test for simulating hot cracking during continuous casting.

(a) Crack initiation and melt flowing into crack

(b) Stop of melt flowing

Fig. 2. Direct observation of crack initiation using in-situ thermal-gradient tensile and high speed camera test.

Fig. 3. Displacement of the various positions within the deformation zone. (stroke: 2 mm, time: 1 sec)
assuming the curve of displacements and the relative positions of the measured points to be linear. The times of initial deformation and crack occurrence could also be determined through tracking the movements of points within the deformation zone.

Figure 4 shows that the critical strain, using the above method, is sharply decreased from 5.3 to 2.2% with increases of Ni content from 0.021 to 3.35% and slowly decreased thereafter. Austenite phases formed during solidification cause larger microsegregation\(^\text{12,13}\) due to much lower diffusivities of solute elements in Austenite phases than in Delta ferrite phases. The sharp decrease in critical strain up to \(\sim 3.5\%\) Ni may correspond to the fact that residence time during solidification within Delta ferrite phase decreases with increases in Ni content and according to the Fe–Ni phase diagram, the Austenite phase starts to form from \(>\sim 4.1\%\) Ni. Figure 5 shows that critical strain is increased from 1.9 to 2.5% by decreasing the cool side temperature of the specimen from 1250 to 950°C. Since the various cool side temperatures were obtained with the same cooling rate, which implies the supersaturation of vacancies\(^7\) of each specimen was equal, the variation in critical strain can be related to the area where liquid coexists with solid. Lower temperature on the cool side results in a smaller area of solid/liquid co-exist region due to a higher temperature gradient.

Figure 6 shows hot cracks fully formed in the solid/liquid co-existing region with identical cool side temperatures of 1100°C for 0.021% Ni and 9.46% Ni steel samples. In the case of 0.021% Ni steel, the crack shape is straight, the length is very short and its direction seems to be parallel to the heat flow, as shown in Fig. 6(a). In the case of 9.46% Ni steel, it is rather zigzag, comparatively lengthy, and doesn’t seem to be parallel to the heat flow, as shown in Fig. 6(b). The reliability of measuring critical strain could be confirmed by the fact that several cracks in one sample initiated at the same time. Figure 6 represents that Ni content greatly affects the crack propagation as well as the initiation of the crack.

Figure 4 represents the crack propagation length with variations in Ni content. Crack length is greatly increased from 1.6 to 4.0 mm with increases in Ni from 0.021–9.46%. Crack length as well as critical strain seem to depend on the length of the brittle region (defined as the region between ZST and ZDT). However, the hot side temperature of higher Ni content steel, which may be lower due to a lower melting point, results in a lower temperature gradient and a shorter length of the brittle region. Therefore, the increase in cracking susceptibility with Ni content does not seem to have any relationship with the length of the brittle region. Figure 5 also shows that decreasing the cool side temperature of the specimen from 1250 to 950°C decreased crack length from 5.1 to 3.1 mm. Since the various cool side temperatures were obtained with the same cooling rate, which implies the supersaturation of vacancies of each specimen was equal, the variations in crack length can be related to the length of the brittle region between ZST and ZDT. Lower cool side temperature results in a shorter length of the brittle region due to a higher temperature gradient.

### 3.2. Metallographic Observations

Figure 7 shows the microstructure of A1 sample quench-

![Fig. 5. The effects of cool side temperature on critical strain and crack length at the solid/liquid coexisting region. (Shell thickness: 20 mm)](image)

![Fig. 6. The effect of Ni content on crack length. (a) Sample A1 (0.021 % Ni), (b) sample A5 (9.46 % Ni).](image)
ed on the hot side after being temperature-gradient heated. The hot side temperature was about liquidus and the cool side temperature was 900°C. Three regions could be characterized depending on the temperature range. From the hot surface, fully melted white region, the elongated grain region that seems to be formed in the solid/liquid coexisting zone, and the equiaxed grains that grow with time, could be confirmed in order of decreasing temperature. Interestingly, rapid grain growth along the direction of heat flow occurred in the solid/liquid coexisting region where hot cracks readily formed. Figure 8 shows the microstructure of A3 sample (higher Ni content), quenched on the hot side, after being temperature-gradient heated and tensile deformed. The hot side temperature was about liquidus and the cool side temperature was 1100°C. Three regions also could be characterized depending on the temperature range. Much larger grains, which imply much faster grain growth, were formed in the solid/liquid coexisting region compared with sample A1. Hot cracks were found to propagate along grain boundaries and resulted in intergranular cracking.

The solute element content near the crack surface was measured using an Auger Electron Microscope with a sputter rate of 0.217 nm/s as shown in Fig. 9. High sulfur content is worthy of note. High carbon and oxygen near the surface may be due to contamination and oxidation, respectively. Highest sulfur was found around ~0.1 μm from the surface and may correspond to the original crack surface before oxidation, if it does not move by the Kirkendall effect.

4. Discussion

4.1. Evaluation of Crack Susceptibility

The new method of direct observation of hot cracking developed in this study has many advantages. First, it is very reliable because shell thickness is kept constant during the experiment, unlike other methods where solidified shell thickness varies and uniform thickness cannot be guaranteed, which may produce local strain concentration and result in underestimation of the critical strain. Second, temperature-gradient and cool side temperature can be readily simulated for continuous casting. Third, unlike other methods where many experiments are required to get one value of critical strain, the new method requires only one experiment to obtain one value of critical strain. However, since the sample is not fully melted across the width and partly solidified during cooling of the cool side, it does not accurately simulate the microstructure solidified during continu-
ous casting. While reliable trends in hot cracking susceptibility can be obtained, the measured values of critical strain need to be compensated for in order to determine true values of critical strain, because crack occurrence is defined as the time it becomes visible, not when it actually occurs. This compensation can be accomplished by assuming that the crack width is proportionally increased with increases in strain.

4.2. Characteristics in Microstructure

Figure 10 shows that austenite grain boundary is enriched by Mn, Ni in the A3 sample heated to 1300°C, held for 5 min and water quenched. Austenite is transformed to liquid instead of delta in case of Ni/11022/3.5 % during heating based on C–Fe–Ni phase diagram. Therefore, in the A3 sample with 3.35 % Ni, austenite grain boundary with higher content of Ni than 3.5 % may be melted into liquid film without austenite-delta transformation as shown in Fig. 8. Subsequently, liquid boundary may retard delta nucleation at austenite grain boundary to peritectic temperature. At higher temperature delta phase may start to nucleate and grow at the expense of austenite phase. Meanwhile, when liquid amount is enough to wet the whole boundary of a delta grain, it is separated from the solid sample and fall into the liquid layer resulting in delta grains partially wetted by liquid. As a result, only 100 % liquid layer at the hot surface and grains with liquid grain boundaries exist in the sample. Figure 11 shows the quenched microstructure of the sample with 1.47 % Ni content that has thicker delta phase layer than that of the sample with 3.35 % Ni, where delta phase layer is very thin as shown in Fig. 8. All grains grow by liquid grain boundary migration, resulting in larger austenite grain and smaller delta grain. The smaller delta grain may be due to the incubation time to nucleation and shorter time for grain growth. The location where liquid starts to form is thought to correspond to the one where sudden increase in austenite grain size is found. Basically, if grain growth occurs within solid phase, grain size should increase exponentially with temperature. Therefore, sudden increase is believed to relate to the liquid film boundary. The grain growth by liquid boundary migration under temperature gradient is discussed in the following chapter.

4.3. Grain Growth under Temperature Gradient and Hot Cracking Susceptibility

The region where grains grow along the heat flow direc-
tion was clearly verified by metallographic examination on the microstructure of samples that were temperature-gradient heated followed by water quenching as shown in Figs. 7 and 8. If the temperature is assumed to be linear through the width of each sample, in the case of sample A1, the real solidus (Zero Ductility Temperature) calculated based on Nagata’s empirical equation corresponds to about 1.73 mm from the fully melted layer in the hot surface. This is nearly equal to the position between the elongated grains and equiaxed grains in Fig. 7. That is, the region of elongated grains may correspond to the solid/liquid coexisting region.

On the other hand, steel containing Ni reveals much larger grains that implies much faster grain boundary migration. While the calculated ZDT locates at about 2.3 mm from the fully melted layer, large elongated grains are observed 3.9 mm from the fully melted layer. That reflects a great difference between the calculated ZDT position and the solid/liquid boundary determined by the microstructure. The region where very large grains develop and cracks propagate reaches lower temperatures than the calculated ZDT. In Fig. 12, crack lengths and ZDT positions according to cool side temperature are represented. The distance from the hot surface to the ZDT position and crack length increases with an increase in cool side temperature. It indicates that crack propagation readily occur when the increase in cool side temperature extends the length of the brittle region (ZST-ZDT).

Meanwhile, high sulfur segregation near the hot crack surface was observed and is shown in Fig. 9. As stated above, the location of highest sulfur may correspond to the original crack surface before oxidation, if we don’t consider the Kirkendall effect. When high sulfur content at the crack surface is considered, the ZDT should be much lower than the calculated one, which may result in crack propagation beyond the ZDT position without serious sulfur segregation. Kim, Hansson and Fredriksson have observed that an Invar alloy with a sulfur content of 14 ppm is more susceptible to hot cracking than an alloy with a sulfur content of 4 ppm. They suggested that S atoms segregated to the boundaries could quantum mechanically weaken the boundaries and the work for crack nucleation decrease as a result of a reduction in the surface tension due to S segregation. Therefore, they suggested that hot cracking could occur even when liquid films are not present along the grain boundaries. However, from the results of this study, in the steel containing Ni, sulfur content segregated to the boundaries seems to be much higher than in any other steel. Therefore, the decrease of solidus of the grain boundary due to S segregation in steel containing Ni should be much larger than that in any other steel. This high S segregation may cause high susceptibility to hot cracking in steel containing Ni. Additional research is warranted to determine why S segregates more in steel with Ni than without.

Another reason for high susceptibility to hot cracking in steel containing Ni can be large elongated grains along the direction of heat flow. In the solid/liquid coexisting region under temperature-gradient, grain boundary can be assumed to be the liquid boundary as depicted in Fig. 13. If local equilibrium is satisfied at solid/liquid boundaries, equilibrium concentration at the grain/liquid boundary is
larger than at the grain 2/liquid boundary due to temperature gradient. Therefore, solute diffusion from the grain 1/liquid boundary to the grain 2/liquid boundary occurs because of the concentration gradient. Instantly, local equilibrium at the solid/liquid boundaries should be broken. At the grain 1/liquid, undercooling due to lower concentration than equilibrium value and at the grain 2/liquid, superheating due to higher concentration, should occur. Consequently, the grain 1/liquid undercooled should be solidified and the grain 2/liquid superheated should be melted, in order to satisfy local equilibrium. It should result in liquid boundary migration toward the higher temperature. The rate of this liquid boundary migration should depend on the velocity of diffusion flow within the liquid boundary. In addition, these liquid boundaries move to the hot surface. Since the liquid boundary where the temperature is lower has a higher concentration, instant concentration gradient should develop within the surface liquid layer as depicted in Fig. 14. That is, the concentration at solid/liquid layer is higher than at liquid/Ar atmosphere due to the influx of high concentration liquid into the surface liquid layer. If local equilibrium is broken due to the increase in concentration at the solid/liquid layer boundary, superheating should occur at this boundary and result in melting to satisfy local equilibrium. If the hot surface is quenched by water spraying, which is carried out for microstructural investigation, unidirectional solidification should occur, starting from the solid/liquid layer boundary. The initial concentration along the distance from the solid/liquid layer decreases before unidirectional solidification. After solidification, the concentration would decrease near the solid/liquid layer boundary due to the decrease in initial concentration, and then increase because solute rejection during solidification is greater than the decrease in initial concentration, as depicted in Fig. 14. Figure 15 shows the Mn redistribution profile near the hot surface that was water-quenched after being temperature-gradient heated. It is verified that the concentration is decreased first and then sharply increased within the marked area that is believed to be a liquid layer before quenching. As predicted above, this solute redistribution

Fig. 13. The mechanism of liquid boundary migration under temperature gradient. (Local equilibrium at the solid/liquid boundary is assumed.)

Fig. 14. Solute redistribution due to s/l interface migration induced by temperature-gradient followed by rapid unidirectional solidification. (Local equilibrium at the solid/liquid layer is assumed.)
profile implies that high concentration liquid boundaries move to the hot surface and result in increasing the thickness of the liquid layer at the hot surface.

As stated above, the rate of liquid boundary migration should depend on the velocity of the solute diffusion within the liquid boundary. In the case of steel containing Ni, equilibrium concentrations of Ni in the solid/liquid coexisting region vary greatly with changes in temperature, as shown in the Fe–Ni phase diagram. Therefore, the Ni concentration gradient is very steep within the liquid boundary and Ni diffusion should take place rapidly due to the large driving force for diffusion. This may cause large elongated grains and result in high cracking susceptibility in steel containing Ni.

It has been reported that grain growth is suppressed by liquid or delta phase in grain boundary. Since there is no driving force to move liquid film due to same chemical free energy between adjacent two solid/liquid interfaces, grain movement may be retarded by this liquid phase. Meanwhile, since there should be driving force for solute diffusion within liquid film under the condition of temperature gradient and local equilibrium at solid/liquid interfaces, liquid film may move resulting in grain growth. Some papers have shown that liquid phase is believed to retard grain growth based on the experimental results that the increase in P content extends the region of fine γ grain and moves transition region from fine γ grain to coarse γ grain to lower temperature. It has also been mentioned that transition occurs at the critical liquid volume fraction, ~0.14. Meanwhile, in the present study on hot cracking under temperature-gradient, when liquid amount is enough to wet the whole boundary of a delta grain, it is separated from the solid sample and fall into the liquid layer near the hot surface. As a result, only small volume fraction of liquid remains in the sample. It indicates that the region of very coarse grain corresponds to that of small volume fraction of liquid and the region of relatively fine grain to that of complete solid phase (Figs. 7 and 8). Therefore, it could be concluded that only small volume fraction of liquid (that is, liquid film) may cause very coarse grain, particularly along the direction of heat flow and after complete solidification grain grows with much lower velocity. In other words, sudden increase in grain size (very coarse grain) is due to small volume fraction of liquid phase (that is, liquid film) and could be explained by the movement of liquid boundary under temperature-gradient. The trial to calculate the velocity of grain growth is needed to verify the effect of alloying elements in the future.

5. Summary

Hot cracking behavior of Fe–Ni alloys has been investigated by direct observation with high-speed camera of hot crack occurrence through temperature-gradient tensile tests. The following results were obtained.

(1) The occurrence of hot cracking was successfully simulated in temperature-gradient heated and tensile deformed samples. Cracking susceptibility was evaluated by estimating the critical strain and crack propagation length from images of cracking recorded by high-speed camera and the program that tracks a movement in any position during hot deformation.

(2) Crack initiation and propagation readily occur when the increase in cool side temperature extends the length of the brittle region (ZST–ZDT). However, the increase of Ni content greatly increases cracking susceptibility despite the decrease in the length of the brittle zone. This indicates that the length of brittle zone alone cannot explain the high cracking susceptibility of Fe–Ni alloys.

(3) Under temperature-gradient, liquid boundaries move toward the hot surface in order to satisfy local equilibrium at solid/liquid boundaries, resulting in large elongated grains within the solid/liquid coexisting region and increasing the thickness of the liquid layer at the hot surface. The rate of liquid boundary migration should depend on the velocity of solute diffusion within the liquid boundary.

(4) High cracking susceptibility of Fe–Ni alloys with short freezing ranges was explained in terms of very large grains developed due to fast migration of liquid boundaries and high sulfur segregation near the grain boundary. This finding deserves further research.

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