Assessment of Hot Ductility with Various Thermal Histories as an Alternative Method of *in situ* Solidification

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Much work has been devoted to improving the hot ductility of steels to make them less prone to transverse cracks during the unbending stage of continuous casting. As a more positive method to avoid cracks, developing a fine grain size within the surface and sub-surface region of a slab through a double phase transformation was proposed. The *in situ* solidification method is usually adopted for assessing the effect of a double phase transformation, resulting from intensive cooling and reheating, on hot ductility. There exist, however, some weak points to simulating the continuous casting process by an *in situ* solidification. In the present work, two kinds of thermal history were adopted during the tensile test to assess the effect of a double phase transformation more precisely, instead of *in situ* solidification, for simulating the unbending operation. Eliminating the perpetual experimental problems of *in situ* solidification tensile test, a solidified slab was used throughout the present work.

KEY WORDS: hot ductility; transverse cracking; double phase transformation; grain refinement.

1. Introduction

Transverse cracking and edge cracking are well known surface problems developed during the continuous casting of steel. These surface defects deteriorate slab properties, and the removal of such defects results in wasted material.1–3)

There has been much work to reduce these defects through various methods, such as control of the chemical composition of steel, surface structure control of the slab, and hot deformation on cooling.4–8)

Several experimental methods have been adopted to simulate the continuous casting process more precisely, inducing fatigue deformation on cooling, *in situ* solidification tensile tests, and tensile tests in an air atmosphere.1,7–9)

Among these methods, the *in situ* solidification tensile test has been known as the best tool to simulate the continuous casting process because this method could carry with it stress development and solidification cracking during solidification. However, *in situ* solidification tensile tests also have some drawbacks, i.e., void formation on solidification and interface phenomena between the mold wall and the specimen, which are perpetual experimental problems. Also, it is almost impossible to measure the temperature of the specimen precisely during tensile tests at high temperature, resulting in inaccurate data.

The hot ductility behavior of steels after solidification is greatly influenced by microstructural constituents, especially austenite grain size.1)

In this study, two kinds of thermal history were adopted to control austenite grain size, in which fine austenite grain size was obtained from the thermal history, due, it is thought, to a double phase transformation from reheating and cooling.

Eliminating the perpetual experimental problems of *in situ* solidification tensile tests, a solidified slab was used throughout the present work, where intensive cooling from the solution treatment temperature to below the Ar3 temperature, followed by reheating to 1100°C, was carried out.

2. Experimental Procedure

Specimens were obtained from an as-cast slab after conventional continuous casting. The composition of the steel examined in this study is given in Table 1. Specimens for hot tensile tests were machined from the slab with a diameter of 8 mm, a total length of 15 mm, and a parallel barrel of 60 mm. The tensile test was performed by a cast simulator in an Ar atmosphere and the strain rate was $5 \times 10^{-4}$/s throughout the experiment. To assess the effect of double phase transformations, the tensile test was performed with two kinds of thermal histories, as indicated in Figure 1.

In thermal history TH0 (Fig. 1(a)), the specimen was heated to 1300°C and held for 5 min for solution treatment. After the solution treatment, the specimen was cooled to

| Table 1. Chemical composition of steels examined (unit: wt%). |
|------------------------------------------------------------|
| C       | Si       | Mn       | P        | S        | Cu       | Nb       | Ni        | Cr        | Ti        | Al        |
|---------|----------|----------|----------|----------|----------|----------|----------|----------|----------|----------|
| 0.085   | 0.26     | 1.47     | 0.008    | 0.002    | 0.12     | 0.013    | 0.24     | 0.01     | 0.015    | 0.035    |
the test temperature with a cooling rate of 10°C/s. Then, a tensile deformation was applied to the specimen, directly followed by water quenching in a chamber. Thermal history TH5, which was designed to induce a double phase transformation, is given in Fig. 1(b). The specimen was cooled from the solution treatment temperature to below the Ar3 temperature, 600°C, at a rate of 20°C/s, which is a higher cooling rate than for thermal history TH0. After holding for 2 min at 600°C, the specimen was reheated to 1100°C, cooled down again to the test temperature, and tensile deformed.

To observe the microstructure of the fractured specimens, longitudinally sectioned specimens were etched with 3% Nital and modified picric acid to reveal the prior austenite grain boundary. The ferrite volume fraction and prior austenite grain size were measured by point counting and the linear intercept method, respectively. To calculate Ae3 temperature, Andrew’s formula was used, and Ar3 temperatures of two different thermal histories were measured by dilatation test. Ae3 and Ar3 temperatures are shown in Table 2.

3. Results and Discussion

3.1. Hot Ductility Behavior

The hot ductility curves of specimens with two different thermal histories are shown in Figure 2. Both specimens show the typical shape of RA (reduction of area) curves. Thermal history TH5 showed higher RA values than the RA values of thermal history TH0. The RA value at the ductility trough was more than 70% in thermal history TH5, about 20% higher than for TH0. The trough region of thermal history TH5 ranged from 750 to 850°C. In the case of thermal history TH0, ductility starts to fall from a higher temperature, 950°C. Thermal history TH0 showed similar hot ductility behavior, having a minimum RA value of 50% at 800°C, and a trough temperature region from 950 to 700°C.

The calculated Ae3 temperature is 792°C for the steel used in present work, which is lower than the temperature representing ductility trough (800°C). And Ar3 temperatures of thermal cycle TH0 and TH5, measured by dilatation test, are 610°C and 749°C, respectively (Table 2). Yuan et al. reported that Ar3 temperature is varied with austenite grain size and cooling rate from solution temperature to test temperature.10 In the present work, the difference in Ar3 temperature between two thermal cycles is also attributed to the difference in cooling rate of thermal history and grain size. The development of ductility trough is likely to be attributable to the formation of deformation induced ferrite. And ductility recovery is related to the formation of transformed ferrite and deformation induced ferrite in the lower temperature region (below 800°C). In the higher temperature region (over 850°C), however, the ductility recovery is not related to the formation of ferrite.

As we shall see later, it seems reasonable to expect that hot ductility would increase with a decrease in grain size resulting from the double phase transformation induced by thermal history TH5. The specimen treated with thermal cycle TH5 experienced a double phase transformation that occurred both by intensive cooling from the solution treatment temperature to below the Ar3 temperature and by reheating to the austenite temperature region.

3.2. Metallographic Observation

3.2.1. Microstructures at Ductility Trough

Figure 3 shows the microstructures obtained from specimens after a tensile test at 800°C, at which the minimum RA values were obtained, as shown in Fig. 2. Both the microstructures consist of deformed austenite and very fine ferrite formed around an austenite grain boundary. Specimen TH5 shows finer austenite grain size than specimen TH0, which is caused by the double phase transformation during thermal history TH5. On the other hand, the ferrite existing as fine globular particles around austenite grains increased in amount in specimen TH5. It is likely that the fine austenite grain size is responsible for the increased amount of ferrite, clearly encountered as a higher RA val-

Table 2. Ae3 and Ar3 temperature measured with two different thermal histories (unit: °C).

| Thermal cycle | Ae3 | Ar3 |
|---------------|-----|-----|
| TH0           | 792 | 610 |
| TH5           | 792 | 749 |

Fig. 1. Thermal cycles applied to the specimens (strain rate: 5×10⁻⁴/s); (a) TH0, (b) TH5.

Fig. 2. Hot ductility values of specimens with thermal histories.
ues than specimen TH0, by leading to easy nucleation of deformation induced ferrite, which then lessens the difference in strength between austenite and ferrite.11) The difference in volume fraction of deformation induced ferrite between specimen TH0 and TH5, measured by point counting method, is shown in Table 3. Same period of time for deformation was maintained isothermally at 800°C before quenching for the measurement of ferrite volume fraction in non-deformed specimen. The volume fractions of ferrite in non-deformed specimen with thermal cycle TH0 and TH5 mark 0.9 and 0.8%, respectively, which shows little amount and little differences in ferrite volume fraction. On the other hand, increased amount of deformation induced ferrite is obtained in both deformed specimens TH0 and TH5 in which the specimen TH5 possesses increased amount of ferrite due to the previously mentioned reason. Therefore, the ductility values are likely to be influenced both by deformation induced ferrite and austenite grain size in this temperature.

It should be emphasized that the cracks formed in specimen TH0 were larger than in specimen TH5, which means that the strain was concentrated at the large austenite grain boundary, resulting in low hot ductility.

3.2.2. Illustration of Double Phase Transformation
In hot tensile tests (TH0), the specimen usually experiences an austenite to ferrite single phase transformation, in which the austenite to ferrite phase transformation occurs once during cooling from the solution temperature, 1300°C. Figure 4 shows a schematic illustration of the single phase transformation that occurred in thermal history TH0.

First of all, full austenite grains appeared at the solution treatment temperature (Fig. 4(a)). After solution treatment, several kinds of precipitation, including alloy carbides, starts to develop during cooling to the Ar3 temperature (Fig. 4(b)), with more precipitates at the grain boundary than in the grain interior. When the temperature of the specimen decreases below the Ar3 temperature (Fig. 4(c)), the size of precipitates formed at an early stage (\(\alpha\)) increases, and fine precipitates (\(\beta\)) are also developed by the solubility decrease during cooling below the Ar3 temperature. A thin ferrite film appeared at the prior austenite grain boundary and also starts to develop allotriomorphically below the Ar3 temperature. As the temperature decreases far below the Ar3 temperature, the amount of transformed ferrite increases, thereby the prior austenite grain boundary is covered with transformed ferrite (Fig. 4(d)).

Figure 5 shows a schematic illustration of the double phase transformation occurring in thermal history TH5. Figure 5(a) represents the same state as Fig. 4(a), which shows the microstructures developed at the solution treatment temperature. A small amount of precipitate developed, compared to Fig. 4(b), during cooling from the solution treatment temperature to the Ar3 temperature (Fig. 5(b)). The distinction between these two conditions is attributable to the difference in cooling rate, which was much higher in thermal history TH5 than in thermal history TH0. While the temperature of the specimen is far below the Ar3 temperature, 600°C, the austenite transforms to ferrite and the amount of precipitate is also increased by a solubility decrease during cooling (Fig. 5(c)). It should be noted that the size of the precipitates is reduced compared to thermal history TH0, resulting from the intensive cooling in thermal history TH5.

As mentioned previously, in thermal history TH5, the temperature of the specimen is reheated to far above the Ae3 temperature, 1 100°C (Fig. 5(d)). The austenite trans-

| Thermal cycle | TH0 | TH5 |
|---------------|-----|-----|
| deformed      | 15  | 22  |
| non-deformed  | 0.9 | 0.8 |

Fig. 3. Microstructures of the near fractured surface of (a) TH0, (b) TH5 at 800°C.

Fig. 4. Schematic illustration of the evolution of the microstructure during a single phase transformation; (a) solution treatment at 1 300°C, (b) above Ar3, (c) below Ar3, (d) far below Ar3.

Table 3. Volume fraction of ferrite near the fractured surface with two different thermal histories at 800°C (unit: %).

(c) = austenite, (b) = prior austenite grain boundary, (a) = precipitates, (c) = precipitates formed above Ar3, (d) = precipitates formed below Ar3, (f) = ferrite.
formed during reheating shows significantly refined microstructures, which result both from double phase transformation and from fine precipitates that act as effective nucleation sites for austenite. When the temperature of the specimen decreased again just below the Ar$_3$ temperature, the fine austenite grain boundary is covered with thin film ferrite, and fine ferrite is also developed at precipitates existing in the austenite grain interior, resulting in a more refined microstructure (Fig. 5(e)). Finally, when the temperature decreases to far below the Ar$_3$ temperature, a very fine microstructure is developed by thermal history TH5, namely a double phase transformation (Fig. 5(f)).

Figure 6 represents the microstructures of quenched specimens at 800°C with the two types of thermal history. The measured grain size is listed in Table 4. The original austenite grain developed at the solution treatment temperature measured 100 $\mu$m in size. The final grain size, however, for thermal history TH5 measured 36 $\mu$m, compared to 68 $\mu$m in TH0. These microstructures are obtained at 800°C, which means the trough temperature of the RA value.

Table 4. Austenite grain size at 800°C without deformation with thermal histories (unit: $\mu$m).

| Original(1300°C) | TH0 | TH5 |
|------------------|-----|-----|
| 100              | 68  | 36  |
3.3. Ductility Recovery

Figure 7 shows the microstructure of the specimen after a tensile test at lower temperature, namely the ductility recovery region, at 700°C. These microstructures consist of white ferrite and austenite that is revealed as a relatively dark phase, with thermal history TH5 being finer than the other. Although both specimens show similar RA values in the lower temperature region (Fig. 2), thermal history TH5 caused the RA value to recover fully at higher temperature, 750°C, resulting from refined austenite grain size, thereby increasing both strain induced ferrite and transformed ferrite at higher temperature, although it is difficult to distinguish in this microstructure (Fig. 7).

The hot ductility is recovered at higher temperature both by destruction of grain boundary ferrite and by dynamic recrystallization of austenite.\textsuperscript{1,12} Figure 8 shows the microstructures of a specimen after a tensile test at higher temperature. A fully recrystallized structure was revealed for specimen TH5 tested at 850°C (Fig. 8(b)), compared with specimen TH0, which was partially recrystallized at the same temperature, 850°C (Fig. 8(a)). The deflection changes of the stress–strain curves, which indicate the occurrence of dynamic recrystallization during the tensile test, are difficult to follow in Figure 9(a). It should be noted that in Fig. 9(a), the presence of load fluctuations on the tensile stress–strain curves is less sensitive and, therefore, the microstructural evidence should also be considered to confirm the occurrence of dynamic recrystallization.\textsuperscript{12,13}

Figure 10 shows the microstructures of specimen ob-
tained from far below the fractured surface at which it experiences less strain than the fractured surface, thereby the dynamic recrystallization is less effective in this region. Measured grain size of specimen TH0 and TH5 is 85 $\mu$m and 33 $\mu$m, respectively, which implies that dynamic recrystallization was hardly occurred by little amount of deformation, comparing to the fractured surface in which the measured grain size of specimen TH0 and TH5 is 20 $\mu$m and 14 $\mu$m, respectively.

The rapid increase in hot ductility due to the occurrence of dynamic recrystallization at lower temperature with thermal history TH5 can be ascribed to the refined austenite which is induced by the double phase transformation.\(^\text{14)}\)

On the other hand, both specimens show a fully recrystallized microstructure at 950°C (Figs. 8(c), 8(d)), and it is clear both in Fig. 9(b), shown as a clear load deflection, and in Fig. 2 as a RA recovery.

5. Conclusion

(1) Thermal history TH5 showed higher RA values. The RA value at the ductility trough was 70% in thermal history TH5, which was about 20% higher than TH0.

(2) Specimen TH5 showed a finer austenite grain size than specimen TH0, which is caused by the double phase transformation during thermal history TH5.

(3) Although both the specimens show similar RA values in the lower temperature region, thermal history TH5 causes the RA value to recover fully at a higher temperature, 750°C, resulting from a refined austenite grain size.

(4) In the higher temperature region, hot ductility is recovered due to the occurrence of dynamic recrystallization, which is caused by the refined austenite that is induced by the double phase transformation.

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