Analysis and control of surface cracks in a B-bearing continuous casting blooms

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Abstract: In this work, normal manufacture of a B-bearing steel was hampered by cracks in steel. In order to control these cracks, the formation mechanism has been examined through a comprehensive analysis of crack morphology, element segregation, high temperature mechanical properties, and precipitates. The high-temperature thermoplastic capabilities of the steel were found to be reduced by boron nitride particles precipitated at grain boundary. This led to the formation of a brittle zone in the straightening zone of the continuous casting process, which in turn caused cracks. Based on the formation mechanism of these cracks, the cracks were successfully controlled by adding an appropriate amount of Ti element to the steel and reducing the charging temperature of the heating furnace.

Keywords: hot ductility, BN precipitates, cracks

1 Introduction

Studies have demonstrated that adding a very little quantity of boron element (0.0005–0.0030%) to steel can increase its mechanical characteristics, strengthening, hardenability, and surface hardness without compromising the steel's toughness and plasticity. As a result, steel that contains boron is frequently utilized in industries including aircraft, automotive, and bridges [1–3]. Many researches studies have been carried out on the effect of boron addition on the steel. Generally, boron not only improves the properties of steel, but also brings challenges to the steelmaking processes. Boron nitride (BN) precipitates generated by boron addition will contribute to the issue of surface fractures and finally be inherited into the steel product due to the faulty smelting and hot rolling process [4,5]. Besides, some researchers believed that the remelting phenomenon caused by boron segregation is the main reason for the quality problems of boron-bearing steel [6]. While some studies revealed that BN co-precipitates with sulfides, and improved the steel’s hot ductility [7,8].

In this study, surface cracks on hot-rolled product impeded regular production. Therefore, the goal of this study is to better understand how cracks form during the continuous casting and hot rolling process, as well as to eliminate the cracks by optimizing the operating parameters.

2 Experimental

The steel is produced via BOF-LF-VD-CC (arc-shaped)-heating furnace-hot rolling process. The chemical composition of the steel is listed in Table 1.

Specimens including cracks which sampled from the radial sections of rolled product were etched with 4% nitric acid ethanol solution for revealing the macro-morphology of cracks (Figure 1). The macro-morphology of cracks under the surface of the continuous casting (CC) round bloom was obtained by sawing the round bloom along the drawing direction with a fire cutter. The bloom was cut into a sample with a thickness of about 30 mm, and was pickled for 20 min at 333–353 K using hydrochloric acid. Thus, the subcutaneous cracks of the round bloom were observed. Thermo-Calc software was used to identify the phases and the precipitates present at a range of temperatures.

The mechanical properties at high temperature were examined via hot tensile tests. Samples for the hot tensile test (10 mm in diameter and 120 mm in gauge length) were
extracted from the round bloom. Tensile tests were conducted using the computerized thermal stress/strain simulator, Gleeble 3800, and the reduction of area (RA) was measured to assess the hot ductility of the steel. As illustrated in Figure 2, the specimens were rapidly heated to 1,473 K at a heating rate of 15 K/s in the first step. In the second stage, the specimens were heated to 1,573 K at a rate of 2 K/s and maintained at this temperature for 1 min before being cooled to the deformation temperature ($T_{\text{DT}}$; ranging from 873 to 1,373 K, total 10 points) at a rate of 6 K/s. The specimens were held at the deformation temperature for 1 min and then strained to failure at a strain rate of $10^{-3}$/s.

Scanning electron microscope (SEM) was performed by Model ZEISS ULTRA 55 to observe the surface morphology of cracks, the fracture surface of the hot tensile tests and the elements distribution of precipitates were identified by energy dispersive spectroscopy (EDS) analysis.

### 3 Results and discussion

#### 3.1 Crack morphology

Figure 3(a) reveals that a portion of the cracks on the rolled product are situated beneath the surface, which needs polishing before these cracks can be detected. Some cracks have extended to the surface of the rolled product, showing as fine surface cracks. The overall extending direction of the cracks is parallel to the rolling direction.

Figure 4 displays a representative microstructure of a surface crack. The direction of crack propagation is observed to be perpendicular to the surface of the rolled product. An appreciable decarburization layer is evident on the surface of the crack, suggesting that the current cracks were generated at a relatively high temperature [9]. Furthermore, the thickness of the decarburization layer is approximately 100 μm and accompanied by coarse ferrite grains, indicating that the cracks have undergone a prolonged thermal history throughout the production.

![Table 1: Chemical compositions of the investigated steel (mass fraction/%)](image)

<table>
<thead>
<tr>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>Mo</th>
<th>Al</th>
<th>B</th>
<th>N</th>
<th>Ti</th>
<th>P</th>
<th>S</th>
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<td>0.52</td>
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<td>1.20</td>
<td>0.40</td>
<td>0.018</td>
<td>0.0010</td>
<td>0.0040</td>
<td>—</td>
<td>0.012</td>
<td>0.001</td>
</tr>
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![Figure 1: Illustration of the sample position in bloom.](image)

![Figure 2: Thermal profile used in hot tensile tests.](image)

![Figure 3: (a) Surface crack and (b) subcutaneous crack morphology of hot-rolled product.](image)
process. The formation of cracks is likely to have occurred in the previous rolling process, i.e., the continuous casting process.

Figure 5 presents a low magnification image of a round bloom after pickling. It is evident that numerous fine cracks are distributed beneath the skin of the bloom, with a maximum length of 4 mm. Closer examination of the crack tip revealed that the crack originated from the grain boundary (Figure 6(a)). It was also noted that the crack deflected at the junction of the three grains, still propagating along the grain boundary (Figure 6(b)). This observation fully confirms that during the production process of the bloom, the strength of its grain boundaries is relatively low, thereby rendering it an essential site for crack initiation and development. It is the subcutaneous cracks of the bloom that cause surface cracks on the subsequent product surfaces.

### 3.2 Hot ductility

The area reduction (RA) is a crucial index for assessing the thermoplasticity of steel, reflecting the capacity of the material to undergo uniform plastic deformation at high temperature. The RA is defined as follows [10,11]:

$$RA = \frac{A_b - A_a}{A_b} \times 100.$$  

where $A_b$ and $A_a$ are, respectively, the original cross-sectional area of the specimen before and after fracture test.

The hot ductility and stress–strain curves of the tested steel are demonstrated in Figure 7. As illustrated in Figure 7(a), at the current test temperature, the steel exhibits two distinct brittleness ranges. The third brittleness range spans 923–1,193 K (within the current test temperature range). With the increase in deformation temperature, the RA value decreases from 73.3% at 873 K to 37.2% at 1,073 K, before gradually rebounding. When the deformation temperature reaches 1,273 K, the RA value peaks at 92.2%. Generally, RA = 60% is considered the threshold between good and poor plasticity [12]. At the current straightening temperature of the CC process (approximately 1,173 K), the steel is in the low section shrinkage region, characterized by poor plasticity, high crack sensitivity, and a high likelihood of crack formation during straightening.
With the temperature rising from 873 to 1,373 K, the tensile strength continuously decreases from 90 to 45 MPa. Furthermore, as shown in Figure 7(b), the maximum stress value on the stress–strain curve decreases gradually with the increase in test temperature. At 1,073 K, when the stress surpasses the yield strength, the stress reaches its maximum value within a small strain range (<0.2%), indicating that the plastic deformation ability of the steel grade is worst at this temperature.

Figure 8 depicts the fracture surface of the investigated steels, tensioned at temperatures ranging from 1,223 to 873 K. Notably, the current temperatures exhibit varying degrees of necking. At a tensile temperature of 1,223 K, the fracture exhibits a distinct necking phenomenon, characterized by numerous deep dimples on the cross-section. It is widely acknowledged that the size and depth of dimples are determined by the material’s relative plasticity during fracture. A larger or deeper dimple size and depth signify higher plasticity [13,14].

At 1,223 K, the dimple size and depth are larger and deeper, indicating better plasticity. As the tensile temperature decreases to 1,173 K, a small amount of quasi-cleavage morphology appears on the fracture surface, besides dimples, suggesting a transition from plastic to brittle deformation. When temperature continues to decrease to 1,073 K, at which point ferrite precipitates in a film-like formation at the grain boundary. Due to the low grain boundary strength and poor plasticity, the tensile fracture assumes a rock-sugar configuration, with a cleavage surface size of approximately 200 μm. This fracture type typifies brittle intergranular fracture. At 973 K, the tensile fracture exhibits distinct cleavage plane morphology, with a significant reduction in the size and number of dimples. When the temperature reaches 873 K, the tensile section demonstrates an irregular surface, with a multitude of intracrystalline dimples of varying sizes. Given the low test temperature, a substantial amount of ferrite and pearlite has been generated, resulting in good overall plasticity.

3.3 Precipitates

Precipitates play a crucial role in the formation of cracks in microalloyed steel. Figure 9 presents the calculated equilibrium precipitation during the solidification process, utilizing the Thermo-Calc software.

The calculations reveal that various complex precipitate forms will be generated during the cooling process of this steel. As illustrated in the figure, MnS begins to precipitate and reaches its maximum amount at approximately 1,323 K, stabilizing thereafter. BN precipitation commences at 1,373 K, while AlN starts to precipitate at 1,283 K, reaching its maximum amount at 1,143 K and remaining stable. BN reaches its maximum precipitation amount at approximately 1,153 K. Subsequently, the B element stabilizes as a more robust M3B2 phase, persisting throughout subsequent cooling processes. It is evident that BN does not stabilize during the molten steel solidification process and is categorized as a transitional phase. The temperature range for BN precipitation existence spans between 1,153 and 1,373 K, with the straightening zone temperature falling within this range.

Owing to the transitional phase of BN precipitates during solidification, they only exist stably near the straightening temperature. To obtain BN precipitates, the sample is heated to its austenitizing temperature (1,173 K) and maintained for 20 min, followed by water quenching. Figure 10 displays an image of BN precipitates obtained after
Figure 8: Tensile fracture morphology of the samples at different temperatures: (a and b) 1,223 K, (c and d) 1,173 K, (e and f) 1,073 K, (g and h) 973 K, and (i and j) 873 K.
As can be discerned from the figure, the surface morphology of BN precipitates is irregular, characterized by sharp corners, and a non-smooth texture. BN precipitation along the grain boundary weakens the strength of the grain boundary, predisposing it to crack formation [15–17].

### 3.4 Micro-segregation of B

The boron content of the steel in this study is 10 ppm, which is relatively low. However, due to the strong grain boundary segregation characteristics of boron, it is prone to forming precipitates at grain boundaries [18,19]. In the actual solidification process, particularly at the end of solidification, segregation leads to an increase in the solute concentration of the liquid phase at the solidification front. When the concentrations of B and N elements exceed the equilibrium value, BN will be generated. In this study, the Scheil Model is employed to evaluate boron segregation.

\[
C_B = C_0(1 - f_s)^k
\]

where \(C_B\) is the solid phase solute concentration, \(C_0\) is the initial solute concentration, \(k\) is the solute equilibrium distribution coefficient (0.001 for boron, 0.48 for nitrogen element [20]), and \(f_s\) is the solidification fraction.

The relationship between the segregation ratio and solidification fraction is presented in Figure 11. The results indicate that the boron element exhibits a significant segregation tendency when the solidification fraction exceeds 0.9. The segregation ratio reaches 9.9 when the solidification fraction is 0.9, and at the end of solidification, the boron content is approximately 99.5 times higher than the initial boron content. This is highly beneficial for the formation of BN precipitates.
3.5 Formation mechanism of cracks

Through the above comprehensive analysis, the formation mechanism of cracks in boron-bearing steel can be elucidated. For the current steel, the B element forms BN precipitates with the N element, which accumulate in significant quantities along the grain boundaries, thereby weakening the ability of the grain boundaries to resist deformation. When the bloom reaches the straightening stage of continuous casting (at which BN precipitation reaches its maximum value exactly), the application of straightening stress results in the generation of small cracks perpendicular to the surface. Some of these cracks are situated below the surface of the bloom, while others have already propagated to the surface of the bloom. Subsequently, during the rolling process, these cracks are stretched and extended, ultimately forming surface cracks and subcutaneous cracks almost parallel to the rolling direction (Figure 12).

3.6 Control of the cracks

The precipitation of compounds, such as BN, within the straightening temperature range, particularly at the grain boundary, has been identified as a crucial factor contributing to the high crack sensitivity of this steel. The longer the precipitation exists, the greater the precipitation amount, and the greater the impact on the high-temperature thermoplasticity of steel. Consequently, the control of BN precipitates is essential for the prevention of cracks.

As illustrated in Figure 13, the addition of 150 ppm Ti to the steel results in the formation of a substantial amount of Ti(C,N). When compared with Figure 9, it becomes evident that due to the high content of Ti in the steel and its strong binding capability with N, the temperature range where BN precipitates is reduced to 1,177–1,273 K, and the maximum precipitation amount also decreases from $17 \times 10^{-4}$ to $6.4 \times 10^{-4}$. Moreover, AlN only precipitates slightly below 923 K. When the Ti content increases to 250 ppm, BN does not form within the entire temperature range, and the AlN formation temperature also decreases to 873 K. This suggests that the addition of a specific amount of Ti to the steel can effectively minimize the precipitation of BN and AlN, thereby curtailing the propensity of casting blooms to develop subcutaneous cracks [21].

The marked improvement in high-temperature plasticity resulting from the addition of Ti can be vividly
observed in Figure 14. After adding 250 ppm Ti into the steel, the thermoplastic properties of the steel are significantly enhanced in the vicinity of the straightening temperature. This leads to an RA value as high as 80%, effectively minimizing the propensity for cracking during the straightening process.

In addition, the presence of this steel as a hypoeutectoid grade leads to the intermittent or filmy distribution of pre-eutectoid ferrite along the original austenite grain boundaries, characterized by a strength of merely one-fourth that of austenite. Consequently, during heating, cracks can easily propagate along these grain boundaries [22]. The surface temperature of the bloom, from the continuous casting process to the heating furnace, was measured to be approximately 0.1 K/s cooling rate. Utilizing the JMatPro software, the continuous cooling transformation curve for this steel can be calculated (Figure 15). Clearly, from the curve, the precipitation temperature of ferrite is at 1,023 K, followed by the eutectoid transformation of austenite to pearlite at 993 K. Finally, when the temperature reaches 953 K, the pearlite transformation is completed.

To preclude the propagation of cracks originating from the grain boundaries during the hot charging process, it is imperative to decrease the charging temperature of the heating furnace below 953 K. This ensures a uniform pearlite microstructure, curtails the propagation of cracks along the grain boundaries, and mitigates the propensity for crack formation in the steel during subsequent heating processes.

When titanium element (250 ppm) is added to the steel and reducing the charging temperature of heating furnace to below 953 K, the quality of the cast bloom and hot rolled products is improved. From the figure, it can be seen that the subcutaneous cracks of the casting bloom have disappeared, and no surface or subcutaneous cracks were observed on the surface of subsequent hot rolled products (Figure 16).
4 Conclusions

The present study delved into the formation mechanism of the cracks by examining the crack morphology, hot ductility, precipitates, and element segregation. The main findings are concluded as follows:

1. The precipitation of BN at grain boundary emerged as the main cause of cracks formation. Owing to the precipitation temperature of BN falling within the straigtening temperature range of the continuous casting process, this steel is susceptible to subcutaneous cracks under the application of straigtening stress. This leads to the formation of surface and subcutaneous cracks during subsequent rolling processes.

2. The addition of an appropriate quantity of Ti element to the steel, coupled with a reduction in the charging temperature of the heating furnace, noticeably enhances the hot ductility and crack resistance of the steel. By implementing these two measures, the crack issue in hot-rolled products has been effectively addressed.

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References