Research Article

Xiong Yang, Wei Yu*, Di Tang, Jiaxin Shi, Yuqian Li, Jia Fan, Donggui Mei, and Qiming Du

Effect of Cooling Rate and Austenite Deformation on Hardness and Microstructure of 960MPa High Strength Steel

<https://doi.org/10.1515/secm-2020-0045>

Received Jul 26, 2020; accepted Oct 22, 2020

Abstract: In order to develop 960MPa grade high strength steel, the effects of cooling rate and austenite deformation on the hardness and the microstructure of high strength steel has been studied by Scanning Electron microscope (SEM), Transmission Electron Microscope (TEM), Gleeble-3500 thermal simulation testing machine and T2500 Vickers hardness tester. The results show that only when the cooling rate was higher than 10°C/s and the final cooling temperature was lower than 250°C , the microstructure mainly consists of martensite, and the strength of high strength steel could be above 960MPa; and austenite deformation could effectively refine the width of martensite lath, thus improving the strength and toughness.

Keywords: high strength steel, martensite, dislocation strengthening, fine grain strengthening

1 Introduction

For reducing self-weight, improving load bearing capacity and reducing manufacturing cost, 960MPa ultra-high strength steel has obtained more and more applications in engineering machinery, coal mining machinery, port machinery and other fields [1].

Most of the 960MPa grade steel has been produced via offline quenching and tempering [2–4]. In order to improve the hardenability, some expensive alloys such as Cr, Mo, Cu and Ni were added [5–8], so the manufacturing costs of the 960MPa grade steel has been increased. In order to reduce the manufacturing cost of 960MPa grade high strength steel, new direct quenching process could be an effective alternative option.

As a part of TMCP, direct quenching (DQ) process refers to the process of quenching plate directly by continuous cooling equipment on the rolling line after rolling. Compared with the traditional quenching process [9], DQ omits reheating process, so the energy and production cost could be saved; moreover, in the DQ process, the temperature of the billet is so high that the solid solution of alloying elements in austenite could be more uniform, so the hardenability has been improved. So the content of expensive alloy elements, carbon equivalent and the cost could be reduced and the welding performance could be improved.

Therefore, in order to develop a 960MPa grade high strength steel, a new composition has been designed and the effect of cooling rate and austenite deformation on the microstructure and hardness of 960MPa grade steel has been studied by gleeble3500 thermal simulator, SEM, TEM and Vickers hardness measurement.

***Corresponding Author: Wei Yu:** Institute of Engineering Technology, University of Science and Technology Beijing, Beijing 100083, China; Email: yuwei9451@126.com

Xiong Yang: Institute of Engineering Technology, University of Science and Technology Beijing, Beijing 100083, China; Hesteel Group Company Limited Handan branch, Technology Center, Handan, Hebei, China, 056015; Email: 99609627@qq.com

Di Tang: Collaborative Innovation Center of Steel Technology, University of Science and Technology Beijing, Beijing 100083, China; Email: tangdi@nercar.ustb.edu.cn

Jiaxin Shi: Institute of Engineering Technology, University of Science and Technology Beijing, Beijing 100083, China; Email: sjx_sjx1992@163.com

Yuqian Li: Hesteel Group Company Limited Handan branch, Technology Center, Handan, Hebei, China, 056015; Email: liyuqian@hbisco.com

Jia Fan: Hesteel Group Company Limited Handan branch, Technology Center, Handan, Hebei, China, 056015; Email: fanjia@hbisco.com

Donggui Mei: Hesteel Group Company Limited Handan branch, Technology Center, Handan, Hebei, China, 056015; Email: meidonggui@hbisco.com

Qiming Du: Hesteel Group Company Limited Handan branch, Technology Center, Handan, Hebei, China, 056015; Email: duqiming@hbisco.com

2 Experimental materials and methods

2.1 Experimental materials

The chemical composition of the 960MPa grade high steel is shown in Table 1. The billet was homogenizing at 1200°C for 1 hour, and then rolled into a 20 mm thick plate. The dumbbell shaped samples used for thermal expansion test were machined from the plate with a size of $\Phi 4\text{mm} \times 12\text{mm}$.

2.2 Experimental methods

The continuous cooling expansion tests were performed by using Gleeble-3500 thermal mechanical simulator in a vacuum state. Two treatments patterns were employed to study the effect of cooling rate and austenite deformation on the microstructure and hardness of the 960MPa steel. Treatment(a) was an undeformed schedule; the specimens were austenitized for 120s at 950°C, and then cooled to room temperature at a constant cooling rate of 1, 3, 5, 8, 10, 15, 20, 30, 35 and 40°C/s, respectively. Treatment(b) was a deformed schedule; the specimens were austenitized for 300s at 1200°C, subsequently deformed by 30% at 1050°C after cooling at 5°C/s from 1200°C, then deformed by 40% at 880°C after cooling at 2°C/s from 1050°C and finally cooled to room temperature at a constant cooling rate of 1, 3, 10, 15 and 20°C/s, respectively.

Different transformation that occur within similar temperatures can often overlap in the dilatometer curve, so microstructural characterization and Vickers hardness measurements could be helpful to better interpret the slope changes in the dilatometry curves and to more accurate transformation.

Samples were prepared by metallographic technique and microstructures were observed by FEI Quanta 600 scanning electron microscopy (SEM) and JSM-2100 transmission electron microscopy (TEM). The SEM were etched by 4 vol.% nital; the TEM samples were cut into thin slices with a thickness of 0.1 mm, after ultrasonic cleaning, the samples were polished to about 50 μm , and then the round plates with diameter of 3 mm were washed and thinned by electrolytic double-jet equipment. The electrolyte composition was 20% perchloric acid alcohol and the voltage was 20V. Vickers hardness were measured on T2500 Vickers hardness tester, the load was 5 kg and the time was 5s. The average value was taken after 3 hardness values were tested.

3 Results and Discussion

3.1 Hardness and microstructure of undeformed specimens

The hardness of undeformed specimen at different cooling rates was shown in Figure 1. The hardness of the sample increased with the increase of cooling rate. The hardness changed obviously from 223HV to 380HV with the cooling rate increased from 1°C/s to 5°C/s. But the hardness increased slowly from 385HV to 410HV when the cooling rate increased from 8°C/s to 20°C/s. The variation of hardness indicated that the microstructure has no obvious change. When the cooling rate continuously increased to 40°C/s, and hardness could increase to 485HV.

The microstructure of undeformed specimen at different cooling rates was shown in Figure 2. The microstructure consists of acicular ferrite and lath bainite when cooled at 1°C/s (as shown in Figure 2a), which makes the hardness only 211HV; the bainite lath has been obviously refined (as shown in Figure 2b) and the content of bainite has also been increased when cooled at 3°C/s, which makes the hardness increased to 353HV. Some martensite lath began to appear when cooled at 5°C/s and 8°C/s, and the hardness increased to 380HV and 385HV respectively.

The microstructure mainly consists of lath martensite when the cooling rate exceeds 10°C/s and the martensite lath would be refined with the increase of cooling rate. But the refinement of martensite lath would not be very obvious when cooled at between 10°C/s and 20°C/s, while the lath would be obviously refined when cooled at 40°C/s.

Figure 1: Hardness of undeformed specimen at different cooling rates

Table 1: Chemical composition of experimental materials (wt%)

C	Si	Mn	Als	Mo	Nb	V	Ti	Cr	B	Bal.
0.149	0.293	1.4	0.0454	0.48	0.026	0.101	0.019	0.46	0.0017	Fe

Figure 2: Microstructure of undeformed samples at different cooling rates

Figure 3: TEM results of undeformed samples

In order to better study the effect of cooling rate on martensite lath, TEM examinations were carried out, and the results were shown in Figure 3. When cooled between 10°C/s and 40°C/s, there were a lot of entangled dislocations distributed within the martensitic lath and the boundary and almost no precipitates were found. Moreover, the width of lath has gradually been refined with the increase of cooling rate. According to the statistics of TEM results, the width of lath was obtained as shown in Table 2. When the cooling rate increased from 10°C/s to 20°C/s, the average width of lath has been refined from 394nm to 300nm, and the average width could be refined to 223nm when cooled at 40°C/s.

Cr is a medium carbide forming element, which can delay the pearlite transformation [10, 11]. Mo element can increase the diffusion activation energy of carbon in austenite and reduce the diffusion coefficient of carbon, it can also effectively promote the formation of acicular ferrite

Table 2: Width of martensite lath

Cooling rate /(°C/s)	10	15	20	40
Width of Lath/nm	259-529	236-427	200-400	173-273
Average Width of Lath/nm	394	340	300	223

and inhibit the transformation of polygonal ferrite and pearlite [12]; Zhang [13] studied that the maximum cooling rate of pearlite transformation of 0.33% Mo and 0.42% Mo steel is 0.5°C/s and 0.14°C/s respectively. Therefore, no pearlite would be transformed in the 0.5%Mo steel when cooled at 0.5°C/s or above. Chen [14] studied that the starting temperatures of ferrite transformation temperature in Mo free, 0.17% Mo and 0.38% Mo steels at cooling rates of 0.5°C/s were 664°C, 630°C and 585°C respectively. Therefore, when Mo content is 0.5% and cooling rate is 1°C/s, the

starting temperature of ferrite transformation should be lower than 585°C. Mo could effectively decrease the ferrite transformation temperature, but has no obvious effect on bainite transformation. When cooled at 1°C/s, bainite and acicular ferrite can be obtained directly in the 0.5%Mo and 0.5%Cr steel and the starting temperature of phase transformation could be 540°C while the hardness was only about 223HV.

When cooled at 3°C/s, the diffusion ability of carbon and alloy elements has been further limited, which promotes the bainite transformation and limits the ferrite transformation; although the transformation microstructure still consists of bainite and ferrite, but the proportion of bainite has been greatly increased, which makes the hardness increased to 353HV; When the cooling rate continued to increase to 5°C/s and 8°C/s, the phase transformation was dominated by bainite and martensite phase transition, which further promoted the hardness to 380HV and 385HV.

The microstructure could be complete martensite when cooled between 10°C/s and 40°C/s. The undercooling degree of transformation increased with the increase of cooling rate, which not only increased the nucleation point of martensite, but also reduced the growth rate of martensite lath. Therefore, the average width of martensitic lath has been refined from 394nm at 10°C/s to 223nm at 40°C/s; moreover, the increase of cooling rate also increases the dislocation density in the martensite; the refinement of lath and the increase of dislocation density were helpful to improve the hardness, The hardness increased from 385HV at 8°C/s to 485HV at 40°C/s.

3.2 Hardness and microstructure of deformed specimens

The hardness of deformed specimen at different cooling rates was shown in Figure 4, which also increased with the increase of cooling rate. The hardness increased obviously from 223HV to 403HV with the cooling rate increased from 1°C/s to 3°C/s, which means that the microstructure have experienced obvious change. But the hardness increased slowly from 412.5HV to 435HV when the cooling rate increased from 10°C/s to 20°C/s, which indicated that the microstructure have almost no obvious change.

The microstructure of the deformed specimen at different cooling rates was shown in Figure 5.

The microstructure consists of acicular ferrite and lath bainite when cooled at 1°C/s, the existence of bainite makes the hardness of 311HV. The microstructure consists of mainly lath bainite and a little martensite when cooled

Figure 4: Hardness variation of deformed at austenite temperature with different cooling rate

at 3°C/s, the bainite lath has obviously refined (Figure 5b). The existence of martensite and the refinement of bainite could be helpful to improve the hardness to 403HV.

The microstructure mainly consists of lath martensite when the cooling rate exceeds 10°C/s and the martensite lath would be refined with the increase of cooling rate. But the refinement of martensite lath was not obvious when cooled at between 10°C/s and 20°C/s, so the hardness of the deformed specimens increased slowly from 412.5HV when cooled at 10°C/s to 435HV when cooled at 20°C/s.

TEM examinations were carried out, and the results were shown in Figure 6. When cooled at 1°C/s, the microstructure consists of lath bainite with a width of 255–445nm, and the entangled dislocations were distributed in the lath. When the cooling rate exceeds 10°C/s, the microstructure mainly consists of martensite lath and a large number of entangled dislocations have been distributed inside and at the interface of martensite lath. The width of the lath has been refined with the increase of cooling rate, which is shown in Table 3. The average width of the lath decreased from 282.5nm to 227.5nm when the cooling rate increased from 10°C/s to 20°C/s.

Table 3: Width of martensite lath of the deformed specimens

Cooling speed/(°C/s)	1	10	15	20
Width of Lath/nm	255-445	279-286	225-250	182-273
Average Width of Lath/nm	355	282.5	237.5	227.5

Figure 5: Microstructure of deformation test steel at different cooling rates

Figure 6: TEM photos of deformed samples

3.3 Phase transition temperature and CCT curve

According to the expansion curve of gleeble3500 test steel and the microstructure and hardness value of the sample, the transformation temperature of the experimental steel has been determined, as shown in Table 4 and Table 5 respectively.

For the undeformed specimens, the phase transformation starting temperature of (bainite and ferrite) were 540°C and 539°C when the cooling rate were 1°C/s and 3°C/s respectively; and the phase transformation starting temperature of bainite was 517°C when the cooled at 5°C/s, and the phase transformation starting temperature of martensite decreased from 448°C to 424°C when the cooling rate increased from 8°C/s to 40°C/s, and the phase transformation finish temperature of martensite also decreased from 304°C to 250°C.

For the deformed specimens, the phase transformation starting temperature of (bainite and ferrite) was 559°C when cooled at 1°C/s; and the phase transformation starting temperature of bainite was 553°C when cooled at 3°C/s, and the phase transformation starting temperature of martensite decreased from 443°C to 435°C when the cooling rate increased from 10°C/s to 20°C/s, and the phase transformation finish temperature of martensite also decreased from 259°C to 242°C.

According to the data in Table 4 and Table 5, the static and dynamic CCT curves were drawn respectively, as shown in Figure 7.

The empirical formula Eq. 1 gives the relationship between phase transformation starting temperature of martensite (M_s) and alloy elements [15].

$$M_s = 561 - 474C - 33Mn - 17Cr - 17Ni - 21Mo \quad (1)$$

Table 4: Transformation temperature, microstructure and hardness of the undeformed specimens at different cooling rate

Cooling rate/°C/s	1	3	5	8	10	15	20	30	35	40
(B+F)s/°C	540	539	/	/	/	/	/	/	/	/
(B+F)f/°C	450	456	/	/	/	/	/	/	/	/
B_s /°C			517	/	/	/	/	/	/	/
B_f /°C			450	/	/	/	/	/	/	/
M_s /°C	/	/	450	448	441	433	428	427	426	424
M_f /°C	/	/	341	304	300	295	262	257	250	250
Microstructure	B+F	B+F	B+M	M	M	M	M	M	M	M
Hardness/HV	223	353	383	395	403	415	427	450	480	485

Table 5: Transformation temperature, microstructure and hardness of the deformed specimens at different cooling rate

Cooling rate/ $^{\circ}\text{C/s}$	1	3	10	15	20
(B+F) _s / $^{\circ}\text{C}$	559	/	/	/	/
(B+F) _f / $^{\circ}\text{C}$	460	/	/	/	/
B _s / $^{\circ}\text{C}$	/	553	/	/	/
B _f / $^{\circ}\text{C}$	/	443	/	/	/
M _s / $^{\circ}\text{C}$	/	443	443	439	435
M _f / $^{\circ}\text{C}$	/	300	259	249	242
Microstructure	B+F	B+M	M	M	M
Hardness/HV	311	403	412.5	433.5	435

Figure 7: CCT curve of high strength steel (a) Static CCT curve; (b) Dynamic CCT curve

According to formula Eq. 1, M_s of the experimental steel was about 426.27°C . According to the data in Table 4, when the cooling rate was no less than 20°C/s , the M_s decreased from 428°C to 424°C , which was consistent with the calculated value; when the cooling rate was lower than 20°C/s , the M_s point was slightly higher than the calculated value.

The content of C and alloy elements that dissolved in the prior austenite was the main factors affecting M_s , M_s decreased with the increase of content of C and alloy elements dissolved in the prior austenite. According to the measured and calculated values of M_s , it could be considered that when the cooling rate was no less than 20°C/s , C and alloy elements were almost completely dissolved in austenite, and the fluctuation of M_s with cooling rate was very small, and the relationship between M_s and alloy element content conforms to formula Eq. 1; when the cooling rate was lower than 10°C/s , some alloy elements has been consumed in ferrite and bainite formed first, thus reduc-

ing the content of alloy dissolved in the austenite, so as to improve M_s slightly.

From the data in Table 4 and Table 5, M_s of deformed austenite could be about $2-7^{\circ}\text{C}$ higher than that of undeformed austenite at the same cooling rate. The research [16] shows that martensitic transformation could nucleate at crystal defects, such as dislocations, stacking faults and twins, these crystal defects have structural fluctuations, and martensite nuclei could be formed under the matching of energy fluctuations. Therefore, the microstructure defects caused by austenite deformation could provide more martensite nucleation, thus refining the martensite grains. The energy required for martensitic transformation were not only contributed from undercooling, but also contributed from the applied stress or strain [17]; therefore, after austenite deformation, not only the martensite nucleations has been increased, but also the energy of martensitic transformation has been added, and in this way, the

martensitic transformation has been promoted, and the M_s has been raised.

3.4 Strengthening mechanism of martensite

Dislocation strengthening, fine grain strengthening, solution strengthening and precipitation strengthening are generally considered as the main strengthening mechanisms of martensite steel [18]. No precipitation of the second phase has been found during the continuous cooling transformation of the experimental steel, so the martensite strengthening could be considered to be composed of dislocation strengthening [19], fine grain strengthening and solution strengthening. The reinforcement increment can be calculated by formula [19, 20] Eq. 2–Eq. 4.

$$\Delta\sigma_{dis} = \alpha M G b \rho^{1/2} \quad (2)$$

$$\Delta\sigma_{GB} = K_{HP} \cdot d^{-1/2} \quad (3)$$

$$\alpha_{ss} = 4570w([C]) + 37w([Mn]) \quad (4)$$

$\Delta\sigma_{dis}$ denotes the increment of dislocation strengthening, where α was a constant, about 0.4, M was about 3, G was the shear modulus, with the value of 7.85×10^4 MPa; b was the burgers vector, with the value of 0.248 nm, and ρ was the dislocation density [21].

$\Delta\sigma_{GB}$ represents the fine grain strengthening increment, according to the research results of Kim [22], Shibata [23] and Wang [24], the KHP value was 0.19–0.21 MPa $m^{-1/2}$. Therefore the value of formula Eq. 3 was 0.20 MPa $m^{-1/2}$. Zhao Jie [20] showed that the width of martensitic lath was an effective structural unit for controlling tensile strength. Therefore, the average width of martensitic lath can be used as d value to calculate the increment of fine grain strengthening.

For low-carbon martensitic steels ($w \leq 0.2\%$), most of the carbon segregate in dislocations, which do not dissolve in the solid solution. The strengthening effect of carbon could be mainly reflected by dislocation strengthening. The increment of solution strengthening of carbon

can be regarded as zero [20]. The amount of solid solution strengthening could be ignored.

So the main strengthening mechanisms of martensite steel studied in this paper mainly consist of dislocation strengthening and fine grain strengthening.

The fine grain strengthening increment of martensite lath under different cooling rates was calculated by formula Eq. 3, as shown in Table 6, and converted according to the ratio of tensile strength to hardness of 1/3.

The strength increment contributed by martensitic lath increases from 334.56 MPa at 10°C/s to 444.7 MPa at 40°C/s , and after austenite deformation the strength increment contributed by martensitic lath increased from 395.1 MPa at 10°C/s to 440.28 MPa at 20°C/s .

Since dislocation density has not been measured in this paper, so the dislocation hardness increment and dislocation density could be calculated by formula Eq. 2, according to the hardness value of the experimental steel and the contribution of martensitic lath hardness calculated in Table 6, and the result was shown in Table 7.

The strengthening increment caused by dislocation didn't vary significantly with the cooling rate when the cooling rate was below 20°C/s ; and the dislocation density increased from $12.79 \times 10^{14} \text{ m}^{-2}$ at 10°C/s to $13.13 \times 10^{14} \text{ m}^{-2}$ at 20°C/s , and the dislocation density could significantly increase to $18.70 \times 10^{14} \text{ m}^{-2}$ at 40°C/s . After deformation, the dislocation density increased from $13.00 \times 10^{14} \text{ m}^{-2}$ at 10°C/s to $13.70 \times 10^{14} \text{ m}^{-2}$ at 20°C/s .

According to the data in Table 6 and Table 7, the main strengthening mechanism of martensite was dislocation strengthening (strength contribution ratio $> 66.7\%$), followed by lath refinement strengthening.

When the cooling rate was between 10°C/s and 20°C/s , the cooling rate and austenite deformation mainly affected the strength by refining the width of martensitic lath (the hardness increment caused by lath refinement accounts for more than 64% of the total hardness increase). When the cooling rate increases to 40°C/s , the cooling rate mainly affected the strength by increasing the dislocation density (the hardness increment caused by dislocation

Table 6: Results of strengthening increment and Hardness increment contributed by Lath width

Cooling rate / $^\circ\text{C/s}$	Undeformed specimen			Deformed specimen		
	width of lath nm	increment of strength MPa	increment of hardness HV	width of lath nm	increment of strength MPa	increment of hardness HV
10	394	334.56	111.52	282.5	395.1	131.7
15	340	360.15	120.05	230.5	437.41	145.8
20	300	383.41	127.8	227.5	440.28	146.76
40	223	444.7	148.23	/	/	/

Table 7: Results of strengthening increment and Hardness increment contributed by dislocation density

Cooling rate / °C/s	Undeformed specimen			Deformed specimen		
	Dislocation density/ m^{-2}	strengthening increment /MPa	Hardness increment/HV	Dislocation density/ m^{-2}	strengthening increment /MPa	Hardness increment/HV
10	12.79×10^{14}	835.44	278.48	13.00×10^{14}	842.40	280.80
15	12.92×10^{14}	839.85	279.95	13.65×10^{14}	863.09	287.70
20	13.13×10^{14}	846.59	282.20	13.70×10^{14}	864.72	288.24
40	18.70×10^{14}	1010.3	336.77	/	/	/

strengthening accounts for 83% of the total hardness increase).

The research of literature [25] shows that the boundaries of martensite block and packet were large angle grain boundaries, and the size change of martensite block and packet played a decisive role in toughness and can be regarded as “effective grain size” to control toughness. Austenite deformation can effectively refine the size of martensite block and packet, which could be helpful to further improve the low-temperature toughness of high-strength steel.

Due to the interaction between high-density dislocations, the resistance of dislocation slip has been increased and the number of movable dislocations has been reduced, which increased the tendency of dislocation accumulation and internal stress, so the crack nucleation and propagation has been promoted [26]. The increase of cooling rate, especially when the cooling rate was higher than 20°C/s , increased the dislocation density rapidly, which was not conducive to the low temperature impact toughness of high strength steel.

Therefore, in order to develop 960MPa high strength steel, the final cooling temperature should be lower than 250°C in the quenching process to ensure enough martensite; the cooling rate should be between 10°C/s and 20°C/s to prevent the high cooling rate from increasing dislocation density and deteriorating the toughness of high-strength steel; in order to further improve the low-temperature toughness of high-strength steel, direct quenching process should be considered to refine martensite through austenite deformation.

4 Conclusion

The effects of cooling rate and austenite deformation on the hardness and the microstructure of high strength steel has been studied by SEM, TEM, Gleeble-3500 thermal simu-

lation testing machine and T2500 Vickers hardness tester in this paper, and the conclusions were as follows:

- With the increase of cooling rate, the microstructure has transformed from ferrite+bainite to martensite; when the cooling rate increased from 10°C/s to 40°C/s , the microstructure of high-strength steel mainly consists of martensite, the average width of martensite lath has been refined from 394nm to 223nm, and the hardness increased from 390HV to 485HV.
- Bainite and martensite phase transformation could be effectively promoted by austenite deformation, and the width of martensite could be effectively refined. After austenite deformation, the average width of martensite lath has been refined from 282.5nm to 227.5nm when the cooling rate increased from 10°C/s to 20°C/s , and the hardness increased from 412.5HV to 435HV.
- The main strengthening mechanism of the martensite is dislocation strengthening, and the second strengthening mechanism is refined strengthening of lath. Cooling rate and austenite deformation can increase the hardness of martensite, and the main mechanism of hardness increase is lath refine grain strengthening.
- In order to develop 960MPa high strength steel, the final cooling temperature should be lower than 250°C and the cooling rate of quenching process should be between 10°C/s and 20°C/s to prevent high cooling rate from increasing dislocation density and deteriorating toughness. The direct quenching process should be considered to refine martensite through austenite deformation.

References

- [1] J. Jia, Y. Q. Yang, H. S. Li, X. J. Wu, G. B. Li, F. Yi, J. B. Nie, GMAW welding technology of Q890 ultra-high strength steel, *J. Electric Welding Machine* 45 (2015) 55-58.
- [2] Q. X. Zhang, W. Tang, D. Luo, Y. L. Li, Y. Q. Yang, Experimental Study on Production of High Toughness Q890E Steel for Engineering Machinery, *J. Metal Materials and Metallurgy Engineering* 44 (2016) 15-19.
- [3] J. Kang, F. Lu, C. Wang, Z. D. Wang, G. D. Wang, Quenching and Tempering Technology of Q960 Steel Used for Engineering Machinery, *J. Materials for Mechanical Engineering* 36 (2012) 7-10+57.
- [4] X. Song, G. J. Wang, G. D. Zhang, K. Wang, H. P. Zheng, X. J. Bai, Research and development of heavy gauge quenched and tempered high strength Q890 steel, *J. Transactions of Materials and Heat Treatment* 35 (2014) 147-152.
- [5] G. Mehran, A. Joseph, M. Juho, B. Timo, L. Jari, Finite element simulation of welding distortions in ultra-high strength steel S960 MC including comprehensive thermal and solid-state phase transformation models, *J. Engineering structures* 219 (2020) 110804.
- [6] D. H. Li, D. R. Yu, Z. G. Zhou, C. Z. Luo, J. J. Gu, J. S. Wei, J. B. Wen, Effects of electrode wire on microstructure and hardness of Q890 steel weld, *J. Heat Treatment of Metals* 37 (2012) 96-98.
- [7] B. S. Xie, Q. W. Cai, W. Yu, J. M. Cao, R. Li, Effect of tempering temperature on mechanical properties of Q960 low carbon bainitic steel, *J. Transactions of Materials and Heat Treatment* 35 (2014) 123-130.
- [8] C. Y. Zhang, Effect of tempering on precipitates and strain hardening behavior of Q960 steel, *J. Heat Treatment of Metals* 43 (2018) 186-191.
- [9] O. Chiaki, Development of Steel Plates by Intensive Use of TMCP and Direct Quenching Processes *J. ISIJ international* 41 (2001) 542-553.
- [10] T. Chen, A. M. Zhao, H. L. Fan, X. G. Li, S. YAO, Continuous cooling curve and phase transformation behavior of wear resistant steel 65MnCr, *J. Transactions of Materials and Heat Treatment* 37 (2016) 144-150.
- [11] W. Y. Ding, F. ZHAO, H. Zhang, S. Ye, C. J. Li, Determination and analysis of CCT curves of CLAM steel, *J. Transactions of Materials and Heat Treatment* 38 (2017) 101-106.
- [12] M. C. Zhao, K. Yang, F. R. Xiao, Y. Y. Shan, Continuous cooling transformation of undeformed and deformed low carbon pipeline steels, *J. Materials science & engineering. A, Structural materials properties, microstructure and processing* 355 (2003) 126-136.
- [13] C. L. Zhang, D. Y. Cai, Y. H. Wang, M. Q. Liu, B. Liao, Y. C. Fan, Effects of deformation and Mo, Nb, V, Ti on continuous cooling transformation in Cu-P-Cr-Ni-Mo weathering steels, *J. Materials Characterization* 59 (2008) 1638-1642.
- [14] J. Chen, S. Tang, Z. Y. Liu, G. D. Wang, Influence of molybdenum content on transformation behavior of high performance bridge steel during continuous cooling, *J. Materials in engineering* 49 (2013) 465-470.
- [15] C. F. Huang, Empirical Formulas for Process Planning for Steel Heat Treatment, *J. Aeronautical Manufacturing Technology* (2000) 53-55.
- [16] Z. C. Liu, C. J. Yuan, Y. P. Ji, H. P. Ren, Nucleation and critical nucleus of martensite phase transformation, *J. Heat Treatment of Metals* 35 (2010) 17-22.
- [17] G. Niu, Y. L. Chen, H. B. Wu, L. M. Xu, Phase transformation rule of V-Ti microalloyed corrosion resistant spring steel during continuous cooling, *J. Transactions of Materials and Heat Treatment* 37 (2016) 155-160.
- [18] F. Tadashi, Matrix Structure of Martensite and Bainite in Steels, *J. Heat Treatment* 24 (2009) 16-21.
- [19] Y. L. Zhao, Study on Microstructure Control and Strengthening Mechanism of 1500MPa Direct Quenched Martensite Steel, D, (Kunming university of science and technology), 2010, pp. 185.
- [20] J. Zhao, H. F. Xu, J. Shi, J. Li, J. Pu, W. Q. Cao, Effect of Grain Refinement Strengthening and Dislocation Strengthening on Strength of Medium Manganese Martensitic Steels, *J. Iron and Steel* 47 (2012) 57-61.
- [21] H. H. Liu, Study on microstructure control and mechanical properties of 718H Plastic Mould steel with thick and large cross section, D, (University of Science and Technology of China), 2019, pp. 183.
- [22] B. Kim, E. Boucard, T. Sourmail, D. San Martín, N. Gey, and P.E.J. Rivera-Díaz-Del-Castillo, The influence of silicon in tempered martensite: Understanding the microstructure-properties relationship in 0.5–0.6wt.% C steels, *J. Acta Materialia* 68 (2014) 169-178.
- [23] S. Akinobu, N. Takashi, Masato Sone, M. Shigekazu, H. Yakichi, Evaluation of the block boundary and sub-block boundary strengths of ferrous lath martensite using a micro-bending test, *J. Materials Science & Engineering A* 527 (2010) 7538-7544.
- [24] Y. W. Chen, B. M. Huang, Y. T. Tsai, S. P. Tsai, C. Y. Chen, J. R. Yang, Microstructural evolutions of low carbon Nb/Mo-containing bainitic steels during high-temperature tempering, *J. Materials Characterization* 131 (2017) 298-305.
- [25] X. Y. Wang, T. Pan, H. Wang, H. Su, X. Y. Li, X. Z. Cao, Investigation of the Toughness of Low Carbon Tempered Martensite in the Surface of Ni-Cr-Mo-B Ultra-heavy Plate Steel, *J. Acta Metallurgica Sinica* 48 (2012) 401-406.
- [26] Y. M. Liu, Deformation Dislocation Structure and Strength of Structural Steels, *J. Journal of Iron and Steel Research* (2007) 1-5.