Research Article

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Effects of composition and strain rate on hot ductility of Cr–Mo-alloy steel in the two-phase region

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Abstract: The hot tensile tests were conducted in this study to investigate the effects of Nb, B, Mo, and V on hot ductility of 25CrMo alloy steel in a temperature range of 650–850°C with strain rates of 0.005 and 0.5 s^{−1}. Besides, the influences of ferrite transformation and

precipitates on hot ductility were also investigated by the use of SEM and TEM. Thermo-Calc and J Mat Pro were used for calculating equilibrium precipitates and CCT curves, respectively. The results indicated that the hot ductility is deteriorated with the addition of 0.04% Nb due to Nb(C,N) particles and ferrite transformation. The addition of B inhibits ferrite transformation and improves hot ductility. The hot ductility is improved with increasing strain rate from 0.005 to 0.5 s^{−1} due to the nucleation and growth behavior of ferrite. The fast strain rate promotes nucleation of ferrite; however, the ferrite has no sufficient time to grow up. The addition of Mo inhibits ferrite transformation and improves hot ductility. The addition of 0.12% V has no obvious effect on ferrite transformation. The hot ductility has deteriorated a little with the addition of 0.12% V due to the solution V that increases stress during hot deformation.

Keywords: Cr–Mo-alloy steel, hot ductility, chemical composition, strain rate

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1 Introduction

25CrMo steel is used for the axle of a high-speed train. To refine microstructure for improving strength and toughness, Nb is added to the steel. B is added to improve the hardenability of large diameter axles. V and Mo elements affect hardenability, phase transformation, and carbides precipitation and finally influence the mechanical properties of steel. The previous study results showed that the austenite grains were refined obviously by the addition of Nb while the hardenability was improved significantly with the addition of B [1]. In addition, the strength was improved significantly with increasing V and Mo content in 25CrMo steel [2].

Good hot ductility is very important in the continuous casting stage and hot rolling. The hot ductility of steel is influenced by many factors such as chemical

composition, strain rate, dynamic recrystallization, second particle, microstructure, and thermal history [3–9]. Some researchers have studied the effect of Nb and B on the hot ductility of steel. The hot ductility trough of Nb-containing steel is usually wide since the Nb-containing particles can act as nucleation sites of ferrite and promote ferrite transformation at prior austenite grain boundary [10–13]. Normally, the hot ductility can be improved with the addition of B in Nb-containing steel. Since B enhances grain boundary cohesion and hinders grain boundary sliding through non-equilibrium segregation [14–16].

The research about the effect of V and Mo on the hot ductility of steel is less. V and Mo affect phase transformation and carbide precipitation, which will influence hot ductility [17–19]. V always deteriorates hot ductility due to the precipitation of V(C,N) particles [4, 17–19]. Mejia et al. [20] found that the hot ductility of TWIP steel was improved in a temperature range of 800–900°C with the addition of 0.3% Mo. However, they did not explain the mechanism.

The hot ductility is mainly affected by intergranular ferrite and second particles. Normally, the hot ductility can be improved with a fast strain rate [4]. The intergranular ferrite has less time to grow up, and the second particles at grain boundary have no sufficient time to precipitate with a fast strain rate; hence, the hot ductility is much better than that of low strain rate. In addition, the intergranular crack has more time to nucleate and grow up with a low strain rate and deteriorate hot ductility. Increasing strain rate could reduce the time for the propagation of voids and therefore postpone final fracture. The grain boundary sliding also can be inhibited with a fast strain rate.

Studying the effects of Nb, B, Mo, V, and strain rate on hot ductility of Cr–Mo steel in the two-phase region is of great significance for developing new alloy steel 25CrMo. However, the synergistic effects of chemical composition and strain rate on hot ductility of Cr–Mo-alloy

steels in a two-phase region have not been systematically studied. Therefore, the present work aims to study the effects of Nb, B, Mo, V, and strain rate on hot ductility of 25CrMo alloy steel. The effect of chemical composition on precipitates characteristic and phase transformation was also investigated. Besides, the relation among precipitates, phase transformation, and the hot ductility of 25CrMo steel was discussed.

2 Materials and methods

The 50 kg vacuum induction furnace was used to produce the experimental steels, and then, the steel ingots were forged into $\Phi 16$ mm steel bars. The chemical compositions of the tested steels are listed in Table 1. Steel 1# and 2# have different contents of Nb to study the effect of Nb on hot ductility of 25CrMo. Steel 2#, 3#, and 4# have different contents of B to study the effect of B on hot ductility of Nb-bearing steel 25CrMo. Steel 3#, 5#, and 6# have different contents of V to study the effect of V on hot ductility of Nb- and B-bearing steel 25CrMo. Steel 6#, 7#, and 8# have different contents of Mo to study the effect of Mo on the hot ductility of Nb- and B-bearing steel 25CrMo.

A Gleeble 3500 (thermomechanical simulator) that can control the temperature of the specimens and the uniaxial load and strain was used to conduct the hot tensile straining according to the Chinese standard GB/T 228.2-2015. Figure 1 shows the thermal schedule used for the investigation of hot ductility. Specimens were heated to 1,200°C and held for 5 min for structure and composition homogenization and then cooled to various test temperatures in the range of 650–850°C with a cooling rate of 3°C s⁻¹. The specimens were held for 1 min at each temperature and strained to fracture with a strain rate of 0.5 or 0.005 s⁻¹, and then quenched by a

Table 1: The chemical compositions of the experimental steels (wt%)

Steel	C	Si	Mn	Cr	Mo	V	Ni	N	O	Nb	B
1#	0.27	0.32	0.74	1.10	0.27	0.06	0.28	0.0015	0.0015	0	0
2#	0.27	0.33	0.71	1.10	0.26	0.06	0.28	0.0014	0.0015	0.04	0
3#	0.27	0.32	0.73	1.08	0.27	0.06	0.28	0.0020	0.0013	0.04	0.0020
4#	0.27	0.32	0.72	1.05	0.27	0.06	0.29	0.0018	0.0016	0.04	0.0036
5#	0.26	0.33	0.71	1.07	0.27	0	0.30	0.0018	0.0016	0.04	0.0020
6#	0.27	0.32	0.70	1.08	0.25	0.12	0.30	0.0017	0.0015	0.04	0.0021
7#	0.26	0.34	0.71	1.07	0	0.12	0.29	0.0017	0.0015	0.04	0.0018
8#	0.26	0.30	0.71	1.06	0.50	0.12	0.28	0.0020	0.0013	0.04	0.0020

Figure 1: The thermal schedule of hot tensile straining.

high-speed Ar flow to room temperature. The influences of Nb and B on the hot ductility of 25CrMo steel were studied. Hot ductility was evaluated according to the Chinese standard GB/T 228.1-2010. The area reduction of tensile sample is a performance index to measure the plastic deformation ability of materials. The percentage of the difference between the cross-sectional area of the necking part and the original cross-sectional area divided by the quotient of the original cross-sectional area is the reduction of area. The larger the value is, the better the plasticity is.

In the ferrite and austenite two-phase temperature region, the influences of chemical composition including Nb, B, Mo, and V on hot ductility of 25CrMo steel were studied. The hot ductility during the straightening process of continuous casting was simulated with a strain rate of 0.005 s^{-1} and the hot ductility during hot rolling was simulated with a strain rate of 0.5 s^{-1} .

The fracture morphology was observed using field emission scanning electron microscopy (FE-SEM, FEI MLA-250). FE-SEM was also used to investigate the microstructures

near fracture surfaces. Thermodynamic calculation of equilibrium precipitation was carried out using Thermo-Calc. The CCT curves were calculated by J Mat Pro software. The chemical composition of precipitates was analyzed by energy dispersive spectrometer (EDS), and the morphology of the precipitates was observed by the use of transmission electron microscopy (TEM, Tecrai G2 F20 S-TWIN) [1–3]. The carbon extraction replicas used to investigate the precipitates morphology with TEM were prepared as follows: first, the samples were etched with 4 vol pct nital-alcohol solution after mechanical polishing, second, plating a carbon film with 20–30 nm thickness on the surface, third, extracting with 4 vol pct nital-alcohol solution, and finally, taking out the carbon films with copper net and drying them for TEM observation.

3 Results and discussion

3.1 The effect of Nb, B, and strain rate on hot ductility

3.1.1 Hot ductility behavior and fracture morphology

With a strain rate of 0.005 s^{-1} , the hot ductility curves of the tested steels deformed with a strain rate of 0.005 s^{-1} are shown in Figure 2(a). The area reduction of the tested steels is all above 60% in the temperature range of 650–850°C. With increasing temperature from 650 to 850°C, the hot ductility of base steel 1# is the best and the hot ductility of the Nb-containing steel 2# is the worst. The results show that the hot ductility has deteriorated with the addition of 0.04% Nb. With the addition of B in

Figure 2: Hot ductility curves with a strain rate of 0.005 s^{-1} (a) and 0.5 s^{-1} (b) of the test steels.

Nb-containing steel, the hot ductility is improved and the hot ductility trough moves from 750 to 700°C. With the strain rate of 0.005 s^{-1} , there is no obvious difference of hot ductility between 0.0020% B-containing steel 3# and 0.0036% B-containing steel 4#. The hot ductility of Nb–B steels is worse than that of base steel 1#.

The hot ductility curves of the tested steels deformed with a strain rate of 0.5 s^{-1} are shown in Figure 2(b). The area reduction of the tested steels is all above 80% in the temperature range of 650–850°C. There is a trough of hot ductility at 750°C of all tested steels. The hot ductility of B-bearing steels 3# and 4# is much better than that of B-free steel 1# or steel 2# in the temperature range of 650–850°C. The results show that the hot ductility is improved with addition of B. There is no obvious difference in hot ductility between base steel 1# and Nb-containing steel 2# with the strain rate of 0.5 s^{-1} .

Comparing the area reduction of the tested steels with strain rates of 0.005 and 0.5 s^{-1} , it is obvious that the hot ductility is improved with increasing strain rate except for base steel 1#, as shown in Figure 3. The hot ductility troughs of base steel 1#, 3#, and 4# move from 700 to 750°C with increasing strain rate. The hot ductility of base steel 1# with a strain rate of 0.5 s^{-1} in the temperature range of 750–850°C is worse than that of 0.005 s^{-1} , as shown in Figure 3. Because the fast strain rate that promotes ferrite transformation moves to a higher temperature.

The peak stress as a function of temperature is shown in Figure 4. The peak stress reduces with increasing deformation temperature for the tested steels. The peak stress increases about 50 MPa with an increasing strain rate from 0.005 to 0.5 s^{-1} . The peak stress of base steel 1# is the lowest when the strain rate is 0.005 s^{-1} . The

solution strengthening increases with the addition of Nb; hence, the peak stress increases. However, the peak stress of Nb–B-containing steel 4# is the lowest when the strain rate is 0.5 s^{-1} .

The hot ductility difference of the tested steels is more obvious at 750°C with a strain rate of 0.005 s^{-1} . The fracture morphologies are compared, as shown in Figure 5(a)–(h). The fractures of the tested steels are ductile except Nb-containing steel 2#, which area reduction is only 61.5% at 750°C. The macroscopic morphology is a cup-shaped fracture or pure shear fracture, as shown in Figure 5(a)–(d). The micromorphology of a fracture surface of steel 1#, 3#, and 4# is fibrous and consists of many dimples. They are mainly composed of deformed shallow dimples, micro-porous concentrated dimples, dimple bands, and a large number of short and curved tear ridges. Besides, there is a small number of river patterns, which indicates that there is a plastic deformation before the fracture [21–25]. The fracture of steel 2# has mixed-rupture characteristics of transgranular and dimples. The transgranular fracture occurs along prior austenite grain boundaries in some regions. However, the effective cleavage planes in the grain interior are not clear since the true cleavage planes have been replaced by unclear and smaller cleavage planes, as shown in Figure 5(b) and (f).

3.1.2 The effect of ferrite on hot ductility

With the strain rate of 0.005 s^{-1} , the area reductions of the tested specimens deformed at 700 or 750°C are low. Therefore, the microstructure of the vertical sections near fractures at the temperatures of 700 and 750°C is investigated, as shown in Figures 6 and 7. As shown in Figure 2(a), with the strain rate of 0.005 s^{-1} , the hot ductility at 750°C has deteriorated obviously with the addition of 0.04% Nb.

With the strain rate of 0.005 s^{-1} , there is ferrite at prior austenite grain boundaries in all tested steels strained at 700°C, as shown in Figure 6. The size of ferrite in Nb-containing steel 2# is the biggest, and the thickness of intergranular ferrite is about 10 μm in Nb-containing steel 2# and that is about 2 μm in other steels. When the deformation temperature is 750°C, there is no ferrite in base steel 1# and Nb–B-containing steel 4#, as shown in Figure 7. There are some small size intergranular ferrites in Nb-containing steel 2# and Nb–B-containing steel 3#. It is indicated that Nb promotes ferrite transformation at a higher temperature, B inhibits ferrite transformation due to an increase in the hardenability of austenite. With the addition of 0.020% B in Nb-containing steel, the ferrite

Figure 3: Hot ductility curves with a strain rate of 0.5 and 0.005 s^{-1} of the base steel 1#.

Figure 4: Peak stress curves with a strain rate of 0.005 s^{-1} (a) and 0.5 s^{-1} (b) of the test steels.

Figure 5: Fractures surface morphology at 750°C with a strain rate of 0.005 s^{-1} observed by SEM: (a and e) steel 1#; (b and f) steel 2#; (c and g) steel 3#; (d and h) steel 4#.

can't be inhibited completely at 750°C . When the B content increases to 0.035%, there is no obvious ferrite in the sample.

The reason for the bad hot ductility of steel 2# at 750°C is that there is ferrite film at the prior austenite grain boundary, as shown in Figure 7(b). During hot deformation, the stress will concentrate on ferrite film at the grain boundary that promotes cracks initiation and propagation. With decreasing deformation temperature, the size of grain boundary ferrite increases and the intragranular ferrite appears in the matrix. The increment of the amount of ferrite makes the stress distribution more uniform; hence, the hot ductility of Nb-containing steel 2# is recovered at 700°C . With the addition of B in

Nb-containing steel, the ferrite transformation is inhibited due to the segregation of B at the grain boundary.

With a strain rate of 0.5 s^{-1} , there is a trough of hot ductility at 750°C of all tested steels, as shown in Figure 2(b). The microstructure near fracture surface strained at 750°C with a strain rate of 0.5 s^{-1} is observed by SEM. There is deformation-induced ferrite at PAG boundaries in steels 1# and steel 2#, as shown in Figure 8(a) and (b). Since the strain rate is fast and there is not enough time for ferrite to grow up, the ferrite size is smaller than $1 \mu\text{m}$. However, there is no obvious ferrite in B-containing steel 3# and steel 4#, as shown in Figure 8(c) and (d). The main microstructure in steel 4# deformed at 750°C is martensite and a small amount of bainite.

Figure 6: Microstructures near fracture surface at 700°C with a strain rate of 0.005 s^{-1} observed by SEM: (a) steel 1#; (b) steel 2#; (c) steel 3#; (d) steel 4#.

Figure 7: Microstructures near fracture surface at 750°C with a strain rate of 0.005 s^{-1} observed by SEM: (a) steel 1#; (b) steel 2#; (c) steel 3#; (d) steel 4#.

Figure 8: Microstructure near fracture surface strained at 750°C with a strain rate of 0.5 s^{-1} observed by FE-SEM: (a) steel 1#; (b) steel 2#; (c) steel 3#; (d) steel 4#.

With the strain rate of 0.5 s^{-1} , the hot ductility is improved, and the hot ductility troughs of steel 1#, 3#, and 4# move from 700 to 750°C. Normally, the hot ductility can be improved with increasing strain rate [3,19–22]. Though the fast strain rate promotes the nucleation of ferrite at a higher temperature, the intergranular ferrites have less time to grow up with a fast strain rate; hence, the hot ductility is much better than that of a low strain rate. In addition, the intergranular cracks have more time to nucleate and grow up with a low strain rate and deteriorate hot ductility. Increasing strain rate could reduce the time for the propagation of voids and postpone final fracture. The grain boundary sliding also can be inhibited with a fast strain rate. The hot ductility trough of base steel 1# moves from 700 to 750°C with an increasing strain rate from 0.005 to 0.5 s^{-1} . With the strain rate of 0.5 s^{-1} , there is some small size ferrite in steel 1# when the deformation temperature is 750°C, as shown in Figures 7(a) and 8(a). It indicates that a fast strain rate promotes ferrite nucleation at a higher temperature but inhibits ferrite's growth rate. The improvement of hot ductility at 750°C of Nb–B-bearing steel is attributed to the restraint of ferrite transformation by the addition of B. There is no obvious ferrite in Nb–B-bearing steel deformed at 750°C.

3.1.3 The effect of precipitates on hot ductility

The precipitation behavior of the second phase in steel is an important factor affecting hot ductility besides intergranular ferrite. Thermo-Calc software was used to calculate equilibrium precipitates in the tested steels, and the results are shown in Figure 9. There are MnS, V(C,N), MC, M_7C_3 , $M_{23}C_6$, and cementite precipitates in base steel 1#. In addition to the above-mentioned phases, there is also Nb(C,N) phase in Nb-containing steel. As shown in Figure 9(c) and (d), the calculated precipitation temperature of the FCC_A1#1 phase (Nb(C,N)) is much higher than that of BN; therefore, Nb can fix some nitrogen and avoid B from the formation of BN to certain extent. There are M_2B and BN phases in the B-bearing steels. With increasing B content, the quantity of the M_2B phase increases and its precipitation temperature rises. The completely dissolved temperatures of M_2B are 915 and 1,080°C for steels 3# and 4#, respectively.

TEM and EDS are used for observing and analyzing precipitates in Nb–B-containing steel strained at 750°C. The size of most Nb(C,N) particles is smaller than 20 nm, as shown in Figure 10. The size of BN precipitates is 200–500 nm, and they have different shapes, as shown in Figure 11. The BN phases always precipitate on Nb(C,N)

Figure 9: Thermodynamic calculation results of Thermo-Calc: (a) steel 1#; (b) steel 2#; (c) steel 3#; (d) steel 4#.

particles, since the precipitation temperature of Nb(C,N) is a little higher than that of BN according to the Thermo-Calc calculation result in Figure 9(c) and (d). The large size BN particles have an adverse effect on hot ductility. It is easy to cause stress concentration and cracks initiation during deformation. Another reason for the bad hot ductility of Nb-bearing steel is that there are a lot of Nb(C,N) particles. The hot plasticity of alloy steel during high-temperature deformation is affected by the second-phase particles' location and distribution. When the fine

second-phase particles locate on the austenite grain boundary, they can not only significantly reduce the grain boundary cohesion but also effectively pin the austenite grain boundary and prevent its dynamic migration. In addition, during the process of grain boundary sliding with stress at high temperature, the second-phase particles distributed on the grain boundary can be used as the nucleation core of microcracks and promote the generation of microcracks. As the strain continues, the intergranular cracks continue to propagate and polymerize with

Figure 10: Nb(C, N) precipitates morphology (a) and EDS analysis result (b) of steel 4# strained at 750 °C.

Figure 11: BN precipitates morphology (a) and EDS analysis result (b) of steel 4#.

each other, resulting in the ultimate fracture of the specimen.

3.2 The effect of Mo and V on hot ductility

The effect of V and Mo on hot ductility with a strain rate of 0.005 s^{-1} is shown in Figure 12. The area reduction is reduced slightly with increasing V content from 0 to 0.12%, as shown in Figure 12(a). The variation of area reduction is more obvious with increasing Mo content from 0 to 0.5%, as shown in Figure 12(b). The hot ductility has deteriorated a little with the addition of 0.12% V, and it is improved with the addition of 0.5% Mo. The peak stress increases with the addition of V or Mo. The solution strengthening increases with the addition of V or Mo; hence, the peak stress increases, as shown in Figure 13.

The fracture morphologies of steels 7# and 8# at 700°C with a strain rate of 0.005 s^{-1} are compared, as shown in Figure 14. The fracture of steel 8# is more ductile than that of steel 7#. The microstructure near the fracture surface is shown in Figures 15 and 16. There is intergranular ferrite in Mo-free steel 7#; however, the ferrite transformation is slowed down with the addition of 0.5% Mo. The main microstructure is martensite in steel 8#. The addition of 0.12% V has no obvious effect on ferrite transformation at 700°C , as shown in Figure 16. There is intergranular ferrite in both 0% V steel and 0.12% V steel, and the ferrite thickness is about $2 \mu\text{m}$.

To understand the effect of Mo and V on phase transformation behavior, J Mat Pro software is used to calculate the CCT curves. The simulation austenitization temperature is 900°C , and austenite grain size is 11.0 ASTM. Tensile stress or plastic deformation of austenite causes lattice distortion and increases dislocation density, which

Figure 12: Hot ductility curves with a strain rate of 0.005 s^{-1} : (a) steel 5# and 6#; (b) steel 7# and 8#.

Figure 13: Peak stress curves with a strain rate of 0.005 s^{-1} : (a) steel 5# and 6#; (b) steel 7# and 8#.

Figure 14: Fractures surface morphology at 700°C with a strain rate of 0.005 s^{-1} observed by SEM: (a and b) steel 7#-0% Mo; (c and d) steel 8#-0.5% Mo.

Figure 15: Microstructures near fracture surface at 700°C with a strain rate of 0.005 s^{-1} observed by SEM: (a) steel 7#-0% Mo; (b) steel 8#-0.5% Mo.

Figure 16: Microstructures near fracture surface at 700°C with a strain rate of 0.005 s^{-1} observed by SEM: (a) steel 5#-0% V; (b) steel 6#-0.12% V.

Figure 17: CCT curves calculated by J Mat Pro software: (a) steel 5#; (b) steel 6#.

Figure 18: CCT curves calculated by J Mat Pro software: (a) steel 7#; (b) steel 8#.

is beneficial to the diffusion of carbon and iron atoms, promotes the nucleation and growth of ferrite, and accelerates the transformation of ferrite. The CCT curves will move to the left with tensile stress or plastic deformation.

There is no obvious difference in the CCT curve with the addition of 0.12% V, as shown in Figure 17. The microstructure and CCT curves show that the addition of 0.12% V in steel has no obvious effect on phase transformation; hence, the ferrite has no significant influence on hot ductility. The hot ductility has deteriorated a little with the addition of 0.12% V due to the solution V and carbides precipitation, which increase stress during hot deformation.

The CCT curve moves to the right obviously with the addition of 0.5% Mo, as shown in Figure 18. The ferrite transformation is inhibited with the addition of 0.5% Mo which is in accordance with the microstructure observation results in Figure 15. The influence mechanism of Mo on hot ductility is that Mo slows down the ferrite transformation due to increasing the hardenability of austenite, which improves the hot ductility.

4 Conclusion

- (1) When the strain rate is 0.005 s^{-1} , with increasing strain temperatures from 650 to 850°C, the hot ductility of base steel 1# is the best and that of the Nb-containing steel 2# is the worst. The reason for the bad hot ductility of steel 2# at 750°C is that there is ferrite film at the prior austenite grain boundary. Nb promotes ferrite transformation at a higher temperature, but B inhibits ferrite transformation due to increasing hardenability of austenite.
- (2) With the strain rate of 0.5 s^{-1} , the hot ductility is improved, and the hot ductility troughs of steels 1#, 3#, and 4# move from 700 to 750°C. The intergranular ferrites have less time to grow up with a fast strain rate; hence, the hot ductility is much better than that with a low strain rate.
- (3) In Nb–B-containing steel, the size of most Nb(C,N) particles is smaller than 10 nm, and the size of BN precipitates is 200–500 nm. The fine second-phase particles locating on the grain boundary not only reduce the grain boundary cohesion but also effectively pin the austenite grain boundary and prevent its dynamic migration. Therefore, the hot ductility has deteriorated obviously with the addition of 0.04% Nb in 25CrMo steel.
- (4) The addition of Mo inhibits ferrite transformation and improves hot ductility. The addition of 0.12% V has

no obvious effect on ferrite transformation. The hot ductility has deteriorated a little with the addition of 0.12% V due to the solution V which increases stress during hot deformation.

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