

HIGH-PERFORMANCE-UNIFORMITY SUBSEA PIPELINE STEEL AND MANUFACTURING METHOD THEREFOR

Technical field

5 The present invention relates to a field of pipeline steel manufacturing, in particular to a high-performance-uniformity subsea pipeline steel and manufacturing method therefor.

Background Art

10 The laying of subsea pipeline has a long history of development internationally. Since the first subsea pipeline was laid by Brown & Root Marine Engineering Company in the Gulf of Mexico, USA, in 1954, a large number of subsea pipelines of various types and diameters have been successfully laid in offshore areas around the world, including the North Sea, the Gulf of Mexico, the Mediterranean Sea, Australia, and the South China Sea. Especially in recent years, with the gradual depletion of onshore oil and gas resources, the exploitation of oil and gas has
15 extended to areas with harsh geological conditions such as oceans and polar region, leading to a stronger demand for high-performance subsea pipeline steels.

 Subsea pipelines are subjected to significant compression, tensile loads, and even bending deformations during the laying of the subsea pipeline. During service, harsh operating conditions such as deep-water high pressure, surging waves, strong bottom currents, and subsea movements
20 will also cause substantial plastic deformations or even fracture in subsea pipelines, resulting in serious environmental disasters. These complex operating conditions impose stricter requirements on deep-water pipelines compared to onshore pipelines. Furthermore, as water depth increases, the demands on raw materials continue to rise.

 Subsea pipeline construction is characterized by high risk and high investment. Compared to
25 onshore pipeline construction, it has high requirements for construction quality, construction environment, and construction technology, etc. To avoid the risks of difficult repair and environmental pollution caused by the failure of subsea pipelines, stricter technical requirements are often put forward to improve the quality of subsea pipelines, and process management is strengthened to improve the reliability of the product. Therefore, isotropy of the steel plate (more
30 uniform strength/toughness distribution in the length/thickness direction) is required, as well as good welding performance, low-temperature fracture toughness, strain aging performance, etc. Furthermore, as the subsea pipeline is subjected to significant axial forces and displacements during laying and service process, its load-bearing capacity and strain capacity can be ensured by controlling longitudinal performance. Therefore, longitudinal tensile properties, especially the
35 longitudinal low yield ratio is required for the steel plate.

The Chinese patent application CN201210123639.X discloses "A Large Wall Thickness Subsea Pipeline Steel Plate and Production Method therefor". The steel plate consists of the following composition in mass percentage of: C: 0.05~0.07%, Si: 0.15~0.25%, Mn: 1.42~1.48%, P≤0.010%, S≤0.002%, Ni: 0.13~0.18%, Nb: 0.043~0.048%, Al: 0.020~0.040%, Ti: 0.014~0.024%, Mo: 0.13~0.18%, and a balance of Fe and unavoidable impurities. The steel plate has characteristics such as high low-temperature impact toughness, good tear resistance, high strength, moderate yield ratio, and good elongation. However, the pipeline steel in this patent can only meet the low-temperature toughness requirement at -20°C but fail to meet the toughness requirement at extremely low temperatures. And it is difficult to meet the tensile properties requirement after aging, especially a low yield ratio requirement. Most importantly, the achievement of performance uniformity of the pipeline steel has not been considered in this patent.

The Chinese patent application CN201310457780.8 discloses "A Low-cost Thick-gauge Subsea Pipeline Steel Plate and Manufacturing Method therefor", which comprises the following composition in mass percentage of: C: 0.04~0.06%, Si: 0.15~0.20%, Mn: 1.43~1.47%, P≤0.008%, S≤0.003%, Ni: 0.10~0.15%, Nb: 0.030~0.040%, Ti: 0.012~0.023%, Al: 0.015~0.025%, N≤0.006%, Nb+V+Ti≤0.15%, and a balance of iron and unavoidable impurities. The steel plate obtained has a dual-phase microstructure, achieves high strength and a low yield ratio, and it is suitable for the high deformation resistance requirements of subsea pipelines and also has good toughness. However, the pipeline steel in this patent can only meet the low-temperature toughness requirement at -30°C but fail to meet the toughness requirement at extremely low temperatures. It is difficult to meet the tensile properties requirement after aging, especially a low yield ratio requirement. Furthermore, the achievement of performance uniformity of the pipeline steel has not been considered in this patent.

The Chinese patent application CN201210492187.2 discloses "A Thick-walled High DWTT Performance X65-70 Subsea Pipeline Steel and Manufacturing Method Therefor", which comprises a composition in mass percentage of: C 0.03~0.05%, Si≤0.25%, Mn 1.47~1.70%, P≤0.010%, S≤0.001%, Ti 0.006~0.010%, Cr 0.10~0.20%, Cu 0.12~0.20%, Ni 0.36~0.45%, Al 0.025~0.045%, Ca 0.0008~0.0025%, N≤0.0035%, O≤0.0025%, Nb 0.040~0.050%, and a balance of Fe and unavoidable impurities; and $C_{eq}=0.34\sim0.040$, $P_{cm}=0.13\sim0.17$; wherein, $C_{eq}=C+Mn/6+(Cr+Mo+V)/5+(Ni+Cu)/15$; $P_{cm}=C+Si/30+Cu/20+Ni/60+Cr/20+Mo/15+V/10$, with a balance of iron and unavoidable impurities. The steel plate has thick specifications, high strength, excellent low-temperature impact toughness, weldability, and good DWTT performance. However, the pipeline steel disclosed in this patent can only meet a low-temperature toughness requirement at -20°C but fail to meet the requirement for low yield ratio and tensile properties after aging. Furthermore, the achievement of performance uniformity of the pipeline steel has not been

considered in this patent.

The Chinese patent application CN201410515400.6 discloses "A Special Thick Specification X70 Pipeline Steel and Manufacturing Method therefor", which comprises a chemical composition in mass percentage of: C: 0.03~0.06%, Si: 0.1~0.30%, Mn: 1.30~1.60%, P≤0.010%, S≤0.0050%,
5 Nb: 0.030~0.050%, Ti: 0.008~0.020%, Ni: 0.10~0.30%, Cr: 0.10~0.30%, Cu: 0.10~0.30%, and a balance of Fe and unavoidable impurity elements. A pipeline steel with large thickness, high strength within a narrow range, excellent low-temperature toughness, and superior resistance to low-temperature aging and HIC is obtained. However, the pipeline steel disclosed in this patent can only meet the low-temperature toughness requirement at -30°C but fail to meet the toughness
10 requirement at extremely low temperatures. Furthermore, the achievement of performance uniformity of the pipeline steel has not been considered in this patent.

The Chinese patent application CN201510383013.6 discloses "A Production Method for Low Yield Ratio Acid-resistant Subsea Pipeline Steel", which comprises a chemical composition of: C: 0.03~0.05%, Si: 0.10~0.30%, Mn: 1.15~1.35%, P≤0.010%, S≤0.0010%, Al: 0.01~0.06%, Nb:
15 0.04~0.07%, Ti: 0.008~0.020%, Ni: 0.10~0.25%, Cr: 0.10~0.30%, and a balance of Fe and unavoidable impurity elements. This patent adopts advanced smelting and continuous casting technology to control the narrow fluctuation range of carbon equivalent and the quantity, size and morphology of inclusions. It also employs a relaxation phase transformation process to control the structure morphology of the steel plate and adopts a new water-cooling control mode to improve
20 the uniformity of water cooling of the steel plate, effectively ensuring that the steel plate has a low yield ratio and excellent acid resistance. However, the achievement of performance uniformity of the pipeline steel has not been considered in this patent.

The Chinese patent application CN202011039510.1 discloses "A Subsea Pipeline Steel Plate Coil with Excellent Low-temperature CTOD Performance and Production Process therefor",
25 which comprises a chemical composition in mass percentage of: C: 0.030~0.070%, Si: 0.10~0.20%, Mn: 1.10~1.60%, Nb: 0.020~0.050%, V: 0.020~0.040%, Ti: 0.010~0.030%, Mo: 0.10~0.20%, Al≤0.06%, S≤0.0080%, P≤0.018%, and a balance of Fe and unavoidable impurity elements, so that the subsea pipeline steel plate coil has excellent CTOD performance. However, the achievement of performance uniformity of the pipeline steel has not been considered in this
30 patent.

A review of existing patent research reveals that, while the strength and low-temperature toughness are considered in current subsea pipeline steels, there has been little research on how to ensure that the pipeline steel not only have high performance but also have good performance uniformity.

Summary

The objective of the present invention is to provide a high-performance-uniformity subsea pipeline steel and manufacturing method therefor. The pipeline steel has high strength and good low-temperature toughness, and at the same time, the uniformity of performance of the pipeline steel can also be ensured.

In order to achieve the above objective, the technical solution of the present invention is as follows:

a high-performance-uniformity subsea pipeline steel, comprises a chemical composition in weight percentage of: C: 0.02~0.06%, Si: 0.10~0.30%, Mn: 1.20~1.65%, $Cu \leq 0.30\%$, $Ni \leq 0.30\%$, Cr: 0.15~0.30%, $Mo \leq 0.15\%$, Nb: 0.03~0.05%, $V \leq 0.05\%$, Ti: 0.008~0.020%, Ca: 0.0012~0.0030%, Al: 0.015~0.045%, and a balance of Fe and other unavoidable impurities; the impurity elements shall be controlled as follow: $P \leq 0.01\%$, $S \leq 0.001\%$, $B \leq 0.0004\%$, $N \leq 0.005\%$, $O \leq 0.0025\%$, and $H \leq 0.0002\%$; and simultaneously satisfying: $Ti/N \geq 2$;

$(Nb/7.7C + V/4.3C) \geq 0.10$;

$1.5 \leq Ca/S \leq 4$;

$0.30 \leq Ceq \leq 0.39$, $Ceq = C + Mn/6 + (Cr + V + Mo)/5 + (Cu + Ni)/15$;

a welding cold crack sensitivity index P_{cm} is ≤ 0.17 , and $P_{cm} = C + Si/30 + (Mn + Cu + Cr)/20 + Ni/60 + Mo/15 + V/10 + 5B$.

Further, the balance is Fe and other unavoidable impurities.

The microstructure of the subsea pipeline steel described in the present invention is ferrite + bainite, wherein, a volume fraction of the ferrite is 20%~40%.

The subsea pipeline steel described in the present invention has a grain size grade of $\geq$ grade 9.

The transverse tensile properties of the subsea pipeline steel described in the present invention are as follows: yield strength $R_{t0.5}$: 415~555MPa, tensile strength R_m : 515~670MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.88 ; and a difference in strength between the head and tail of the steel plate is ≤ 50 MPa;

longitudinal tensile: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.86 ; and a difference in strength between the head and tail of the steel plate is ≤ 50 MPa;

after a strain aging test with a 3% pre-strain and holding at 250°C for 1 hour, the longitudinal tensile properties of the steel plate are as follows: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.88 .

A Charpy impact energy of the subsea pipeline steel described in the present invention at -45°C is ≥ 250 J, and a difference in Charpy impact energy between the thickness center of the steel

plate and the near-surface of the steel plate is $\leq 50\text{J}$; a SA of DWTT test at -30°C is $\geq 85\%$, and a difference in SA between a single-sided thinned DWTT sample and full-thickness DWTT sample is $\leq 5\%$; a hardness is $\leq 230\text{HV}10$, wherein, a difference between the surface hardness of the steel plate and center hardness of the steel plate is $\leq 25\text{HV}10$, preferably a difference between the surface hardness of the steel plate and center hardness of the steel plate is $\leq 15\text{HV}10$.

The characteristics of the composition design of the subsea pipeline steel of the present invention are as follow:

1. By adopting an alloy composition system of low C and low Mn, and controlling $0.30 \leq \text{Ceq} \leq 0.39$ and $\text{Pcm} \leq 0.17$, good weldability is achieved under the premise of the steel meeting the requirement for strength and toughness. At the same time, central segregation and banded structures are improved, and the isotropy of material performance is guaranteed.

2. The amount of impurity elements P, S, B, N, O, and H is controlled at extremely low levels, so that the number of inclusions is decreased, the degree of segregation is reduced, and the strength and toughness is ensured, and it is conducive to ensure excellent low-temperature toughness of the pipeline steel in particularly.

3. The addition of alloy elements such as Cu, Ni, Cr, and Mo exerts a solid solution strengthening effect, simultaneously affects the phase transformation during a cooling process of the material, and improves the isotropy of the steel plate properties.

4. By adding trace amount of V-Nb-Ti and strictly limiting $\text{Ti/N} \geq 2$ and $(\text{Nb}/7.7\text{C} + \text{V}/4.3\text{C}) \geq 0.10$, micro-alloy carbonitrides are formed, grains are refined, precipitation are strengthened, C and N atoms are fixed, and the effects of strain aging on material performance are reduced.

C: A most basic strengthening element, carbon dissolves in a steel to form interstitial solid solution, and exerts a solid solution strengthening effect. When C forms carbides precipitation with strong carbide-forming elements, it helps to precipitation strengthening. However, with the increase of C content, the ductility, toughness and weldability of steel are decreased significantly, and central segregation in the continuous-casting slab occurs, which is not conducive to the uniformity of the performance of steel plate in the thickness direction. Furthermore, after strain aging of the steel plate, free C atoms may entangle with dislocations, hinder the movement of dislocations, and increase the yield strength, which is not conducive to the maintenance of a low yield ratio. Therefore, the content of C is controlled at $0.02\% \sim 0.06\%$ in the present invention.

Mn: A solid solution strengthening element, Mn is a most important and economical strengthening element in the steel to compensate the loss of strength caused by the decrease of the content of C. Mn is also an element that expands the γ -phase, and Mn may improve the hardenability of steel, reduce the $\gamma \rightarrow \alpha$ phase transformation temperature of steel, help to obtain

fine phase transformation products, and improve the toughness of steel. However, Mn is also an element that is prone to segregate. When the content of Mn is too high, central segregation in the continuous casting slab tends to occur, hard-phase structures are prone to be formed during a cooling process after rolling, a low-temperature toughness of the material is reduced, and it is also not conducive to the uniformity of the performance of steel plate in the thickness direction. Therefore, the content of Mn is controlled at 1.20~1.65% in the present invention.

Si: A solid solution strengthening element, and it is also a deoxidation element in the steel. However, when the content of Si is too high, it may deteriorate the welding performance of steel, especially the toughness of a welding heat-affected region, and it is not conducive to the removal of iron oxide scale during a hot rolling process. Therefore, the content of Si is controlled at 0.10~0.30% in the present invention.

Cu and Ni: Solid solution strengthening elements, and they may improve the toughness of the steel. They also help to improve the hardenability of steel and delaying the transformation of pearlite during a cooling process of the steel plate. However, when the addition of Cu and Ni is too high, it is not conducive to the toughness and welding performance of the steel plate. Therefore, the content of Cu and Ni is each controlled at $\leq 30\%$ in the present invention.

Cr and Mo: Solid solution strengthening elements, and important elements for improving the hardenability of the steel. They promote the formation of acicular ferrite, improve the toughness of steel, and are conducive to the uniformity of microstructure and performance across the full thickness of the steel plate; however, when the content is too high, they may promote the formation of low-temperature phase transformation structures, and are not conducive to the low-temperature toughness. Therefore, the content of Cr is controlled at 0.15~0.30%; and Mo is controlled at $\leq 0.15\%$ in the present invention.

Nb: One of the important elements in low-carbon micro-alloying steel. Solid-soluted Nb precipitates as Nb(C,N) particles via strain-induced precipitation during a hot rolling process, which may delay austenite recrystallization, increase the austenite recrystallization temperature of steel, make a two-stage controlled rolling possible, prevent the formation of mixed grains, help to obtain a uniform and fine structure, and improve the strength and toughness of the material while the isotropy of the steel plate properties is ensured. Furthermore, during a cooling process, solid-soluted Nb precipitates dispersedly as the second-phase particles NbC in the matrix, contributing to precipitation strengthening. However, when the content of Nb is too high, it cannot be fully solid-solubilized, which not only fails to exert its effect but also increases production costs. And it may lead to premature precipitation of NbC at a high temperature, and larger NbC precipitates at central segregation regions tend to be formed, which is not conducive to the toughness. Therefore, the content of Nb is controlled at 0.03~0.05% in the present invention.

V: One of the important micro-alloying elements, it tends to combine with C in the steel to form VC, having a relatively high precipitation strengthening effect and a relatively weak grain refinement effect. However, when the content of V is too high, the precipitation particles will coarsen significantly, and the low-temperature toughness of steel is reduced. In the present invention, V mainly helps to strength precipitation to make up for strength reduction caused by the deficiency of other alloys. Therefore, the content of V is controlled at $\leq 0.05\%$ in the present invention.

In addition, in order to fully utilize the carbon fixation effect of Nb and V, and reduce the impact of free carbon on strain aging properties, it is necessary to control $(Nb/7.7C+V/4.3C) \geq 0.10$. In some embodiments, $(Nb/7.7C+V/4.3C)$ is controlled at 0.10-0.50, such as 0.10-0.47.

Ti: Ti is a strong carbonitride-forming element that may play a role in nitrogen fixation. During a slab continuous casting process, it may form fine and high-temperature stable TiN precipitates, thereby improving the toughness of the material. In addition, the undissolved carbonitrides of Ti may prevent the growth of austenite grains during heating process of the steel, and TiN and TiC precipitated in high-temperature austenite zone during a rough rolling process may effectively inhibit the growth of austenite grains. Furthermore, TiN and TiC particles in steel may significantly prevent grain growth in the heat-affected region during a welding process, thereby improving the welding performance of the steel plate. Therefore, the content of Ti is controlled at 0.008~0.020% in the present invention; meanwhile, to fully utilize Ti's nitrogen fixation effect, it is necessary to control $Ti/N \geq 2$. In some embodiments, Ti/N is controlled at 2.0~5.0.

Ca: Ca is mainly used to modify MnS inclusions, resulting in the morphology of the inclusions spheroidized and uniform distributed, helping to improve low-temperature toughness. However, when the content of Ca is too high, it may tend to form CaO inclusions, which is not conducive to toughness. Therefore, the content of Ca is controlled at 0.0012%~0.0030% in the present invention. Meanwhile, to fully utilize the modifying effect of Ca on sulfides, Ca/S is controlled at $1.5 \leq Ca/S \leq 4$, resulting in elongated MnS inclusions convert into dispersed spherical CaS inclusions, and a low-temperature toughness of the material is improved.

Al: Al is mainly used for deoxidation of steel. Adding an appropriate amount of Al helps to refine grains and improve the strength and toughness of steel. However, with the increase of the content of Al, oxides of Al will be formed in the steel, thereby reducing the toughness of the base material and the welding heat-affected region. Therefore, the content of Al is controlled at 0.015%~0.045% in the present invention.

S, P, B, N, O, H: Unavoidable impurity elements in steel. P is an element which is prone to segregation; and S tends to form sulfides in steel, which both may significantly reduce the low-

temperature toughness of steel. B tends to precipitate at grain boundaries, resulting in a decrease in the plasticity and toughness of the material. When the content of N and H is too high, cracks tend to occur in a continuous-casting slab. When the content of O is too high, various oxides are formed, thereby reducing hot workability, corrosion resistance, and toughness. Therefore, the contents of the impurity elements are controlled as follows: $P \leq 0.01\%$, $S \leq 0.001\%$, $B \leq 0.0005\%$, $N \leq 0.005\%$, $O \leq 0.0025\%$, and $H \leq 0.0002\%$ in the present invention.

Pcm: An indicator of welding cold crack sensitivity. When Pcm exceeds a certain threshold, the hardenability of steel becomes too great, resulting in a significant decrease in the toughness of the welding region. Pcm is controlled at ≤ 0.17 in the present invention. In some embodiments, Pcm is controlled at 0.12~0.17.

By adopting an alloy composition system of low C and low Mn in the present invention, center segregation is reduced under the premise of the steel meeting the requirement for strength. By combining the effects of Cr, Mo, Cu, and Ni on enhancing hardenability and inhibiting ferrite transformation, the ferrite content is controlled at 20~40%, center segregation is reduced further, good uniformity in the thickness direction of the steel plate is ensured, a difference in SA% between a single-sided thinned DWTT sample and a full-thickness DWTT sample is $\leq 5\%$, and a difference between the surface hardness and center hardness of the steel plate is $\leq 25\text{HV}10$. Furthermore, by combining C, Mn, Cu, Ni, Cr, Mo, and V, the Ceq is controlled at $0.30 \leq \text{Ceq} \leq 0.39$, ensuring the strength performance of the steel plate.

Typically, micro-alloying elements such as Nb and V are added in pipeline steel to strengthen grain refinement and precipitation. However, in the present invention, by utilizing the fixation effect of Nb and V on C atoms, $(\text{Nb}/7.7\text{C} + \text{V}/4.3\text{C}) \geq 0.10$ is designed. By maximizing the fixation effect of Nb and V on C atoms, the influence of free carbon on strain aging properties is reduced, and the unconventional performance requirement for tensile properties of the subsea pipeline steel after strain aging is satisfied. After a strain aging test with a 3% pre-strain and holding at 250°C for 1 hour for the subsea pipeline steel described in the present invention, the longitudinal tensile properties of the steel plate are as follows: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.88 .

The content of ferrite is controlled at 20%~40% in the present invention. it is crucial for a suitable proportion of ferrite obtained to meet the performance requirement of a subsea pipeline steel. When the content of ferrite is too low, a low yield ratio requirement can't be satisfied; when the content of ferrite is too high, strength of the steel is insufficient. Therefore, in terms of composition design, by adding the alloy elements such as Cu, Ni, Cr, and Mo, a certain degree of austenite stability is obtained, and the amount of ferrite transformation may be controlled within a desired range.

The manufacturing method of the high-performance-uniformity subsea pipeline steel described in the present invention comprises the following steps:

1) Smelting and continuous casting

5 The above composition is smelt and continuous cast into a slab; during the continuous casting process, a superheat of the tundish is controlled at 15~40°C, and a casting speed of the continuous casting slab is controlled at 0.40~1.10m/min, ensuring that the solidification end of the continuous casting slab is located in the horizontal section of a continuous casting caster; and an electromagnetic stirring process is adopted in the continuous casting mold;

2) Controlled rolling

10 A heating temperature is 1120~1190°C, a start rolling temperature of rough rolling is 1000 - 1100°C, a finish rolling temperature of rough rolling is 930~980°C, a reduction ratio of rough rolling is $\geq 50\%$, and a spread ratio is 1.20-1.80; a start rolling temperature of finish rolling is 790~880°C, a finish rolling temperature of finish rolling is 780-850°C, and a reduction ratio of finish rolling is $\geq 70\%$;

15 3) Controlled cooling

A relaxation time is $\geq 30s$, a start cooling temperature is 730~780°C, a cooling rate is 10~40°C/s, a stop cooling temperature is 350~550°C, and a temperature difference between the upper and lower surfaces of the steel plate after stop cooling is $\leq 30^\circ C$;

4) After stop cooling, the steel plate is air-cooled or stack-cooled to room temperature.

20 Preferably, in step 4), when a stop cooling temperature is $\leq 450^\circ C$, the steel plate is tempered at 450~650°C for 10~30min.

Preferably, in step 1), a fluctuation value of the superheat of adjacent slabs is $\leq 10^\circ C$ and/or a fluctuation of the casting speed of adjacent slabs is $\leq 10\%$.

25 Preferably, in step 1), a dynamic soft reduction process is carried out at the solidification end of the continuous casting slab, with a reduction amount of 2~3mm.

Preferably, in step 3), a temperature difference of water entry between the head and tail of the steel plate is $\leq 30^\circ C$.

30 Preferably, in step 4), after the steel plate is cooled to room temperature, the head and tail portions of the steel plate are cropped, with a cropping amount of $\geq 250mm$ each, such as 250~350mm. In some embodiments, a cropping amount of the head is 250~350mm, and a cropping amount of the tail is 250~300mm.

In the manufacturing method described in the present invention:

35 To reduce center segregation in the continuous-casting slab and improve the low-temperature toughness and isotropy of properties in the steel plate, the following requirements should be satisfied in the continuous-casting process as follow:

(1) a superheat of the tundish is controlled at 15~40°C, and a fluctuation value of the superheat of adjacent slabs is $\leq 10^\circ\text{C}$;

(2) a casting speed of the continuous casting slab is controlled at 0.40~1.10m/min, with a casting speed controlled constant, and a fluctuation of the casting speed of adjacent slabs is $\leq 10\%$, ensuring that the solidification end of the continuous casting slab is located at the horizontal section of a continuous casting caster;

(3) an electromagnetic stirring process is used in the continuous casting mold section ;

(4) a dynamic soft reduction technology is adopted at the solidification end of the continuous casting slab.

A superheat of the tundish and a casting speed of the continuous-casting slab are controlled, and a fluctuation range of the casting speed is strictly limited in the present invention, so that the continuous casting process is stable, and the solidification end of the continuous-casting slab is located in the horizontal section of a continuous casting caster, and the effectiveness of a dynamic soft reduction is maximized, thereby ensuring good control of center segregation in the slab. Meanwhile, electromagnetic stirring is adopted in a continuous casting process, effectively resulting in the uniform distribution of elements, thereby reducing the segregation tendency and ensuring the uniformity of the performance of the pipeline steel.

A heating temperature of the slab is 1120~1190°C. when a heating temperature is too low, the alloy carbides fail to be fully dissolved, resulting in fail to fully play a role of solid solution strengthening and precipitation strengthening, and it is not conducive to the control of center segregation. When a heating temperature is too high, grain coarsening occurs, and it has an adverse effect on the low-temperature toughness of the steel plate.

A start rolling temperature of rough rolling is 1000~1100°C, and a finish rolling temperature of rough rolling is 930~980°C. A reduction ratio of rough rolling is $\geq 50\%$, and a spread ratio is 1.20~1.80. Rough rolling is carried out in a fully recrystallized austenite zone to avoid the occurrence of mixed grains. A sufficient reduction ratio of rough rolling can fully refine the original austenite grains, and is beneficial to the strength and toughness of the steel plate. At the same time, precise control of the spread ratio is conducive to control of plate shape and reduces the performance difference between the transverse and longitudinal directions of the steel plate. In some embodiments, a reduction ratio of rough rolling is 50~80%, such as 50~74%.

A start rolling temperature of finish rolling is 790~880°C, a finish rolling temperature of finish rolling is 780~850°C, and a reduction ratio of finish rolling is $\geq 70\%$. Finish rolling is carried out in a non-recrystallized austenite zone. A sufficient reduction ratio of finish rolling may fully flatten the austenite grains, and provide more nucleation sites for subsequent phase transformation, thereby refining the final structure and improving the strength and toughness of the steel plate. In

some embodiments, a reduction ratio of finish rolling is 70~85%.

A relaxation time is ≥ 30 s, a start cooling temperature is 730~780°C, a cooling rate is 10~40°C/s, and a stop cooling temperature is 350~550°C. An appropriate cooling rate and stop cooling temperature may ensure the formation of a dual-phase structure of ferrite + bainite, guaranteeing both the strength and toughness of the steel plate while achieving a low yield ratio. A certain relaxation time and start cooling temperature may ensure that a certain amount of proeutectoid ferrite is generated in the structure of steel plate, thereby reducing a yield ratio while maintaining strength. A temperature difference of water entry between the head and tail of the steel plate is $\leq 30^\circ\text{C}$, and reducing the temperature difference of water entry between the head and tail of the steel plate is beneficial for ensuring the isotropy of the steel plate properties. A temperature difference between the upper and lower surfaces of the steel plate after cooling is $\leq 30^\circ\text{C}$. Reducing the temperature difference between the upper and lower surfaces of the steel plate after cooling is beneficial for improving the isotropy of properties in the thickness direction. In some embodiments, a relaxation time is 30~60s.

A relatively high finish rolling temperature, a relatively low finish cooling temperature, and a relatively fast cooling rate are beneficial for inhibiting the enrichment of alloy elements towards the Center of thickness of the plate during a phase transformation process, thereby further reducing the impact of center segregation.

After stop cooling, the steel plate is cooled to room temperature. Preferably, it may be cooled to room temperature by stack-cooled.

Preferably, for the steel plate with a stop cooling temperature of $\leq 450^\circ\text{C}$, tempering may be performed at 450~650°C for 10~30 minutes.

After stop cooling, the steel plate may usually be cooled to room temperature by air cooling on a cold bed; or the steel plates after rolling are stacked together and cooled to room temperature.

A cooling rate of the stack-cooled is relatively slow, so that the steel plate may stay at a high temperature for a longer period of time, which is conducive to the precipitation of alloy carbides and fully play a role of precipitation strengthening and carbon fixation. For the steel plate with a lower stop cooling temperature, the precipitation time of alloy carbides is insufficient. By utilizing short-term high-temperature tempering, it gives full play to the role of alloy precipitation strengthening and carbon fixation, and it is beneficial for the decomposition of hard phase MA constituents and improving low-temperature strength and toughness.

A cropping amount of the head and tail of the finished steel plate after rolling is ≥ 250 mm. Due to significant differences in deformation conditions, cooling conditions between the head/tail section and the main body section of the steel plate, there are significant differences in structure and performance between the head/tail section and the main body section of the steel plate. By

ensuring a certain cropping amount of the head and tail, the influence of deformation and cooling independent of positions is eliminated, which is beneficial for achieving the isotropy of the steel plate properties.

Compared with the prior art, the present invention has the following advantages:

5 In terms of component design, by adding a relatively low amount of C and Mn elements, controlling the amount of impurity elements such as P, S, B, N, O, and H at an extremely low levels, and controlling $0.30 \leq C_{eq} \leq 0.39$ and $P_{cm} \leq 0.17$, in combination with the addition of alloy elements such as Cr, Mo, Cu, and Ni in the present invention, central segregation is improved, phase transformation behavior during a cooling process of the material is affected, the stability of
10 austenite is increased, and the ferrite phase transformation process is delayed, and the requirement of good uniformity of the performance in steel plate is satisfied on the basis of ensuring the mechanical performance of the steel plate.

By utilizing the addition of trace amounts of V-Nb-Ti, and strictly limiting $Ti/N \geq 2$ and $(Nb/7.7C + V/4.3C) \geq 0.10$, micro-alloying carbonitrides are formed, grains are refined and
15 precipitation are strengthened, C and N atoms are fixed, and the effects of strain aging on material performance are reduced. After strain aging, the steel still has high performance. After a strain aging test with a 3% pre-strain and holding at 250°C for 1 h, the steel plate has the following longitudinal tensile properties: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.88 .

20 Based on the composition design, a superheat of the tundish, a continuous casting speed, and a fluctuation range are controlled strictly in a continuous casting process, a dynamic soft reduction process is adopted, so that the solidification end of the continuous casting slab is located in the horizontal section of a continuous casting caster, the effect of dynamic soft reduction is maximized, thereby reducing the segregation tendency and ensuring the uniformity of pipeline steel properties.
25 However, there are no specific requirements or low requirements for a superheat of the tundish, a continuous casting speed, and a fluctuation range in the conventional continuous casting process. Parameter settings are primarily determined by matching production rhythms between the previous and subsequent processes. Therefore, the solidification end of the continuous casting slab is often located in the curved section of the continuous casting caster, so that it limits the effectiveness of
30 dynamic soft reduction applied in the horizontal section to improve the quality of the internal slab and reduce central segregation, ultimately failing to meet the requirements for high performance and uniformity.

In addition, by using a two-stage controlled rolling technology and spread ratio control in a recrystallized austenite zone and non-recrystallized austenite zone, and ensuring sufficient
35 compression ratio during the two-stage rolling process, grain refinement is fully achieved, and

mixed crystal structure is avoided to occur. A rolling temperature, a relaxation time, and a start cooling temperature are controlled to control the ferrite transformation process from both thermodynamic and kinetic aspects, thereby forming a dual phase structure of ferrite and bainite. And the proportion of ferrite is accurately controlled at 20-40%, so that the steel plate has excellent mechanical performance such as strength and toughness, and isotropy, and the issue of mismatched between the strength and uniformity of the performance of subsea pipeline steel is solved.

Description of the drawings

Fig.1 shows the microstructure photograph of the steel in Example 1 of the present invention.

Detailed Description

The present invention will be further explained and illustrated below with reference to the drawings.

Table 1 and 2 list the chemical composition of the steel in the Examples of the present invention, with a balance of Fe and unavoidable impurities; Table 3 and 4 list the specific process parameters.

Tensile properties test: Tensile properties tests were conducted on the steel plates obtained from each example of the present invention at room temperature according to ASTM A370-2021. A yield strength, a tensile strength, and a yield ratio of the steel plates of each example were tested. And the longitudinal tensile properties of the steel plates after a strain aging with 3% pre-strain and holding at 250°C for 1 hour were tested. The specific results are shown in Table 5.

Low-temperature toughness test: According to ASTM E23-2018, impact tests were conducted on the test steel plates at -45°C using V-notched samples, and the Charpy impact energy at -45°C for the steel plate of each examples was measured. The specific results are shown in Table 6. According to API RP 5L3-2014, drop weight tear tests (DWTT) were conducted on the test steel plates at -30°C using full-thickness V- notched samples and single-sided thinned thickness V-notched samples respectively, and the SA value of DWTT test at -30°C for the steel plate of each examples was tested. The specific results are shown in Table 7.

Hardness test: According to ASTM E92, the hardness of the steel plate near the upper surface, the hardness of the steel plate at the thickness center, and the hardness of the steel plate near the lower surface were measured under a load of 10kgf. The specific results are shown in Table 8.

The testing method of grain size was in accordance with ASTM E112.

As can be seen from Figure 1, the microstructure of the steel plate obtained by using the composition and process described in the present invention is ferrite + bainite, with a ferrite content of 20~40%.

As shown in Table 5, the transverse tensile properties of the steels of the present invention are as follows: yield strength $R_{t0.5}$: 415~555MPa, tensile strength R_m : 515~670MPa, and a yield ratio $R_{t0.5}/R_m$ of 0.88, and a difference in strength between the head and tail of the steel plate is ≤ 50 MPa.

5 Longitudinal tensile properties: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratios $R_{t0.5}/R_m$ of 0.86, and a difference in strength between the head and tail of the steel plate is ≤ 50 MPa.

10 After a strain aging test with a 3% pre-strain and holding at 250°C for 1 hour, the longitudinal tensile properties of the steel plate are as follows: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of 0.88.

As can be seen from Table 6, the steel plate obtained in the present invention has a Charpy impact energy of ≥ 250 J at -45°C, and a difference in Charpy impact energy between the thickness center of the steel plate and the near-surface of the steel plate is ≤ 50 J.

15 As can be seen from Table 7, the steel plate obtained in the present invention has a SA of $\geq 85\%$ in the DWTT test at -30°C. Specifically, a difference in SA between a single-sided thinned DWTT sample (Test was conducted in accordance with API RP 5L3 with the temperature progressively lowered) and a full-thickness DWTT sample is $\leq 5\%$.

20 As can be seen from Table 8, the steel plate obtained in the present invention has a hardness of ≤ 230 HV10. Specifically, a difference between the near-surface hardness and center hardness of the steel plate is ≤ 25 HV10.

25 In Comparative Example 1, $Ti/N=1.5$ and $Ca/S=1.12$, neither of which meets the requirements of the present invention: $Ti/N \geq 2$ and $1.5 \leq Ca/S \leq 4$. Moreover, the contents of P, S, and N are relatively high, and the casting speed waveform is not within the control range specified in the present invention during process control, which cannot satisfy the condition that the solidification end of the continuous casting slab is located in the horizontal section of a continuous casting caster. Furthermore, a fluctuation value of the superheat, a temperature difference of water entry between the head and tail of the steel plate, and a temperature difference between the upper and lower surfaces of the steel plate after cooling are all not within the control range specified in the present invention. Although the strength of the steel plate obtained meet the requirements, a difference in strength between the head and tail of the steel plate is too large, exceeding 50 MPa, and the yield ratio is relatively high. Furthermore, a difference in Charpy impact energy between the Center of thickness and near-surface of the steel plate is 66, and a difference in SA between the single-sided thinned DWTT sample and the full-thickness DWTT sample is 14, and a difference between the surface hardness and the center hardness of the steel plate is 35. Obviously, 35 the uniformity of the performance of the steel plate obtained in Comparative Example 1 is poor.

In Comparative Example 2, $Nb/7.7C+V/4.3C=0.09$, which fails to meet the requirement of the present invention: $(Nb/7.7C+V/4.3C) \geq 0.10$. Furthermore, the contents of B and O are relatively high, and a heating temperature, a rough rolling temperature of finish rolling, and a relaxation time in process control are not within the range specified in the present invention. As a result, a Charpy impact energy at -45°C is less than 250J, a SA of DWTT test at -30°C is less than 85%, and the low-temperature toughness of the steel plate is poor.

In Comparative Example 3, $Ceq=0.40$, $Pcm=0.18$, and $Nb/7.7C+V/4.3C=0.07$, none of which meets the requirements of the present invention: $0.30 \leq Ceq \leq 0.39$, $Pcm \leq 0.17$, and $(Nb/7.7C+V/4.3C) \geq 0.10$. A difference in SA between the single-sided thinned DWTT sample and the full-thickness DWTT sample of the steel plate obtained is 13, and a difference in Charpy impact energy between the Center of thickness and near-surface of the steel plate is 115J. The uniformity of the performance of the steel plate obtained in Comparative Example 3 is poor.

Table 1 (unit: weight percentage)

	C	Si	Mn	P	S	Cu	Ni	Cr	Mo	V	Nb	Ti	Ca	Alt	B	N	O	H
Ex.1	0.02	0.10	1.45	0.0075	0.0010	0.30	0.30	0.16	0.11	0.020	0.030	0.010	0.0015	0.0450	0.0004	0.0050	0.0025	0.00020
Ex.2	0.03	0.15	1.20	0.0880	0.0008	0.10	0	0.30	0.15	0.030	0.039	0.008	0.0026	0.0280	0.0003	0.0030	0.0021	0.000170
Ex.3	0.05	0.18	1.57	0.0098	0.0009	0	0.08	0.24	0	0	0.040	0.012	0.0014	0.0334	0.0003	0.0038	0.0018	0.00015
Ex.4	0.025	0.22	1.22	0.0076	0.0007	0.25	0.10	0.19	0.04	0.015	0.033	0.018	0.0021	0.0220	0.0004	0.0043	0.0017	0.00017
Ex.5	0.040	0.20	1.38	0.0083	0.0008	0.17	0.26	0.15	0.08	0.040	0.037	0.015	0.0028	0.0190	0.0003	0.0037	0.0023	0.00016
Ex.6	0.031	0.19	1.24	0.0063	0.0006	0.19	0.15	0.26	0	0.037	0.045	0.011	0.0019	0.0330	0.0003	0.0040	0.0019	0.00015
Ex.7	0.035	0.30	1.65	0.0100	0.0008	0.13	0.14	0.18	0.12	0.050	0.035	0.020	0.0030	0.0150	0.0004	0.0040	0.0022	0.00018
Ex.8	0.041	0.23	1.23	0.0070	0.0006	0.20	0.17	0.25	0.06	0	0.041	0.013	0.0016	0.0287	0.0002	0.0039	0.0019	0.00016
Ex.9	0.060	0.27	1.62	0.0084	0.0008	0.05	0.06	0.20	0	0.010	0.050	0.016	0.0012	0.0310	0.0003	0.0033	0.0020	0.00019
Comp.Ex.1	0.04	0.19	1.38	0.0145	0.0025	0.17	0.26	0.15	0.08	0.040	0.038	0.008	0.0028	0.0190	0.0003	0.0055	0.0023	0.00016
Comp.Ex.2	0.05	0.18	1.57	0.0098	0.0009	0.00	0.08	0.24	0.00	0.000	0.033	0.012	0.0014	0.0334	0.0006	0.0038	0.0035	0.00015
Comp.Ex.3	0.07	0.27	1.70	0.0084	0.0008	0.05	0.06	0.20	0	0.010	0.020	0.016	0.0012	0.0310	0.0003	0.0033	0.0020	0.00019

Table 2

	Ceq	Pcm	Ti/N	Nb/7.7C+V/4.3C	Ca/S
Ex.1	0.36	0.13	2.0	0.43	1.50
Ex.2	0.33	0.12	2.7	0.40	3.25
Ex.3	0.36	0.15	3.2	0.10	1.56
Ex.4	0.30	0.12	4.2	0.31	3.00
Ex.5	0.35	0.15	4.1	0.35	3.50
Ex.6	0.32	0.13	2.7	0.47	3.17
Ex.7	0.39	0.15	5.0	0.46	3.75
Ex.8	0.33	0.14	3.3	0.13	2.67
Ex.9	0.38	0.17	4.8	0.14	1.50
Comp.Ex.1	0.35	0.15	<u>1.5</u>	0.36	<u>1.12</u>
Comp.Ex.2	0.36	0.15	3.2	<u>0.09</u>	1.56
Comp.Ex.3	<u>0.40</u>	<u>0.18</u>	4.8	<u>0.07</u>	1.50

Note: $Ceq = C + Mn/6 + (Cr + V + Mo)/5 + (Cu + Ni)/15$; $Pcm = C + Si/30 + (Mn + Cu + Cr)/20 + Ni/60 + Mo/15 + V/10 + 5B$.

Table 3

	A superheat of the tundish °C	A fluctuation value of the superheat °C	A casting speed of the continuous casting slab m/min	A fluctuation of the casting speed %	A heating temperature °C	A start rolling temperature of rough rolling °C	A finish rolling temperature of rough rolling °C	A reduction ratio of finish rolling %	A spread ratio	A start rolling temperature of finish rolling °C	A finish rolling temperature of finish rolling °C	A reduction ratio of finish rolling %
Ex.1	40	5	1.00	9	1130	1070	940	58	1.40	810	810	73
Ex.2	20	6	1.10	8	1170	1010	945	63	1.60	820	820	82
Ex.3	25	5.5	0.55	5	1152	1020	960	65	1.20	790	790	75
Ex.4	30	10	0.40	7	1190	1090	975	60	1.80	870	820	70
Ex.5	15	7	0.70	4	1140	1060	970	55	1.42	830	800	78
Ex.6	23	4.5	1.05	10	1180	1100	980	70	1.50	880	850	79
Ex.7	18	8	0.80	6	1120	1000	930	50	1.30	800	780	80
Ex.8	28	5.7	0.65	5	1160	1040	950	74	1.48	860	840	78
Ex.9	35	9	0.90	5.5	1138	1080	938	69	1.70	850	830	85
Comp.Ex. 1	25	<u>12</u>	0.70	<u>15</u>	1140	1060	970	55	1.42	830	800	78
Comp.Ex. 2	25	5	0.55	5	<u>1200</u>	1040	<u>910</u>	65	1.20	790	790	75
Comp.Ex. 3	35	9	0.90	5	1138	1080	938	69	1.70	850	830	85

Table 4

	A relaxation time s	A start cooling temperature °C	A temperature difference of water entry between the head and tail of the steel plate °C	A cooling rate °C/s	A stop cooling temperature °C	A temperature difference between the upper and lower surfaces of the steel plate after stop cooling °C	A cooling method of the steel plate	A tempering temperature °C	A tempering time min	A cropping amount of the head of the steel plate mm	A cropping amount of the tail of the steel plate mm
Ex.1	50	760	18	30	380	8	stack- cooled	-	-	300	250
Ex.2	55	770	23	18	400	18	stack- cooled	550	20	250	250
Ex.3	35	745	15	20	460	10	stack- cooled	-	-	350	300
Ex.4	60	730	20	25	550	20	stack- cooled	-	-	300	250
Ex.5	45	750	18	15	445	25	air- cooled	650	10	300	250
Ex.6	33	775	28	40	500	12	air- cooled	-	-	350	300
Ex.7	40	780	30	10	350	30	stack- cooled	450	30	250	250
Ex.8	30	765	25	33	450	15	stack- cooled	-	-	350	300
Ex.9	42	740	26	26	420	5	air- cooled	-	-	350	300
Comp.Ex.1	45	750	<u>51</u>	15	450	<u>42</u>	air- cooled	-	-	300	250
Comp.Ex.2	<u>10</u>	760	15	20	460	10	stack- cooled	-	-	<u>100</u>	<u>50</u>
Comp.Ex.3	42	740	26	26	420	5	air- cooled	-	-	350	300

Table 5

	Sample position	Transverse tensile				Longitudinal tensile				Longitudinal tensile after strain aging		
		Yield strength $R_{t0.5}$ MPa	Tensile strength R_m MPa	A difference in strength between the head and tail of the steel plate MPa	Yield ratio $R_{t0.5}/R_m$	Yield strength $R_{t0.5}$ MPa	Tensile strength R_m MPa	A difference in strength between the head and tail of the steel plate MPa	Yield ratio $R_{t0.5}/R_m$	Yield strength $R_{t0.5}$ MPa	Tensile strength R_m MPa	Yield ratio $R_{t0.5}/R_m$
Ex.1	The head of the steel plate	460	560	12/20	0.82	457	558	8/19	0.82	459	566	0.81
	The tail of the steel plate	448	540		0.83	449	539		0.83	466	565	0.82
Ex.2	The head of the steel plate	500	590	50/50	0.85	485	575	5/9	0.84	488	588	0.83
	The tail of the steel plate	450	540		0.83	480	566		0.85	489	577	0.85
Ex.3	The head of the steel plate	513	597	17/17	0.86	486	564	14/10	0.86	495	568	0.87
	The tail of the steel plate	496	580		0.86	472	554		0.85	486	566	0.86
Ex.4	The head of the steel plate	438	530	23/15	0.83	428	518	28/13	0.83	430	520	0.83
	The tail of the steel plate	415	515		0.81	405	505		0.80	405	505	0.80
Ex.5	The head of the steel plate	495	600	28/12	0.83	485	583	50/50	0.83	479	586	0.82
	The tail of the steel plate	467	588		0.79	435	533		0.82	468	568	0.82
Ex.6	The head of the steel plate	498	568	3/4	0.88	488	566	1/2	0.86	499	570	0.88
	The tail of the steel plate	495	572		0.87	489	568		0.86	495	572	0.87

Ex.7	The head of the steel plate	555	670	50/20	0.83	530	650	26/6	0.82	530	650	0.82
	The tail of the steel plate	505	650		0.78	504	644		0.78	515	648	0.79
Ex.8	The head of the steel plate	513	581	13/12	0.88	493	570	8/7	0.86	509	588	0.87
	The tail of the steel plate	500	569		0.88	485	563		0.86	503	577	0.87
Ex.9	The head of the steel plate	535	660	20/50	0.81	528	639	11/6	0.83	525	648	0.81
	The tail of the steel plate	515	610		0.84	517	633		0.82	530	635	0.83
Comp.Ex. 1	The head of the steel plate	542	610	<u>55/52</u>	<u>0.89</u>	521	601	<u>54/50</u>	<u>0.87</u>	<u>544</u>	611	<u>0.89</u>
	The tail of the steel plate	<u>487</u>	558		0.87	467	551		0.85	495	588	0.84
Comp.Ex. 2	The head of the steel plate	559	617	<u>57/16</u>	<u>0.91</u>	523	597	35/25	<u>0.88</u>	<u>565</u>	604	<u>0.94</u>
	The tail of the steel plate	502	633		0.79	488	622		0.78	528	633	0.83
Comp.Ex. 3	The head of the steel plate	596	<u>680</u>	11/31	0.88	<u>578</u>	<u>661</u>	5/16	<u>0.87</u>	<u>595</u>	<u>667</u>	<u>0.89</u>
	The tail of the steel plate	585	649		<u>0.90</u>	<u>573</u>	645		<u>0.89</u>	<u>588</u>	650	<u>0.90</u>

Table 6

	Sample position	CVN1 J	CVN1 J	CVN1 J	CVN-avg J	Difference
Ex.1	Center of thickness	250	268	278	265	50
	Near-surface	327	318	301	315	
Ex.2	Center of thickness	330	345	317	331	19
	Near-surface	359	340	351	350	
Ex.3	Center of thickness	411	414	402	409	43
	Near-surface	465	438	453	452	
Ex.4	Center of thickness	315	312	330	319	30
	Near-surface	346	342	358	349	
Ex.5	Center of thickness	320	343	331	331	18
	Near-surface	342	352	354	349	
Ex.6	Center of thickness	429	424	422	425	13
	Near-surface	453	434	427	438	
Ex.7	Center of thickness	336	314	322	324	19
	Near-surface	354	322	354	343	
Ex.8	Center of thickness	430	427	411	423	18
	Near-surface	457	439	428	441	
Ex.9	Center of thickness	314	307	309	310	36
	Near-surface	358	337	343	346	
Comp.Ex.1	Center of thickness	<u>188</u>	<u>255</u>	<u>223</u>	222	<u>66</u>
	Near-surface	286	279	299	288	
Comp.Ex.2	Center of thickness	<u>195</u>	<u>225</u>	<u>231</u>	217	37
	Near-surface	<u>232</u>	<u>214</u>	<u>241</u>	229	
Comp.Ex.3	Center of thickness	<u>235</u>	<u>247</u>	<u>211</u>	231	<u>115</u>
	Near-surface	325	338	321	346	

Table 7

	Sample type	SA %	SA %	SA-avg %	Difference
Ex.1	A full-thickness sample	89	88	89	1
	A single-sided thinned sample	89	91	90	
Ex.2	A full-thickness sample	85	85	85	5
	A single-sided thinned sample	88	92	90	
Ex.3	A full-thickness sample	90	87	89	2
	A single-sided thinned sample	90	91	91	
Ex.4	A full-thickness sample	86	86	86	4
	A single-sided thinned sample	89	91	90	
Ex.5	A full-thickness sample	90	90	90	4
	A single-sided thinned sample	92	95	94	
Ex.6	A full-thickness sample	95	93	94	2
	A single-sided thinned sample	95	96	96	
Ex.7	A full-thickness sample	86	88	87	5
	A single-sided thinned sample	95	88	92	
Ex.8	A full-thickness sample	92	93	93	4
	A single-sided thinned sample	98	95	97	
Ex.9	A full-thickness sample	87	90	89	2
	A single-sided thinned sample	94	87	91	
Comp.Ex.1	A full-thickness sample	<u>67</u>	<u>81</u>	<u>74</u>	<u>14</u>
	A single-sided thinned sample	88	87	88	
Comp.Ex.2	A full-thickness sample	<u>69</u>	<u>77</u>	<u>73</u>	4
	A single-sided thinned sample	<u>78</u>	<u>76</u>	<u>77</u>	
Comp.Ex.3	A full-thickness sample	<u>68</u>	<u>72</u>	<u>70</u>	<u>13</u>
	A single-sided thinned sample	<u>80</u>	86	<u>83</u>	

Table 8

	Near the upper surface			Center of Thickness			Near the lower surface			Difference	Grain size
	hardness	hardness	hardness	hardness	hardness	hardness	hardness	hardness	hardness	hardness	hardness
Ex.1	210	203	194	211	208	194	200	196	199	4	10
E9x.2	197	195	196	202	203	207	191	199	200	8	9
Ex.3	218	213	214	200	219	212	213	215	218	5	11
Ex.4	203	208	215	210	190	196	202	191	201	5	9
Ex.5	198	198	203	214	214	192	196	203	195	8	10
Ex.6	201	210	203	197	190	207	202	204	203	6	11
Ex.7	205	207	218	217	230	215	215	213	209	9	9
Ex.8	201	205	200	204	201	190	208	207	205	6	10
Ex.9	208	211	202	196	195	207	220	212	212	12	11
Comp.Ex.1	197	201	200	233	238	231	195	205	199	<u>35</u>	<u>9</u>
Comp.Ex.2	223	218	219	205	220	216	218	221	223	7	8
Comp.Ex.3	211	213	215	228	218	215	198	210	205	12	9

What is claimed is:

1. A high-performance-uniformity subsea pipeline steel, wherein chemical components of the high-performance-uniformity subsea pipeline steel, in weight percentage, are: C: 0.02~0.06%, Si: 0.10~0.30%, Mn: 1.20~1.65%, $Cu \leq 0.30\%$, $Ni \leq 0.30\%$, Cr: 0.15~0.30%, $Mo \leq 0.15\%$, Nb: 0.03~0.05%, $V \leq 0.05\%$, Ti: 0.008~0.020%, Ca: 0.0012~0.0030%, Al: 0.015~0.045%, and a balance of Fe and unavoidable impurities, wherein, the impurity elements shall be controlled as follow: $P \leq 0.01\%$, $S \leq 0.001\%$, $B \leq 0.0004\%$, $N \leq 0.005\%$, $O \leq 0.0025\%$, and $H \leq 0.0002\%$; and simultaneously satisfying:

$$Ti/N \geq 2;$$

$$(Nb/7.7C + V/4.3C) \geq 0.10;$$

$$1.5 \leq Ca/S \leq 4;$$

$$0.30 \leq Ceq \leq 0.39, Ceq = C + Mn/6 + (Cr + V + Mo)/5 + (Cu + Ni)/15;$$

$$Pcm \leq 0.17, Pcm = C + Si/30 + (Mn + Cu + Cr)/20 + Ni/60 + Mo/15 + V/10 + 5B.$$

2. The high-performance-uniformity subsea pipeline steel according to claim 1, wherein the balance is Fe and unavoidable impurities.

3. The high-performance-uniformity subsea pipeline steel according to claim 1 or 2, wherein a microstructure of the subsea pipeline steel is ferrite + bainite, wherein a volume fraction of the ferrite is 20-40%.

4. The high-performance-uniformity subsea pipeline steel according to any one of claims 1-3, wherein a grain size grade of the subsea pipeline steel is $\geq$ grade 9.

5. The high-performance-uniformity subsea pipeline steel according to any one of claims 1-4, wherein:

transverse tensile properties of the subsea pipeline steel are as follows: yield strength $R_{t0.5}$: 415~555MPa, tensile strength R_m : 515~670MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.88 ; and a difference in yield strength between the head and tail of the steel plate and a difference in tensile strength between the head and tail of the steel plate are both ≤ 50 MPa;

longitudinal tensile: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.86 ; and a difference in yield strength between the head and tail of the steel plate and a difference in tensile strength between the head and tail of the steel plate are both ≤ 50 MPa;

after a strain aging test with a 3% pre-strain and holding at 250°C for 1 hour, the longitudinal tensile properties of the steel plate are as follows: yield strength $R_{t0.5}$: 405~530MPa, tensile strength R_m : 505~650MPa, and a yield ratio $R_{t0.5}/R_m$ of ≤ 0.88 .

6. The high-performance-uniformity subsea pipeline steel according to any one of claims 1-

5, wherein a Charpy impact energy of the subsea pipeline steel at -45°C is $\geq 250\text{J}$, and a difference in Charpy impact energy between the thickness center of the steel plate and the near-surface of the steel plate is $\leq 50\text{J}$; a SA of the DWTT test at -30°C is $\geq 85\%$, and a difference in SA between a single-sided thinned DWTT sample and a full-thickness DWTT sample is $\leq 5\%$; a hardness is $\leq 230\text{HV}10$, wherein, a difference between the surface hardness of the steel plate and center hardness of the steel plate is $\leq 25\text{HV}10$.

7. A manufacturing method for the high-performance-uniformity subsea pipeline steel according to any one of claims 1-6, comprising the following steps:

1) smelting and continuous casting

the composition according to claim 1 or 2 is smelted and continuously cast into a slab; during the continuous casting process, a superheat of the tundish is controlled at $15\sim 40^{\circ}\text{C}$, and a casting speed of the continuous casting slab is controlled at $0.40\sim 1.10\text{m/min}$, ensuring that the solidification end of the continuous casting slab is located in the horizontal section of a continuous casting caster, and an electromagnetic stirring process is adopted in the continuous casting mold;

2) controlled rolling

a heating temperature is $1120\sim 1190^{\circ}\text{C}$, a start rolling temperature of rough rolling is $1000\sim 1100^{\circ}\text{C}$, a finish rolling temperature of rough rolling is $930\sim 980^{\circ}\text{C}$, a reduction ratio of rough rolling is $\geq 50\%$, and a spread ratio is $1.20\sim 1.80$; a start rolling temperature of finish rolling is $790\sim 880^{\circ}\text{C}$, a finish rolling temperature of finish rolling is $780\sim 850^{\circ}\text{C}$, and a reduction ratio of finish rolling is $\geq 70\%$;

3) controlled cooling

a relaxation time is $\geq 30\text{s}$, a start cooling temperature is $730\sim 780^{\circ}\text{C}$, a cooling rate is $10\sim 40^{\circ}\text{C/s}$, a stop cooling temperature is $350\sim 550^{\circ}\text{C}$, and a temperature difference between the upper and lower surfaces of the steel plate after stop cooling is $\leq 30^{\circ}\text{C}$;

4) after stop cooling, the steel plate is air-cooled or stack-cooled to room temperature.

8. The manufacturing method according to claim 7, wherein in step 1), a fluctuation value of the superheat of adjacent slabs is $\leq 10^{\circ}\text{C}$ and/or a fluctuation of the casting speed of adjacent slabs is $\leq 10\%$.

9. The manufacturing method according to claim 7 or 8, wherein in step 1), a dynamic soft reduction process is adopted at the solidification end of the continuous casting slab, with a reduction amount of $2\sim 3\text{mm}$.

10. The manufacturing method according to claim 7 or 8, wherein in step 1), a reduction ratio of rough rolling is $50\sim 80\%$, and a reduction ratio of finish rolling is $70\sim 85\%$.

11. The manufacturing method according to claim 7, wherein in step 3), a temperature

difference of water entry between the head and tail of the steel plate is $\leq 30^{\circ}\text{C}$.

12. The manufacturing method according to claim 7, wherein in step 3), a relaxation time is 30~60s.

13. The manufacturing method according to claim 7, wherein in step 4), when a stop cooling
5 temperature is $\leq 450^{\circ}\text{C}$, the steel plate is tempered at $450\sim 650^{\circ}\text{C}$ for 10~30min.

14. The manufacturing method according to claim 7 or 13, wherein in step 4), after the steel plate is cooled to room temperature, the head and tail portions of the steel plate are cropped, and a cropping amount of both the head and tail is $\geq 250\text{mm}$.