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(71) Applicant: **Baoshan Iron & Steel Co., Ltd.**
Shanghai 201900 (CN)

(72) Inventors:
- **YAO, Liandeng**
  Shanghai 201900 (CN)
- **ZHAO, Sixin**
  Shanghai 201900 (CN)
- **WANG, Pengjian**
  Shanghai 201900 (CN)
- **MIAO, Yuchuan**
  Shanghai 201900 (CN)

(74) Representative: **Kuhnen & Wacker**
Patent- und Rechtsanwaltsbüro
Prinz-Ludwig-Straße 40A
85354 Freising (DE)

(54) STEEL PLATE WITH YIELD STRENGTH AT 890MPA LEVEL AND LOW WELDING CRACK SENSITIVITY AND MANUFACTURING METHOD THEREFOR

(57) A steel plate with yield strength at an 890MPa level and low welding crack sensitivity and a manufacturing method therefor. Weight percentages of components thereof are: C: 0.06-0.13wt.%, Si: 0.05-0.70wt.%, Mn: 1.20-2.30wt.%, Mo: 0-0.25wt.%, Nb: 0.03-0.11wt.%, Ti: 0.002-0.050wt.%, Al: 0.02-0.15wt.%, and B: 0-0.0020wt.%, where 2Si+3Mn+4Mo≤8.5, and others being Fe and inevitable impurities. The use of controlled thermo-mechanical rolling and cooling technologies is beneficial to improvement of steel plate strength, plasticity and toughness. The yield strength of the steel plate is greater than 890MPa, the tensile strength is greater than 950MPa, the charpy impact work Akv(-20°C) is greater than or equal to 120J, and the welding crack sensitivity indicator Pcm is less than or equal to 0.25%.

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Description

FIELD OF THE INVENTION

[0001] The present invention relates to a steel plate with a high strength and low welding crack sensitivity, and in particular, the present invention relates to a steel plate with a yield strength at an 890 Mpa level and low welding crack sensitivity and a method for manufacturing the same.

BACKGROUND

[0002] The steel for high strength mechanical equipment and engineering construction requires a relatively high strength and an excellent toughness, wherein contribution to the strength from various factors can be expressed by the following formula:

$$\sigma = \sigma_f + \sigma_p + \sigma_{sl} + \sigma_d$$

wherein σ_f is fine grain strengthening, σ_p is precipitation strengthening, σ_{sl} is solid solution strengthening, and σ_d is dislocation strengthening. The thermo-mechanical treatment of the steel plate is usually done by a controlled rolling and controlled cooling process (TMCP). The refinement of the microstructures or the formation of the high strength structures such as ultrafine bainite can be realized by controlling the deformation rate and cooling rate, thus improving the yield strength of the steel.

[0003] Currently, the composition of the low-carbon and high-strength steel produced using TMCP is mainly Mn-Ni-Nb-Mo-Ti and Si-Mn-Cr-Mo-Ni-Cu-Nb-Ti-Al-B systems.

[0004] For example, the chemical composition of a low-alloy and high-strength steel produced by the TMCP process in two temperature stages disclosed in the international publication no. WO 99/05335 is as follows (wt.%): C: 0.05-0.10%, Mn: 1.7-2.1%, Ni: 0.2-1.0%, Mo: 0.25-0.6 Mo%, Nb: 0.01-0.10%, Ti: 0.005-0.03%, P ≤ 0.015%, and S ≤ 0.003%.

[0005] Also for example, the chemical composition of a superlow carbon bainitic steel disclosed in the Chinese patent publication no. 1521285 is as follows (wt.%): C: 0.01-0.05%, Si: 0.05-0.55%, Mn: 1.0-2.2%, Ni: 0.0-1.0%, Mo: 0.0-0.5%, Cr: 0.0-0.7%, Cu: 0.0-1.8%, Nb: 0.015-0.070%, Ti: 0.005-0.03%, B: 0.0005-0.005%, and Al: 0.015-0.07%.

[0006] The alloying element designs of the above two types of the steels disclosed are an Mn-Ni-Nb-Mo-Ti and an Si-Mn-Cr-Mo-Ni-Cu-Nb-Ti-Al-B system respectively; since Mo and Ni are both precious metals, the production costs of such steel plates are relatively high from the analysis of the type and the total amount of the alloying elements added.

SUMMARY OF THE INVENTION

[0007] An object of the present invention is to provide a steel plate with a yield strength at an 890 Mpa level and low welding crack sensitivity and a method for manufacturing the same, using the type of steel of an Si-Mn-Nb-Mo-V-Ti-Al-B system, by the controlled thermo-mechanical rolling and cooling technologies, without tempering, and the steel plate has a welding crack sensitivity index Pcm ≤ 0.25%, a yield strength of greater than 890 MPa, a tensile strength of greater than 950 MPa, a Charpy impact energy Akv (-20°C) ≥ 120 J, a plate thickness of up to 60 mm, has a good low-temperature toughness and weldability, and is a low-carbon superfine bainite lath steel plate with low welding crack sensitivity.

[0008] To achieve the above-mentioned object, the technical solution of the present invention consists in:

A steel plate with a yield strength at an 890 Mpa level and low welding crack sensitivity, wherein the steel plate has the following components in weight percentage: C of 0.06-0.13 wt.%, Si of 0.05-0.70 wt.%, Mn of 1.20-2.30 wt.%, Mo of 0-0.25 wt.%, Nb of 0.03-0.11 wt.%, Ti of 0.002-0.050 wt.%, Al of 0.02-0.15 wt.%, and B of 0-0.0020 wt.%, with 2Si+3Mn+4Mo ≤ 8.5, the balance being Fe and inevitable impurities; and the steel plate meets the welding crack sensitivity index Pcm ≤ 0.25%.

[0009] In the composition design of the present invention:

C: C can enlarge an austenitic area, and carbon in a supersaturated ferrite structure formed in the quenching can increase the strength. C has an adverse impact on welding performance. The higher the content of C, the poorer the welding performance; for a bainitic steel produced using a TMCP process, the lower the content of C, the better the toughness, and a lower carbon content can results in a high toughness steel plate having a greater thickness; therefore, the content of C in the present invention is controlled at 0.06 to 0.13%.

[0010] Si: Si cannot be formed into a carbide in the steel, but exists in the bainite ferrite or austenite in the form of a solid solution. It can improve the strength of the bainite austenite or ferrite in the steel. The solid solution strengthening effect of Si is stronger than that of Mn, Nb, Cr, W, Mo and V. Si can reduce the diffusion rate of carbon in the austenite, and makes the ferrite CCT curve and pearlite C curve shift to the right, thus facilitating the formation of a bainite structure in a continuous cooling process. In the inventive steel, no more than 0.70% of Si is added, which facilitates to improve the matching relationship between the strength and toughness of the steel.

[0011] Mn: Mn and Fe can form a solid solution, which improves the strength and hardness of the bainite ferrite and austenite in the steel. Mn can enlarge the austenitic area in the iron-carbon equilibrium phase diagram, so that the ability of the steel to form a stable austenite structure is second only to that of Ni, which strongly increases the hardenability of the steel. When the Mn content is relatively high, it has the tendency to grain coarsening of the steel. In the present invention, 1.20-2.30% of Mn is added, and the speed of the ferrite and pearlite transform is slowed, which is beneficial for the formation of the refined bainite structure, and impart the steel with a certain strength.

[0012] Mo and Cr: Mo and Cr are ferritizing elements, which reduce the austenitic area. Mo and Cr are in a solid solution in austenite and ferrite to increase the strength, improve the hardenability of the steel and prevent the temper brittleness. Mo is a very expensive element, and the present invention does not require the tempering treatment; in the present invention, only no more than 0.25% of Mo and no more than 0.20% of Cr are added to achieve the purpose of reducing the cost.

[0013] Nb: in the present invention, a relatively high amount of Nb is added in order to, on the one hand, achieve the purpose of refining crystal grains and increasing the thickness of the steel plate, and on the other hand to increase the non-recrystallization temperature of the steel, which facilitates the use of a relatively high finish rolling temperature in the rolling process, thus accelerating the rolling speed and increasing the production efficiency. In addition, since the grain refining effect is strengthened, a thicker steel plate can be produced. In the present invention, 0.03-0.10 wt.% of Nb is added to give consideration to the solid solution strengthening effect and the fine grain strengthening effect of Nb.

[0014] Ti: Ti is a ferritizing element, which reduces the austenitic area significantly. The carbide of Ti, i.e. TiC, is relatively stable, and can inhibit the growth of the crystal grains. Ti, solid solved in austenite, is favourable to improve the hardenability of the steel. Ti can reduce the first type of temper brittleness at 250-400°C; however, the present invention does not require the tempering, so the addition amount of Ti can be reduced. In the present invention, an amount of 0-0.050 wt.% is added, which forms fine carbonitride to precipitate out, thus refining the Bainite laths.

[0015] Al: Al can increase the driving force of the phase change in the transition from austenite to ferrite and is an element which can intensively reduce the phase circle of the austenite. Al interacts with N in the steel to form fine and diffusive AlN, which precipitates out and can inhibit the growth of the crystal grains, thus achieving the purpose of refining crystal grains and improving the low temperature toughness of the steel. A too high content of Al will have an adverse impact on the hardenability and welding property of the steel. In the present invention, no more than 0.15% of Al is added to refine the crystal grains, improving the toughness and ensuring the welding property of the steel plate.

[0016] B: B can dramatically increase the hardenability of the steel, in the present invention, 0-0.002% of B is added so that one can relatively easily obtain a high strength bainite structure from steel under certain cooling conditions.

[0017] The contents of the three elements, Si, Mn and Mo, should comply with the following relationship: 2Si+3Mn+4Mo ≤ 8.5, to meet that the steel plate of the present invention has a good welding property. Specifically, it can be ensure that the steel plate having a thickness of 60 mm or less has no cracks upon welding at relatively low preheating temperature (normal temperature to 50°C) conditions.

[0018] The steel plate having a maximum thickness of 60 mm is produced using the chemical composition designed in the present invention and by reasonably using the action of various alloying elements.

[0019] The welding crack sensitivity index Pcm of the steel plate with low welding crack sensitivity can be determined according to the following formula:

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

[0020] The welding crack sensitivity index Pcm is an indicator for judging the weld cold cracking inclination of the steel, wherein the smaller the Pcm, the better the weldability, and conversely, the worse the weldability. Good weldability means that the occurrence of weld cracking is not easy during welding; in contrast, cracks easily occur in the steel having poor weldability; in order to prevent cracking, steel is preheated before welding; the better the weldability, the lower the preheating temperature required, inversely, a higher preheating temperature is required. According to the stipulations of the Chinese ferrous metallurgy industry standards YB/T 4137-2005, a Pcm value for the type of steel of trademark Q800CF should be lower than 0.28%. The superfine bainite lath steel plate with a high strength and low welding crack sensitivity involved in the present invention has a welding crack sensitivity of lower than 0.20%, and has an excellent welding property.

[0021] A method for manufacturing a steel plate with a yield strength at 890 Mpa level and low welding crack sensitivity of the present invention comprises the steps of:

1) smelting and casting

[0022] The following chemical components were smelt and casted to a continuous casting billet or steel ingot of a thickness not less than 4 times of the thickness of the finished steel plate; wherein the steel plate has the following components in weight percentage: C of 0.06-0.13 wt.%, Si of 0.05-0.70 wt.%, Mn of 1.20-2.30 wt.%, Mo of 0-0.25 wt.%, Nb of 0.03-0.11 wt.%, Ti of 0.002-0.050 wt.%, Al of 0.02-0.15 wt.%, and B of 0-0.0020 wt.%, with 2Si+3Mn+4Mo ≤ 8.5, the balance being Fe and inevitable impurities; and the steel plate meets the welding crack sensitivity index Pcm ≤ 0.25%;

2) heating and rolling

[0023] The heating temperature is 1050-1180°C, and the holding time is 120 to 180 minutes;
the rolling is divided into a first stage of rolling and a second stage of rolling;
during the first stage of rolling, the start rolling temperature is 1050-1150°C, and when the thickness of the rolled piece reached 2-3 times of the thickness of the finished steel plate, it is stayed in the roller bed until the temperature reached 800-860°C;
during the second stage of rolling, the pass deformation rate is 10-28%, and the finish rolling temperature is 780-840°C;

3) Cooling

[0024] The steel plate is cooled to 220-350°C at a speed of 15-30°C/S, and air cooled after being out of water.

[0025] further, in step 3), the air cooling is cooling in packed formation or in a cold bed. In the manufacturing method of the present invention:

(1) rolling process

[0026] When the thickness of the rolled piece reached 2-3 times of the thickness of the finished steel plate, it is stayed in the roller bed until the temperature reached 800-860°C. For the steel containing Nb, the non-recrystallizing temperature is about 950-1050°C. It is firstly rolled at a relatively high temperature, and there is a certain dislocation density in the austenite. During the relaxation process of lowering the temperature of the rolled billet to 800-860°C, a recovery and static recrystallization process inside the austenite crystal grains occur, thus refining the austenite crystal grains. In the relaxation process, individual precipitation and complex precipitation of carbonitrides of Nb, V and Ti occur simultaneously. The precipitated carbonitrides pin the dislocation and subgrain boundary movements, reserves a lot of dislocation in the austenite crystal grains, and provides a lot of nucleation sites for the formation of bainite during the cooling process. Rolling at 800-860°C greatly increases the dislocation density in the austenite. The carbonitride precipitated at the dislocation inhibits the coarsing of the deformed crystal grains. Due to the precipitating effect induced by deformation, a relatively large pass deformation rate will facilitate the formation of finer and more diffusive precipitates. The precipitates from high density dislocation and fine diffusion provide high density of nucleation sites for bainite, and the pining effect of the second phase particles on the bainite growth interface inhibits the growth and coarsing of the bainite laths, which is beneficial for both the strength and toughness of the steel.

[0027] The finish rolling temperature is controlled in the low temperature section of the non-recrystallization zone, and at the same time, this temperature zone is close to the phase transmission point Ar3, i.e. the finish rolling temperature is 780-840°C, and finishing rolling within this temperature range can increase the defects in the austenite by increasing the deformation and inhibiting the recovery, thus providing higher energy accumulation for the bainite phase change without bringing about a too high load to the roller, being suitable for producing a thick plate.

(2) cooling process

[0028] After the rolling is complete, the steel plate is sent to an accelerated cooling device, and cooled to 450-550°C at a rate of 15-30°C/s. A faster cooling speed can avoid the formation of ferrite and pearlite, and directly enters the bainite transition area of the CCT curve. The phase change driving force of the bainite can be expressed by:

$$\Delta G = \Delta G_{chem} + \Delta G_d$$

wherein ΔG_{chem} is a chemical driving force, and ΔG_d is a strain storage energy resulting from defects. A faster cooling speed results in the overcooling of the austenite, increases the driving force of a chemical phase change, and increase the driving force of the bainite nucleation when considered by combining the strain storage energy ΔG_d caused during the rolling process. Due to the high dislocation density in the crystal grains, the nucleation sites of bainite increase. Considered by combining both the thermodynamic and dynamic factors, the bainite can nucleate at a very large speed. A faster cooling speed enables the bainite transformation to be completed quickly and inhibits the coarsing of the bainite ferrite laths. Air cooling in packed formation at 450-550°C can enable a more complete precipitation of the carbide of V in the ferrite, thus enhancing the contribution of the precipitation strengthening to the strength. Therefore, the matrix structure composed mainly of the refined bainite can be obtained by the heat treatment process of the present invention, so as to produce steel plates having a higher strength and a good toughness.

[0029] The steel for high strength mechanical equipment and engineering construction requires a relatively high strength and an excellent toughness, wherein contribution to the strength from various factors can be expressed by the following formula:

$$\sigma = \sigma_f + \sigma_p + \sigma_{sl} + \sigma_d$$

wherein σ_f is fine grain strengthening, σ_p is precipitation strengthening, σ_{sl} is solid solution strengthening, and σ_d is dislocation strengthening. The thermo-mechanical treatment of the steel plate is usually done by a controlled rolling and controlled cooling process (TMCP). The refinement of the microstructures or the formation of the high strength structures such as ultrafine bainite can be realized by controlling the deformation rate and cooling rate, thus improving the yield strength of the steel. In the composition of the present invention, a microalloy element Nb is added, and during the heat treatment Nb may form a carbonitride, which has a precipitation strengthening effect. Nb in a solid solution in the matrix has a solid solution strengthening effect. During the heat treatment, modified TMCP and Relaxation Precipitation Controlling (RPC) technologies are used to form a stable dislocation network, and diffusive and fine second phase particles precipitate out at the dislocation and subgrain boundary; the refinement of the bainite lath is achieved by promoting the nucleation and inhibiting the growth, and a combined action of dislocation strengthening, precipitation strengthening and fine grain strengthening is formed, thus improving the strength and roughness of the steel, its principle mechanism being as follows:

the steel plate fully deforms in the recrystallization zone, such that a high defect accumulation occurs in the deformed austenite, thus greatly increasing the dislocation density in the austenite. Recovery and recrystallization occurring during the rolling refine the original austenite crystal grains. Dislocation within the crystals will be re-arranged during the controlled cooling relaxation after rolling and deforming. Since a hydrostatic pressure field exits in the edge dislocation, interstitial atoms such as B will enrich to the dislocation, grain boundary and subgrain boundary, and reduce the dislocation mobility. The high density dislocation resulting from the deformation will evolve during the recovery to form a stable dislocation network. During the relaxation, the microalloy elements such as Nb, V, Ti precipitate out at the grain boundary, subgrain boundary and dislocations in the form of carbonitrides of different stoichiometric ratios such as (Nb,V,Ti)x(C,N)y. The second phase particles, such as the precipitated carbonitrides, pin the dislocations and subgrain boundary within the crystal grains and stabilize the substructures, such as dislocation walls and the like. After relaxation, the dislocation density of the steel is further increased by rolling. After relaxation, when the deformed austenite is accelerated cooled, the deformed austenite crystal grains with dislocation and carbonitride precipitation configuration at the beginning of the phase change is different from the circumstance that after deformation, no relaxation occurs and there are a lot of dislocations disorderly distributed. Firstly, a subgrain boundary with a certain orientation difference is a preferred nucleation site, and if a second phase, which has a heterophasic interface with the matrix, precipitates out nearby, this will be more advantageous for the new phase nucleation during phase change. After relaxation, a lot of new phase crystal grains will nucleate within the original austenite crystal grains. Secondly, since after relaxation, a certain amount of dislocations move to the subgrain boundary, the orientation difference between the subgrains is increased to a certain extent. After the medium temperature transformed product, such as bainite, nucleates at the subgrain boundary, it is hindered by the front subgrain boundary during the growth. When the bainite ferrite forms, its phase change interface is daggled by the precipitated second phase carbonitride particles, which inhibits its growth process. The TMCP + RPC process results in a high density dislocation network structure, and the second phase precipitation particle points provide a lot of potential nucleation sites for the nucleation of the bainite ferrite The daggling effect of the second phase particles to the moving interface and the evolved subgrain boundary have an inhibiting effect on the growth of the bainite. The combined effect of promoting the nucleation and inhibiting the growth in the process refines the bainite ferrite laths of the final structure.

[0030] With regard to high strength steels used in the mechanical structure and engineering construction, no preheating or a little preheating is required before welding, without the generation of crack, which mainly solves the welding construction problem of large steel structures. The only method to reduce Pcm is to reduce the addition amounts of carbon and alloying elements; however, for the high strength steel produced by a quenching + tempering process, reducing the addition amounts of carbon and alloying elements will inevitably lead to the reduction of the steel strength, while the use of the modified TMCP + RPC process in the present invention can remedy such a defect. The composition system used in the present invention ensures that the steel plate has a high strength and a low-temperature toughness, and at the same time a welding crack sensitivity index Pcm ≤ 0.20%, and has an excellent welding property.

[0031] Advantageous effects of the present invention lie in:

1. The content of C is greatly reduced by reasonably designing the chemical composition, replacing part of Mo with cheap alloy elements such as Mn, replacing the precipitation strengthening effect of Cu with the precipitation strengthening effect of precipitated fine particles of carbonitride of Nb, without adding noble elements such as Ni; the content of the alloy element is low, the cost of the raw materials is relatively low, the welding crack sensitivity is relatively low and no preheating is required before welding.
2. The steel plate of the present invention does not require any additional thermal treatment, thus simplifying the manufacturing procedure and reducing the manufacture cost of the steel.
3. Due to the reasonable composition and process design, the process system is relatively loose in view of the implementing effects and the steel plate can be produced stably in a medium, and thick steel plate production line.
4. The steel plate with low welding crack sensitivity of the present invention has a yield strength of greater than 890 MPa, a tensile strength of greater than 950 MPa, a Charpy impact energy Akv (-20°C) ≥ 100 J, and a plate thickness of up to 60 mm. The steel plate has a welding crack sensitivity index Pcm ≤ 0.25%, and has an excellent welding property.
5. A thick plate having a maximum thickness of 60 mm can be produced by the present invention.

PREFERRED EMBODIMENTS OF THE INVENTION

[0032] The present invention is described by the following examples in further detail. These examples are only intended to describe the preferred embodiments of the invention, but not to limit the scope of the invention in any way.

[0033] Table 1 is the chemical composition (wt.%) of the steel plate of the examples of the present invention and the Pcm (%) values. Table 2 is the mechanical property of the steel plate of the examples of the present invention. Table 3 is the test (small Tekken test) results of the welding property of the steel plate with an 890 Mpa level and low welding crack sensitivity of Example 1 of the present invention.

Example 1

[0034] The chemical components as shown in Table 2 are smelt in an electric furnace or a converter and casted to a continuous casting billet or steel ingot, which is then heated to 1110°C for a holding time of 120 min and is subjected to a first stage of rolling in a middle, and thick rolling mill, wherein the start rolling temperature is 1050°C, when the thickness of the rolled piece is 60 mm, it is stayed in the roller bed until the temperature reached 850°C, and then a second stage of rolling is performed, wherein the pass deformation rate in the second stage of rolling is 15-28%, the finish rolling temperature is 830°C, and the thickness of the finished steel plate is 20 mm. After the rolling is complete, the steel plate is sent to an accelerated cooling (ACC) device, and cooled to 300°C at a rate of 30°C/s, followed by cooling in packed formation or in a cold bed after being out of water.

Example 2

[0035] It is performed as in Example 1,wherein the heating temperature is 1050°C and holding time is 240 min; the start rolling temperature in the first stage of rolling is 1040°C, and the thickness of the rolled piece is 90 mm; the start rolling temperature in the second stage of rolling is 840°C, the pass deformation rate is 15-20%, the finish rolling temperature is 810°C, and the thickness of the finished steel plate is 30 mm; and the cooling rate of the steel plate is 25°C/S, and the final temperature is 350°C.

Example 3

[0036] It is performed as in Example 1,wherein the heating temperature is 1150°C and the holding time is 150 min; the start rolling temperature in the first stage of rolling is 1080°C, and the thickness of the rolled piece is 120 mm; the start rolling temperature in the second stage of rolling is 830°C, the pass deformation rate is 10-15%, the finish rolling temperature is 820°C, and the thickness of the finished steel plate is 40 mm; and the cooling rate of the steel plate is 20°C/S, and the final temperature is 330°C.

Example 4

[0037] It is performed as in Example 1,wherein the heating temperature is 1120°C and the holding time is 180 min; the start rolling temperature in the first stage of rolling is 1070°C, and the thickness of the rolled piece is 150 mm; the start rolling temperature in the second stage of rolling is 830°C, the pass deformation rate is 10-20%, the finish rolling temperature is 800°C, and the thickness of the finished steel plate is 50 mm; and the cooling rate of the steel plate is 15°C/S, and the final temperature is 285°C.

Example 5

[0038] It is performed as in Example 1,wherein the heating temperature is 1130°C and the holding time is 180 min; the start rolling temperature in the first stage of rolling is 1080°C, and the thickness of the rolled piece is 150 mm; the start rolling temperature in the second stage of rolling is 840°C, the pass deformation rate is 10-15%, the finish rolling temperature is 810°C, and the thickness of the finished steel plate is 60 mm; and the cooling rate of the steel plate is 15°C/S, and the final temperature is 220°C.

Example 6

[0039] It is performed as in Example 1,wherein the heating temperature is 1120°C and the holding time is 180 min; the start rolling temperature in the first stage of rolling is 1050°C, and the thickness of the rolled piece is 120 mm; the start rolling temperature in the second stage of rolling 820°C, the pass deformation rate is 15-25%, the finish rolling temperature is 780°C, and the thickness of the finished steel plate is 40 mm; and the cooling rate of the steel plate is 20°C/S, and the final temperature is 300°C.

Table 1 unit: weight percentage

Examples	C	Si	Mn	Nb	Al	Ti	Cr	Mo	B	Fe	Pcm
1	0.09	0.35	1.80	0.070	0.02	0.015	0.16	0.25	0.0018	the balance	0.21 7
2	0.06	0.70	2.25	0.045	0.06	0.020	0	0	0.0010	the balance	0.20 1
3	0.08	0.40	2.06	0.085	0.04	0.050	0.20	0.10	0.0011	the balance	0.21 8
4	0.13	0.55	1.20	0.110	0.15	0	0.16	0.25	0.0015	the balance	0.18 3
5	0.06	0.05	1.45	0.065	0.07	0.020	0.12	0.20	0.0010	the balance	0.24 1
6	0.10	0.15	1.90	0.095	0.09	0.008	0.15	0.22	0.0020		0.23 2

Table 2

Examples	Yield strength MPa	Tensile strength MPa	Elongation %	-20°C Longitudinal impact energy J		
1	940 965	1050 1065	16.0 16.5	189	216	204
2	950 975	1060 1070	15.9 15.2	208	190	209
3	955 960	1058 1065	15.0 15.0	195	202	212
4	945 940	1154 1150	16.1 16.3	191	208	206
5	988 995	1169 1173	15.0 14.5	207	201	224
6	910 915	1067 1082	17.3 17.3	202	210	210

[0040] From tables 1 and 2, it can be seen that the Pcm of the steel plate with a yield strength at an 890 Mpa level and low welding crack sensitivity involved in the present invention is ≤ 0.25%, the yield strength is larger than 890 MPa, the tensile strength is larger than 950 MPa, the Charpy impact energy Akv (-20°C) is ≥ 120 J, and the plate thickness can be up to 60 mm, and the steel plate has an excellent low-temperature toughness and weldability.

[0041] The steel plate of Example 1 of the present invention is tested for the welding property (small Tekken test), under conditions of room temperature and 50°C, and no crack is observed (see table 3), indicating that the type of steel of the present invention has an excellent welding property, and generally does not require preheating when welding.

Table 3

Test temp eratu re	Examples	Surface crack rate %	Root crack rate %	section crack rate %	Environ mental temperat ure	Relat ive humi dity
RT	1	0	0	0	22°C	60%
	2	0	0	0		
	3	0	0	0		
50°C	4	0	0	0		
	5	0	0	0		

Claims

1. A steel plate with a yield strength at an 890 Mpa level and low welding crack sensitivity, wherein the steel plate has the following components in weight percentage: C of 0.06-0.13 wt.%, Si of 0.05-0.70 wt.%, Mn of 1.20-2.30 wt.%, Mo of 0-0.25 wt.%, Nb of 0.03-0.11 wt.%, Ti of 0.002-0.050 wt.%, Al of 0.02-0.15 wt.%, and B of 0-0.0020 wt.%, with 2Si+3Mn+4Mo ≤ 8.5, the balance being Fe and inevitable impurities; and the steel plate meets the welding crack sensitivity index Pcm ≤ 0.25%.

2. A method for manufacturing a steel plate with a yield strength at 890 Mpa level and low welding crack sensitivity, comprising the steps of:

   1) smelting and casting
   the following chemical components were smelt and casted to a continuous casting billet or steel ingot of a thickness not less than 4 times of the thickness of the finished steel plate; the continuous casting billet or steel ingot has the following chemical components in weight percentage: C of 0.06-0.13 wt.%, Si of 0.05-0.70 wt.%, Mn of 1.20-2.30 wt.%, Mo of 0-0.25 wt.%, Nb of 0.03-0.11 wt.%, Ti of 0.002-0.050 wt.%, Al of 0.02-0.15 wt.%, and B of 0-0.0020 wt.%, with 2Si+3Mn+4Mo ≤ 8.5, the balance being Fe and inevitable impurities; and the steel plate meets the welding crack sensitivity index Pcm ≤ 0.25%;
   2) heating and rolling
   the heating temperature is 1050-1180°C, and the holding time is 120 to 180 minutes;
   the rolling is divided into a first stage of rolling and a second stage of rolling; during the first stage of rolling, the start rolling temperature is 1050-1150°C, and when the thickness of the rolled piece reached 2-3 times of the thickness of the finished steel plate, it is stayed in the roller bed until the temperature reached 800-860°C; during the second stage of rolling, the pass deformation rate is 10-28%, and the finish rolling temperature is 780-840°C;
   3) Cooling

   the steel plate is cooled to 220-350°C at a speed of 15-30°C/S, and air cooled after being out of water.

3. The method for manufacturing the steel plate with a yield strength at 890 Mpa level and low welding crack sensitivity of claim 2, **characterized in that**, in step 3), the air cooling is cooling in packed formation or in a cold bed.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/CN2015/070729

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/14 (2006.01) i, C21D 7/13 (2006.01) i

According to International Patent Classification (IPC) or to both national classification and IPC

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)

C22C 38; C21D 7

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)

WPI, EPODOC, CNKI, CPRS: strength, weld+, Si, Mn, Mo, Nb, Ti, Al, B, silicon, manganese, molybdenum, niobium, titanium, aluminium, aluminum, boron

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
PX	CN 103898406 A (BAOSHAN IRON & STEEL CO., LTD.) 02 July 2014 (02.07.2014) claims 1-3	1-3
Y	CN 101481774 A (BAOSHAN IRON & STEEL CO., LTD.) 15 July 2009 (15.07.2009) claims 1-6, description, paragraph [0016], and embodiment 3	1-3
Y	CN 102618793 A (BAOSHAN IRON & STEEL CO., LTD.) 01 August 2012 (01.08.2012) description, paragraphs [0038]-[0042], and embodiment 4	1-3
Y	CN 103484768 A (WUHAN IRON & STEEL GROUP CORP.) 01 January 2014 (01.01.2014) claims 1 and 2	1-3
Y	CN 101942616 A (UNIV BEIJING SCI & TECHNOLOGY) 12 January 2011 (12.01.2011) claims 1-4	1-3
Y	JP 2010070845 A (KIMA) 02 April 2010 (02.04.2010) the abstract	1-3

☐ Further documents are listed in the continuation of Box C. ☒ See patent family annex.

* Special categories of cited documents:

"A" document defining the general state of the art which is not considered to be of particular relevance

"E" earlier application or patent but published on or after the international filing date

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"O" document referring to an oral disclosure, use, exhibition or other means

"P" document published prior to the international filing date but later than the priority date claimed

"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention

"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone

"Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art

"&"document member of the same patent family

Date of the actual completion of the international search	Date of mailing of the international search report
05 April 2015	22 April 2015

Name and mailing address of the ISA	Authorized officer
State Intellectual Property Office of the P. R. China No. 6, Xitucheng Road, Jimenqiao Haidian District, Beijing 100088, China Facsimile No. (86-10) 62019451	WU, Chenchen Telephone No. (86-10) 62084743

Form PCT/ISA /210 (second sheet) (July 2009)

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/CN2015/070729

Patent Documents referred in the Report	Publication Date	Patent Family	Publication Date
CN 103898406 A	02 July 2014	None	
CN 101481774 A	15 July 2009	CN 101481774 B	24 November 2010
CN 102618793 A	01 August 2012	CN 102618793 B	20 November 2013
CN 103484768 A	01 January 2014	None	
CN 101942616 A	12 January 2011	CN 101942616 B	03 October 2012
JP 2010070845 A	02 April 2010	KR 20100032490 A	26 March 2010
		CN 101676430 A	24 March 2010
		KR 1095911 B1	21 December 2011
		KR 20110111357 A	11 October 2011
		KR 1094310 B1	19 December 2011

Form PCT/ISA /210 (patent family annex) (July 2009)

REFERENCES CITED IN THE DESCRIPTION

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Patent documents cited in the description

- WO 9905335 A **[0004]**
- CN 1521285 **[0005]**