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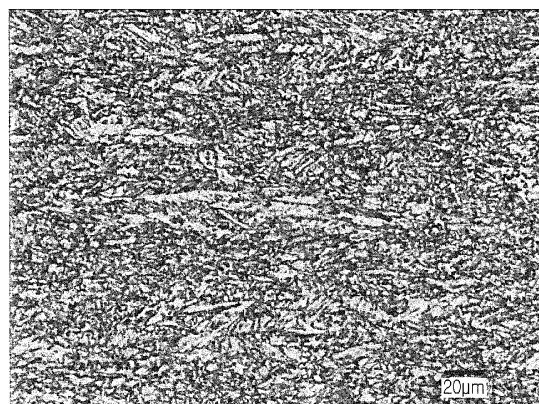
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(54) **HIGH-STRENGTH STRUCTURAL STEEL SHEET EXCELLENT IN HOT RESISTANCE AND MANUFACTURING METHOD THEREOF**

(57) The present invention relates to a high-strength structural steel sheet having excellent heat treatment resistance, comprising in percentage by weight: 0.03 to 0.07% of C; 0.05 to 0.2% of Si; 1.6 to 2.3% of Mn; 0.008% or less of P; 0.002% or less of S; 0.025% or less of Al; 0.1 to 0.4% of Cu; 1.4 to 2.3% of Ni; 0.08 to 0.2% of Mo; 0.01 to 0.025% of Nb; 0.008 to 0.02% of Ti; 0.001 to 0.008% of N; and the remainder being Fe and unavoidable impurities, wherein a microstructure within 10 mm or less from a surface comprises, by volume fraction, 80% or more of acicular ferrite and 20% or less of polygonal ferrite.

[FIG. 2]



Description

[Technical Field]

5 **[0001]** The present disclosure relates to a high-strength structural steel sheet having excellent heat treatment resistance and a manufacturing method thereof.

[Background Art]

10 **[0002]** The exteriors of ships and marine and architectural structures have both planar and curved surfaces.

[0003] Since planar processing is determined during a plate material forming process, the exterior is formed without performing a separate process when ships or marine structures are constructed, but in the case of a curved surface forming process, a plate material is processed. In order to perform this process, linear heating, an operation of heating a surface of steel sheet, is performed.

15 **[0004]** Bending work, based on linear heating, utilizes properties in which a heated portion is deformed due to constraint from a non-heated region around the heated portion when the heated portion contracts due to cooling after being heat-expanded.

[0005] In order to apply such linear heating, a surface of the steel sheet is heated to a temperature of about 600 to 900°C or water-cooled after heating, and thus, after the linear heating, physical properties of the steel sheet may be lowered. When a steel material is heated to an austenite start transformation temperature, if the steel material is heated to above the transformation temperature or a recrystallization temperature due to annealing of dislocations, or the like, the material is mainly deteriorated due to the growth of a crystal grain.

20 **[0006]** Further, the steel sheet may be embrittled due to a heat cycle of heating and cooling of the surface of the steel sheet, resulting in deterioration of toughness.

25 **[0007]** Therefore, development of a high-strength structural steel sheet having excellent heat treatment resistance, which has excellent yield strength, tensile strength, and impact toughness even after linear heating, and a manufacturing method thereof are required.

[Disclosure]

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[Technical Problem]

[0008] An aspect of the present disclosure is to provide a high-strength structural steel sheet having excellent heat treatment resistance, which has excellent yield strength, tensile strength, and impact toughness, even after linear heating, and a manufacturing method thereof are required.

35 **[0009]** Meanwhile, the object of the present disclosure is not limited to the above description. It will be understood by a person skilled in the art to which the present disclosure pertains that there is no difficulty in understanding additional problems of the present disclosure.

40 [Technical Solution]

[0010] According to an aspect of the present disclosure, a high-strength structural steel sheet having excellent heat treatment resistance includes: by weight percent (wt%), 0.03 to 0.07% of C, 0.05 to 0.2% of Si, 1.6 to 2.3% of Mn, 0.008% or less of P, 0.002% or less of S, 0.025% or less of Al, 0.1 to 0.4% of Cu, 1.4 to 2.3% of Ni, 0.08 to 0.2% of Mo, 0.01 to 0.025% of Nb, 0.008 to 0.02% of Ti, 0.001 to 0.008% of N, the remainder being Fe, and unavoidable impurities, wherein a microstructure within 10 mm of the surface contains 80% or more of acicular ferrite and 20% or less of polygonal ferrite.

45 **[0011]** According to an aspect of the present disclosure, a method of manufacturing a high-strength structural steel sheet having excellent heat treatment resistance includes: reheating a slab including, by weight percent (wt%), 0.03 to 0.07% of C, 0.05 to 0.2% of Si, 1.6 to 2.3% of Mn, 0.008% or less of P, 0.002% or less of S, 0.025% or less of Al, 0.1 to 0.4% of Cu, 1.4 to 2.3% of Ni, 0.08 to 0.2% of Mo, 0.01 to 0.025% of Nb, 0.008 to 0.02% of Ti, 0.001 to 0.008% of N, the remainder being Fe, and unavoidable impurities; performing non-recrystallization region rolling on the reheated slab at 750 to 850°C; and performing cooling to a cooling end temperature of 380 to 440°C at a cooling rate of 10°C/sec. or faster after the non-recrystallization region rolling.

50 **[0012]** The foregoing technical solutions do not fully enumerate all of the features of the present disclosure. The foregoing and other objects, features, aspects and advantages of the present disclosure will become more apparent from the following detailed description of the present disclosure when taken in conjunction with the accompanying drawings.

[Advantageous Effects]

[0013] As set forth above, according to an exemplary embodiment in the present disclosure, the high-strength structural steel sheet, which has excellent yield strength, tensile strength, and low-temperature impact toughness even after linear heating by having excellent heat treatment resistance, as well as before the linear heating, and a manufacturing method thereof may be provided.

[Description of Drawings]

[0014]

FIG. 1 is a schematic view illustrating an example of curved surface forming by linear heating.

FIG. 2 is a photograph of a cross-sectional structure of a steel sheet of Inventive Example 1 at a depth of 10 mm from a surface thereof.

[Best Mode for invention]

[0015] Hereinafter, exemplary embodiments of the present disclosure will now be described in detail with reference to the accompanying drawings. The exemplary embodiments may, however, be exemplified in many different forms and should not be construed as being limited to the specific exemplary embodiments set forth herein. Also, these exemplary embodiments are provided so that this disclosure will be thorough and complete, and will fully convey the scope of the inventive concept to those skilled in the art.

[0016] The inventors of the present disclosure have found that when the high-strength structural steel sheet is linearly heated to be bent to have a curved surface in the appearance of ships, marine structures and building structures, physical properties of the steel sheet may be lowered after the linear heating.

[0017] Since the surface of the steel sheet is heated to 600 to 900°C during linear heating, a base structure and a grain boundary are softened, a crystal grain is grown, carbide (Fe_3C) is coarsened, causing a phenomenon in which both strength and toughness are degraded.

[0018] Further, the heating to the austenite start transformation temperature causes deterioration of the material due to annealing of dislocations, and the like, and when the steel sheet is heated to above the transformation temperature or above the recrystallization temperature, the material is mainly degraded due to the growth of the crystal grains.

[0019] In order to solve the problems, the inventors of the present disclosure have found that when a microstructure within 10 mm or less from a surface includes, by volume fraction, 80% or more of acicular ferrite and 20% or less of polygonal ferrite through low temperature rolling and strong cooling by lowering an Ar_3 temperature by adding large amounts of Mn and Ni, growth of a crystal grain after linear heating may be prevented and growth of a crystal grain and formation of coarse carbide may be prevented using a grain boundary pinning effect of precipitates of NbC and Mo_2C , whereby, the inventors noticed that, a high-strength structural steel sheet having excellent yield strength, tensile strength, and impact toughness even after linear heating, as well as before the linear heating, and a manufacturing method thereof can be provided, thus completing the present disclosure.

[0020] Hereinafter, the high-strength structural steel sheet having excellent heat treatment resistance according to an aspect of the present disclosure will be described.

[0021] A high-strength structural steel sheet having excellent heat treatment resistance according to an aspect of the present disclosure includes, by weight percent (wt%), 0.03 to 0.07% of C, 0.05 to 0.2% of Si, 1.6 to 2.3% of Mn, 0.008% or less of P, 0.002% or less of S, 0.025% or less of Al, 0.1 to 0.4% of Cu, 1.4 to 2.3% of Ni, 0.08 to 0.2% of Mo, 0.01 to 0.025 % of Nb, 0.008 to 0.02% of Ti, 0.001 to 0.008% of N, the remainder being Fe, and unavoidable impurities, and a microstructure within 10 mm of the surface contains 90% or more of acicular ferrite.

C: 0.03 to 0.07% by weight % (hereinafter, the unit of each element content is wt%)

[0022] C is a very important element for securing strength.

[0023] In order to secure sufficient strength, 0.03% or more of C may be added. If C is excessively added, a coarse carbide may be formed during cooling after the linear heating to lower impact toughness, and thus, an upper limit of C is preferably 0.07%.

Si: 0.05 to 0.2%

[0024] Si is a useful element as a deoxidizing agent, but if the content of C is excessive, toughness may be lowered. For deoxidation, the Si content is preferably 0.05% or more. However, if the Si content exceeds 0.2%, toughness may

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be lowered. Therefore, the Si content is preferably 0.05 to 0.2%.

Mn: 1.6 to 2.3%

5 **[0025]** Mn has an effect of improving strength as a solid solution strengthening element and improving grain refinement and toughness of a base material. In addition, formation of polygonal ferrite may be minimized by low temperature rolling and strong cooling by lowering the Ar_3 temperature.

10 **[0026]** In order to sufficiently exhibit the above effect, Mn may be added in an amount of 1.6% or more. If, however, Mn is excessively added, a nonmetallic inclusion of MnS is formed at a central portion, and the MnS inclusion may be stretched after rolling to significantly lower low temperature toughness. Therefore, an upper limit thereof is preferably 2.3%.

P: 0.008% or less

15 **[0027]** P is an element advantageous for improving strength and corrosion resistance but since it is an element which greatly hinders impact toughness, it is advantageous to keep it as low as possible, and therefore, an upper limit thereof is preferably 0.008%.

S: 0.002% or less

20 **[0028]** Since S forms MnS, or the like, to greatly deteriorate impact toughness, it is advantageous to add it as little as possible, so that the upper limit is preferably 0.002%.

Al: 0.025% or less

25 **[0029]** Al, an element for effectively performing deoxidation, is preferably controlled to 0.005 to 0.025%. It is not necessary to specifically control a lower limit, but Al may be included by 0.005% or more for deoxidation.

Cu: 0.1 to 0.4%

30 **[0030]** Cu, a solid solution strengthening and precipitation strengthening element, is an element capable of improving strength while minimizing a decrease in toughness of the base material. In order to achieve a sufficient strength improvement effect, Cu is preferably contained in an amount of 0.1% or more. Meanwhile, the excessive addition of Cu may cause defects on a surface of the steel due to hot brittleness, so an upper limit thereof is preferably 0.4% or less.

35 Ni: 1.4 to 2.3%

40 **[0031]** Ni is an element capable of simultaneously improving both strength and toughness of a base material. In addition, formation of polygonal ferrite may be minimized by low temperature rolling and strong cooling by lowering the Ar_3 temperature.

[0032] If the Ni content is less than 1.4%, the above-mentioned effect is insufficient. If the Ni content is more than 2.3%, curability may be increased and impact toughness may be lowered due to formation of bainite. Therefore, the Ni content is preferably 1.4 to 2.3%.

45 Mo: 0.08 to 0.2%

50 **[0033]** Mo is an element which effectively increases strength by the addition of a small amount, and Mo may be added in an amount of 0.08% or more in order to form precipitates of fine Mo-C series after linear heating to prevent deterioration of strength. However, since coarse precipitates may occur due to excessive addition of Mo, an upper limit thereof is preferably 0.2% or less.

Nb: 0.01 to 0.025%

55 **[0034]** Nb dissolved in the steel sheet prior to linear heating is precipitated in the form of NbC or NbCN during the linear heating to improve strength of the base material. Nb is important to maintain strength after linear heating, and at least 0.01% must be added in order to effectively demonstrate the addition effect of Nb. However, excessive addition of Nb may coarsen the precipitates, and therefore, an upper limit thereof is preferably 0.025% or less.

Ti: 0.008 to 0.02%

[0035] Ti forms a nitride with N to prevent crystal grains from growing at a high temperature. In order to sufficiently secure such an effect, the Ti content is 0.008% or more. Meanwhile, excessive Ti addition may lower impact toughness due to coarsening of the Ti precipitates, and thus, an upper limit thereof is preferably 0.02%.

N: 0.001 to 0.008%

[0036] N is an element which forms a precipitate together with Ti, Nb and Al to improve strength and toughness by making an austenite structure finer during reheating.

[0037] If the N content is less than 0.001%, the above-mentioned effect may not be sufficiently obtained. Meanwhile, when the N content exceeds 0.008%, surface cracking may occur at a high temperature and the residual N may exist in an atomic state to reduce toughness. Therefore, the N content is preferably 0.001 to 0.008%.

[0038] The remainder of the steel sheet of the present disclosure is iron (Fe). However, in the ordinary manufacturing process, impurities that are not intended may be inevitably mixed from a raw material or a surrounding environment, so the impurities may not be excluded. These impurities are known to any person skilled in the art of manufacturing, and thus, are not specifically mentioned in this disclosure.

[0039] Hereinafter, a microstructure of a high-strength structural steel sheet having excellent heat treatment resistance according to one aspect of the present disclosure will be described.

[0040] According to an aspect of the present disclosure, a microstructure within 10 mm of the surface of the high-strength structural steel sheet having excellent heat treatment resistance includes 80% or more of an acicular ferrite and 20% or less of polygonal ferrite, by volume fraction.

[0041] Since crystal grains of the polygonal ferrite are easily grown by heating, if the polygonal ferrite is present in an amount of more than 20 volume % in the microstructure within 10 mm, crystal grains may grow, a coarse carbide may be formed, and a base structure may be degraded.

[0042] A thickness of the steel sheet is preferably 40 mm or less. If the thickness of the steel sheet exceeds 40 mm, it is difficult to apply bending by linear heating. Here, a minimum thickness of the steel sheet may be 12 mm.

[0043] By controlling the alloy composition and microstructure as described above, the steel sheet has yield strength of 500 MPa or greater, tensile strength of 600 MPa or greater, and impact toughness of 100 J or greater at -40°C. Accordingly, the steel sheet may be preferably used in ships, marine and building structures, and the like.

[0044] When the high-strength structural steel sheet having excellent heat treatment resistance having the alloy composition and microstructure as described above is subjected to linear heating, one or more of a Mo₂C precipitate and a NbC precipitate is precipitated.

[0045] In an example of curving (bending) by linear heating illustrated in FIG. 1, bending based on linear heating uses properties that a heated portion is deformed by constraint from a surrounding non-heated region when a heated portion contracts due to cooling after thermally expanded.

[0046] Also, in the case of linear heating, the surface of the steel sheet is generally heated to 600 to 900°C. Mo dissolved in the steel sheet at a relatively low temperature of 600 to 800°C is precipitated as Mo₂C upon cooling after the linear heating and Nb dissolved in the steel sheet at a relatively high temperature of 800°C or higher is precipitated as NbC upon cooling after the linear heating.

[0047] The Mo₂C precipitate or NbC precipitate is precipitated to grain boundaries to inhibit growth of crystal grains (pinning effect) and prevent formation of a coarse carbide. Further, as C is greatly consumed as a precipitate, generation and coarsening of carbide may be prevented. Here, a size of the Mo₂C precipitate and the NbC precipitate is preferably 2 to 20 nm.

[0048] Therefore, although the steel sheet according to the present disclosure is linearly heated to 600 to 900°C and subjected to curving, the steel sheet has a yield strength of 500 MPa or greater, tensile strength of 600 MPa or greater, and an impact toughness of 100 J or more at -40°C.

[0049] Hereinafter, a method of manufacturing a high-strength structural steel sheet having excellent heat treatment resistance, as another aspect of the present disclosure, will be described.

[0050] The method of manufacturing a high-strength structural steel sheet having excellent heat treatment resistance, as another aspect of the present disclosure, includes: reheating a slab having the above-described alloy composition; performing non-recrystallization region rolling on the reheated slab at 750 to 850°C; and performing cooling to a cooling end temperature of 380 to 440°C at a cooling rate of 10°C/sec. or faster after the non-recrystallization region rolling.

Reheating

[0051] The slab having the above-described alloy composition is reheated. A reheating temperature of the slab is not limited but is preferably 1100 to 1200°C.

Non-recrystallization region rolling

[0052] The reheated slab is subjected to non-recrystallization region rolling at 750 to 850°C. This is to make crystal grains finer.

[0053] In order to make the crystal grains finer, the non-recrystallization rolling must be performed at a lowest temperature right above the A_{r3} temperature. In the present disclosure, since the A_{r3} temperature is sufficiently low by increasing the content of Mn and Ni, non-recrystallization region rolling is preferably performed at 750°C or higher. If the non-recrystallization region rolling temperature is higher than 850°C, grain refinement is difficult to weaken toughness. Therefore, an upper limit of the non-recrystallization region rolling temperature is preferably 850°C, and a more preferable upper limit is 800°C.

Cooling

[0054] After the non-recrystallization region rolling, the steel sheet is cooled to a cooling end temperature of 380 to 440°C at a cooling rate of 10°C/sec. or faster.

[0055] By controlling the cooling step as described above, the microstructure within 10 mm of the surface of the steel sheet may include at least 80% or more of acicular ferrite and 20% or less of polygonal ferrite by a volume fraction.

[0056] When the cooling rate is less than 10°C/sec. or a cooling finish temperature is higher than 440°C, cooling is not sufficiently performed to form a large amount of polygonal ferrite, cause grain growth, and deteriorate the matrix during linear heating.

[0057] Meanwhile, linear heating the cooled steel sheet at a temperature of 600 to 900°C to perform bending may be additionally performed.

[0058] By the above-mentioned linear heating, it is possible to bend the steel sheet, and since Mo_2C precipitates and NbC precipitates are precipitated upon cooling after the linear heating, grain growth may be inhibited (pinning effect) and formation of coarse carbide may be prevented.

[Mode for invention]

[0059] Hereinafter, the present disclosure will be described more specifically by way of examples.

[0060] Molten steel having the alloy composition as illustrated in Table 1 below were prepared, and thereafter, steel slabs were prepared using continuous casting. Inventive steels A, B, and C are steel sheets satisfying the component range defined in the present disclosure, and comparative steels D, F, G, and H are steel sheets having a component range which exceeds or falls short of the component range of the present disclosure. C of comparative steel D is outside the component range of the present disclosure, Mo of comparative steel E is outside the component range of the present disclosure, Nb of the comparative steel F is outside the component range of the present disclosure, and Ni and Mn of comparative steel G are outside the component range of the present disclosure.

[0061] The inventive steels and comparative steels were rolled and cooled under the manufacturing conditions shown in Table 2 to prepare thick steel sheets. Specifically, rolling end temperatures were 780°C and 880°C, and cooling end temperatures were 400°C and 600°C. The manufactured steel sheets were cut into a size available for linear heating, and linear heating for bending was performed under four temperature conditions (600°C, 700°C, 800°C and 900°C).

[0062] Table 3 shows mechanical properties of base materials manufactured under the above-mentioned conditions and mechanical properties after linear heating.

[0063] Tensile strength of the base material was measured by taking a JIS1B sample from a full thickness of the steel sheets in a direction perpendicular to a rolling direction and performing a tension test at room temperature. As for low temperature toughness of the base material, a sample was taken in a direction perpendicular to a rolling direction from a portion 2mm directly under from a surface portion of the steel sheet to manufacture a V-notch sample, Charpy impact test was performed on the test sample at -40°C three times and an average value is illustrated in Table 3.

[0064] In addition, the microstructure within 10 mm of the surface of the steel sheet was observed and a volume fraction of polygonal ferrite is shown in Table 3. A structure other than the polygonal ferrite was acicular ferrite.

[Table 1]

Classific ation	C	Si	Mn	P	S	Al	Ni	Cu	Mo	Nb	Ti	N
Inventive steel A	0.042	0.086	1.95	0.0055	0.0015	0.011	1.71	0.274	0.13	0.015	0.010	0.0038
Inventive steel B	0.054	0.116	1.83	0.0057	0.0012	0.010	1.75	0.249	0.125	0.021	0.012	0.0042
Inventive steel C	0.045	0.123	1.92	0.0062	0.0011	0.011	1.82	0.254	0.119	0.013	0.013	0.0039
Comparative steel D	0.126	0.123	1.88	0.0061	0.0012	0.010	1.80	0.249	0.121	0.017	0.011	0.0049
Comparative steel E	0.052	0.118	1.91	0.0052	0.0014	0.015	1.68	0.261	0.052	0.02	0.012	0.0051
Comparative steel F	0.046	0.121	1.89	0.0075	0.0013	0.012	1.76	0.248	0.125	0.008	0.010	0.0045
Comparative steel G	0.048	0.119	1.25	0.0065	0.0013	0.013	0.65	0.253	0.132	0.018	0.012	0.0042
In Table 1, the unit of each element content is wt%.												

[Table 2]

Classification	Steel type	FM start temperature	FM end temperature	Cooling start temperature	Cooling finish temperature	Cooling rate	Linear heating temperature
Inventive example 1	Inventive steel A	795	770	732	432	11.3	600
							700
							800
							900
Comparative example 1	Inventive steel A	885	864	832	412	13.2	600
							700
							800
							900
Inventive example 2	Inventive steel B	789	761	726	408	12.5	600
							700
							800
							900
Comparative example 2	Inventive steel B	796	769	733	603	9.8	600
							700
							800
							900
Inventive example 3	Inventive steel C	792	775	740	410	11.8	600
							700
							800
							900
Comparative example 3	Comparative steel D	801	779	746	408	13.5	600
							700
							800
							900

(continued)

Classification	Steel type	FM start temperature	FM end temperature	Cooling start temperature	Cooling finish temperature	Cooling rate	Linear heating temperature
Comparative example 4	Comparative steel E	795	776	742	411	12.5	600
							700
							800
							900
Comparative example 5	Comparative steel F	803	776	738	403	12.8	600
							700
							800
							900
Comparative example 6	Comparative steel G	795	768	734	411	11.6	600
							700
							800
							900

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[0065] In Table 2, the unit of temperature is °C, the unit of the cooling rate is °C/sec., the FM start temperature refers to the non-recrystallization region rolling start temperature, and the FM end temperature refers to the non-recrystallization region rolling end temperature.

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[Table 3]

Classification	Steel type	Base material			After linear heating				Average impact toughness
		Yield strength	Tensile strength	Polygonal ferrite (volume %)	Impact toughness average	Yield strength	Tensile strength	Individual impact toughness	
Inventive example 1	Inventive steel A	527	665	7.5	278	514	661	267/262/253	260
						513	649	265/288/302	285
						532	642	277/322/326	308
						507	627	194/130/184	170
Comparative example 1	Inventive steel A	525	654	24.5	79	509	646	45/56/128	76
						502	635	79/65/84	76
						500	624	56/102/28	62
						498	612	57/58/38	51
Inventive example 2	Inventive steel B	538	672	6.7	247	534	668	245/167/158	190
						529	659	264/286/302	284
						518	642	268/231/188	229
						510	621	154/186/109	150
Comparative example 2	Inventive steel B	510	623	31.8	183	507	612	203/264/197	221
						498	601	174/123/184	160
						488	596	156/89/205	150
						482	584	142/76/87	102
Inventive example 3	Inventive steel C	542	671	7.9	281	538	664	221/265/287	258
						531	659	234/212/264	237
						528	640	198/187/234	206
						513	627	265/188/203	219

(continued)

Classification	Steel type	Base material					After linear heating			
		Yield strength	Tensile strength	Polygonal ferrite (volume %)	Impact toughness average		Yield strength	Tensile strength	Individual impact toughness	Average impact toughness
Comparative example 3	Comparative steel D	587	689	22.1	134		567	678	75/68/32	58
							552	670	15/78/54	49
							542	652	103/28/36	56
							523	641	28/64/28	40
Comparative example 4	Comparative steel E	524	625	35.5	226		514	615	52/105/39	65
							508	611	64/103/154	107
							501	602	51/136/121	103
							492	598	25/38/65	43
Comparative example 5	Comparative steel F	536	628	38.4	193		523	613	156/123/158	146
							512	607	126/154/130	137
							503	598	78/123/162	121
							494	588	58/28/42	43
Comparative example 6	Comparative steel G	506	624	42	133		503	611	120/175/56	117
							493	608	89/45/37	57
							483	594	35/48/56	47
							479	578	59/21/34	38

[0066] In Table 3, the unit of yield strength and tensile strength is MPa, and the unit of impact toughness is J.

[0067] When the mechanical properties of the base material and the mechanical properties after the linear heating are compared in Table 3, Inventive Examples 1 to 3, which all satisfied the alloy composition and manufacturing conditions according to the present disclosure, achieve the target yield strength, tensile strength and impact toughness at -40°C.

[0068] Specifically, Inventive Examples 1 to 3 have the characteristics of yield strength of 500 MPa or greater, tensile strength of 600 MPa or greater, impact toughness of 100 J or greater at -40°C both before and after linear heating.

[0069] Comparative Example 3 uses comparative steel D in which the C content is excessive. It can be seen that strength is significantly higher than a target level but impact toughness is significantly reduced. This is considered to be due to formation of coarse carbide, which causes fracture in the impact test.

[0070] Comparative Example 4 uses comparative steel E in which the Mo component was insufficient and Comparative Example 5 uses comparative steel F in which the Nb component is insufficient. It can be seen that strength is remarkably reduced and impact toughness is lowered. This is because the amount of dissolved Mo and Nb is small and the amount of Mo and Nb is not sufficient to form a precipitate after linear heating. If Mo and Nb are added excessively, toughness may rather be deteriorated due to coarse precipitates, and thus, Mo and Nb must be added in the control range of the present disclosure.

[0071] Comparative Example 6 uses comparative steel G in which Mn and Ni components are insufficient. Since a sufficiently low Ar₃ temperature was not secured, a large amount of polygonal ferrite was formed during rolling at a low temperature, degrading strength and toughness.

[0072] Comparative Example 1 satisfied the alloy composition of the present disclosure, but the rolling temperature exceeded 850°C to lower impact toughness. Comparative Example 3 satisfied the alloy composition of the present disclosure, but the cooling conditions were out of the range of the present disclosure, resulting in an increase in a polygonal ferrite fraction to degrade strength and toughness due to the growth of crystal grain after linear heating.

[0073] While example exemplary embodiments have been shown and described above, it will be apparent to those skilled in the art that modifications and variations could be made without departing from the scope of the present disclosure as defined by the appended claims.

Claims

1. A high-strength structural steel sheet having excellent heat treatment resistance, the high-strength structural steel sheet comprising:

by weight percent (wt%), 0.03 to 0.07% of C, 0.05 to 0.2% of Si, 1.6 to 2.3% of Mn, 0.008% or less of P, 0.002% or less of S, 0.025% or less of Al, 0.1 to 0.4% of Cu, 1.4 to 2.3% of Ni, 0.08 to 0.2% of Mo, 0.01 to 0.025% of Nb, 0.008 to 0.02% of Ti, 0.001 to 0.008% of N, the remainder being Fe, and unavoidable impurities, wherein a microstructure within 10 mm of the surface contains 80% or more of acicular ferrite and 20% or less of polygonal ferrite.

2. The high-strength structural steel sheet of claim 1, wherein a thickness of the steel sheet is 40 mm or less.

3. The high-strength structural steel sheet of claim 1, wherein the steel sheet has yield strength of 500 MPa or greater, tensile strength of 600 MPa or greater, and impact toughness of 100 J or greater at -40°C.

4. The high-strength structural steel sheet of claim 1, wherein after the steel sheet is linearly heated to 600 to 900°C, the steel sheet has yield strength of 500 MPa or greater, tensile strength of 600 MPa or greater, and impact toughness of 100 J or greater at -40°C.

5. A method of manufacturing a high-strength structural steel sheet having excellent heat treatment resistance, the method comprising:

reheating a slab including, by weight percent (wt%), 0.03 to 0.07% of C, 0.05 to 0.2% of Si, 1.6 to 2.3% of Mn, 0.008% or less of P, 0.002% or less of S, 0.025% or less of Al, 0.1 to 0.4% of Cu, 1.4 to 2.3% of Ni, 0.08 to 0.2% of Mo, 0.01 to 0.025% of Nb, 0.008 to 0.02% of Ti, 0.001 to 0.008% of N, the remainder being Fe, and unavoidable impurities;

performing non-recrystallization region rolling on the reheated slab at 750 to 850°C; and

performing cooling to a cooling end temperature of 380 to 440°C at a cooling rate of 10°C/sec. or faster after

the non-recrystallization region rolling.

6. The method of claim 5, further comprising:

5 linearly heating the cooled steel sheet to 600 to 900°C and subsequently performing bending.

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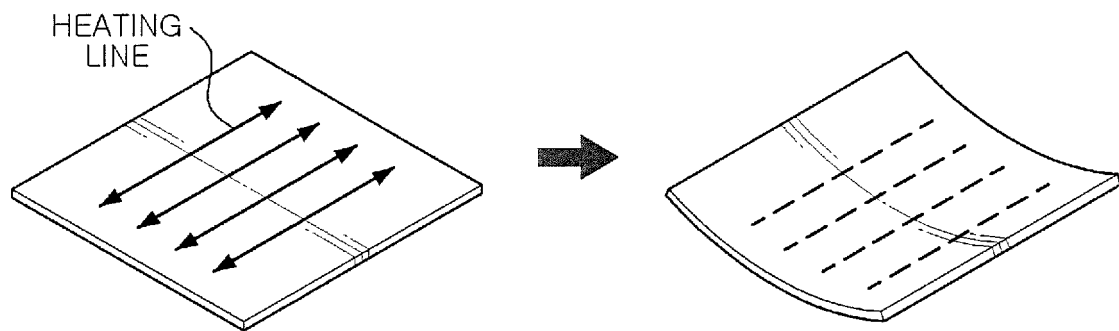
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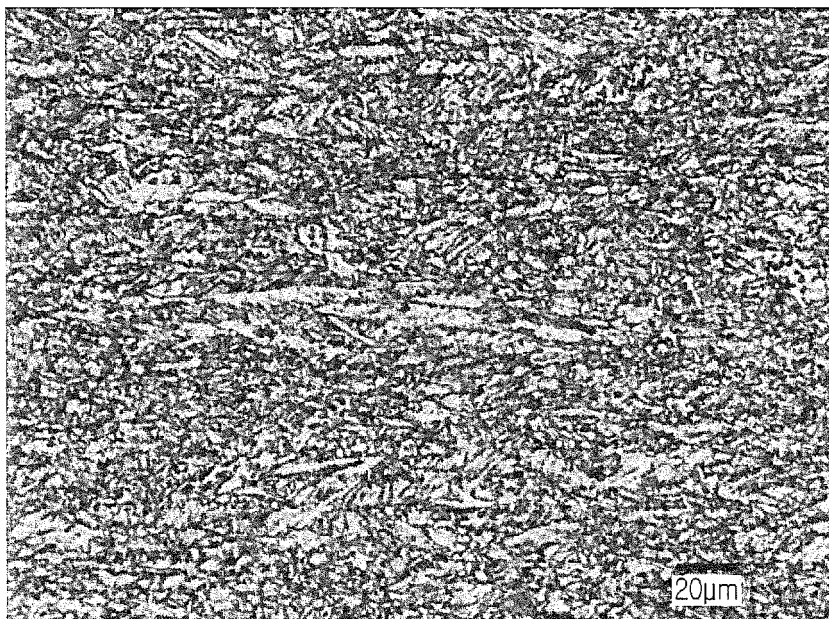
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[FIG. 1]



[FIG. 2]



INTERNATIONAL SEARCH REPORT

International application No.

PCT/KR2016/014964

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/16(2006.01)i, C22C 38/14(2006.01)i, C22C 38/12(2006.01)i, C22C 38/08(2006.01)i, C22C 38/06(2006.01)i, C22C 38/04(2006.01)i, C22C 38/02(2006.01)i, C22C 38/00(2006.01)i, C21D 8/02(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 38/16; C22C 38/06; C21D 8/02; C22C 38/00; C21D 8/00; C22C 38/12; C22C 38/14; C22C 38/08; C22C 38/04; C22C 38/02

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

Korean Utility models and applications for Utility models: IPC as above

Japanese Utility models and applications for Utility models: IPC as above

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

eKOMPASS (KIPO internal) & Keywords: manganese, nickel, copper, molybdenum, niobium, titanium, ferrite, nucleating agent, plate forming

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	JP 63-183123 A (KOBE STEEL LTD.) 28 July 1988 See page 2, tables 1-3 and claim 2.	1-6
A	JP 2007-056348 A (NIPPON STEEL CORP.) 08 March 2007 See paragraph [0124] and claims 1-7.	1-6
A	JP 11-152540 A (SUMITOMO METAL IND. LTD.) 08 June 1999 See paragraphs [0031]-[0034] and claims 1-5.	1-6
A	JP 2010-121149 A (NIPPON STEEL CORP.) 03 June 2010 See paragraphs [0089]-[0092] and claims 1-7.	1-6
A	KR 10-2011-0022308 A (HYUNDAI STEEL COMPANY) 07 March 2011 See paragraphs [0069]-[0079] and claim 1.	1-6

☐ 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

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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

23 MARCH 2017 (23.03.2017)

Date of mailing of the international search report

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.

PCT/KR2016/014964

Patent document cited in search report	Publication date	Patent family member	Publication date
JP 63-183123 A	28/07/1988	NONE	
JP 2007-056348 A	08/03/2007	JP 04751137 B2	17/08/2011
JP 11-152540 A	08/06/1999	NONE	
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KR 10-2011-0022308 A	07/03/2011	NONE	

Form PCT/ISA/210 (patent family annex) (January 2015)