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(54) **ULTRA-HIGH STRENGTH COLD-ROLLED STEEL SHEET HAVING EXCELLENT HOLE-EXPANDABILITY AND METHOD FOR MANUFACTURING SAME**

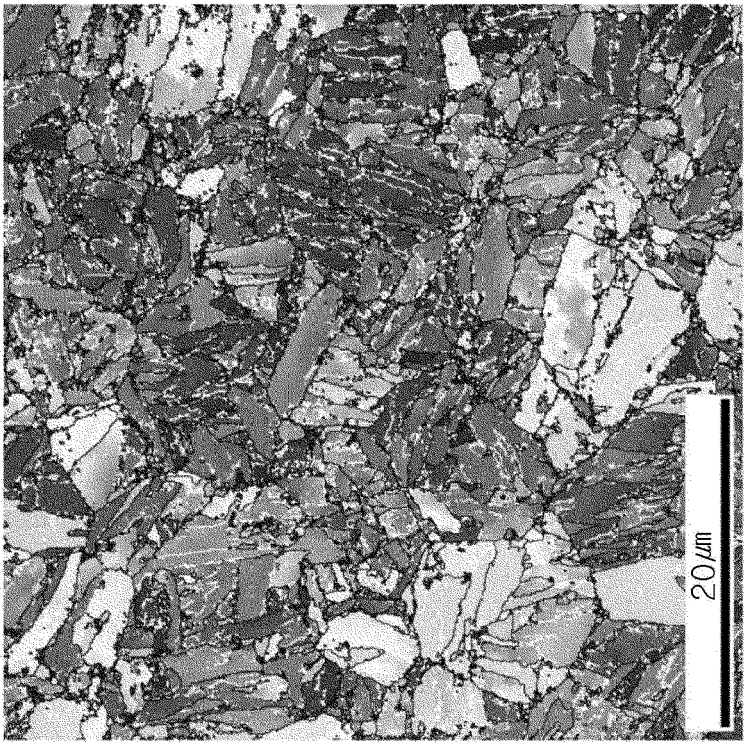
(57) The present invention relates to an ultra-high strength cold-rolled steel sheet having excellent hole expandability and a method for manufacturing same and, more specifically, to an ultra-high strength cold-rolled

steel sheet having excellent hole expandability which can be mainly used for structural members and to ensure safety in the event of an automobile collision, and to a method for manufacturing same.

EP 4 407 062 A1

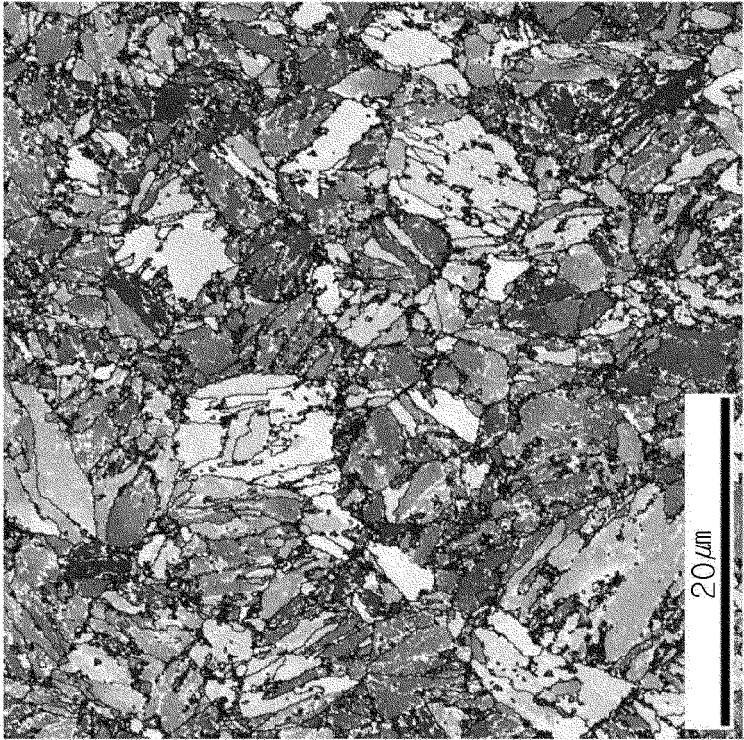
[FIG. 2]

COMPARATIVE EXAMPLE 5



Boundaries: Rotation Angle			
Min	Max	Fraction	
2?	15?	0.350	
15?	65?	0.650	
Number		Length	
44179		3.53mm	
81868		6.55mm	

INVENTIVE EXAMPLE 5



Boundaries: Rotation Angle			
Min	Max	Fraction	
2?	15?	0.321	
15?	65?	0.679	
Number		Length	
54606		4.37mm	
115604		9.25mm	

Description

Technical Field

5 **[0001]** The present invention relates to an ultra-high strength cold-rolled steel sheet having excellent expandability and a method for manufacturing the same and, more specifically, to an ultra-high strength cold-rolled steel sheet having excellent hole expandability which can mainly be used for automobile collision and structural members, and to a method for manufacturing the same.

10 Background Art

[0002] A steel sheet for automobiles should be lighter to preserve the global environment, while the steel sheet for automobiles needs to meet the conflicting goal of securing collision safety for the safety of passengers. To this end, various steel sheets for automobiles such as dual phase (DP) steel, transformation induced plasticity (TRIP) steel, 15 complex (CP) steel, and the like, are being developed. However, tensile strength that can be realized in advanced high strength steel (AHSS) is limited to about 1200Mpa. Accordingly, when manufacturing structural members to secure collision safety, a hot press forming method securing final strength through rapid cooling (water cooling) through direct contact with a die after being formed at high temperature, is in the spotlight, but expansion of application thereof is not significant due to high facility investment costs and high heat treatment and processing costs.

20 **[0003]** Meanwhile, a roll forming method, which is more productive than general press forming and hot press forming, is a method of producing complex shapes through multi-stage roll forming, which is generally applied to forming parts of ultra-high strength materials with low elongation, and the application thereof is also expanding. A steel sheet applied to such a roll forming method is mainly manufactured in a continuous annealing facility equipped with a water cooling facility. However, there is a disadvantage in that shape quality is deteriorated due to deviation in temperatures thereof 25 in a width direction and a length direction during water cooling, so that there is a disadvantage in that workability is deteriorated and material deviation for each location occurs when the roll forming method is applied. Therefore, there is a need to devise an alternative to a rapid cooling method through water cooling.

Summary of Invention

30 Technical Problem

[0004] An aspect of the present disclosure is to provide an ultra-high strength cold-rolled steel sheet having excellent hole expandability and a method for manufacturing the same.

35 Solution to Problem

[0005] According to an aspect of the present disclosure, provided is an ultra-high strength cold-rolled steel sheet having excellent hole expandability, the ultra-high strength cold-rolled steel sheet having excellent hole expandability 40 including by weight: C: 0.2-0.4%, Si: 0.5% or less (excluding 0%), Mn: 1.0-2.0%, P: 0.03% or less (excluding 0%), S: 0.015% or less (excluding 0%), Al: 0.1% or less (excluding 0%), Cr: 0.5% or less (excluding 0%), Mo: less than 0.2% (excluding 0%), Ti: 0.1% or less (excluding 0%), Nb: 0.1% or less (excluding 0%), B: 0.005% or less (excluding 0%), N: 0.01% or less (excluding 0%), with a remainder of Fe and other unavoidable impurities, wherein a microstructure includes a single-phase structure of tempered martensite structure or a mixed structure of martensite and tempered martensite, 45 wherein the microstructure has an F_{HAGB} of 60% or more by area and L_{HAGB} of 8 mm or more, per unit area of $45 \mu\text{m} \times 45 \mu\text{m}$.

[0006] where F_{HAGB} is a fraction of grains with high-hardness angle grain boundaries, and L_{HAGB} is a total length of grain boundaries with high-hardness angle grain boundaries, wherein the high-hardness grain boundaries refer to grain boundaries with a mismatch angle between adjacent grains of 15° or more.

50 **[0007]** According to another aspect of the present disclosure, provided is a method for manufacturing an ultra-high cold-rolled steel sheet having excellent hole expandability, the method for manufacturing an ultra-high cold-rolled steel sheet having excellent hole expandability including operations of: heating a steel slab including by weight: C: 0.2-0.4%, Si: 0.5% or less (excluding 0%), Mn: 1.0-2.0%, P: 0.03% or less (excluding 0%), S: 0.015% or less (excluding 0%), Al: 0.1% or less (excluding 0%), Cr: 0.5% or less (excluding 0%), Mo: less than 0.2% (excluding 0%), Ti: 0.1% or less 55 (excluding 0%), Nb: 0.1% or less (excluding 0%), B: 0.005% or less (excluding 0%), N: 0.01% or less (excluding 0%), with a remainder of Fe and other unavoidable impurities, to a temperature in a range of 1100 to 1300°C; finish hot rolling the heated steel slab at a temperature of A_r3 or higher; coiling the hot-rolled steel sheet at a temperature of 720°C or lower; cold rolling the coiled hot-rolled steel sheet to obtain a cold-rolled steel sheet; performing an annealing heat

treatment on the cold-rolled steel sheet at a temperature in a range of 780 to 900°C; slowly cooling the annealed heat-treated cold-rolled steel sheet to a temperature in a range of 650 to 750°C at a cooling rate of 5°C/s or less; rapidly cooling the slowly-cooled cold-rolled steel sheet to a temperature of 150°C or lower at a cooling rate of 40 °C/s or more; and reheating and performing an overaging heat treatment on the rapidly-cooled cold-rolled steel sheet at a temperature in a range of 180 to 240°C.

Advantageous Effects of Invention

[0008] As set forth above, according to an aspect of the present disclosure, an ultra-high cold-rolled steel sheet having excellent hole expandability and a tensile strength of 1470 MPa or more and a method for manufacturing the same may be provided.

Brief description of drawings

[0009]

FIG. 1 is a photograph of Inventive Example 5 and Comparative Example 5 according to an embodiment of the present disclosure observed with an optical microscope.

FIG. 2 is a photograph of Inventive Example 5 and Comparative Example 5 according to an embodiment of the present disclosure obtained by measuring a microstructure using electron backscattering diffraction attached a scanning electron microscope, and then analyzing high-hardness angle grain boundaries and low-hardness angle grain boundaries.

Best Mode for Invention

[0010] Hereinafter, an ultra-high strength cold-rolled steel sheet having excellent hole expandability according to an embodiment of the present disclosure will be described. First, an alloy composition of the present disclosure will be described. A content of the alloy composition described below refers to by weight unless otherwise stated.

Carbon (C): 0.2 to 0.4%

[0011] Carbon (C) is an element added to secure strength of martensite, and it is preferable that C is preferably added in an amount of 0.2% or more for the above effect. However, if the C content exceeds 0.4%, weldability may be poor. Therefore, it is preferable that the C content is in a range of 0.2 to 0.4%. A lower limit of the C content is more preferably 0.21%, and even more preferably 0.22%. An upper limit of the C content is more preferably 0.3%, even more preferably 0.29%, and most preferably 0.28%.

Silicon (Si): 0.5% or less (excluding 0%)

[0012] Silicon (Si), a ferrite stabilizing element, has a disadvantage of weakening strength by promoting formation of ferrite during slow cooling after annealing in a continuous annealing furnace with a slow cooling section. In addition, there is a risk of causing dent defects due to surface thickening and oxidation due to Si during annealing. Therefore, it is preferable that the Si content is in a range of 0.5% or less. The Si content is more preferably 0.4% or less, and even more preferably 0.3% or less.

Manganese (Mn): 1.0 to 2.0%

[0013] Manganese (Mn) is an element which suppresses ferrite formation and facilitates austenite formation. When the Mn content is less than 1.0%, ferrite is easily formed during slow cooling, while when the Mn content exceeds 2.0%, bending workability, delayed fracture resistance, and weldability may be reduced. Therefore, the Mn content is preferably in a range of 1.0 to 2.0%. A lower limit of the Mn content is more preferably 1.3%, and even more preferably 1.5%.

Phosphorus (P): 0.03% or less (excluding 0%)

[0014] Phosphorus (P) is an impurity element, and if a P content exceeds 0.03%, weldability decreases and a risk of steel brittleness increases, and a possibility of causing dent defects increases, so an upper limit of the P content is limited to be 0.03%. The P content is more preferably 0.025% or less, and even more preferably 0.02% or less.

Sulfur (S): 0.015% or less (excluding 0%)

[0015] Sulfur (S), like P, is an impurity element that impairs ductility and weldability of a steel sheet. If the S content exceeds 0.015%, a possibility of impairing the ductility and weldability of the steel sheet is high, so an upper limit of the S content is preferably limited to be 0.015%. The S content is more preferably 0.01% or less, and even more preferably 0.005% or less.

Aluminum (Al): 0.1% or less (excluding 0%)

[0016] Aluminum (Al) is an element that expands a ferrite transformation section, and when using a continuous annealing process with a slow cooling section as in the present disclosure, there is a disadvantage of promoting ferrite formation, and high-temperature hot rolling properties may be reduced due to AlN formation. Therefore, an upper limit thereof is limited to 0.1%. The Al content is more preferably 0.07% or less, and even more preferably 0.05% or less.

Chromium (Cr): 0.5% or less (excluding 0%)

[0017] Chromium (Cr) is an alloy element that facilitates securing a low-temperature transformation structure by suppressing ferrite transformation, and when using a continuous annealing process with slow cooling as in the present disclosure, there is an advantage of suppressing ferrite formation. However, when the Cr content exceeds 0.5%, delayed fracture resistance may deteriorate, carbides such as CrC, or the like may be formed, to impair hole expandability and bending workability, and costs may increase due to excessive alloy input. Therefore, the Cr content is preferably in a range of 0.5% or less. The Cr content is more preferably 0.4% or less, and even more preferably 0.3% or less.

Molybdenum (Mo): less than 0.2% (excluding 0%)

[0018] Molybdenum (Mo) has an effect of improving quenchability of steel, an effect of generating fine carbides containing Mo, serving as a hydrogen trap site, and an effect of improving delayed fracture resistance by refining martensite. However, when the Mo content is 0.2% or more, phosphatability may be deteriorated, and there is a problem of increased cost, so it is preferable to limit the range. Therefore, the Mo content is preferably in a range of less than 0.2%. A lower limit of the Mo content is more preferably 0.03%, even more preferably 0.05%, and most preferably 0.1%.

Titanium (Ti): 0.1% or less (excluding 0%)

[0019] Titanium (Ti) is a nitride forming element, and is an element scavenging by precipitating N in steel into TiN. When Ti is not added, there is a possibility that cracks may occur during continuous casting due to AlN formation. However, if the Ti content exceeds 0.1%, the strength of martensite may be reduced by additional carbide precipitation in addition to removal of dissolved N, and hole expandability and bending workability may be impaired by the formation of carbides and nitrides such as TiC and TiN. Therefore, the Ti content is preferably in a range of 0.1% or less. The Ti content is more preferably 0.07% or less, and even more preferably 0.05% or less. Meanwhile, for the scavenging effect and suppression of AlN formation, Ti may be added in a chemical equivalent amount of $48/14 \times [N]$ or more.

Niobium (Nb): 0.1% or less (excluding 0%)

[0020] Niobium (Nb) is an element that segregates at austenite grain boundaries and suppresses coarsening of austenite grains during annealing heat treatment. However, when the Nb content exceeds 0.1%, precipitation of carbides and nitrides increases, which reduces workability of a base material, and the cost increases as an alloy input amount becomes excessive. Therefore, the Nb content is preferably in a range of 0.1% or less. The Nb content is more preferably 0.08% or less, and even more preferably 0.06% or less.

Boron (B): 0.005% or less (excluding 0%)

[0021] Boron (B) is an element that suppresses formation of ferrite, and accordingly, in the present disclosure, B has an advantage of suppressing the formation of ferrite during cooling after annealing. However, if the B content exceeds 0.005%, ductility may decrease. Therefore, the B content is preferably in a range of 0.005% or less. The B content is more preferably 0.004% or less, and most preferably 0.003% or less.

Nitrogen (N): 0.01% or less (excluding 0%)

[0022] Nitrogen (N) is an element which is an impurity element, and if a N content exceeds 0.01%, it greatly increases a risk of cracks occurring during continuous casting due to AlN formation, or the like, so it is preferable to limit an upper limit thereof to 0.01%. The N content is more preferably 0.008% or less, and most preferably 0.006% or less.

[0023] The remaining component of the present disclosure is iron (Fe). However, since in the common manufacturing process, unintended impurities may be inevitably incorporated from raw materials or the surrounding environment, the component may not be excluded. Since these impurities are known to any person skilled in the common manufacturing process, the entire contents thereof are not particularly mentioned in the present specification.

[0024] Meanwhile, the cold-rolled steel sheet of the present disclosure may further include one or more of Cu: 0.5% or less and Ni: 0.5% or less.

Copper (Cu): 0.5% or less

[0025] Copper (Cu) improves corrosion resistance, and has an effect of suppressing hydrogen intrusion by being coated on a surface of the steel sheet. However, if the Cu content exceeds 0.5%, it may cause surface defects. Therefore, the Cu content is preferably in a range of 0.5% or less. The Cu content is more preferably 0.4% or less, and even more preferably 0.3% or less.

Nickel (Ni): 0.5% or less

[0026] Nickel (Ni), like Cu, also improves corrosion resistance, and serves to reduce surface defects that tend to occur due to addition of Cu. However, when the Ni content exceeds 0.5%, scale generation within a heating furnace may become uneven, which may cause surface defects. Therefore, the Ni content is preferably in a range of 0.5% or less. The Ni content is more preferably 0.4% or less, and even more preferably 0.3% or less.

[0027] In addition, the cold-rolled steel sheet of the present disclosure may further include Sb: 0.05% or less.

Sb: 0.05% or less

[0028] Sb is an element that contributes to high strength and improved delayed fracture resistance by suppressing oxidation and nitridation of a surface layer. However, if the Sb content exceeds 0.05%, castability may deteriorate and delayed fracture resistance may deteriorate. Therefore, the Sb content is preferably in a range of 0.05% or less. The Sb content is more preferably 0.04% or less, and even more preferably 0.03% or less.

[0029] Hereinafter, a microstructure of an ultra-high-strength cold-rolled steel sheet with excellent hole expandability according to an embodiment of the present disclosure will be described.

[0030] The microstructure of the cold-rolled steel sheet of the present disclosure preferably includes a single-phase structure of tempered martensite or a mixed structure of martensite and tempered martensite. As described above, since the microstructure includes a tempered martensite a single-phase structure of tempered martensite or a mixed structure of martensite and tempered martensite, an effect of high yield strength and excellent hole expandability may be obtained. It is more preferable that the microstructure of the present disclosure is a single-phase structure of tempered martensite, but since tempering does not occur completely during the manufacturing process, the microstructure of the present disclosure may include a mixed structure of martensite and tempered martensite. In the present disclosure, a fraction of the mixed structure of martensite and tempered martensite is not particularly limited, but for example, the mixed structure may have a fraction of tempered martensite of 80% or more by area, and more preferably 90% or more by area.

[0031] In addition, the microstructure of the present disclosure preferably has an F_{HAGB} of 60% or more by area and an L_{HAGB} of 8 mm or more per unit area of $45 \mu\text{m} \times 45 \mu\text{m}$. In this case, the F_{HAGB} is a fraction of grains with high-hardness angle grain boundaries, the L_{HAGB} is a total length of grain boundaries with high-hardness angle grain boundaries, and the high-hardness grain boundaries refer to grain boundaries having a mismatch angle between adjacent grains of 15° or more. When the F_{HAGB} is less than 60% by area or the L_{HAGB} is less than 8 mm, there is a disadvantage in that hole expandability is inferior.

[0032] Meanwhile, the cold-rolled steel sheet of the present disclosure may have an average particle size of prior austenite of $6 \mu\text{m}$ or less. When the average particle size of the prior austenite exceeds $6 \mu\text{m}$, there may be a disadvantage in that hole expandability and bending workability are inferior.

[0033] The cold-rolled steel sheet of the present disclosure provided as described above may have a tensile strength (TS) of 1470 MPa or more, and a value of a tensile strength (TS) (MPa) $\times$ a hole expansion ratio (HER) (%) of 73,500 MPa-% or more, so that ultra-high strength and excellent hole expandability may be secured at the same time.

[0034] Hereinafter, a method for manufacturing an ultra-high-strength cold-rolled steel sheet with excellent hole expandability according to an embodiment of the present disclosure will be described.

[0035] First, a slab satisfying the above-described alloy composition is heated to a temperature in a range of 1100 to 1300°C. When the heating temperature is less than 1100°C, a problem in that a hot rolling load increases rapidly may occur, and when the heating temperature exceeds 1300°C, an amount of surface scales increases, which may result in material loss. Therefore, the heating temperature of the steel slab is preferably in a range of 1100 to 1300°C.

[0036] Thereafter, the heated steel slab is subjected to finish hot rolling at a temperature of Ar₃ or higher to obtain a hot-rolled steel sheet. The Ar₃ temperature is a temperature at which ferrite begins to appear when austenite is cooled. When the finish rolling temperature is less than Ar₃, rolling in a dual phase of ferrite and austenite or ferrite region is performed to create a mixed structure, and there may be concerns about malfunction of a hot rolling equipment due to fluctuations in the hot rolling load. The finish hot rolling temperature is more preferably 800°C or higher, even more preferably 850°C or higher, and most preferably 900°C or higher.

[0037] Thereafter, the hot rolled steel sheet is coiled at a temperature of 720°C or lower. When the coiling temperature exceeds 720°C, an oxide film may be excessively generated on a surface of the steel sheet, which may cause defects. As the coiling temperature is lowered, the strength of the hot-rolled steel sheet increases, which may cause a disadvantage of increasing a rolling load of cold rolling, which is a post-process, but since this is not a factor that makes actual production impossible, in the present disclosure, a lower limit of the coiling temperature is not specifically limited. The coiling temperature is more preferably 700°C or lower, even more preferably 680°C or lower, and most preferably 650°C or lower.

[0038] Thereafter, the coiled hot-rolled steel sheet is cold-rolled to obtain a cold-rolled steel sheet. In the present disclosure, there is no particular limitation on the cold rolling process, and all processes commonly used in the technical field may be used. Meanwhile, a pickling process may be further performed prior to the cold rolling process.

[0039] Thereafter, the cold-rolled steel sheet is subjected to an annealing heat treatment at a temperature in a range of 780 to 900°C. When the annealing heat treatment temperature is less than 780°C, the strength may be reduced due to formation of a large amount of ferrite. In addition, when connected with other steel materials annealed at a temperature of 800°C or higher, material deviation may occur due to temperature gradients at the top and end of the steel material of the present disclosure. On the other hand, when the annealing heat treatment temperature exceeds 900°C, durability of a continuous annealing furnace may deteriorate, resulting in a difficulty in product production. Therefore, the annealing heat treatment temperature is preferably in a range of 780 to 900°C. A lower limit of the annealing heat treatment temperature is more preferably 800°C, even more preferably 820°C, and most preferably 840°C. An upper limit of the annealing heat treatment temperature is more preferably 880°C, and even more preferably 860°C.

[0040] Thereafter, the annealed heat-treated cold-rolled steel sheet is slowly cooled to a temperature in a range of 650 to 750°C at a cooling rate of 5°C/s or less. In general, a continuous annealing furnace has a slow cooling section after annealing heat treatment. That is, after the annealing heat treatment process described above, slow cooling is performed for a certain period. Typically, in the case of a continuous annealing furnace including a slow cooling section, there is a slow cooling section of 100 to 200 m after annealing, and a soft phase such as ferrite is formed by slow cooling after annealing at a high temperature, resulting in a disadvantage in that it is difficult to manufacture ultra-high steel. For example, when there is a slow cooling section of 160 m, when a sheet-passing speed of steel sheet is 160 m per minute, a time maintained in the slow cooling section is 60 seconds, and when an annealing temperature is 830°C and a final temperature of the slow cooling section is 60 seconds. When the temperature is 650°C, a cooling rate in the slow cooling section is very low at 3°C/s. Thereby, a possibility of generating a soft phase such as ferrite is very high. Meanwhile, in order to increase the cooling rate to be higher than 5°C/sec. during slow cooling after the annealing, an additional cooling device should be introduced, which may cause problems such as manufacturing costs, equipment replacement, or the like.

[0041] Accordingly, in the present disclosure, the slowly-cooled cold-rolled steel sheet is rapidly cooled to a temperature of 150°C or lower at a cooling rate of 40°C/s or more. Through the rapid cooling process, the microstructure may be transformed into martensite. When the rapid cooling rate is less than 40°C/s or a rapid cooling end temperature exceeds 150°C, martensite transformation may not be sufficiently achieved, and it may be difficult to secure the microstructure to be intended to be obtained in the present disclosure. The rapid cooling rate is more preferably 50°C/s or more, more preferably 60°C/s or more, and most preferably 70°C/s or more. The rapid cooling end temperature is more preferably 140°C or lower, and even more preferably 130°C or lower.

[0042] Thereafter, the rapidly-cooled cold-rolled steel sheet is reheated and is subjected to an overaging heat treatment at a temperature in a range of 180 to 240°C. Through the reheating and overaging heat treatment, martensite obtained through the above-described rapid cooling process may be transformed into tempered martensite. When the reheating and overaging heat treatment temperature is less than 180°C, tempering is not sufficiently performed, resulting in a disadvantage in that low yield strength is low and insufficient toughness may not be achieved, and when the reheating and overaging heat treatment temperature exceeds 240°C, a large amount of carbides are precipitated and coarsened, resulting in a disadvantage in that bending workability may be inferior. A lower limit of the reheating and overaging heat treatment temperature is more preferably 190°C, and even more preferably 200°C. An upper limit of the reheating and overaging heat treatment temperature is more preferably 230°C, and even more preferably 220°C. Meanwhile, the overaging heat treatment may be performed for more than 400 seconds. When the overaging heat treatment time is less than 400 seconds, there is a disadvantage in that the yield strength is low because tempering is not sufficiently

achieved. Meanwhile, in the present disclosure, the overaging heat treatment time is not specifically limited, but it is difficult to exceed 1000 seconds due to the characteristics of continuous annealing equipment. The lower limit of the overaging heat treatment time is more preferably 500 seconds, and even more preferably 600 seconds.

Mode for Invention

[0043] Hereinafter, the present disclosure will be specifically described through the following Examples. However, it should be noted that the following examples are only for describing the present disclosure by illustration, and not intended to limit the right scope of the present disclosure. The reason is that the right scope of the present disclosure is determined by the matters described in the claims and reasonably inferred therefrom.

(Example 1)

[0044] Molten steel having the alloy composition shown in Table 1 below was cast into an ingot and then sized and rolled to prepare a steel slab. The steel slab was heated to a temperature of 1200°C, maintained for 1 hour, and then was subjected to finish hot rolling at a temperature of 900°C, charged into a furnace, which is pre-heated to a temperature of 550°C, maintained for 1 hour, and then furnace-cooled to simulate hot rolling. After pickling the hot-rolled steel sheet prepared as described above, cold rolling was performed at a cold rolling reduction rate of 50%, and then was subjected to an annealing heat treatment, slow cooling, rapid cooling, reheating, and overaging heat treatment under the conditions shown in Table 2 below to form a cold-rolled steel sheet.

[0045] A microstructure and mechanical properties of the cold-rolled steel sheet prepared as described above were measured, which were shown in Table 3 below.

[0046] In this case, the microstructure was measured using an optical microscope, and an average grain size of prior austenite was measured using [Relational Expression 1] $F_m = (F_k \times 10^6) / ((0.67n + z) \times V^2)$. In Relational Expression 1, F_m is an average grain size of prior austenite, F_k is a total area of an image of the microstructure, Z is the number of grains inside a circle, n is the number of grains spanning the circle, and V is magnification when measuring the microstructure. When measuring the average grain size of prior austenite, a circle with a diameter of 140 μm was drawn on the image of the microstructure measured with an optical microscope at 1000 $\times$ magnification.

[0047] In addition, F_{HAGB} and L_{HAGB} were obtained by measuring a microstructure within a measurement area of 45 $\mu\text{m} \times 45 \mu\text{m}$ and a measurement interval of 0.75 μm using electron backscattering diffraction (EBSD), and then analyzing the same based on a threshold of 15° using TSL-OIM software.

[0048] A tensile strength (TS) and yield strength (YS) were measured by collecting a JIS No. 5 tensile test sample in a direction perpendicular to a rolling direction and then performing a tensile test at a strain rate of 0.01/s.

[0049] A hole expansion ratio (HER) was measured according to ISO 16630 standards. A sample dimension was 120mm $\times$ 120mm, and an initial hole diameter was 10mm based on a clearance standard of 12%. A punching holding load was 20 tons, and a test speed was 12 mm/min.

[0050] For R/t (bending characteristics), the cold-rolled steel sheet was processed into a sample having a width of 100mm $\times$ a length of 30 mm, then was subjected to a 90° bending test, and then a minimum bending radius R at which no cracks occur was divided by a thickness (t) of a test sample by checking cracks in a bending portion using a microscope, to obtain an R/t value.

[Table 1]

Steel type	Alloy composition (weight %)												
	C	Si	Mn	P	S	Al	Cr	Mo	Ti	Nb	Cu	Ni	B
No.													
Comparative Steel 1	0.15	0.5	2.5	0.01	0.002	0.012	0.1	0.01	0.025	0.01	0.01	0.01	0.001
Comparative Steel 2	0.18	0.1	3.5	0.01	0.002	0.025	0.1	0.01	0.025	0.04	0.01	0.01	0.002
Comparative Steel 3	0.22	0.01	0.9	0.01	0.002	0.025	0.1	0.01	0.025	0.01	0.01	0.01	0.002
Comparative Steel 4	0.23	0.5	2.5	0.01	0.002	0.025	0.01	0.05	0.025	0.04	0.01	0.01	0.002
Inventive Steel 1	0.24	0.5	2.0	0.01	0.002	0.025	0.3	0.05	0.025	0.04	0.01	0.01	0.002
Inventive Steel 2	0.24	0.1	1.9	0.01	0.002	0.025	0.3	0.05	0.025	0.04	0.01	0.01	0.002
Inventive Steel 3	0.25	0.1	1.7	0.01	0.002	0.025	0.1	0.10	0.025	0.01	0.01	0.01	0.002
Inventive Steel 4	0.24	0.2	1.9	0.01	0.002	0.025	0.1	0.05	0.025	0.02	0.01	0.01	0.002

[Table 2]

Division	Steel type No.	Annealing temperature (°C)	Slowcooling rate (°C/s)	Slowcooling end temperature (°C)	Rapidcooling rate (°C/s)	Rapidcooling end temperature (°C)	Reheating temperature (°C)	Overaging heat treatment temperature (°C)	Overaging heat treatment time (sec.)
Comparative Example 1	Comparative Steel 1	850	2.1	700	39	200	230	230	649
Comparative Example 2	Comparative Steel 1	870	2.4	700	43	150	230	230	649
Comparative Example 3	Comparative Steel 1	870	2.4	700	47	100	230	230	649
Comparative Example 4	Comparative Steel 2	850	2.9	650	39	150	230	230	649
Comparative Example 5	Comparative Steel 3	850	2.1	700	43	150	230	230	649
Comparative Example 6	Comparative Steel 4	850	2.1	700	39	200	230	230	649
Comparative Example 7	Comparative Steel 4	850	2.1	700	43	150	230	230	649
Comparative Example 8	Comparative Steel 4	850	2.1	700	47	100	230	230	649
Comparative Example 9	Inventive Steel 1	850	2.1	700	39	200	230	230	649
Inventive Example 1	Inventive Steel 1	850	2.1	700	43	150	230	230	649

(continued)

Division	Steel type No.	Annealing temperature (°C)	Slowcooling rate (°C/s)	Slowcooling end temperature (°C)	Rapidcooling rate (°C/s)	Rapidcooling end temperature (°C)	Reheating temperature (°C)	Overaging heat treatment temperature (°C)	Overaging heat treatment time (sec.)
Inventive Example 2	Inventive Steel 1	850	2.1	700	47	100	230	230	649
Comparative Example 10	Inventive Steel 2	850	2.1	700	39	200	230	230	649
Inventive Example 3	Inventive Steel 2	850	2.1	700	43	150	230	230	649
Inventive Example 4	Inventive Steel 2	850	2.1	700	47	100	230	230	649
Comparative Example 11	Inventive Steel 3	850	2.1	700	39	200	230	230	649
Inventive Example 5	Inventive Steel 3	850	2.1	700	43	150	230	230	649
Inventive Example 6	Inventive Steel 3	850	2.1	700	47	100	230	230	649
Comparative Example 12	Inventive Steel 4	850	2.1	700	39	200	230	230	649
Inventive Example 7	Inventive Steel 4	850	2.1	700	43	150	230	230	649
Inventive Example 8	Inventive Steel 4	850	2.1	700	47	100	230	230	649

[Table 3]

Divisio n	Microst ructure	F _{HAGB} (area %)	L _{HAGB} (mm)	F _m (μm)	YS (MPa)	TS (MPa)	Yield ratio (YS/TS)	HER (%)	R/t	TS -HER (MPa ·%)
Compara tive Example 1	TM+FM+B	62	8.6	5.3	1065	1270	0.84	49	3.8	62230
Compara tive Example 2	TM	65	8.2	5.9	1168	1324	0.88	52	4.2	68848
Compara tive Example 3	TM	61	9.0	5.7	1175	1357	0.87	51	4.2	69207
Compara tive Example 4	TM+F+B	63	5.4	5.3	1153	1598	0.72	42	4.5	67116
Compara tive Example 5	TM	65	6.5	6.8	1032	1426	0.72	28	3.5	39928
Compara tive Example 6	TM+FM+B	60	8.9	5.3	1016	1553	0.65	38	3.7	59014
Compara tive Example 7	TM	64	9.4	5.4	1236	1587	0.78	32	4.5	50784
Compara tive Example 8	TM	61	9.3	5.8	1247	1579	0.79	35	4.3	55265
Compara tive Example 9	TM+FM+B	63	8.9	5.2	1105	1549	0.71	47	3.4	72803
Inventi veExample 1	TM	64	9.4	4.9	1223	1543	0.79	55	3.8	84865
Inventi veExamp le2	TM	62	9.3	5.5	1246	1538	0.81	58	3.8	89204
Compara tive Example 10	TM+FM+B	63	9.1	5.2	1095	1524	0.72	48	4.0	73152
Inventi veExamp le3	TM+FM	65	9.2	5.3	1245	1543	0.81	53	4.0	81779
Inventi veExamp le4	TM+FM	62	9.4	5.1	1256	1552	0.81	59	4.0	91568

(continued)

Divisio n	Microst ructure	F_{HAGB} (area %)	L_{HAGB} (mm)	F_m (μm)	YS (MPa)	TS (MPa)	Yield ratio (YS/TS)	HER (%)	R/t	TS -HER (MPa ·%)
Compara tive Example 11	TM+FM+B	61	7.6	7.2	1125	1554	0.72	46	3.5	71484
Inventi veExamp le5	TM	64	9.1	5.6	1255	1543	0.81	53	4.0	81779
Inventi veExamp le6	TM	63	9.4	5.4	1261	1552	0.81	59	4.0	91568
Compara tive Example 12	TM+FM+B	61	9.2	5.4	1145	1561	0.73	42	3.5	65562
Inventi veExamp le7	TM	68	9.3	5.2	1255	1545	0.81	57	4.0	88065
Inventi veExamp le8	TM	63	8.9	5.2	1263	1552	0.81	61	3.5	94672
TM: tempered martensite, FM: fresh martensite, B: bainite, F: ferrite										

[0051] As can be seen from Tables 1 to 3, in Inventive Examples 1 to 8 satisfying the conditions of the present disclosure, it can be seen that an excellent tensile strength and a value of tensile strength $\times$ hole expansion ratio are illustrated. On the other hand, in Comparative Examples 1 to 12 not satisfying the conditions of the present disclosure, it can be seen that the tensile strength or the value of tensile strength $\times$ hole expansion ratio desired by the present disclosure may not be secured.

[0052] FIG. 1 is a photograph of Inventive Example 5 and Comparative Example 5 observed with an optical microscope. As can be seen from FIG. 1, in Inventive Example 5, an average grain size of prior austenite is fine, while in Comparative Example 5, an average grain size of prior austenite is relatively large.

[0053] FIG. 2 is a photograph illustrating a microstructure of Inventive Example 5 and Comparative Example 5 using electron backscattering diffraction attached to a scanning electron microscope and then analyzing high-hardness angle grain boundaries and low-hardness angle grain boundaries. As can be seen from FIG. 2, in Inventive Example 5, it can be seen that F_{HAGB} and L_{HAGB} had high values, while in Comparative Example 5, F_{HAGB} and L_{HAGB} had low values.

[0054] While example embodiments have been illustrated 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. An ultra-high strength cold-rolled steel sheet having excellent hole expandability, comprising by weight:

C: 0.2-0.4%, Si: 0.5% or less (excluding 0%), Mn: 1.0-2.0%, P: 0.03% or less (excluding 0%), S: 0.015% or less (excluding 0%), Al: 0.1% or less (excluding 0%), Cr: 0.5% or less (excluding 0%), Mo: less than 0.2% (excluding 0%), Ti: 0.1% or less (excluding 0%), Nb: 0.1% or less (excluding 0%), B: 0.005% or less (excluding 0%), N: 0.01% or less (excluding 0%), with a remainder of Fe and other unavoidable impurities, wherein a microstructure comprises a single-phase structure of tempered martensite or a mixed structure of martensite and tempered martensite, wherein the microstructure has an F_{HAGB} of 60% or more by area and L_{HAGB} of 8 mm or more, per unit area of

45 μm $\times$ 45 μm .

where F_{HAGB} is a fraction of grains with high-hardness angle grain boundaries, and L_{HAGB} is a total length of grain boundaries with high-hardness angle grain boundaries, wherein the high-hardness grain boundaries refer to grain boundaries with a mismatch angle between adjacent grains of 15° or more.

2. The ultra-high strength cold-rolled steel sheet having excellent hole expandability of claim 1, wherein the cold-rolled steel sheet further comprises one or more of Cu: 0.5% or less and Ni: 0.5% or less.

3. The ultra-high strength cold-rolled steel sheet having excellent hole expandability of claim 1, wherein the cold-rolled steel sheet further comprises Sb: 0.05% or less.

4. The ultra-high strength cold-rolled steel sheet having excellent hole expandability of claim 1, wherein the cold-rolled steel sheet has an average particle size of prior austenite of 6 μm or less.

5. The ultra-high strength cold-rolled steel sheet having excellent hole expandability of claim 1, wherein the cold-rolled steel sheet has a tensile strength (TS) of 1470MPa or more and a value of tensile strength (TS) (MPa) and hole expansion rate (HER)(%) of 73500MPa% or more.

6. A method for manufacturing an ultra-high cold-rolled steel sheet having excellent hole expandability, comprising operations of:

heating a steel slab including by weight: C: 0.2-0.4%, Si: 0.5% or less (excluding 0%), Mn: 1.0-2.0%, P: 0.03% or less (excluding 0%), S: 0.015% or less (0%) (excluding 0%), Al: 0.1% or less (excluding 0%), Cr: 0.5% or less (excluding 0%), Mo: less than 0.2% (excluding 0%), Ti: 0.1% or less (excluding 0%), Nb: 0.1% or less (excluding 0%), B: 0.005% or less (excluding 0%), N: 0.01% or less (excluding 0%), with a remainder of Fe and other unavoidable impurities, to a temperature in a range of 1100 to 1300°C;

finish hot rolling the heated steel slab at a temperature of Ar3 or higher;

coiling the hot-rolled steel sheet at a temperature of 720°C or lower;

cold rolling the coiled hot-rolled steel sheet to obtain a cold-rolled steel sheet;

performing an annealing heat treatment on the cold-rolled steel sheet at a temperature in a range of 780 to 900°C;

slowly cooling the annealed heat-treated cold-rolled steel sheet to a temperature in a range of 650 to 750°C at a cooling rate of 5°C/s or less;

rapidly cooling the slowly-cooled cold-rolled steel sheet to a temperature of 150°C or lower at a cooling rate of 40 °C/s or more; and

reheating and performing an overaging heat treatment on the rapidly-cooled cold-rolled steel sheet at a temperature in a range of 180 to 240°C.

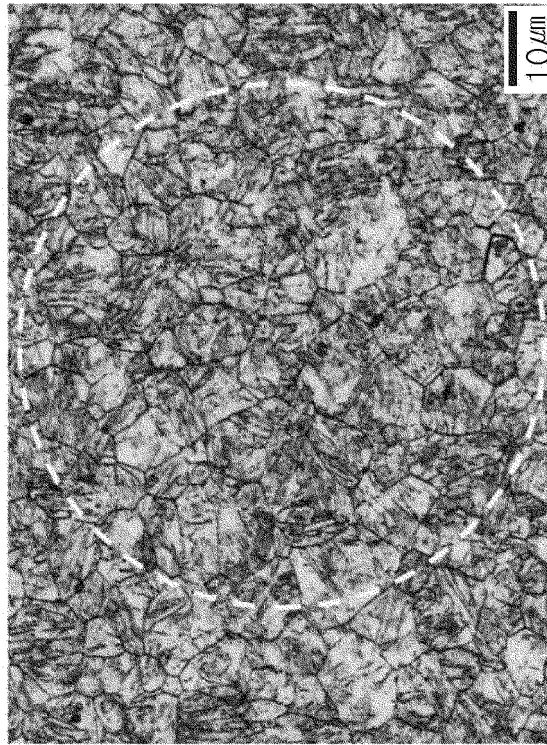
7. The method for manufacturing an ultra-high cold-rolled steel sheet having excellent hole expandability of claim 6, wherein the steel slab further comprises one or more of Cu: 0.5% or less and Ni: 0.5% or less.

8. The method for manufacturing an ultra-high cold-rolled steel sheet having excellent hole expandability of claim 6, wherein the steel slab further comprises, Sb: 0.05% or less.

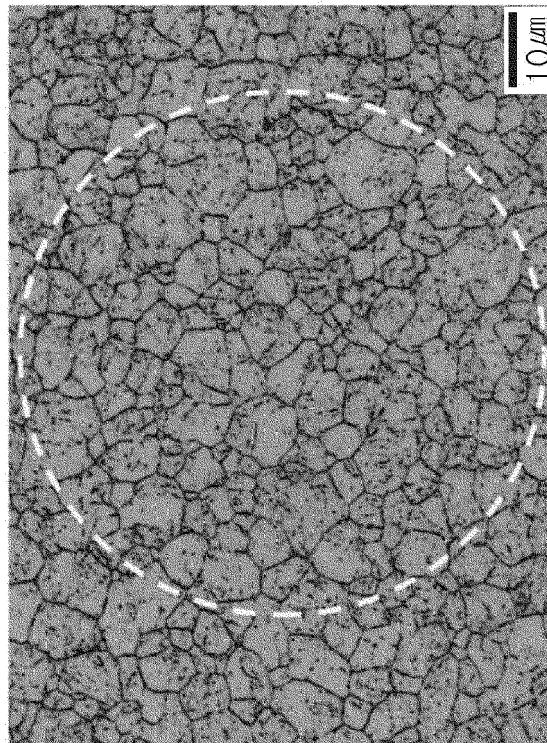
9. The method for manufacturing an ultra-high cold-rolled steel sheet having excellent hole expandability of claim 6, wherein the overaging heat treatment is performed for more than 400 seconds.

[FIG. 1]

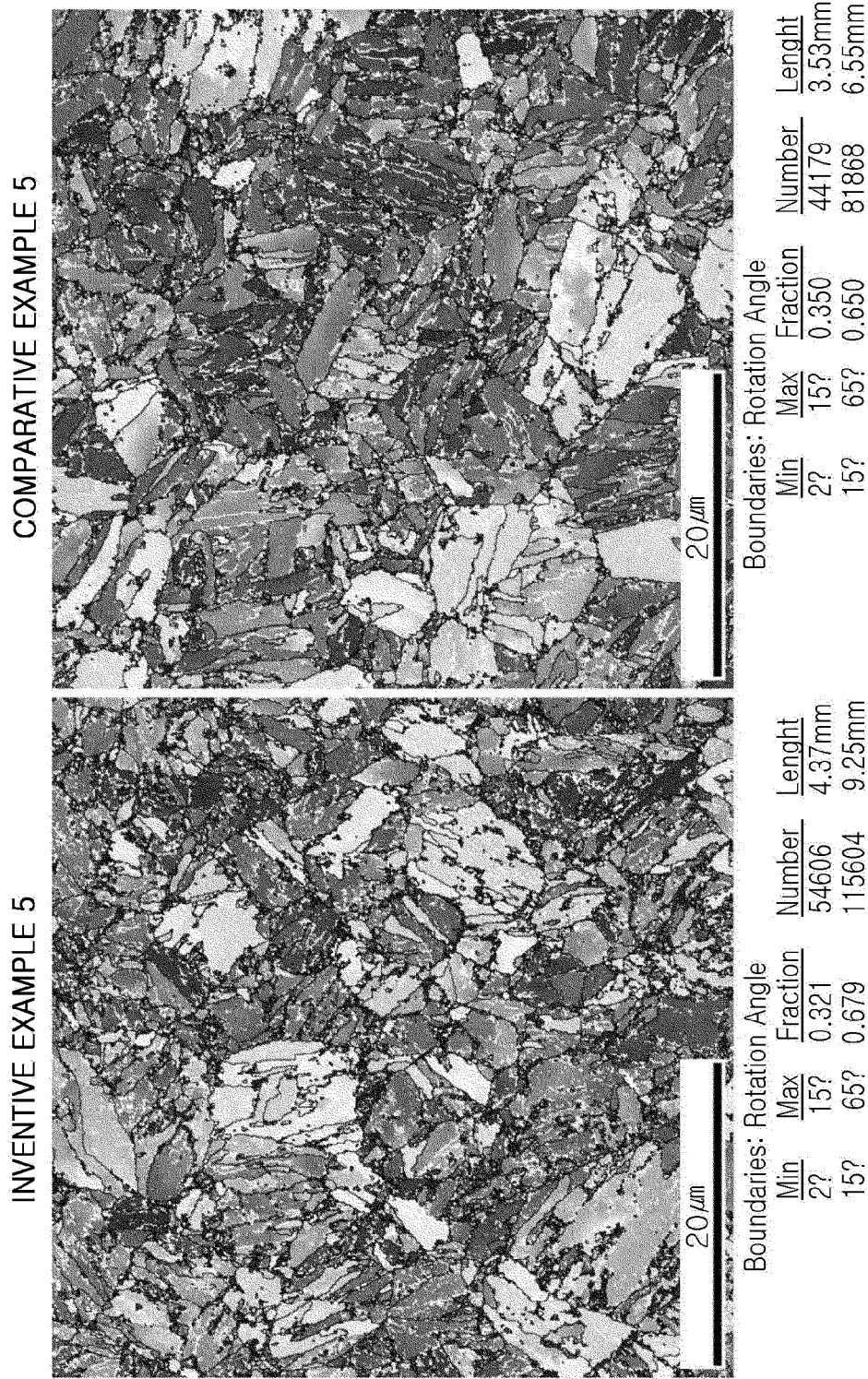
COMPARATIVE EXAMPLE 5



INVENTIVE EXAMPLE 5



[FIG. 2]



INTERNATIONAL SEARCH REPORT

International application No.

PCT/KR2022/014244

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/38(2006.01)i; C22C 38/22(2006.01)i; C22C 38/28(2006.01)i; C22C 38/26(2006.01)i; C22C 38/32(2006.01)i;
C22C 38/00(2006.01)i; C22C 38/58(2006.01)i; C22C 38/42(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 38/38(2006.01); C21D 1/18(2006.01); C21D 8/02(2006.01); C21D 9/46(2006.01); C22C 38/00(2006.01);
C22C 38/02(2006.01); C22C 38/04(2006.01); C22C 38/60(2006.01)

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: 마르텐사이트(martensite), 템퍼링(tempering), 냉각(cooling), 냉연(cold rolled), 티
타늄(titanium), 니오븀(niobium), 보론(boron), 소둔(annealing)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	WO 2021-070951 A1 (NIPPON STEEL CORPORATION et al.) 15 April 2021 (2021-04-15) See paragraphs [0044], [0054] and [0056] and claims 1 and 7.	1-9
A	CN 113166828 A (ARCELORMITTAL) 23 July 2021 (2021-07-23) See paragraphs [0063]-[0065] and claim 1.	1-9
A	CN 111748745 A (BAOSHAN IRON & STEEL CO., LTD.) 09 October 2020 (2020-10-09) See claims 1-4 and 6.	1-9
A	KR 10-2020-0121872 A (NIPPON STEEL CORPORATION) 26 October 2020 (2020-10-26) See paragraph [0094] and claim 1.	1-9
A	KR 10-2020-0062428 A (HYUNDAI STEEL COMPANY) 04 June 2020 (2020-06-04) See claims 1-2.	1-9

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

* Special categories of cited documents:	"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
"A" document defining the general state of the art which is not considered to be of particular relevance	"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
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"E" earlier application or patent but published on or after the international filing date	"&" document member of the same patent family
"L" document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)	
"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	

Date of the actual completion of the international search 28 December 2022	Date of mailing of the international search report 28 December 2022
Name and mailing address of the ISA/KR Korean Intellectual Property Office Government Complex-Daejeon Building 4, 189 Cheongsaro, Seo-gu, Daejeon 35208 Facsimile No. +82-42-481-8578	Authorized officer Telephone No.

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.

PCT/KR2022/014244

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
WO 2021-070951 A1	15 April 2021	CN 114867880 A	05 August 2022
		EP 4043595 A1	17 August 2022
		KR 10-2022-0060552 A	11 May 2022
		US 2022-0333221 A1	20 October 2022
CN 113166828 A	23 July 2021	BR 112021006139 A2	29 June 2021
		CA 3115028 A1	25 June 2020
		EP 3899067 A1	27 October 2021
		JP 2022-510873 A	28 January 2022
		KR 10-2021-0072070 A	16 June 2021
		US 2022-0017985 A1	20 January 2022
		WO 2020-128574 A1	25 June 2020
		WO 2020-128811 A1	25 June 2020
		ZA 202101976 B	23 February 2022
CN 111748745 A	09 October 2020	CN 111748745 B	18 March 2022
KR 10-2020-0121872 A	26 October 2020	CN 111655885 A	11 September 2020
		CN 111655885 B	19 November 2021
		EP 3778951 A1	17 February 2021
		EP 3778951 A4	27 October 2021
		JP 6477978 B1	06 March 2019
		KR 10-2460598 B1	31 October 2022
		US 11180837 B2	23 November 2021
		US 2021-010118 A1	14 January 2021
		WO 2019-186930 A1	03 October 2019
KR 10-2020-0062428 A	04 June 2020	None	

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