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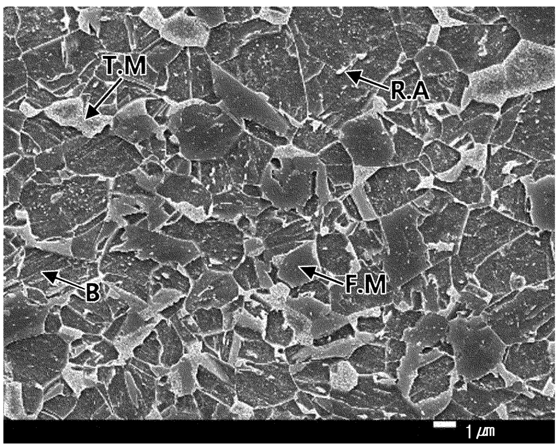
(54) ULTRA-HIGH-STRENGTH COLD-ROLLED STEEL SHEET HAVING EXCELLENT YIELD STRENGTH AND BENDING PROPERTIES AND METHOD FOR MANUFACTURING SAME

(57) The present invention relates to an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties and a method for manufacturing same and, more particularly, to an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, which can be used as a structural component for automobiles such as a member, a seat rail, and a pillar, and a method for manufacturing same.

[Fig. 1]



Description

Technical Field

[0001] The present disclosure relates to an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, and a method for manufacturing the same, and more particularly, to an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, which can be used as a structural member for automobiles such as a member, a seat rail, a pillar, and the like, and a method for manufacturing the same.

Background Art

[0002] Recently, as safety regulations for passengers and pedestrians have been strengthened, as provision of safety devices is mandatory, and there is a problem in that a weight of a vehicle body increases, as opposed to weight reductions improving fuel efficiency of the vehicle. Consumers are increasingly interested in hybrid and electric vehicles that are eco-friendly and have high fuel efficiency. In order to produce such eco-friendly and safe vehicles, it is necessary to reduce the weight of structures of the vehicle body and to secure stability of a material of the vehicle body. However, a hybrid vehicle has been provided with various devices such as an electric engine, an electric battery, and a secondary fuel storage tank as well as a conventional gasoline engine. In addition, as driver convenience facilities are continuously added, the weight of the vehicle body is increasing. Accordingly, in order to realize weight reduction of the vehicle body, it is essential to develop a material which is thin yet has excellent strength, ductility, and bending properties. Therefore, in order to solve these problems, it is necessary to develop a giga-grade steel sheet capable of securing high strength having a tensile strength of 980 MPa or more and high ductility.

[0003] Meanwhile, as regulations for impact stability of automobiles have expanded, a high-strength steel having excellent yield strength is being employed as a material of structural members such as a member, a seat rail, a pillar, and the like, for improving impact resistance of a vehicle body. The structural member has a characteristic that is advantageous in absorbing impact energy as a yield strength compared to a tensile strength, that is, a yield ratio (yield strength/tensile strength) becomes higher. However, in general, as the strength of the steel sheet increases, an elongation decreases, resulting in a problem of deterioration in molding processability. Therefore, there is a demand for the development of materials with improved high yield ratio, formability, and bending properties, which are major physical properties during part processing.

[0004] A representative manufacturing method for increasing the yield strength is to use water cooling during continuous annealing. That is, after being soaked in an annealing process, immersion in water and tempering, so that it is possible to manufacture a steel sheet obtained by transforming a microstructure thereof from martensite into tempered martensite. As representative prior art of such a method, there is provided Patent Document 1. Patent Document 1 discloses a technology of manufacturing a steel material, and in Patent Document 1, a steel material having 0.18 to 0.3 of carbon is continuously annealed and then water cooled to room temperature, followed by an overaging treatment at a temperature of 120 to 300°C for 1 to 15 minutes, and the steel material has a martensite volume fraction of 80 to 97% and a balance of ferrite. As described above, when an ultra-strength steel is manufactured by a tempering method after water cooling, there is a problem in that a yield ratio is very high but shape quality of a coil deteriorates due to a deviation in temperature in a width direction and a length direction. Therefore, problems such as material defects depending on parts to be processed during roll forming, workability deterioration, and the like.

[0005] Provided is Patent Document 2 as a prior art for improving the processibility of the high tensile strength steel sheet. Patent Document 2 discloses a steel sheet formed of a composite structure mainly composed of tempered martensite, and in Patent Document 2, in order to improve processability, fine precipitated Cu particles having a particle size of 1 to 100 mm are dispersed in the structure. However, in Patent Document 1, in order to precipitate good fine Cu particles, red heat brittleness due to Cu may occur due to excessive addition of Cu content of 2 to 5%, and there is also a problem in that manufacturing costs are excessively increased.

[0006] Meanwhile, Patent Document 3 mainly discloses a steel sheet having ferrite as a base structure and 2 to 10% by area of pearlite as a microstructure, and having improved strength through precipitation strengthening and crystal grain refinement through the addition of carbon nitride forming elements such as Ti, or the like. Patent Document 3 has an advantage of easily obtaining high strength compared to low manufacturing costs, but has a disadvantage in that high-temperature annealing should be performed in order to secure ductility by causing sufficient recrystallization by rapidly increasing a recrystallization temperature due to fine precipitates. In addition, an existing precipitation strengthening steel, which is strengthened by precipitating carbonitrides on a ferrite base, has a problem in that it is difficult to obtain a high-strength steel of 600 MPa grade or higher.

[0007] Therefore, by solving the above-described problems, there is a demand for the development of a steel material having ultra-high strength capable of cold forming while exhibiting high yield strength and bending properties.

[Prior art Document]

[0008]

(Patent Document 1) Japanese Patent Registration No. 2528387

(Patent Document 2) Japanese Unexamined Patent Publication No. 2005-264176

(Patent Document 3) Korean Patent Publication No. 2015-0073844

Summary of Invention

Technical Problem

[0009] An aspect of the present disclosure is to provide an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties and a method for manufacturing the same.

Solution to Problem

[0010] According to an aspect of the present disclosure, provided is an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, the ultra-high-strength cold-rolled steel sheet including, by weight: C: 0.03 to 0.12%, Si: 0.03 to 0.30%, Mn: 2.1 to 2.9%, Al: 0.005 to 0.07%, Nb: 0.01 to 0.08%, Ti: 0.005 to 0.08%, B: 0.0005 to 0.005%, Cr: 0.7 to 1.4%, Mo: 0.005 to 0.10%, N: 0.008% or less (excluding 0%), with a balance of Fe and inevitable impurities, satisfying the following Relational Expressions 1 to 3, wherein a microstructure thereof includes by area, fresh martensite: 4 to 19%, a sum of tempered martensite and bainite: 78 to 95%, and retained austenite: 0.2 to 2.0%, wherein an average grain size of the microstructure is 0.5 to 6 um,

[Relational Expression 1] $0.18 \le C+Si/30+Mn/20+2P+4S \le 0.30$

[Relational Expression 2] $180 \le 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb \le 270$

[Relational Expression 3] $700 \le 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb/C+Si/30+Mn/20+2P+4S \le 1200$

where, a content of alloy components described in the Relational Expressions 1 to 3 refers to weight %.

[0011] According to another aspect of the present disclosure, provided is a method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, the method including operations of: heating a slab including, by weight: C: 0.03 to 0.12%, Si: 0.03 to 0.30%, Mn: 2.1 to 2.9%, Al: 0.005 to 0.07%, Nb: 0.01 to 0.08%, Ti: 0.005 to 0.08%, B: 0.0005 to 0.005%, Cr: 0.7 to1.4%, Mo: 0.005-0.10%, N: 0.008% or less (excluding 0%), with a balance of Fe and inevitable impurities, satisfying the following Relational Expressions 1 to 3; finish hot rolling the heated slab so that a finish rolling exit temperature is Ar3+50°C to Ar3+150°C; cooling the hot-rolled steel sheet to a temperature within a range of Ms+50°C to Ms+300°C and winding the same; cold rolling the wound hot-rolled steel sheet to obtain a cold-rolled steel sheet; continuously annealing the cold-rolled steel sheet at a temperature within a range of Ar3+10°C to Ar3+70°C; soaking the continuously annealed cold-rolled steel sheet for 50 to 200 seconds; primarily cooling the soaked cold-rolled steel sheet to a temperature of 620 to 700°C at a cooling rate of 1 to 10°C/s; secondarily cooling the primarily-cooled cold-rolled steel sheet to a temperature of 360 to 420°C at a cooling rate of 5 to 50°C/sec; overaging the secondarily-cooled cold-rolled steel sheet for 250 to 650 seconds, and then terminating the same at a temperature of 320 to 400°C, wherein the following Relational Expressions 4 to 6 are satisfied during the secondary cooling and the overaging treatment,

[Relational Expression 1] $0.18 \le C+Si/30+Mn/20+2P+4S \le 0.30$

[Relational Expression 2] $180 \le 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb \le 270$

[Relational Expression 3] $700 \le 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb/C+Si/30+Mn/20+2P+4S \le 1200$

[Relational Expression 4] $10 \le A \le 70$

[Relational Expression 5] $30 \le B \le 100$

[Relational Expression 6] $2.5 \le$ overaging treatment time/B ≤ 14

where, in the above Relational Expressions 1 to 3, a content of alloy components described is weight %, and in the above Relational Expressions 4 to 6, A is a Ms-secondary cooling end temperature (°C), and B is an overaging end temperature (°C).

Advantageous Effects of Invention

[0012] As set forth, according to an aspect of the present disclosure, an ultra-high-cold-rolled steel sheet having excellent yield strength and bending properties, and a method for manufacturing the same may be provided.

Brief description of drawings

[0013] FIG. 1 is a photograph of a microstructure of Inventive Example 6 observed by SEM according to an embodiment of the present disclosure.

Best Mode for Invention

[0014] Hereinafter, an ultra-high strength cold-rolled steel sheet having excellent yield strength and bending properties according to an embodiment of the present disclosure will be described. First, the 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): 0.03 to 0.12%

[0015] Carbon (C) is a very important element added for solid solution strengthening. In addition, C is bonded to a precipitation element to form fine precipitates, thereby contributing strength improvement. When the content of C is less than 0.03%, it is very difficult to secure a desired strength. On the other hand, when the content of C exceeds 0.12%, as martensite is excessively formed during cooling an increase in hardenability, the strength may rapidly increase, and bending properties may deteriorate, so that it may be difficult to obtain an HER, R/t, and a 3-point bending maximum angle, which are intended to be obtained in the present disclosure. In addition, weldability is deteriorated, thereby increasing a possibility of occurrence of welding defects during parts processing by customers. Therefore, the content of C is preferably in a range of 0.03 to 0.12%. A lower limit of the content of C is more preferably 0.04%, and even more preferably 0.05%. An upper limit of the content of C is more preferably 0.10%, and even more preferably 0.09%.

Silicon (Si): 0.03 to 0.30%

[0016] Silicon (Si) is one of the five major elements of steel, and a small amount of Si is naturally added during a manufacturing process of steel. Si contributes to an increase in strength, and suppresses formation of carbides, so that carbon is not formed as carbides during an annealing soaking heat treatment and cooling. In addition, this carbon is distributed and accumulated in retained austenite, so that an austenite phase remains at room temperature and is advantageous in securing elongation. When the Si content is less than 0.03%, it may be difficult to sufficiently secure the above-mentioned effects. On the other hand, when the content of Si exceeds 0.30%, a solid solution strengthening effect may be increased and elongation may be reduced, and surface scale defects may be caused to deteriorate plating surface quality and formation treatment properties. Therefore, the content of Si is preferably in a range of 0.03 to 0.30%. A lower limit of the content of Si is more preferably 0.04%, and even more preferably 0.05%. An upper limit of the content of Si is more preferably 0.25%, and even more preferably 0.20%.

Manganese (Mn): 2.1 to 2.9%

[0017] Manganese (Mn) is an element which completely precipitates sulfur in steel as MnS to prevent hot brittleness due to formation of FeS and solid solution strengthening of steel. When the content of Mn is less than 2.1%, it is difficult to secure a target level of strength in the present disclosure. On the other hand, when the content of Mn exceeds 2.9%, problems such as weldability, hot rolling, and the like, are likely to occur, and at the same time, hardenability may be increased to excessively form martensite, which may result in a decrease in elongation. In addition, there is a problem in that a Mn-Band (band of Mn oxide) are formed in the microstructure, which increases a risk of processing cracks and sheet breakage, and there is a problem in that the Mn oxide is eluted on the surface during annealing, thereby greatly impairing plating properties. Therefore, the content of Mn is preferably in a range of 2.1 to 2.9%. A lower limit of the content of Mn is more preferably 2.2%, and even more preferably 2.3%. An upper limit of the content of Mn is more preferably 2.8%, and even more preferably 2.7%.

Aluminum (Al): 0.005 to 0.07%

[0018] Aluminum (Al) is an element added for deoxidation during steelmaking. When the content of Al is less than 0.005%, it is difficult to obtain a sufficient deoxidation effect, and when the content of Al exceeds 0.07%, Al reacts with oxygen (O) in molten steel to form oxides (inclusions) having a high melting point, which may cause nozzle clogging. In addition, a shape of the inclusion formed in this manner is sharp, so that bending properties may be inferior. Therefore, the content of Al is preferably 0.005 to 0.07%. A lower limit of the content of Al is more preferably 0.010%, and even more preferably 0.020%. An upper limit of the content of Al is more preferably 0.06%, and even more preferably 0.05%.

Niobium (Nb): 0.01 to 0.08%

[0019] Niobium (Nb) is an element which is segregated at austenite grain boundaries to suppress coarsening of austenite crystal grains during an annealing heat treatment, and to contribute to increase strength by forming fine carbides. When the content of Nb is less than 0.01%, the above-described effect is insufficient. On the other hand, when the content of Nb exceeds 0.08%, coarse carbides are precipitated, strength and elongation may be reduced by reducing an amount of carbon dissolved in steel, and manufacturing costs may increase. Therefore, the content of Nb is preferably in a range of 0.01 to 0.08%. A lower limit of the content of Nb is more preferably 0.02%, and even more preferably 0.03%. An upper limit of the content of Nb is more preferably 0.07%, and even more preferably 0.06%.

Titanium (Ti): 0.005 to 0.08%

[0020] Titanium (Ti) is an element for forming fine carbides, thereby contributing to securing yield strength and tensile strength. In addition, Ti is an element for forming nitrides, and having an effect of suppressing AlN precipitation by precipitating N in steel as TiN, thereby reducing a risk of occurrence of cracks during continuous casting. When the content of Ti is less than 0.005%, it may be difficult to obtain the above-described effect. On the other hand, when the content of Ti exceeds 0.08%, coarse carbides are precipitated, strength and elongation may be reduced by reducing an amount of dissolved carbon in steel, and nozzle clogging may be caused during casting. Therefore, the content of Ti is preferably in a range of 0.005 to 0.08%. A lower limit of the content of Ti is more preferably 0.007%, and even more preferably 0.01%. An upper limit of the content of Ti is more preferably 0.07%, and even more preferably 0.06%.

Boron (B): 0.0005 to 0.005%

[0021] Boron (B) is an element greatly contributing to securing hardenability of a steel material, and is preferably added in an amount of 0.0005% or more to obtain such an effect. However, when the content of B exceeds 0.005%, boron carbides are formed at a grain boundary to provide ferrite nucleation sites, which may rather deteriorate the hardenability. Therefore, the content of B is preferably in a range of 0.0005 to 0.005%. A lower limit of the B content is more preferably 0.0010%, and even more preferably 0.0015%. An upper limit of the B content is more preferably 0.0045%, and even more preferably 0.004%.

Chromium (Cr): 0.7 to 1.4%

[0022] Chromium (Cr) is an element improving hardenability and increasing strength of steel. When the content of Cr is less than 0.7%, it may be difficult to secure a target level of strength. On the other hand, when the content of Cr exceeds 1.4%, ductility of a steel sheet may decrease. Therefore, the content of Cr is preferably in a range of 0.7 to 1.4%. A lower limit of the content of Cr is more preferably 0.75%, and even more preferably 0.8%. An upper limit of the content of Cr is more preferably 1.3%, and even more preferably 1.2%.

Molybdenum (Mo): 0.005 to 0.10%

[0023] Molybdenum (Mo) is an element forming carbides, and perform a role of improving yield strength and tensile strength by finely maintaining a size of precipitates when combined with carbon nitride forming elements such as Ti, Nb, V, and the like. In addition, Mo has an advantage of capable of controlling a yield ratio by forming the hardenability of steel to finely form martensite at a grain boundary. In order to obtain the above-described effect, Mo is preferably added in an amount of 0.0005% or more. However, molybdenum (Mo) is an expensive element, and as the content of Mo increases, there is a disadvantage in manufacturing, so that it is preferable to properly control the content of Mo. When the content of Mo exceeds 0.10%, there is a problem in that ductility of the steel is rather deteriorated due to an excessive crystal grain refinement effect and a solid solution strengthening effect, as well as economic feasibility is deteriorated due to a rapid increase in manufacturing costs. Therefore, the content of Mo is preferably in a range of 0.005 to 0.10%. A lower limit of the content of Mo is more preferably 0.007%, and even more preferably 0.01%. An upper limit of the content of Mo is more preferably 0.08%, and even more preferably 0.06%.

Nitrogen (N): 0.008% or less (excluding 0%)

[0024] Nitrogen (N)is an element that is inevitably contained in the manufacturing process, but is an element that contributes to improving the strength of steel through the formation of carbon nitrides. However, when the content of N exceeds 0.008%, a risk of occurrence of brittleness is greatly increased, and an excess amount of N remaining after forming TiN can consume B, which should contribute to hardenability, in a form of BN. Therefore, the content of N is preferably 0.008% or less. The content of N is more preferably 0.007% or less, and even more preferably 0.006% or less.

[0025] Meanwhile, the cold-rolled steel sheet of the present disclosure preferably satisfies the above-described alloy components and the following Relational Expressions 1 to 3. Thereby, it is possible to manufacture an ultra-high-strength steel sheet having a tensile strength of 980 MPa or more, which has very excellent bending processability, targeted by the present disclosure.

[Relational Expression 1] $0.18 \leq C+Si/30+Mn/20+2P+4S \leq 0.30$

[0026] The Relational Expression 1 is a component Relational Expression for securing strength and weldability. When a value of Relational Expression 1 is less than 0.18, it is difficult to a strength targeted by the present disclosure, and when a value of Relational Expression 1 exceeds 0.30, weldability may inferior. Therefore, the value of Relational Expression 1 preferably has a range of 0.18 to 0.30. A lower limit of the value of the Relational Expression 1 is more preferably 0.19, and even more preferably 0.20. An upper limit of the value of Relational Expression 1 is more preferably 0.28, and even more preferably 0.26.

[Relational Expression 2] $180 \leq 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb \leq 270$

[0027] The Relational Expression 2 is a Relational Expression related to a hardenability index for securing hardenability. When a value of the Relational Expression 2 is less than 180, it is difficult to secure a strength targeted by the present disclosure due to insufficient hardenability, and when a value of the Relational Expression 2 exceeds 270, the hardenability may be excessively increased, resulting in poor bending properties and formability. Therefore, the value of the Relational Expression 2 is preferably in a range of 180 to 270. A lower limit of the value of the Relational Expression 2 is more preferably 190, and even more preferably 200. An upper limit of the value of the Relational Expression 2 is more preferably 260, and even more preferably 250.

[Relational Expression 3] $700 \leq 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb/C+Si/30+Mn/20+2P+4S \leq 1200$

[0028] The Relational Expression 3 is a component Relational Expression for simultaneously securing strength, hardenability and weldability, which are the targets of the present disclosure. When a value of the Relational Expression 3 is less than 700, not only weldability may be inferior, but also it is difficult to secure a strength targeted by the present disclosure due to insufficient hardenability, and when a value of the Relational Expression 3 exceeds 1200, hardenability may be excessively increased, resulting in inferior bending properties and formability. Therefore, the value of the Relational Expression 3 is preferably in a range of 700 to 1200. A lower limit of the value of the Relational Expression 3 is more preferably 700, and even more preferably 800. An upper limit of the value of the Relational Expression 3 is more preferably 1150, and even more preferably 1100.

[0029] 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.

[0030] However, thereamong, since phosphorus and sulfur are impurities, commonly mentioned, a brief description thereof is as follows.

Phosphorus (P): 0.04% or less (excluding 0%)

[0031] Phosphorus (P) is an element which may be segregated at a grain boundary and/or an interphase grain boundary to cause brittleness. Therefore, the content of P should be controlled to be as low as possible, and it is preferable to limit the content of P to be 0.04% or less. The content of P is more preferably limited to be 0.03% or less, and more preferably limited to be 0.02% or less.

Sulfur (S): 0.005% or less (excluding 0%)

[0032] Sulfur (S), as an impurity, may be segregated during MnS non-metallic inclusions in steel and casting solidification and can cause high-temperature cracking. Therefore, the content of S should be controlled to be as low as possible, and it is preferable to limit the content of S to be 0.005% or less. The content of S is more preferably limited to be 0.004% or less, and more preferably limited to be 0.003% or less.

[0033] In addition, the impurities may include at least one of Sb, Mg, Sn, Sb, Zn, and Pb as tramp elements, and a total amount thereof may be 0.1% or less by weight. The tramp element is an impurity element derived from scrap, or the like used as a raw material in a steelmaking process, and when the total amount thereof exceeds 0.1%, it may cause surface cracks of the slab, and deteriorate surface quality of the steel sheet.

[0034] Hereinafter, a microstructure of an ultra-high strength cold-rolled steel sheet having excellent yield strength and bending properties according to an embodiment of the present disclosure will be described.

[0035] The microstructure of the cold-rolled steel sheet according to the present disclosure preferably includes, by area, fresh martensite: 4 to 19%, a sum of tempered martensite and bainite: 78 to 95%, and retained austenite: 0.2 to 2.0%. The microstructure of the cold-rolled steel sheet of the present disclosure includes tempered martensite (hereinafter, also referred to as 'TM') and bainite (hereinafter, referred to as 'B') as a main structure. Since the tempered martensite and bainite are not easy to be distinguished in terms of a microstructure, in the present disclosure, a fraction of the sum of the tempered martensite and bainite is controlled. When the fraction of the sum of the tempered martensite and bainite is less than 78%, it is difficult to secure the target strength, and when the fraction of the sum of the tempered martensite and bainite exceeds 95%, bending properties and elongation may be inferior. The fresh martensite (hereinafter, also referred to as 'FM') is a structure, advantageous for securing strength. When a fraction of the fresh martensite is less than 4%, it is difficult to secure the target strength, and when the fraction of the fresh martensite exceeds 19%, bending properties and elongation may be inferior. The retained austenite (hereinafter, also referred to as 'RA') is a structure, advantageous for securing elongation. When a fraction of the retained austenite is less than 0.2%, it may be difficult to sufficiently obtain the above effect, and when the fraction of the retained austenite exceeds 2.0%, it may be transformed into martensite during processing, resulting in inferior HER or bending properties.

[0036] Meanwhile, the microstructure may further include 10% or less of ferrite. The ferrite structure is a structure that can be formed inevitably in the manufacturing process, but also has a positive function. For example, the ferrite may contribute to securing elongation. However, when a fraction of the ferrite exceeds 10%, it may be difficult to secure the strength desired by the present disclosure. The fraction of the ferrite is more preferably 7% or less, and even more preferably 5% or less. Meanwhile, it is preferable that the microstructure has an average grain size of 0.5 to 6 um. The finer the average grain size is, the more favorable it is to secure physical properties such as strength, HER, and the like. However, in order to control the average grain size to be less than 0.5 um, an input amount of Nb, Ti, Mo, V, and the like, which are effective in grain refinement, may be excessively increased, resulting in an increase in manufacturing costs. When the average grain size exceeds 6 um, it is difficult to secure the strength targeted by the present disclosure, and HER and bending properties may be significantly inferior. Therefore, the average grain size preferably has a range of 0.5 to 6.0 μm. A lower limit of the the average grain size is more preferably 1.0 um, and even more preferably 1.5 um. An upper limit of the average grain size is more preferably 5.5 μm, and even more preferably 5.0 μm.

[0037] The cold-rolled steel sheet of the present disclosure provided as described above may have a yield strength (YS): 800 to 980 MPa, a tensile strength (TS): 980 to 1180 MPa, an elongation (EL): 4 to 12%, a yield ratio (YS / TS): 0.70 to 0.95, a hole expansion ratio (HER): 35 to 80%, a R/t: 0.8 or less, and a 3-point bending maximum angle: 90 to 140°. The yield strength is more preferably 820 to 960MPa, more preferably 850 to 950MPa. The tensile strength is more preferably 1000 to 1170MPa, more preferably 1020 to 1160MPa. The elongation is more preferably 5 to 11%, and even more preferably 6 to 10%. The yield ratio is more preferably 0.72 to 0.92, and even more preferably 0.75 to 0.90. The hole expansion ratio is more preferably 40 to 75%, and even more preferably 45 to 70%. The R/t is more preferably 0.15 to 0.70, and even more preferably 0.20 to 0.60. The 3-point bending maximum angle is more preferably 95 to 135 °, and more preferably 100 to 130 °.

[0038] In addition, the cold-rolled steel sheet of the present disclosure may have a hardness (HvBM) of 300 to 400 Hv. The hardness of the base material is more preferably 310 to 390 Hv, and even more preferably 320 to 380 Hv. In addition, hardness of a fusion zone (HvFZ) of a weld zone formed after welding may be 350 to 450 Hv. When the hardness of the fusion zone (HvFZ) of the weld zone is less than 350 Hv, sufficient hardness of the fusion zone may not be secured, and thus a strength of the weld zone may be low. On the other hand, when the hardness of the fusion zone (HvFZ) of the weld zone exceeds 450 Hv, the hardness of the fusion zone is too high, and crack generation susceptibility is increased, so the strength of the weld zone and impact absorption energy may be lowered. The hardness of the cold-rolled steel sheet, that is, the hardness corresponding to the base material after welding (HvBM) is better as it is similar to the hardness of the fusion zone (HvFZ), and therefore, the ratio thereof (HvFZ/HvBM) is preferably 1.30 or less.

[0039] Hereinafter, an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties according to an embodiment of the present disclosure will be described.

[0040] First, a slab satisfying the above-described alloy composition is heated. In the present disclosure, the slab heating temperature is not particularly limited, but for example, the slab may be heated at a temperature within a range of 1100 to 1300°C. When the slab heating temperature is lower than 1100°C, the slab temperature is low, which may cause a rolling load during rough rolling, and when the slab heating temperature is higher than 1300°C, the structure may be coarsened, and there may be a disadvantage such as increased power costs. A lower limit of the slab heating temperature is more preferably 1125°C, and even more preferably 1150°C. An upper limit of the slab heating temperature is more preferably 1275°C, and even more preferably 1250°C. Meanwhile, the slab may have a thickness of 230 to 270 mm.

[0041] Thereafter, the heated slab is finish-rolled so that a finish-rolling exit temperature is Ar3+50°C to Ar3+150°C to obtain a hot-rolled steel sheet. When the finish-rolling exit temperature is lower than Ar3+50°C, there is a high possibility in that hot deformation resistance is rapidly increased. When the finish-rolling exit temperature is higher than Ar3+150°C, there is a high possibility in that too thick oxide scales may be generated and the microstructure of the steel sheet may be coarsened. Therefore, the finish-rolling exit temperature preferably has a range of Ar3+50°C to Ar3+150°C. A lower limit of the finish-rolling exit temperature is more preferably Ar3+60°C, and even more preferably Ar3+70°C. An upper limit of the finish-rolling exit temperature is more preferably Ar3+140°C, and even more preferably Ar3+140°C. Meanwhile, the Ar3 means a temperature at which austenite is transformed into austenite during heating, and a value thereof can be obtained by, for example, a formula such as $910-203C^{1/2}+44.7Si+31.5Mo-30Mn-11Cr+700P+400Al+400Ti$.

[0042] Thereafter, the hot-rolled steel sheet is cooled to Ms+50°C to Ms+300°C and then wound. When the winding temperature is lower than Ms+50°C, excessive martensite or bainite may be generated, resulting in an excessive increase in strength of the hot-rolled steel sheet, and a problem such as shape defects, or the like due to load during cold rolling may occur. On the other hand, when the winding temperature is higher than Ms+300°C, pickling properties may be deteriorated due to an increase in surface scales. Therefore, the winding temperature preferably has a range of Ms+50°C to Ms+300°C. A lower limit of the winding temperatureis more preferably Ms+60°C, and even more preferably Ms+70°C. An upper limit of the winding temperature is more preferably Ms+290°C, and even more preferably Ms+270°C. Meanwhile, after the winding, the wound hot-rolled steel sheet may be cooled to room temperature at a cooling rate of 0.1°C/s or less. The Ms means a temperature at which martensite starts to be transformed during cooling, and a value thereof may be obtained by, for example, a formula such as $539-423C-30.4Mn-7.5Si+30Al$.

[0043] Thereafter, the cold-rolled steel sheet is obtained by cold-rolling the wound and cooled hot-rolled steel sheet. The cold rolling may be performed at a reduction ratio of 40 to 70%. When the cold reduction ratio is less than 40%, a driving force for recrystallization may be weakened, and problems are likely to occur in obtaining good recrystallized grains, and there is a disadvantage in that shape correction is very difficult. When the cold reduction ratio exceeds 70%, there is a high possibility in that cracks may occur at an edge portion of the steel sheet, and a rolling load may rapidly be increased. Therefore, the cold rolling is preferably performed at a reduction ratio of 40 to 70%. Meanwhile, prior to the cold rolling, pickling may be performed to remove scales or impurities attached to the surface thereof.

[0044] Thereafter, the cold-rolled steel sheet is continuously annealed in a temperature range of Ar3+10°C to Ar3+70°C. When the continuous annealing temperature is lower than Ar3+10°C, it may not be sufficiently transformed into austenite, so it is difficult to obtain a fraction of martensite or bainite desired by the present disclosure in a subsequent process, and it may be difficult to secure strength. On the other hand, When the continuous annealing temperature is higher than Ar3+70°C, an austenite grain size may be coarsened, so that it may be difficult to secure the target bending properties. Therefore, the continuous annealing temperature preferably has a range of Ar3+10°C to Ar3+70°C. A lower limit of the continuous annealing temperature is more preferably Ar3+20°C, and even more preferably Ar3+30°C. An upper limit of the continuous annealing temperature is more preferably Ar3+60°C, and even more preferably Ar3+50°C.

[0045] Thereafter, the continuously annealed cold-rolled steel sheet is subjected to a soaking treatment for 50 to 200 seconds. This is to secure a sufficient fraction of austenite at an annealing temperature suggested by the present disclosure along with recrystallization and grain growth of the cold-rolled structure. When the soaking treatment time is less than 50 seconds, reverse transformation into austenite does not sufficiently occur, so that a fraction of ferrite in a final structure increases, making it difficult to secure the target strength. On the other hand, when the soaking treatment time exceeds 200 seconds, the austenite grain size may be coarsened, so that bending properties of the final product may be inferior. A lower limit of the soaking treatment time is more preferably 55 seconds, and even more preferably 60 seconds. An upper limit of the soaking treatment time is more preferably 190 seconds, and even more preferably 180 seconds.

[0046] Thereafter, the soaking-treated cold-rolled steel sheet is primarily cooled to 620 to 700°C at a cooling rate of 1 to 10°C/s. The primary cooling operation is to increase ductility and strength of the steel sheet by securing an equilibrium carbon concentration of ferrite and austenite. When the primary cooling end temperature is lower than 630°C or is higher than 700°C it is difficult to secure the ductility and strength targeted by the present disclosure. When the cooling rate is less than 1°C/s, ferrite transformation is accelerated, so that there is a disadvantage in that it is difficult to secure a fraction of the target microstructure, and when the cooling rate is exceeds 1°C/s, there is a disadvantage in that it is difficult to secure an elongation due to excessive martensitic transformation.

[0047] Thereafter, the primarily-cooled cold-rolled steel sheet is secondarily cooled to 360 to 420°C at a cooling rate of 5 to 50°C/ sec. The secondary cooling is one of control factors that are considered to be important in the present disclosure, and the secondary cooling end temperature is a very important condition to simultaneously secure strength, ductility and bending properties. When the secondary cooling end temperature is lower than 360°C, it is difficult to secure ductility due to an excessive increase in a fraction of martensite, and when the secondary cooling end temperature is higher than 420°C, it is difficult to secure sufficient martensite, so that it is difficult to secure the target strength. Therefore, the secondary cooling end temperature, which is one of the important control factors for securing the target physical properties in the present disclosure, preferably has a range of 360 to 420°C. A lower limit of the secondary cooling end temperature is more preferably 365°C, and even more preferably 370°C. An upper limit of the secondary cooling end temperature is more preferably 410°C, and even more preferably 405°C. When the secondary cooling rate is less than 5°C/s, due to slow cooling rate, ferrite transformation occurs preferentially before martensite and bainite transformation, so there is a disadvantage in that an appropriate amount of fraction of the microstructure to be obtained by the present disclosure may not be obtained, and when the secondary cooling rate exceeds 50°C/s, passing ability may be reduced due to a problem of shape defects due to an excessive cooling rate, and sheet breakage may occur. A lower limit of the secondary cooling rate is more preferably 7.5°C/s, and more preferably 10°C/s. An upper limit of the secondary cooling rate is more preferably 47.5°C/s, and more preferably 45°C/s.

[0048] Meanwhile, it is important to precisely control a difference between an Ms temperature and a secondary cooling end temperature in order to secure fractions of tempered martensite and bainite, which are important microstructures, to a target level. More specifically, it is preferable to satisfy the following Relational Expression 4. When the difference between Ms and the secondary cooling end temperature, that is, a value of A is less than 10°C, it may be difficult to secure the target strength due to low martensite or bainite transformation, and when the value of A exceeds 70°C, it may be difficult to secure ductility due to an excessive increase in the fraction of the martensite due to a long residence time in a martensite region. Therefore, the difference between the Ms and the secondary cooling end temperature, that is, the value of A is preferably 10 to 70°C. A lower limit of the value of A is more preferably 15°C, and even more preferably 20°C. An upper limit of the value of A is more preferably 65°C, and even more preferably 60°C. Meanwhile, Ms means a temperature at which martensitic transformation starts, and a value thereof can be obtained by the following Expression 1.

[Relational Expression 4] $10 \le A \le 70$

where, A is an Ms-secondary cooling end temperature (°C)

[0049] Thereafter, the secondarily-cooled cold-rolled steel sheet is subjected to an overaging treatment for 250 to 650 seconds, and then ends at a temperature within a range of 320 to 400°C. Thereafter, the overaging treatment is preferably performed at a temperature, equal to or higher than the temperature at the time of the secondary cooling. The overaging treatment is a process for promoting transformation of fresh martensite generated at the end of secondary cooling into tempered martensite, through which high yield strength and bending properties can be stably secured. Therefore, the overaging treatment is a very important factor in order to secure high bending processability to be obtained in the present disclosure, and in the present invention, the overaging treatment time is precisely controlled to be in a range of 250 to 650 seconds. When the overaging treatment time is less than 250 seconds, transformation from fresh martensite into tempered martensite may occur in a small amount, resulting in poor bending processability. On the other hand, when the overaging treatment time exceeds 650 seconds, it may be difficult to secure the tensile strength desired by the present disclosure due to reduced productivity and excessive tempered martensitic transformation. A lower limit of the overaging treatment time is more preferably 260 seconds, and even more preferably 270 seconds. An upper limit of the overaging treatment time is more preferably 600 seconds, and even more preferably 550 seconds. When the overaging treatment end temperature is lower than 320°C, it may be difficult to secure an elongation due to excessive fresh martensitic transformation, and bending properties may be inferior. When the overaging treatment end temperature is higher than 400°C, transformation from fresh martensite into tempered martensite may occur in a small amount, resulting in inferior bending properties. A lower limit of the overaging treatment end temperature is more preferably 325°C, and even more preferably 330°C. An upper limit of the overaging treatment end temperature is more preferably 395°C, and even more preferably 380°C. Meanwhile, when it is desired to further improve the HER and bending properties, an overaging treatment may be additionally performed by reheating after cooling is performed to the secondary cooling end temperature after the overaging treatment.

[0050] Meanwhile, it is important to precisely control a difference between the Ms temperature and the overaging treatment end temperature in order to secure a fraction of the tempered martensite, which is an important microstructure in the present invention, to a target level. More specifically, it is preferable to satisfy the following Relational Expression 5. When the difference between the Ms temperature and the overaging treatment end temperature, that is, a value of B is lower than 30°C, it may be difficult to secure the target strength due to insufficient martensitic transformation. When the value of B is higher than 100°C, it may be difficult to secure target elongation and bending properties due to excessive fresh martensitic transformation. Therefore, it is preferable that the difference between the Ms temperature and the overaging treatment end temperature, that is, the value of B is in a 30 to 100°C. A lower limit of the value of B is more preferably 35°C, and even more preferably 40°C. An upper limit of the value of B is more preferably 95°C, and even more preferably 90°C. An upper limit of the value of B is more preferably 95°C, and even more preferably 90°C.

[Relational Expression 5] $30 \le B \le 100$

where, B is an Ms-overaging treatment end temperature (°C)

[0051] In addition, in the present invention, for the target fraction of microstructure and mechanical properties, it is preferable to satisfy the following Relational Expression 6 during the secondary cooling and overaging treatment.

[Relational Expression 6] 2.5 ≤ Overaging treatment time/B ≤ 14

[0052] The relational expression 6 is to precisely control the microstructure targeted by the present disclosure to secure the target physical properties. When a value of Relational Expression 6 is less than 2.5, an overaging holding time may be short or an overaging treatment end temperature may be low, so that it may be difficult to secure the target elongation or bending properties due to excessive fresh martensitic transformation. On the other hand, when the value of the Relational Expression exceeds 14, it may be difficult to secure the target physical properties due to the long overaging holding time or the high overaging treatment end temperature, making it difficult to secure the target fraction of microstructure. Therefore, the value of Relational Expression 6 is preferably in a range of 2.5 to 14. A lower limit of the value of the Relational Expression 6 is more preferably 3.0, and even more preferably 3.5. An upper limit of the value of the Relational Expression 6 is more preferably 12, and even more preferably 10.

[0053] Meanwhile, in the present disclosure, after the overaging treatment, an operation of temper rolling the overaged cold-rolled steel sheet at an elongation of 0.1 to 2.0°, may be further included. In general, in the case of temper rolling, an increase in yield strength of at least 50 MPa or more occurs with little increase in tensile strength. When the elongation is less than 0.1%, it may be difficult to control the shape, and when the elongation exceeds 2.0%, workability may be greatly unstable due to a high elongation work.

Mode for Invention

[0054] 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)

[0055] After preparing molten steel having the alloy composition illustrated in Table 1 below, the molten steel was prepared into a slab having a thickness of 250 mm, heated at a temperature of 1200°C for 12 hours and then was subjected to finish hot rolling under the conditions illustrated in Table 2 below, and then wound to prepare a hot-rolled steel sheet. After pickling was performed on the hot-rolled steel sheet prepared as described above, cold rolling was performed at a cold rolling reduction ratio of 50% to manufacture a cold-rolled steel sheet. The cold-rolled steel sheet was subjected to continuous annealing, soaking treatment, primary and secondary cooling, and overaging treatment under the conditions illustrated in Table 3 below to manufacture a final cold-rolled steel sheet.

[0056] After measuring a microstructure, average grain size, and mechanical properties of the final cold-rolled steel sheet manufactured as described above, the results thereof are illustrated in Table 4 below.

[0057] The microstructure and average grain size were measured using an Electron BackScatter Diffraction (EBSD) instrument.

[0058] Among the mechanical properties, a tensile strength (TS), yield strength (YS), and elongation (EL) were obtained by taking tensile specimens in a rolling horizontal direction, which were measured by a tensile test. In this case, a gauge length was 80 mm, and a width of the tensile specimen was 20mm.

[0059] Hardness in a fusion zone (HvFZ) and hardness of a base material (HvBS) were measured by performing Bead On Plate (BOP) welding on the cold-rolled steel sheet using a CO_2 welder under the conditions of 6kW-3min, and then measured 5 times at a 1/4t (t=thickness point) under a load of 500 gf using a Vickers harness tester and then averaged.

[Table 1]

STEEL TYPE No.	ALLOY COMPOSITION (WEIGHT %)														
	C	Si	Mn	P	S	Al	Nb	Ti	B	Cr	Mo	N	EXP RES SIO N 1	EXP RES SIO N2	EXP RES SIO N 3
INVENTIVE STEEL 1	0.08	0.15	2.35	0.010	0.0020	0.030	0.040	0.03	0.0020	0.85	0.05	0.004	0.231	205	888
INVENTIVE STEEL 2	0.07	0.15	2.45	0.015	0.002 2	0.035	0.05 4	0.02	0.0025	0.95	0.03	0.00 4	0.23 6	229	971
INVENTIVE STEEL 3	0.09	0.12	2.35	0.011	0.0021	0.039	0.04 5	0.02	0.0023	1.01	0.02	0.00 4	0.24 2	222	918
INVENTIVE STEEL 4	0.07	0.11	2.40	0.015	0.0015	0.041	0.051	0.03	0.0025	0.98	0.05	0.00 3	0.23 0	227	989
INVENTIVE STEEL 5	0.10	0.15	2.30	0.012	0.0021	0.035	0.04 5	0.02	0.0020	0.85	0.03	0.00 4	0.25 2	212	841
COMPARATIVE STEEL 1	0.02	0.35	2.41	0.010	0.0021	0.032	0.03 0	0.02	0.0025	0.91	0.03	0.00 4	0.181	172	951
COMPARATIVE STEEL 2	0.06	0.15	1.56	0.011	0.002 5	0.042	0.05 0	0.02	0.0020	0.72	0.05	0.00 3	0.175	174	995
COMPARATIVE STEEL 3	0.07	0.25	2.10	0.012	0.0021	0.037	0.03 0	0.01	0.0020	0.45	0.01	0.00 4	0.216	148	684
COMPARATIVE STEEL 4	0.07	0.11	2.45	0.015	0.0018	0.035	0.00 5	0.01	0.0025	0.85	0.03	0.00 4	0.23 3	156	668
COMPARATIVE STEEL 5	0.05	0.12	2.01	0.012	0.002 4	0.041	0.04 5	0.02	0.0005	0.55	0.01	0.00 4	0.188	161	858
COMPARATIVE STEEL 6	0.14	0.12	2.60	0.015	0.002 0	0.035	0.015	0.02	0.0018	0.85	0.01	0.00 4	0.312	187	600
COMPARATIVE STEEL 7	0.15	0.25	3.50	0.02	0.002 5	0.040	0.00 5	0.03	0.0030	1.50	0.20	0.00 4	0.38 3	280	731
COMPARATIVE STEEL 8	0.06	0.15	1.80	0.02	0.002 4	0.042	0.041	0.02	0.0025	1.50	0.30	0.00 4	0.20 5	257	1255

(continued)

STEEL TYPE No.	ALLOY COMPOSITION (WEIGHT %)													
	C	Si	Mn	P	S	Al	Nb	Ti	B	Cr	Mo	N	EXP RES SIO N 1	EXP RES SIO N2	EXP RES SIO N 3
[EXPRESSION 1] C+Si/30+Mn/20+2P+4S [EXPRESSION 2] 48.8+49logC+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb [EXPRESSION 3] 48.8+49logC+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb/C+Si/30+Mn/20+2P+4S															

[Table 2]

DIVISION	STEEL TYPE No.	Ar3 (°C)	Ms (°C)	FINISH ROLLING TEMPERATURE (°C)	WINDING TEMPERATURE (°C)	THICKNESS OF HOT-ROLLED MATERIAL (mm)	THICKNESS OF COLD-ROLLED MATERIAL (mm)
INVENTIVE EXAMPLE 1	INVENTIVE STEEL 1	812	433	921	667	3.2	1.6
INVENTIVE EXAMPLE 2	INVENTIVE STEEL 2	812	435	915	687	3.2	1.6
INVENTIVE EXAMPLE 3	INVENTIVE STEEL 3	805	430	916	680	3.2	1.6
INVENTIVE EXAMPLE 4	INVENTIVE STEEL 4	819	437	901	678	3.2	1.6
INVENTIVE EXAMPLE 5	INVENTIVE STEEL 5	806	427	925	681	3.2	1.6
COMPARATIVE EXAMPLE 1	COMPARATIVE STEEL 1	843	456	905	669	3.2	1.6
COMPARATIVE EXAMPLE 2	COMPARATIVE STEEL 2	846	466	931	675	3.2	1.6
COMPARATIVE EXAMPLE 3	COMPARATIVE STEEL 3	827	445	925	685	3.2	1.6
COMPARATIVE EXAMPLE 4	COMPARATIVE STEEL 4	808	435	915	675	3.2	1.6
COMPARATIVE EXAMPLE 5	COMPARATIVE STEEL 5	837	457	921	681	3.2	1.6
COMPARATIVE EXAMPLE 6	COMPARATIVE STEEL 6	785	401	928	692	3.2	1.6
COMPARATIVE EXAMPLE 7	COMPARATIVE STEEL 7	769	368	917	662	3.2	1.6
COMPARATIVE EXAMPLE 8	COMPARATIVE STEEL β	845	459	909	681	3.2	1.6
Ar3 = 910-203C$^{1/2}$+44.7 7Si +31.5 5Mo-30Mn -11Cr+700P+400Al +400Ti Ms = 539-423C-30.4M n-7.5Si + 30Al							

[Table 3]

DIVISION	STEEL TYPE No.	ANNEAL -ING TEMPERA-TURE (°C)	SOAK-ING TIME (SEC)	PRIMA-RY COOL-ING RATE (°C/SEC)	PRIMARY COOLING END TEM-PERATURE (°C)	SECOND-ARY COOL-ING RATE (°C/ SEC)	SECOND-ARY COOL-ING END TEMPERA -TURE (°C)	OVERAGING TREATMENT		EXPRES-SION 4	EXPRES-SION 5	EXPRES-SION 6
								TIME (SEC.)	END TEM-PERA -TURE (°C)			
INVENTIVE EXAMPLE 1	INVENTIVE STEEL 1	845	80	3.2	651	11.2	395	400	362	38	71	5.6
INVENTIVE EXAMPLE 2	INVENTIVE STEEL 2	851	79	3.5	654	12.1	397	400	371	38	64	6.3
INVENTIVE EXAMPLE 3	INVENTIVE STEEL 3	835	78	3.1	662	11.6	399	400	360	31	70	5.7
INVENTIVE EXAMPLE 4	INVENTIVE STEEL 4	839	82	3.0	671	11.4	402	400	381	35	56	7.2
INVENTIVE EXAMPLE 5	INVENTIVE STEEL 5	847	81	3.0	654	10.5	385	400	361	42	66	6.1
COMPARA-TIVE EXAM-PLE 1	COMPARA-TIVE STEEL 1	867	83	3.1	645	12.1	385	400	368	71	88	4.6
COMPARA-TIVE EXAM-PLE 2	COMPARA-TIVE STEEL 2	857	84	3.6	655	11.2	415	400	391	51	75	5.3
COMPARA-TIVE EXAM-PLE 3	COMPARA-TIVE STEEL 3	841	87	3.2	661	11.7	415	400	385	30	60	6.7
COMPARA-TIVE EXAM-PLE 4	COMPARA-TIVE STEEL 4	834	85	3.3	657	11.8	405	400	374	30	61	6.5
COMPARA-TIVE EXAM-PLE 5	COMPARA-TIVE STEEL 5	857	79	3.0	651	11.2	417	400	401	40	56	7.1

(continued)

DIVISION	STEEL TYPE No.	ANNEAL -ING TEMPERA-TURE (°C)	SOAK-INGTIME (SEC)	PRIMA-RY COOL-ING RATE (°C/SEC)	PRIMARY COOLING END TEM-PERATURE (°C)	SECOND-ARY COOL-ING RATE (°C/ SEC)	SECOND-ARY COOL-ING END TEMPERA -TURE (°C)	OVERAGING TREATMENT		EXPRES-SION 4	EXPRES-SION 5	EXPRES-SION 6
								TIME (SEC.)	END TEM-PERA -TURE (°C)			
COMPARA-TIVE EXAM-PLE 6	COMPARA-TIVE STEEL 6	821	76	3.1	653	11.4	387	400	359	14	42	9.5
COMPARA-TIVE EXAM-PLE 7	COMPARA-TIVE STEEL 7	818	81	3.8	652	11.2	392	400	367	-24	1	400
COMPARA-TIVE EXAM-PLE 8	COMPARA-TIVE STEEL 8	861	82	3.6	657	11.7	391	400	367	68	92	4.3

[EXPRESSION 4] A = Ms- SECONDARY COOLING END TEMPERATURE
[EXPRESSION 5] B = Ms- OVERAGING TREATMENT END TEMPERATURE
[EXPRESSION 6]OVERAGING TREATMENT END TEMPERATURE / B

[Table 4]

DIVISION	STEEL TYPE No.	FRACTION OF MICROSTRUCTURE (AREA %)				AVERAGE GRAIN SIZE (μm)	YS (MPa)	TS (MPa)	YR	EL (%)	HvBM (Hv)	HVFZ (Hv)	HvFZ/HvBM
		FM	TM + B	F	RA								
INVENTIVE EXAMPLE 1	INVENTIVE STEEL 1	9.3	90.4	0	0.3	2.1	891	1092	0.82	8.5	335	415	1.24
INVENTIVE EXAMPLE 2	INVENTIVE STEEL 2	9.2	90.6	0	0.2	2.3	892	1105	0.81	8.3	336	425	1.26
INVENTIVE EXAMPLE 3	INVENTIVE STEEL 3	9.3	90.1	0	0.6	2.2	904	1116	0.81	7.0	337	414	1.23
INVENTIVE EXAMPLE 4	INVENTIVE STEEL 4	9.4	90.1	0	0.5	2.5	928	1126	0.82	7.1	338	414	1.22
INVENTIVE EXAMPLE 5	INVENTIVE STEEL 5	9.1	90.2	0	0.7	2.3	913	1123	0.81	7.3	329	415	1.26
COMPARATIVE EXAMPLE 1	COMPARATIVE STEEL 1	7.2	74.5	18	0.3	3.2	762	971	0.78	10.8	285	330	1.16
COMPARATIVE EXAMPLE 2	COMPARATIVE STEEL 2	6.9	73.6	19	0.5	2.9	650	956	0.68	13.2	271	320	1.18
COMPARATIVE EXAMPLE 3	COMPARATIVE STEEL 3	5.3	73.4	21	0.3	2.3	561	902	0.62	15.2	264	390	1.48
COMPARATIVE EXAMPLE 5	COMPARATIVE STEEL 4	7.2	80.6	12	0.2	5.2	778	1099	0.71	8.70	254	395	1.56
COMPARATIVE EXAMPLE 5	COMPARATIVE STEEL 5	6.2	80.3	13	0.5	2.5	648	1042	0.62	10.9	269	343	1.28
COMPARATIVE EXAMPLE 6	COMPARATIVE STEEL 6	23.1	76.7	0	0.2	2.1	923	1253	0.74	5.9	356	553	1.55
COMPARATIVE EXAMPLE 7	COMPARATIVE STEEL 7	24.2	75.1	0	0.7	2.3	890	1320	0.67	6.2	389	582	1.50
COMPARATIVE EXAMPLE 8	COMPARATIVE STEEL B	22.9	76.5	0	0.6	2.2	1123	1236	0.91	4.5	352	401	1.14
FM: FRESH MARTENSITE						TM: TEMPERED MARTENSITE				B: BAINITE			

(continued)

DIVISION	STEEL TYPE No.	FRACTION OF MICROSTRUCTURE (AREA %)				AVERAGE GRAIN SIZE (μm)	YS (MPa)	TS (MPa)	YR	EL (%)	HvBM (Hv)	HVFZ (Hv)	HvFZ/HvBM
		FM	TM + B	F	RA								

RA: RETAINED AUSTENITE F: FERRITE

[0060] As can be seen from Tables 1 to 4, in the case of Inventive Examples 1 to 5 satisfying the alloy composition and manufacturing conditions proposed by the present disclosure, it can be confirmed that the present disclosure may secure the microstructure to be obtained by the present disclosure may be secured and have excellent mechanical properties.
[0061] On the other hand, in the case of Comparative Examples 1 to 8, not satisfying the alloy composition propose by the present disclosure and not satisfying some manufacturing conditions, it can be seen that the mechanical properties are poor since the microstructure to be obtained by the present disclosure is not secured.

(Example 2)

[0062] After preparing molten steel having the alloy composition of Inventive Steel 1 described in Example 1, the molten steel was prepared into a slab having a thickness of 250 mm, heated at a temperature of 1200°C for 12 hours and then was subjected to finish hot rolling under the conditions illustrated in Table 5 below, and then wound to prepare a hot-rolled steel sheet. After pickling was performed on the hot-rolled steel sheet prepared as described above, cold rolling was performed at a cold rolling reduction ratio of 50% to manufacture a cold-rolled steel sheet. Such a cold-rolled steel sheet was subjected to continuous annealing, soaking treatment, primary and secondary cooling, and overaging treatment under the conditions illustrated in Table 6 below to manufacture a final cold-rolled steel sheet.
[0063] After measuring a microstructure, average grain size, and mechanical properties of a final cold-rolled steel sheet manufactured as described above, the results thereof were illustrated in Table 7 below.
[0064] The microstructure and average grain size were measured using an Electron BackScatter Diffraction (EBSD) instrument.
[0065] Among the mechanical properties, a tensile strength (TS), yield strength (YS), and elongation (EL) were obtained by taking tensile specimens in a rolling horizontal direction, which were measured by a tensile test. In this case, a gauge length was 80 mm, and a width of the tensile specimen was 20 mm.
[0066] Among the mechanical properties, HER was measured according to the ISO 16330 standard, and holes were sheared with a clearance of 12% using a punch having a diameter of 10 mm.
[0067] Among the mechanical properties, R/t is a value obtained by dividing R (a limit bending radius)by a thickness of the steel sheet. In this case, the R was obtained by taking a test piece having a width of 30 mm × length of 35 mm in a horizontal direction (long axis) to a rolling direction, grinding one surface thereof by 0.2 mm, performing a bending test by a V block method in accordance with JIS Z 2248 so that the ground surface is not in contact with a punch, and variously changing a bending radius at that time from 0 to 5 mm, and obtaining a minimum bending radius at which bending may be performed without breaking the material.
[0068] Among the mechanical properties, three-point bending maximum angle was measured three times for each specimen in accordance with the Verband Der Automobilindustrie (VDA) standard, and then an average value thereof was measured.

[Table 5]

DIVISION	STEEL TYPE No.	Ar3 (°C)	Ms (°C)	FINISH ROLLING TEMPERATURE (°C)	WINDINGTEM PERATURE (°C)	THICKNESS OF HOT-ROLLED MATERIAL (mm)	THICKNESS OF COLD-ROLLED MATERIAL (mm)
INVENTIVE EXAMPLE 6	INVENTIVE STEEL 1	812	433	915	651	3.2	1.6
INVENTIVE EXAMPLE 7				921	667	3.2	1.6
INVENTIVE EXAMPLE 8				917	655	3.2	1.6
INVENTIVE EXAMPLE 9				920	680	3.2	1.6
INVENTIVE EXAMPLE 10				917	674	3.2	1.6
COMPARATIVE EXAMPLE 9				820	-	-	-
COMPARATIVE EXAMPLE 10				918	430	3.2	-
COMPARATIVE EXAMPLE 11				925	651	3.2	1.6
COMPARATIVE EXAMPLE 12				917	687	3.2	1.6
COMPARATIVE EXAMPLE 13				918	687	3.2	1.6
COMPARATIVE EXAMPLE 14				921	679	3.2	1.6
COMPARATIVE EXAMPLE 15				916	680	3.2	1.6
COMPARATIVE EXAMPLE 16				921	689	3.2	1.6
$Ar3 = 910-203C^{1/2}+44.7Si+31.5Mo-30Mn-11Cr+700P+400Al+400Ti$ $Ms = 539-423C-30.4Mn-7.5Si+30Al$							

[Table 6]

DIVISION	STEEL TYPE No.	ANNEALING TEMPERA-TURE (°C)	SOAKING TIME (SEC.)	PRIMARY COOLING RATE (°C/ SEC)	PRIMARY COOLING END TEMPERATURE (°C)	SECONDARY COOLING RATE (°C/SEC)	SECONDARY COOLING END TEMPERATURE (°C)	OVERAGING TREATMENT		EXPRESSION 4	EXPRESSION 5	EXPRESSION 6
								TIME (SEC.)	END TEMPERATURE (°C)			
INVENTIVE EXAMPLE 6	INVENTIVE STEEL 1	838	85	4.2	650	11.6	410	400	375	23	58	6.9
INVENTIVE EXAMPLE 7		847	86	3.6	651	12.1	395	400	365	38	68	5.9
INVENTIVE EXAMPLE 8		851	84	3.2	657	11.6	390	400	362	43	71	5.6
INVENTIVE EXAMPLE 9		861	85	3.4	659	12.5	415	400	382	18	51	7.8
INVENTIVE EXAMPLE 10		855	81	3.2	651	10.9	405	400	371	28	62	6.5
COMPARATIVE EXAMPLE 9		-	-	-	-	-	-	-	-	-	-	-
COMPARATIVE EXAMPLE 10		-	-	-	-	-	-	-	-	-	-	-
COMPARATIVE EXAMPLE 11		800	84	3.2	651	11.5	403	400	370	30	63	6.3
COMPARATIVE EXAMPLE 12		899	81	3.5	650	12.1	415	400	382	18	51	7.8
COMPARATIVE EXAMPLE 13		845	82	0.3	650	12.1	405	400	379	28	54	7.4

(continued)

DIVISION	STEEL TYPE No.	ANNEALING TEMPERA-TURE (°C)	SOAKING TIME (SEC.)	PRIMARY COOLING RATE (°C/ SEC)	PRIMARY COOLING END TEMPERATURE (°C)	SECONDARY COOLING RATE (°C/SEC)	SECONDARY COOLING END TEMPERATURE (°C)	OVERAGING TREATMENT		EXPRESSION 4	EXPRESSION 5	EXPRESSION 6
								TIME (SEC.)	END TEMPERATURE (°C)			
COMPARATIVE EXAMPLE 14		854	80	3.4	654	2.5	421	400	392	12	41	9.8
COMPARATIVE EXAMPLE 15		847	79	3.9	651	11.8	520	400	492	-87	-59	-6.8
COMPARATIVE EXAMPLE 16		852	78	3.2	657	11.5	320	400	285	113	148	2.7

[EXPRESSION 4] A = Ms- SECONDARY COOLING END TEMPERATURE
[EXPRESSION 5] B = Ms- OVERAGING TREATMENT END TEMPERATURE
[EXPRESSION 6]OVERAGING TREATMENT END TEMPERATURE / B

[Table 7]

DIVISION	STEEL TYPE No.	FRACTION OF MICROSTRUCTURE (AREA %)				AVERAGE GRAIN SIZE (μm)	YS (MPa)	TS (MPa)	YR	EL (%)	HER (%)	R/t	3-POINT BENDING (°)
		FM	TM + B	F	RA								
INVENTIVE EXAMPLE 6	INVENTIVE STEEL 1	8.9	90.4	0	0.7	2.6	899	1097	0.82	8.3	54	0.31	107
INVENTIVE EXAMPLE 7		9.5	90.0	0	0.5	2.4	905	1105	0.82	8.9	57	0.31	109
INVENTIVE EXAMPLE 8		11.0	88.2	0	0.8	2.3	921	1125	0.82	8.1	61	0.31	112
INVENTIVE EXAMPLE 9		7.5	91.8	0	0.7	2.6	889	1079	0.82	9.7	54	0.31	108
INVENTIVE EXAMPLE 10		10.1	89.3	0	0.6	2.5	900	1101	0.82	8.6	55	0.31	109
COMPARATIVE EXAMPLE 9		-	-	-	-	-	-	-	-	-	-	-	-
COMPARATIVE EXAMPLE 10		-	-	-	-	-	-	-	-	-	-	-	-
COMPARATIVE EXAMPLE 11		7.6	70.9	21	0.5	3.6	760	1025	0.74	11.2	31	0.94	87
COMPARATIVE EXAMPLE 12		15.1	84.0	0	0.9	6.7	951	1135	0.84	5.20	42	0.94	80
COMPARATIVE EXAMPLE 13		6.8	74.7	18	0.5	4.5	771	1051	0.73	10.9	35	0.63	88
COMPARATIVE EXAMPLE 14		2.5	96.9	0	0.6	3.1	923	975	0.95	11.9	61	0.31	121
COMPARATIVE EXAMPLE 15		21.0	78.3	0	0.7	2.1	780	1250	0.62	5.6	38	0.94	86
COMPARATIVE EXAMPLE 16		3.0	96.7	0	0.3	2.4	997	1135	0.88	4.8	67	0.94	88

FM: FRESH MARTENSITE TM: TEMPERED MARTENSITE B: BAINITE
RA: RETAINED AUSTENITE

[0069] As can be seen from Tables 5 to 7, in the case of Inventive Examples 6 to 10 satisfying the alloy composition and manufacturing conditions proposed by the present disclosure, by securing a type and fraction of the microstructure and an average grain size, it can be seen that mechanical properties (tensile property, HER, bending properties), to be obtained by the present disclosure.

[0070] On the other hand, in Comparative Examples 9 to 16 satisfying the alloy composition proposed by the present disclosure, but not satisfying the manufacturing conditions, it can be seen that the microstructure or average grain size targeted by the present disclosure may not be secured, so that mechanical properties are inferior. In particular, in the case of Comparative Examples 9 and 10, a finish rolling temperature and winding temperature did not satisfy the conditions of the present disclosure, respectively, and thus sheet breakage occurred.

[0071] FIG. 1 is a photograph of a microstructure of Inventive Example 6 observed by SEM. As can be seen from FIG. 1, it can be confirmed that the microstructure targeted by the present disclosure is properly formed in Inventive Example 6.

(Example 3)

[0072] After preparing molten steel having the alloy composition of Inventive Steel 2 described in Example 1, the molten steel was prepared into a slab having a thickness of 250 mm, heated at a temperature of 1200°C for 12 hours and then was subjected to finish hot rolling under the conditions illustrated in Table 8 below, and then wound to prepare a hot-rolled steel sheet. After pickling was performed on the hot-rolled steel sheet prepared as described above, cold rolling was performed at a cold rolling reduction ratio of 50% to manufacture a cold-rolled steel sheet. Such a cold-rolled steel sheet was subjected to continuous annealing, soaking treatment, primary and secondary cooling, and overaging treatment under the conditions illustrated in Table 9 below to manufacture a final cold-rolled steel sheet.

[0073] After measuring a microstructure, average grain size, and mechanical properties of a final cold-rolled steel sheet manufactured as described above, the results thereof were illustrated in Table 10 below.

[0074] The microstructure and average grain size were measured using an Electron BackScatter Diffraction (EBSD) instrument.

[0075] Among the mechanical properties, the 3-point bending maximum angle was measured three times for each specimen in accordance with the Verband Der Automobilindustrie(VDA) standard, and then the average value thereof was measured.

[Table 8]

DIVISION	STEEL TYPE No.	Ar3 (°C)	Ms (°C)	FINISH ROLLING TEMPERATURE (°C)	WINDING TEMPERA -TURE (°C)	THICKNESS OF HOT-ROLLED MATERIAL (mm)	THICKNESS OF COLD-ROLLED MATERIAL (mm)
INVENTIVE EXAMPLE 11	INVENTIVE STEEL 2	812	435	925	675	3.2	1.6
INVENTIVE EXAMPLE 12				920	681	3.2	1.6
INVENTIVE EXAMPLE 13				916	679	3.2	1.6
INVENTIVE EXAMPLE 14				920	682	3.2	1.6
INVENTIVE EXAMPLE 15				918	679	3.2	1.6
COMPARATIVE EXAMPLE 17				925	681	3.2	1.6
COMPARATIVE EXAMPLE 18				931	678	3.2	1.6
COMPARATIVE EXAMPLE 19				925	681	3.2	1.6
COMPARATIVE EXAMPLE 20				927	678	3.2	1.6

$Ar3 = 910-203C^{1/2}+44.7Si+31.5Mo-30Mn-11Cr+700P+400Al+400Ti$

$Ms = 539-423C-30.4Mn-7.5Si+30Al$

[Table 9]

DIVISION	STEEL TYPE No.	ANNEALING TEMPERATURE (°C)	SOAKING TIME (SEC.)	PRIMARY COOLING RATE (°C/SEC)	PRIMARY COOLING END TEMPERATURE (°C)	SECONDARY COOLING RATE (°C/SEC)	SECONDARY COOLING END TEMPERATURE (°C)	OVERAGING TREATMENT		EXPRESSION 4	EXPRESSION 5	EXPRESSION 6
								TIME (SEC.)	END TEMPERATURE (°C)			
INVENTIVE EXAMPLE 11	INVENTIVE STEEL 2	851	84	3.2	649	11.2	395	374	362	40	73	5.1
INVENTIVE EXAMPLE 12		849	80	3.2	651	11.2	385	361	365	50	70	5.2
INVENTIVE EXAMPLE 13		847	79	3.2	661	11.2	395	360	369	40	66	5.5
INVENTIVE EXAMPLE 14		849	89	3.2	658	11.2	387	355	369	48	66	5.4
INVENTIVE EXAMPLE 15		852	81	3.2	649	11.2	394	360	369	41	66	5.5
COMPARATIVE EXAMPLE 17		853	80	3.2	647	11.2	390	195	335	45	100	2.0
COMPARATIVE EXAMPLE 18		849	90	3.2	649	11.2	384	225	310	51	125	1.8
COMPARATIVE EXAMPLE 19		851	81	3.2	651	11.2	395	315	300	40	135	2.3
COMPARATIVE EXAMPLE 20		850	82	3.2	652	11.2	402	325	290	33	145	2.2

[EXPRESSION 4] A = Ms- MS- SECONDARY COOLING END TEMPERATURE
[EXPRESSION 5] B = Ms- OVERAGING TREATMENT END TEMPERATURE
[EXPRESSION 6]OVERAGING TREATMENT END TEMPERATURE / B

[Table 10]

DIVISION	STEEL TYPE No.	FRACTION OF MICROSTRUCTURE (AREA %)				AVERAGE GRAIN SIZE (μm)	YS (MPa)	TS (MPa)	YR	EL (%)	3-POINT BENDING (°)
		FM	TM + B	F	RA						
INVENTIVE EXAMPLE 11	발명강 2	9.5	90.0	0	0.5	2.3	891	1092	0.82	8.0	109
INVENTIVE EXAMPLE 12		10.1	89.3	0	0.6	2.2	905	1115	0.81	8.2	108
INVENTIVE EXAMPLE 13		9.2	90.1	0	0.7	2.4	899	1110	0.81	8.0	110
INVENTIVE EXAMPLE 14		10.5	88.7	0	0.8	2.6	914	1116	0.82	7.1	108
INVENTIVE EXAMPLE 15		11.0	88.0	0	1.0	2.5	915	1120	0.82	7.5	106
COMPARATIVE EXAMPLE 17		25.3	73.7	0	1.0	2.0	965	1269	0.76	5.7	79
COMPARATIVE EXAMPLE 18		26.0	73.2	0	0.8	2.1	975	1275	0.76	5.6	75
COMPARATIVE EXAMPLE 19		24.1	75.2	0	0.7	2.3	956	1265	0.76	6.1	78
COMPARATIVE EXAMPLE 20		26.5	72.6	0	0.9	2.5	967	1287	0.75	5.4	74
FM: FRESH MARTENSITE TM: TEMPERED MARTENSITE B: BAINITE RA: RETAINED AUSTENITE F: FERRITE											

[0076] As can be seen from Tables 8 to 10, in Inventive Examples 11 to 15 satisfying the alloy composition and manufacturing conditions proposed by the present disclosure, it can be seen that the microstructure targeted by the present disclosure may be secured, so that bending properties to be obtained by the present disclosure are secured.

[0077] On the other hand, in Comparative Examples 17 to 20 satisfying the alloy composition proposed by the present disclosure, but not satisfying the overaging treatment condition and Relational Expressions 5 and 6 among manufacturing conditions, it can be seen that the microstructure targeted in the present disclosure may not be secured, so that bending properties is inferior.

Claims

1. An ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, comprising by weight:

   C: 0.03 to 0.12%, Si: 0.03 to 0.30%, Mn: 2.1 to 2.9%, Al: 0.005 to 0.07%, Nb: 0.01 to 0.08%, Ti: 0.005 to 0.08%, B: 0.0005 to 0.005%, Cr: 0.7 to 1.4%, Mo: 0.005 to 0.10%, N: 0.008% or less (excluding 0%), with a balance of Fe and inevitable impurities, satisfying the following Relational Expressions 1 to 3,
   wherein a microstructure thereof includes by area,
   fresh martensite: 4 to 19%, a sum of tempered martensite and bainite: 78 to 95%, and retained austenite: 0.2 to 2.0%,
   wherein an average grain size of the microstructure is 0.5 to 6 um,

   [Relational Expression 1] $0.18 \leq C+Si/30+Mn/20+2P+4S \leq 0.30$

   [Relational Expression 2] $180 \leq 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb \leq 270$

   [Relational Expression 3] $700 \leq 48.8+49\log C+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb/C+Si/30+Mn/20+2P+4S \leq 1200$

   where, a content of alloy components described in the Relational Expressions 1 to 3 refers to weight %.

2. The ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 1, wherein the impurities further comprise: P: 0.04% or less (excluding 0%) and S: 0.005% or less (excluding 0%).

3. The ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 1, wherein the impurities comprise at least one of Sb, Mg, Sn, Sb, Zn, and Pb, and a total content thereof is 0.1% or less by weight.

4. The ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 1, wherein the microstructure further comprises 10% or less of ferrite.

5. The ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 1, wherein the cold-rolled steel sheet has a yield strength (YS) of 800 to 980 MPa, a tensile strength (TS) of 980 to 1180 MPa, an elongation (EL) of 4 to 12%, a yield ratio (YS/TS) of 0.70 to 0.95, a hole expansion ratio (HER) of 35 to 80%, R/t of 0.8% or less, and a 3-point bending maximum angle of 90 to 140°.

6. The ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 1, wherein the cold-rolled steel sheet has hardness (HvBM) of 300 to 400 Hv, a hardness of a fusion zone (HvFZ) of

a weld zone formed after welding of 350 to 450 Hv, wherein HvFZ / HvBM is 1.30 or less.

7. A method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties, comprising operations of:

heating a slab including, by weight: C: 0.03 to 0.12%, Si: 0.03 to 0.30%, Mn: 2.1 to 2.9%, Al: 0.005 to 0.07%, Nb: 0.01 to 0.08%, Ti: 0.005 to 0.08%, B: 0.0005 to 0.005%, Cr: 0.7 to 1.4%, Mo: 0.005 to 0.10%, N: 0.008% or less (excluding 0%), with a balance of Fe and inevitable impurities, satisfying the following Relational Expressions 1 to 3;
finish hot rolling the heated slab so that a finish rolling exit temperature is Ar3+50°C to Ar3+150°C;
cooling the hot-rolled steel sheet to a temperature within a range ofMs+50°C to Ms+300°C and winding the same;
cold rolling the wound hot-rolled steel sheet to obtain a cold-rolled steel sheet;
continuously annealing the cold-rolled steel sheet at a temperature within a range of Ar3+10°C to Ar3+70°C;
soaking the continuously annealed cold-rolled steel sheet for 50 to 200 seconds;
primarily cooling the soaked cold-rolled steel sheet to a temperature of 620 to 700°C at a cooling rate of 1 to 10°C/s;
secondarily cooling the primarily-cooled cold-rolled steel sheet to a temperature of 360 to 420°C at a cooling rate of 5 to 50°C/sec;
overaging the secondarily-cooled cold-rolled steel sheet for 250 to 650 seconds, and then terminating the same at 320 to 400°C,
wherein the following Relational Expressions 4 to 6 are satisfied during the secondary cooling and the overaging treatment,

[Relational Expression 1] $0.18 \leq C+Si/30+Mn/20+2P+4S \leq 0.30$

[Relational Expression 2] $180 \leq 48.8+49logC+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb \leq 270$

[Relational Expression 3] $700 \leq 48.8+49logC+35.1Mn+25.9Si+76.5Cr+105.9Mo+1325Nb/C+Si/30+Mn/20+2P+4S \leq 1200$

[Relational Expression 4] $10 \leq A \leq 70$

[Relational Expression 5] $30 \leq B \leq 100$

[Relational Expression 6] $2.5 \leq$ overaging treatment time/B ≤ 14

where, in the above Relational Expressions 1 to 3, a content of alloy components described is weight %, and in the above Relational Expressions 4 to 6, A is a Ms-secondary cooling end temperature (°C), and B is an overaging end temperature (°C).

8. The method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 7, wherein the slab is performed at a temperature within a range of 1100 to 1300°C.

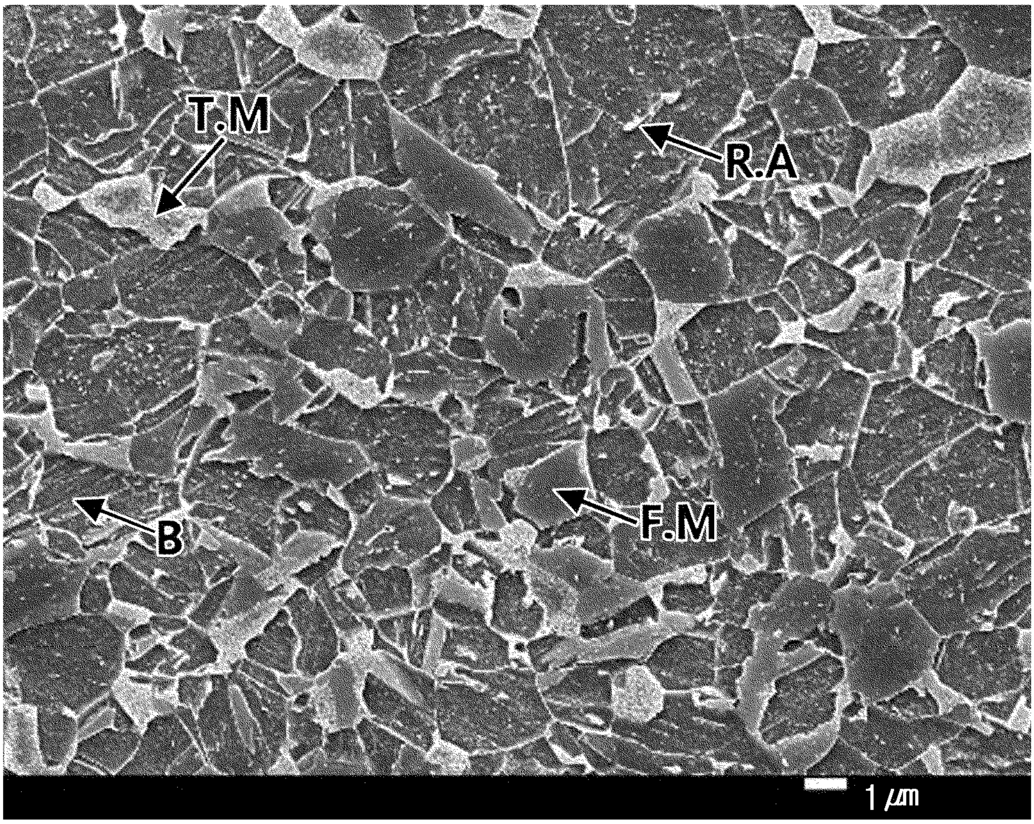
9. The method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 7, wherein the slab has a thickness of 230 to 270 mm.

10. The method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 7, further comprising
cooling the wound hot-rolled steel sheet to room temperature at a cooling rate of 0.1°C/s or less.

11. The method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 7, wherein the cold rolling is performed at a reduction ratio of 40 to 70%.

12. The method for manufacturing an ultra-high-strength cold-rolled steel sheet having excellent yield strength and bending properties of claim 7, further comprising
temper rolling the overaged cold-rolled steel sheet at an elongation of 0.1 to 2.0%, after the overaging treatment.

[Fig. 1]



INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2021/017889

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/38(2006.01)i; **C22C 38/28**(2006.01)i; **C22C 38/22**(2006.01)i; **C22C 38/00**(2006.01)i; **C21D 9/46**(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); B21B 3/00(2006.01); C21D 8/02(2006.01); C21D 9/46(2006.01); C22C 38/00(2006.01); C22C 38/14(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: 냉연강판(cold rolled steel sheet), 초고강도(ultra high strength), 항복강도(yeild strength), 굽힘(bending), 마르텐사이트(martensite), 베이나이트(bainite), 오스테나이트(austenite)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	JP 2014-196557 A (KOBE STEEL LTD.) 16 October 2014 (2014-10-16) See paragraphs [0048]-[0049], [0052], [0055] and [0071], claims 1 and 3-5 and table 3A.	1-12
A	KR 10-2014-0047960 A (POSCO) 23 April 2014 (2014-04-23) See paragraph [0071] and claims 1-10.	1-12
A	KR 10-2013-0046941 A (HYUNDAI STEEL COMPANY) 08 May 2013 (2013-05-08) See claims 1-11 and tables 1 and 2.	1-12
A	KR 10-1546134 B1 (HYUNDAI STEEL COMPANY) 21 August 2015 (2015-08-21) See claims 1-5 and tables 1 and 3.	1-12
A	JP 2007-100114 A (JFE STEEL K.K.) 19 April 2007 (2007-04-19) See paragraph [0027], claims 1-2 and tables 1 and 2.	1-12

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

* Special categories of cited documents:
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"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
22 March 2022	**23 March 2022**

Name and mailing address of the ISA/KR

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/KR2021/017889

Patent document cited in search report			Publication date (day/month/year)	Patent family member(s)			Publication date (day/month/year)
JP	2014-196557	A	16 October 2014	JP	6291289	B2	14 March 2018
KR	10-2014-0047960	A	23 April 2014	KR	10-1449134	B1	08 October 2014
KR	10-2013-0046941	A	08 May 2013	KR	10-1344537	B1	26 December 2013
KR	10-1546134	B1	21 August 2015	KR	10-2015-0014736	A	09 February 2015
JP	2007-100114	A	19 April 2007	JP	5234876	B2	10 July 2013

Form PCT/ISA/210 (patent family annex) (July 2019)

REFERENCES CITED IN THE DESCRIPTION

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- JP 2528387 B **[0008]**
- JP 2005264176 A **[0008]**
- KR 20150073844 **[0008]**