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(54) **HOT-ROLLED STEEL SHEET AND METHOD FOR PRODUCING SAME**
HEISSGEWALZTES STAHLBLECH UND VERFAHREN ZUR HERSTELLUNG DAVON
TÔLE D'ACIER LAMINÉE À CHAUD ET SON PROCÉDÉ DE PRODUCTION

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Description

TECHNICAL FIELD

5 **[0001]** The disclosure relates to a hot rolled steel sheet having high strength such as a tensile strength (TS) of 780 MPa or more, excellent stretch flangeability and blanking workability, and excellent manufacturing stability and suitable for structural-use steel material such as material for parts of transport machinery including vehicles and construction steel material. The disclosure also relates to a method of manufacturing the hot rolled steel sheet.

10 BACKGROUND

[0002] To reduce CO₂ emissions for global environment protection, an ever-present important issue for the automotive industry is to improve automotive fuel efficiency by lightening automotive bodies while maintaining the strength of automotive bodies. An effective way of lightening automotive bodies while maintaining their strength is to strengthen steel sheets as material for automotive parts to thus reduce the thickness of steel sheets. For example, automotive suspension parts in which thick steel sheets tend to be used are expected to be lightened considerably by reducing the thickness of steel sheets through strengthening.

[0003] Typically, automotive suspension parts such as lower control arms are formed by burring, and so require steel sheets to have excellent stretch flangeability. Much research and development have been conducted for hot rolled steel sheets having both strength and workability, and various techniques have been proposed. For example, it is known that high tensile strength and excellent stretch flangeability can both be achieved by making the metallic microstructure a substantially ferrite single-phase microstructure and precipitating fine carbides in the grains of the ferrite phase.

[0004] As such a technique, JP 2012-26034 A (PTL 1) discloses a hot rolled steel sheet whose strength is improved while maintaining stretch flangeability, by making the steel sheet microstructure a ferrite single-phase microstructure having excellent workability with low dislocation density and dispersing and precipitating fine carbides in the ferrite to achieve strengthening by precipitation.

[0005] Burring is typically performed using a steel sheet blanked in a predetermined shape. In actual mass production of parts, the part blanking clearance usually varies due to a temperature increase or wear of the tool caused by continuous pressing. In the case where the clearance varies, defects such as cracking or chipping may occur in the punched end surface. This has raised demand for a steel sheet that maintains excellent blanking workability regardless of the variations of the blanking conditions.

[0006] As such a steel sheet, for example, JP 2014-205888 A (PTL 2) discloses a high strength hot rolled steel sheet whose mass-production blanking workability is improved by setting the volume fraction of bainite phase to more than 92%, setting the average spacing of bainite laths to 0.60 μm or less, and setting the number ratio of Fe-based carbides precipitated in the grains to all Fe-based carbides to 10% or more.

[0007] Furthermore, WO 2014/171062 A1 (PTL 3) discloses a high strength hot rolled steel sheet with a microstructure mainly composed of a bainite phase, the bainite having an average lath interval of 0.45 μm or less.

CITATION LIST

40 Patent Literatures

[0008]

45 PTL 1: JP 2012-26034 A
PTL 2: JP 2014-205888 A
PTL 3: WO 2014/171062 A1

SUMMARY

50 (Technical Problem)

[0009] The steel sheet described in PTL 1 has both high strength and excellent stretch flangeability. However, since the steel sheet microstructure is a substantially ferrite single-phase microstructure, there is hardly any inclusion that serves as a void origin when blanking the steel sheet. Accordingly, in the steel sheet described in PTL 1, the punched end surface may become rough when conditions such as clearance and a blank holder vary.

[0010] The steel sheet described in PTL 2 has excellent blanking workability, by controlling the hot rolling conditions so that the steel sheet microstructure is mainly composed of predetermined bainite. Such a bainite microstructure,

however, tends to vary in mechanical properties such as tensile strength due to variations in coiling temperature. It is often not easy to keep uniform steel sheet temperature throughout the length and width of the coil during cooling after hot rolling. The steel sheet described in PTL 2 may thus vary greatly in mechanical properties, leading to lower manufacturing stability.

[0011] It could be helpful to provide a hot rolled steel sheet having high strength such as a tensile strength (TS) of 780 MPa or more, excellent stretch flangeability and blanking workability, and excellent manufacturing stability, together with an advantageous method of manufacturing the hot rolled steel sheet.

(Solution to Problem)

[0012] We carefully examined a method that can strengthen a steel sheet while maintaining workability and especially stretch flangeability, provide excellent blanking workability, and reduce variations in mechanical properties caused by variations in manufacturing conditions.

[0013] To improve the stretch flangeability of a steel sheet, it is effective to uniformize the strength in the metallic microstructure, as mentioned above. Available techniques include a strengthening technique of forming a ferrite single-phase microstructure to achieve solid solution strengthening or strengthening by precipitation and a strengthening technique of forming a bainite single-phase microstructure to achieve microstructure strengthening. However, in the steel sheet having the ferrite single-phase microstructure, there is hardly any inclusion that serves as a void origin when blanking the steel sheet, so that the punched end surface may become rough when conditions such as clearance and a blank holder vary.

[0014] The steel sheet having the bainite single-phase microstructure has excellent stretch flangeability. The steel sheet having the bainite single-phase microstructure also has excellent blanking workability, because many Fe-based carbides are present in the bainite microstructure and serve as a void origin during blanking. However, since the bainite microstructure varies greatly in mechanical properties such as strength depending on the transformation temperature, there is a possibility that the mechanical properties of the steel sheet vary greatly due to variations in hot rolling conditions such as coiling temperature.

[0015] We then considered reducing the influence of the variations of the hot rolling conditions by tempering a microstructure mainly composed of bainite or bainite and martensite.

[0016] Tempering a bainite or martensite microstructure typically enables a significant reduction of the variations of the mechanical properties caused by the variations of the hot rolling conditions, but also leads to a significant decrease in steel sheet strength. Besides, since Fe-based carbide morphology in tempered bainite or tempered martensite phase varies depending on the annealing conditions, the steel sheet may not be able to have excellent blanking workability depending on the annealing conditions.

[0017] In view of this, we carefully examined a technique of preventing such a decrease in steel sheet strength and achieving excellent stretch flangeability and blanking workability when tempering the microstructure mainly composed of bainite or bainite and martensite.

[0018] We consequently discovered that dispersing and precipitating MC-type carbides such as TiC inside and at the boundaries of laths inhibits the coarsening of laths during annealing and the disappearance of laths resulting from recovery, so that high steel sheet strength can be maintained even after annealing. We also discovered that excellent blanking workability is achieved by ensuring that Fe-based carbides with an aspect ratio of 5 or less make up at least a predetermined proportion in Fe-based carbides precipitated inside and at the boundaries of laths.

[0019] Upon further examination, we discovered that the aforementioned steel sheet microstructure can be stably obtained particularly by adding 0.03% or more Ti and appropriately adjusting heat hysteresis in the annealing.

[0020] MC-type carbides are carbides, such as TiC, NbC, VC, and (Ti, Mo)C, with an atom ratio between an M element (for example, Ti, Nb, V, or Mo) and C of approximately 1:1. The M element need not be of one type, and a complex carbide containing a plurality of metal elements is applicable. A N-containing carbonitride or complex carbonitride is applicable, too.

[0021] Upon further examination, we also discovered that, by appropriately controlling heat hysteresis when cooling the steel sheet from the maximum heating temperature to the room temperature in the annealing, the formation of the balance other than tempered martensite phase and tempered bainite phase, especially the formation of martensite phase, coarse pearlite phase, and retained austenite phase, is suppressed, as a result of which excellent stretch flangeability can be achieved in addition to high strength and excellent blanking workability.

[0022] The disclosure is based on these discoveries and further studies.

[0023] We thus provide:

A hot rolled steel sheet according to claims 1 and 2, and a method of manufacturing a hot rolled steel sheet according to claims 3 and 4.

(Advantageous Effect)

[0024] It is possible to obtain a hot rolled steel sheet that has high strength such as a tensile strength (TS) of 780 MPa or more and excellent stretch flangeability and blanking workability and whose variations in mechanical properties caused by variations in manufacturing conditions are reduced, which is suitable for structural-use steel material such as material for parts of transport machinery including vehicles and construction steel material. This widens the range of uses of hot rolled steel sheets, and has an industrially significant advantageous effect.

BRIEF DESCRIPTION OF THE DRAWINGS

[0025] In the accompanying drawings:

FIG. 1 is a schematic diagram illustrating an example of a microstructure in which tempered bainite phase and tempered martensite phase have laths as a substructure and Fe-based carbides precipitate and MC-type carbides disperse and precipitate inside and at the boundaries of the laths.

DETAILED DESCRIPTION

Detailed description is given below.

[0026] The chemical composition of a hot rolled steel sheet according to the disclosure is described first. The unit of the content of each element in the chemical composition is "mass%", which is simply expressed as "%" below unless otherwise noted.

C: 0.03% or more and 0.20% or less

[0027] C improves the strength of the steel, and promotes the formation of bainite and martensite during hot rolling. The C content therefore needs to be 0.03% or more. If the C content is more than 0.20%, equivalent carbon content is excessively high, which causes a decrease in weldability of the steel sheet. The C content is therefore 0.03% or more and 0.20% or less. The C content is preferably 0.04% or more. The C content is preferably 0.18% or less. The C content is more preferably more than 0.05%. The C content is more preferably 0.15% or less.

Si: 0.4% or less

[0028] Typically, Si is actively used in a high strength steel sheet as an effective element that improves the steel sheet strength without decreasing ductility (elongation). If the Si content is more than 0.4%, however, Si forms oxides on the steel sheet surface during heat treatment, and degrades coating adhesion property. The Si content is therefore 0.4% or less. The Si content is preferably 0.3% or less. The Si content is more preferably 0.2% or less. The Si content may be reduced to an impurity level, and may be 0%.

Mn: 0.5% or more and 2.0% or less

[0029] Mn is an element that dissolves and contributes to higher strength of the steel. Mn also promotes the formation of bainite and martensite during hot rolling, by improving quench hardenability. To achieve such effects, the Mn content needs to be 0.5% or more. If the Mn content is more than 2.0%, austenite becomes excessively stable, causing the microstructure of the steel sheet to excessively contain martensite and retained austenite. This decreases stretch flangeability. The Mn content is therefore 0.5% or more and 2.0% or less. The Mn content is preferably 0.8% or more. The Mn content is preferably 1.8% or less. The Mn content is more preferably 1.0% or more. The Mn content is more preferably 1.7% or less.

P: 0.03% or less

[0030] P is a harmful element that segregates to grain boundaries to decrease elongation, induce cracking during working, and degrade anti-crash property. The P content is therefore 0.03% or less. Excessive dephosphorization, however, leads to longer refining time and higher cost, and so the P content is preferably 0.002% or more.

S: 0.03% or less

[0031] S exists as MnS or TiS in the steel, and facilitates the formation of voids when blanking the hot rolled steel

sheet. S also serves as a void origin during working, and causes a decrease in stretch flangeability. The S content is therefore desirably as low as possible, and is 0.03% or less. The S content is preferably 0.01% or less. Excessive desulfurization, however, leads to longer refining time and higher cost, and so the S content is preferably 0.0002% or more.

5 Al: 0.1% or less

[0032] Al is an element that acts as a deoxidizing material. To achieve such effects, the Al content is desirably 0.01% or more. If the Al content is more than 0.1%, Al remains in the steel sheet as Al oxide. Such Al oxide tends to coagulate and be coarsened, causing a decrease in stretch flangeability. The Al content is therefore 0.1% or less.

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N: 0.01% or less

[0033] N exists as coarse TiN in the steel, and facilitates the formation of coarse voids when blanking the hot rolled steel sheet. N also serves as an origin of coarse voids during working, and causes a decrease in stretch flangeability. The N content is therefore desirably as low as possible, and is 0.01% or less. The N content is preferably 0.006% or less. Excessive denitrification, however, leads to longer refining time and higher cost, and so the N content is preferably 0.0005% or more.

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Ti: 0.03% or more and 0.15% or less

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[0034] Ti is a necessary element to form MC-type carbides to thus inhibit lath coarsening in the annealing and strengthen the steel sheet. MC-type carbides also enhance the steel sheet strength by strengthening by precipitation. If the Ti content is less than 0.03%, such effects are insufficient, and lath coarsening and lower precipitation amount cause a decrease in steel sheet strength, making it difficult to achieve desired steel sheet strength (tensile strength of 780 MPa or more). If the Ti content is more than 0.15%, central segregation is noticeable, causing a decrease in blanking workability. The Ti content is therefore 0.03% or more and 0.15% or less. The Ti content is preferably 0.04% or more. The Ti content is preferably 0.14% or less. The Ti content is further preferably 0.05% or more. The Ti content is further preferably 0.13% or less.

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[0035] While the basic components have been described above, the hot rolled steel sheet may optionally contain one or more of V: 0.01% or more and 0.3% or less, Nb: 0.01% or more and 0.1% or less, and Mo: 0.01% or more and 0.3% or less, for higher strength.

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V: 0.01% or more and 0.3% or less

[0036] V forms MC-type carbides and contributes to higher strength of the steel sheet by lath coarsening inhibition in the annealing and strengthening by precipitation, as with Ti. To achieve such effects, the V content needs to be 0.01% or more. If the V content is more than 0.3%, central segregation is noticeable, causing a decrease in blanking workability. Accordingly, the V content is preferably 0.01% or more. The V content is preferably 0.3% or less. The V content is more preferably 0.01% or more. The V content is more preferably 0.2% or less. The V content is further preferably 0.01% or more. The V content is further preferably 0.15% or less. V may form MC-type carbides by itself, or form complex carbides with Ti, Nb, and Mo. Such carbide composition does not affect the advantageous effects of the disclosure at all.

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Nb: 0.01% or more and 0.1% or less

[0037] Nb forms MC-type carbides and contributes to higher strength of the steel sheet by lath coarsening inhibition in the annealing and strengthening by precipitation, as with Ti. To achieve such effects, the Nb content needs to be 0.01% or more. If Nb is excessively added to be more than 0.1% in content, Nb does not dissolve in the heating furnace during hot rolling. The effects thus saturate, and the alloy cost increases. Accordingly, the Nb content is preferably 0.01% or more. The Nb content is preferably 0.1% or less. The Nb content is more preferably 0.01% or more. The Nb content is more preferably 0.08% or less. The Nb content is further preferably 0.01% or more. The Nb content is further preferably 0.06% or less. Nb may form MC-type carbides by itself, or form complex carbides with Ti, V, and Mo. Such carbide composition does not affect the advantageous effects of the disclosure at all.

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Mo: 0.01% or more and 0.3% or less

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[0038] Mo, when added in combination with Ti, forms MC-type complex carbides and contributes to higher strength of the steel sheet by lath coarsening inhibition in the annealing and strengthening by precipitation, as with Ti. To achieve such effects, the Mo content needs to be 0.01% or more. If the Mo content is more than 0.3%, central segregation is

noticeable, causing a decrease in blanking workability. Accordingly, the Mo content is preferably 0.01% or more. The Mo content is preferably 0.3% or less. Mo may form complex carbides with Nb and V. Such carbide composition does not affect the advantageous effects of the disclosure at all.

[0039] The hot rolled steel sheet may optionally contain B: 0.0002% or more and 0.010% or less, for improved quench hardenability during hot rolling.

B: 0.0002% or more and 0.010% or less

[0040] B is an element that segregates to austenite grain boundaries and inhibits the formation and growth of ferrite to improve quench hardenability and promote the formation of bainite and martensite. To achieve such effects, the B content is preferably 0.0002% or more. If the B content is more than 0.010%, hard iron boride forms and causes a decrease in stretch flangeability. Accordingly, in the case of adding B, the B content is preferably 0.0002% or more and 0.010% or less. The B content is more preferably 0.0002% or more. The B content is more preferably 0.0050% or less. The B content is further preferably 0.0004% or more. The B content is further preferably 0.0030% or less.

[0041] In addition to the composition described above, the hot rolled steel sheet may contain one or more of REM, Zr, As, Cu, Ni, Sn, Pb, Ta, W, Cr, Sb, Mg, Ca, Co, Se, Zn, and Cs so that their total content is 1.0% or less.

[0042] The components other than those described above are Fe and incidental impurities.

[0043] The reasons for limiting the microstructure in the hot rolled steel sheet are given below.

Total area ratio of tempered bainite phase and tempered martensite phase: 70% or more

[0044] The hot rolled steel sheet has a microstructure mainly composed of tempered bainite and tempered martensite having both high strength and excellent blanking workability. If the total area ratio of tempered bainite phase and tempered martensite phase is less than 70%, the hot rolled steel sheet cannot have desired high strength and blanking workability.

Here, the ratio of each of tempered bainite phase and tempered martensite is not individually defined because tempered bainite and tempered martensite after annealing are microstructures not distinguishable from each other. This is a major factor that can reduce variations in mechanical properties after annealing in the case where the manufacturing conditions during hot rolling vary. The total area ratio of tempered bainite phase and tempered martensite phase is therefore 70% or more. The total area ratio of tempered bainite phase and tempered martensite phase is preferably 75% or more. The total area ratio of tempered bainite phase and tempered martensite phase is more preferably 80% or more. The total area ratio of tempered bainite phase and tempered martensite phase may be 100%.

Total area ratio of coarse pearlite phase, martensite phase, and retained austenite phase: 10% or less

[0045] As mentioned above, the microstructure of the hot rolled steel sheet is mainly composed of tempered bainite and tempered martensite, with the balance other than tempered bainite and tempered martensite being, for example, Fe-based carbides, coarse pearlite, fine pearlite, degenerate pearlite, bainite, martensite, and retained austenite. Of these, particularly in the case where coarse pearlite, martensite, and retained austenite are present in the metallic microstructure, stretch flangeability decreases noticeably. The total area ratio of coarse pearlite phase, martensite phase, and retained austenite phase is therefore 10% or less. The total area ratio of coarse pearlite phase, martensite phase, and retained austenite phase is preferably 8% or less. The total area ratio of coarse pearlite phase, martensite phase, and retained austenite phase is more preferably 5% or less. The total area ratio of coarse pearlite phase, martensite phase, and retained austenite phase may be 0%.

[0046] Here, coarse pearlite has a lamellar spacing of 0.2 μm or more, fine pearlite has a lamellar spacing of less than 0.2 μm , and degenerate pearlite is a phase in which pearlite lamellar is not clearly observable. The lamellar spacing can be measured by microstructure observation using a scanning electron microscope.

[0047] The balance other than tempered bainite phase, tempered martensite phase, coarse pearlite phase, martensite phase, and retained austenite phase is, for example, ferrite phase, degenerate pearlite phase, and fine pearlite phase. A total area ratio of such balance of 30% or less is allowable.

Average width of laths which tempered bainite phase and tempered martensite phase have as substructure: 1.0 μm or less

[0048] For strengthening by tempered bainite phase and tempered martensite phase, it is important that tempered bainite phase and tempered martensite phase have fine laths with an average width of 1.0 μm or less as their substructure. FIG. 1 is a schematic diagram illustrating an example of a microstructure in which tempered bainite phase and tempered martensite phase have laths as their substructure and Fe-based carbides precipitate and MC-type carbides disperse and precipitate inside and at the boundaries of the laths. If the laths disappears as a result of recovery or the average width of the laths is more than 1.0 μm , predetermined high strength cannot be achieved. The average width of laths

which tempered bainite phase and tempered martensite phase have as their substructure is therefore 1.0 μm or less. The average width of laths is preferably 0.8 μm or less. The average width of laths is more preferably 0.6 μm or less. No lower limit is placed on the average width of laths, yet the lower limit is typically about 0.1 μm .

- 5 Proportion of Fe-based carbides with an aspect ratio of 5 or less in Fe-based carbides precipitated inside and at the boundaries of laths: 80% or more

10 **[0049]** Fe-based carbides precipitated inside and at the boundaries of laths as illustrated in FIG. 1 serve as a void origin during blanking, thus contributing to improved blanking workability. This effect is particularly high with Fe-based carbides having an aspect ratio of 5 or less. By setting the proportion of such Fe-based carbides to 80% or more, excellent blanking workability can be achieved. The proportion of Fe-based carbides with an aspect ratio of 5 or less in Fe-based carbides precipitated inside and at the boundaries of laths is therefore 80% or more. The proportion is preferably 85% or more. No upper limit is placed on the proportion, yet the upper limit may be 100%.

15 **[0050]** Fe-based carbides are θ carbide (cementite), ε carbide, and the like. An alloying element may be dissolved in the carbides. The aspect ratio is the ratio of the major axis length and minor axis length of Fe-based carbides precipitated inside and at the boundaries of laths.

Average particle size of MC-type carbides dispersed and precipitated inside and at the boundaries of laths: 20 nm or less

20 **[0051]** MC-type carbides finely dispersed and precipitated inside and at the boundaries of laths as illustrated in FIG. 1 inhibit lath coarsening by a pinning effect when annealing the steel sheet and also inhibit lath disappearance resulting from recovery, thus contributing to higher strength. If the average particle size of MC-type carbides is more than 20 nm, the number of particles of MC-type carbides contributing to pinning is insufficient and so the pinning effect is insufficient, causing a decrease in steel sheet strength. If the average particle size of MC-type carbides is 20 nm or less, a sufficient

25 number of particles of MC-type carbides exhibit the pinning effect, to prevent a decrease in steel sheet strength. The average particle size of MC-type carbides dispersed and precipitated inside and at the boundaries of laths of tempered bainite phase and tempered martensite phase is therefore 20 nm or less. The average particle size is preferably 15 nm or less. No lower limit is placed on the average particle size, yet the lower limit is typically about 1 nm. The proportion of MC-type carbides with a particle size of more than 50 nm is preferably 10% or less.

30 **[0052]** It is also important that the hot rolled steel sheet has an average dislocation density in the following range.

Average dislocation density: $1.0 \times 10^{14} \text{ m}^{-2}$ or more and $5.0 \times 10^{15} \text{ m}^{-2}$ or less

35 **[0053]** The variations of the hot rolled steel sheet caused by the variations of the hot rolling conditions are reduced by tempering the steel sheet having bainite and martensite microstructure. If the average dislocation density of the steel sheet after annealing is more than $5.0 \times 10^{15} \text{ m}^{-2}$, the tempering of the steel sheet is insufficient and the influence of the variations of the hot rolling conditions cannot be reduced sufficiently. In the case where tempering is sufficient, the average dislocation density is typically $1.0 \times 10^{14} \text{ m}^{-2}$ or more. The average dislocation density is therefore $1.0 \times 10^{14} \text{ m}^{-2}$ or more and $5.0 \times 10^{15} \text{ m}^{-2}$ or less. The average dislocation density is preferably $1.0 \times 10^{14} \text{ m}^{-2}$ or more. The

40 average dislocation density is preferably $2.0 \times 10^{15} \text{ m}^{-2}$ or less.

[0054] A method of manufacturing the hot rolled steel sheet according to the disclosure is described below.

[0055] The method of manufacturing the hot rolled steel sheet includes: hot rolling a steel raw material having the chemical composition described above, whereby the steel raw material is heated to an austenite single phase region, subjected to rough rolling and finish rolling to obtain a steel sheet, and the steel sheet is cooled and coiled after the

45 finish rolling; pickling the steel sheet after the hot rolling; and then continuous annealing the steel sheet, wherein in the hot rolling, a finisher delivery temperature is 850 °C or more and 1000 °C or less, an average cooling rate to 500 °C after the finish rolling is 30 °C/s or more, and a coiling temperature is 500 °C or less, and in the continuous annealing, a maximum heating temperature of the steel sheet is 700 °C or more and $(A_3 \text{ point} + A_1 \text{ point})/2$ or less, a time during which a temperature of the steel sheet is 600 °C or more and 700 °C or less in heating the steel sheet to the maximum

50 heating temperature is 20 s or more and 1000 s or less, a time during which the temperature of the steel sheet is more than 700 °C is 200 s or less, an average cooling rate to 530 °C when cooling the steel sheet from the maximum heating temperature is 8 °C/s or more and 25 °C/s or less, and a time of holding the steel sheet in a temperature range of 470 °C or more and 530 °C or less after the cooling stops is 10 s or more. The method may further include performing a coating or plating treatment on the steel sheet, after the continuous annealing.

55 **[0056]** The method of obtaining the steel raw material by steelmaking is not limited, and any known steelmaking process such as a converter steelmaking process or an electric furnace steelmaking process may be used. After steelmaking, continuous casting is preferably performed to yield a slab (steel raw material) in terms of productivity and the like. The slab may be yielded by a known casting method such as ingot casting and blooming, thin slab continuous

casting, or the like.

[0057] The obtained steel raw material is subjected to hot rolling, in which the steel raw material is subjected to rough rolling and finish rolling. Before the rough rolling, the steel raw material is heated in the austenite single phase region. If the steel raw material before the rough rolling is not heated in the austenite single phase region, the remelting of Ti carbide and the like present in the steel raw material does not progress, and fine MC-type carbides do not precipitate during the annealing after the hot rolling. Accordingly, the steel raw material is heated to 1150 °C or more, before the rough rolling. No upper limit is placed on the heating temperature, yet an excessively high heating temperature leads to a considerable decrease in yield rate due to the oxidation of the slab surface, and so the heating temperature is typically 1350 °C or less. In the case where the temperature of the cast steel raw material (slab) is in the austenite single phase region when hot rolling the steel raw material, the steel raw material may be subjected to hot direct rolling without being heated or after being heated for a short time.

[0058] The reasons for limiting the manufacturing conditions in the hot rolling are given below.

Finisher delivery temperature: 850 °C or more and 1000 °C or less

[0059] If the finisher delivery temperature is low, ferrite transformation is promoted during the cooling after the rolling, causing a decrease in the bainite and martensite ratio of the hot rolled steel sheet after the hot rolling. This makes it impossible to obtain a predetermined tempered bainite and tempered martensite ratio after the annealing. Accordingly, the finisher delivery temperature needs to be 850 °C or more. The finisher delivery temperature is preferably 880 °C or more. If the finisher delivery temperature is more than 1000 °C, the surface characteristics of the steel sheet degrade. The finisher delivery temperature is therefore 1000 °C or less. The finisher delivery temperature is preferably 970 °C or less. Each of the temperatures such as the finisher delivery temperature and the coiling temperature mentioned here is the temperature of the steel sheet surface.

Cooling rate to 500 °C after finish rolling: 30 °C/s or more

[0060] When cooling the steel sheet after the finish rolling, if the cooling rate is insufficient, ferrite cannot be suppressed adequately, causing a decrease in the bainite and martensite ratio of the hot rolled steel sheet after the hot rolling. This makes it impossible to obtain a predetermined tempered bainite and tempered martensite ratio after the annealing. Accordingly, the cooling rate to 500 °C after the finish rolling needs to be 30 °C/s or more. The cooling rate is preferably 50 °C/s or more. No upper limit is placed on the cooling rate, yet the upper limit is typically about 300 °C/s.

Coiling temperature: 500 °C or less

[0061] Appropriately adjusting the coiling temperature is important in controlling the steel sheet microstructure after the hot rolling. If the coiling temperature is more than 500 °C, the lath width of bainite increases. This makes it impossible to obtain a predetermined lath width of tempered bainite after the annealing. No lower limit is placed on the coiling temperature, yet an excessively low coiling temperature merely leads to higher cooling cost, and so the coiling temperature is preferably 0 °C or more. The coiling temperature is more preferably 200 °C or more.

[0062] After the hot rolling, the hot rolled steel sheet is subjected to pickling and then to continuous annealing. The reasons for limiting the manufacturing conditions in the continuous annealing are given below.

Maximum heating temperature of steel sheet: 700 °C or more and $(A_3 \text{ point} + A_1 \text{ point})/2$ or less

[0063] Appropriately adjusting the maximum heating temperature of the steel sheet in the continuous annealing is important in sufficiently reducing the influence of the variations of the manufacturing conditions in the hot rolling caused by the annealing and achieving desired high strength. If the maximum heating temperature of the steel sheet is less than 700 °C, the dislocation density in bainite and martensite is difficult to be controlled within an appropriate range, and so the influence of the variations of the manufacturing conditions in the hot rolling cannot be reduced sufficiently. Besides, if the heating temperature of the steel sheet is less than 700 °C, the aspect ratio of Fe-based carbides inside and between laths tends to be high, which makes it difficult to set the proportion of Fe-based carbides with an aspect ratio of 5 or less to be in a desired range. If the maximum heating temperature of the steel sheet is more than $(A_3 \text{ point} + A_1 \text{ point})/2$, MC-type carbides coarsen noticeably, as a result of which lath coarsening in bainite and martensite cannot be inhibited adequately. Moreover, austenitizing is promoted, causing a decrease in the bainite and martensite ratio. This makes it impossible to obtain a desired tempered bainite and tempered martensite ratio. The maximum heating temperature of the steel sheet in the continuous annealing is therefore 700 °C or more and $(A_3 \text{ point} + A_1 \text{ point})/2$ or less. The maximum heating temperature is preferably 700 °C or more. The maximum heating temperature is preferably $\{(A_3 \text{ point} + A_1 \text{ point})/2\} - 10$ °C or less.

[0064] The A_1 point and the A_3 point can be calculated according to the following expressions.

$$A_1 \text{ point} = 751 - 26.6 \times [\%C] + 17.6 \times [\%Si] - 11.6 \times [\%Mn] + 22.5 \times [\%Mo] + 233 \times [\%Nb] - 39.7 \times [\%V] - 57 \times [\%Ti] - 895 \times [\%B] - 169 \times [\%Al]$$

$$A_3 \text{ point} = 937 - 476.5 \times [\%C] + 56 \times [\%Si] - 19.7 \times [\%Mn] + 38.1 \times [\%Mo] + 124.8 \times [\%V] + 136.3 \times [\%Ti] - 19 \times [\%Nb] + 3315 \times [\%B]$$

where [%X] denotes the content of an X element in steel (mass%).

[0065] Time during which the steel sheet temperature is 600 °C or more and 700 °C or less in heating the steel sheet to the maximum heating temperature: 20 s or more and 1000 s or less

[0066] In heating the steel sheet to the maximum heating temperature, it is important to appropriately control heat hysteresis in imparting desired high strength and excellent blanking workability to the steel sheet. The pinning effect of MC-type carbides is used to inhibit lath coarsening, as mentioned above. To achieve the pinning effect, MC-type carbides need to be sufficiently dispersed in bainite and martensite before lath coarsening starts. According to our study, the precipitation of MC-type carbides begins to occur noticeably at 600 °C or more. Meanwhile, lath coarsening and disappearance are noticeable at more than 700 °C. Hence, lath coarsening and disappearance can be inhibited by holding the steel sheet temperature in the temperature range of 600 °C or more and 700 °C or less for a predetermined time so that MC-type carbides precipitate sufficiently. For sufficient precipitation of MC-type carbides, the holding time in this temperature range needs to be 20 s or more. If the holding time in the temperature range is insufficient, lath coarsening starts before MC-type carbides precipitate sufficiently, so that the pinning effect is insufficient and the laths coarsen. The holding time is preferably 35 s or more. The holding time is more preferably 50 s or more.

[0067] If the holding time of the steel sheet temperature in the temperature range of 600 °C or more and 700 °C or less is more than 1000 s, Fe-based carbides precipitated inside and between laths dissolve again and move to prior austenite grain boundaries, packet grain boundaries, block grain boundaries, and the like. Thus, Fe-based carbides inside and between laths that effectively contribute to improved blanking workability no longer exist. Accordingly, to obtain a steel sheet having excellent blanking workability, the holding time of the steel sheet temperature in the temperature range of 600 °C or more and 700 °C or less needs to be 1000 s or less. The holding time is preferably 800 s or less. The holding time is more preferably 500 s or less. The steel sheet temperature mentioned here is the temperature of the steel sheet surface.

Time during which the steel sheet temperature is more than 700 °C: 10 s or more and 200 s or less

[0068] When the steel sheet temperature is in the temperature range of more than 700 °C, lath coarsening is noticeable. The pinning effect of MC-type carbides finely dispersed and precipitated is used to prevent the movement of lath boundaries and inhibit lath coarsening, as mentioned above. The steel sheet strength is maintained in this way. If the holding time in this temperature range is excessively long, however, lath coarsening cannot be inhibited. Accordingly, the holding time of the steel sheet temperature in the temperature range of more than 700 °C is 200 s or less, in terms of preventing lath coarsening. The holding time is preferably 180 s or less. The holding time is more preferably 150 s or less. If the time during which the steel sheet temperature is more than 700 °C is less than 10 s, the ductility of the steel sheet decreases to some extent, and so the holding time is 10 s or more.

Average cooling rate to 530 °C when cooling the steel sheet from the maximum heating temperature: 8 °C/s or more and 25 °C/s or less

[0069] When cooling the steel sheet after heating it to the maximum heating temperature in the continuous annealing, it is important to appropriately control the cooling rate in achieving excellent stretch flangeability. Particularly in the case where the average cooling rate to 530 °C is less than 8 °C/s, pearlite transformation cannot be suppressed during the cooling, as a result of which coarse pearlite forms in a predetermined amount or more. This decreases stretch flangeability. If the average cooling rate is excessively high, holding the steel sheet in the temperature range of 470 °C or more and 530 °C or less for a predetermined time as mentioned below is difficult. The average cooling rate to 530 °C when cooling the steel sheet from the maximum heating temperature is therefore 25 °C/s or less.

Holding time in the temperature range of 470 °C or more and 530 °C or less: 10 s or more

[0070] In the continuous annealing, it is important to hold the steel sheet in the temperature range of 470 °C or more and 530 °C or less for a predetermined time after the aforementioned controlled cooling, in achieving excellent stretch flangeability. If the holding temperature after the cooling stops is more than 530 °C, coarse pearlite forms, causing a decrease in stretch flangeability. If the holding temperature after the cooling stops is less than 470 °C, the transformation from austenite to bainite delays. As a result, C concentrates in the non-transformed austenite region to stabilize austenite, hampering the completion of the transformation. In the subsequent cooling, non-transformed austenite transforms to martensite or remains in the steel sheet microstructure as retained austenite, so that stretch flangeability decreases. In the case where the steel sheet is held in the temperature range of 470 °C or more and 530 °C or less for 10 s or more, the transformation of most austenite to bainite completes, with it being possible to reduce the proportion of martensite which forms during the subsequent cooling to a predetermined range. Accordingly, the holding time in the temperature range of 470 °C or more and 530 °C or less after the controlled cooling stops is 10 s or more. The holding time is preferably 20 s or more. The holding time is more preferably 30 s or more. No upper limit is placed on the holding time, yet the holding time is typically 300 s or less.

[0071] Holding the steel sheet in the temperature range of 470 °C or more and 530 °C or less completes the control of the steel sheet microstructure. The subsequent cooling conditions are not limited, and the steel sheet may be cooled to room temperature by any cooling method.

[0072] Even in the case of reheating the steel sheet to 700 °C or less after holding the steel sheet in the temperature range of 470 °C or more and 530 °C or less, desired steel sheet microstructure can still be obtained as long as the total holding time in the temperature range of 600 °C or more and 700 °C or less is 1000 s or less.

[0073] For example, after holding the steel sheet in the temperature range of 470 °C or more and 530 °C or less, the steel sheet may be immersed in a zinc pot to yield a hot-dip galvanized steel sheet. The steel sheet may then be further heated to yield a galvannealed steel sheet. The hot dip coating is not limited to zinc, and may be a coating of aluminum, an aluminum alloy, or the like.

[0074] After the continuous annealing, the steel sheet may be subjected to temper rolling either continuously in the annealing line or using another line according to a conventional method.

[0075] The hot rolled steel sheet manufactured as described above may be electrogalvanized or hot-dip galvanized. The hot rolled steel sheet according to the disclosure is suitable not only as a steel sheet for automotive suspension parts but also for press forming typically performed at ordinary temperature, and has excellent heat resistance. Hence, the hot rolled steel sheet manufactured as described above is also suitable as a blank sheet for a warm forming process of heating a steel sheet to 400 °C to 700 °C before pressing and then immediately press forming the steel sheet.

EXAMPLES

[0076] Molten steels having the compositions listed in Table 1 were each obtained by steelmaking and subjected to continuous casting by a typically known technique, to yield a slab (steel raw material) with a thickness of 300 mm. The slab was heated to the temperature in Table 2, rough rolled, and finish rolled at the finisher delivery temperature in Table 2. After completing the finish rolling, the steel sheet was cooled at the average cooling rate in Table 2, and coiled at the coiling temperature in Table 2, to obtain a hot rolled steel sheet with a sheet thickness of 3.2 mm. The hot rolled steel sheet was then pickled by a typically known technique, and annealed in a continuous annealing line under the conditions in Table 2. Some of the steel sheets were subjected to hot-dip galvanizing treatment and optionally further subjected to alloying treatment in the continuous annealing line, thus yielding hot-dip galvanized steel sheets and galvannealed steel sheets.

Table 1

Steel No.	Chemical composition (mass%)													Remarks
	C	Si	Mn	P	S	Al	N	Ti	V	Nb	Mo	B	Others	
1	0.024	0.06	0.7	0.025	0.002	0.04	0.0032	0.095	-	-	-	-	-	Comparative steel
2	0.060	0.03	0.6	0.027	0.002	0.04	0.0057	0.090	-	-	-	-	-	Conforming steel
3	0.140	0.05	0.7	0.026	0.003	0.03	0.0029	0.056	-	-	-	-	-	Conforming steel
4	0.081	0.31	0.4	0.014	0.002	0.05	0.0035	0.096	-	-	-	-	-	Comparative steel
5	0.128	0.17	1.2	0.020	0.002	0.04	0.0064	0.089	-	-	-	-	-	Conforming steel
6	0.145	0.29	1.8	0.015	0.002	0.03	0.0037	0.082	-	-	-	-	-	Conforming steel
7	0.177	0.07	2.5	0.025	0.002	0.03	0.0064	0.107	-	-	-	-	-	Comparative steel
8	0.099	0.15	1.1	0.021	0.001	0.05	0.0027	0.024	-	-	-	-	-	Comparative steel
9	0.147	0.30	1.6	0.014	0.003	0.03	0.0060	0.062	-	-	-	-	-	Conforming steel
10	0.161	0.23	1.4	0.018	0.001	0.03	0.0047	0.113	-	-	-	-	-	Conforming steel
11	0.053	0.30	1.6	0.015	0.002	0.06	0.0029	0.195	-	-	-	-	-	Comparative steel
12	0.114	0.16	1.1	0.021	0.003	0.04	0.0056	0.048	-	-	-	-	-	Conforming steel
13	0.163	0.10	0.9	0.024	0.003	0.03	0.0030	0.050	-	-	-	-	-	Conforming steel
14	0.183	0.11	0.9	0.023	0.003	0.02	0.0031	0.045	-	-	-	-	-	Conforming steel
15	0.049	0.19	1.2	0.019	0.002	0.06	0.0049	0.097	-	-	-	-	-	Conforming steel
16	0.153	0.25	1.4	0.017	0.001	0.03	0.0051	0.134	-	-	-	-	-	Conforming steel
17	0.163	0.19	1.2	0.019	0.003	0.03	0.0035	0.060	-	-	-	-	-	Conforming steel
18	0.111	0.16	1.1	0.021	0.002	0.04	0.0037	0.072	-	-	-	-	-	Conforming steel
19	0.056	0.15	1.1	0.021	0.002	0.06	0.0032	0.089	-	-	-	-	-	Conforming steel
20	0.112	0.22	1.3	0.018	0.001	0.04	0.0037	0.123	-	-	-	-	-	Conforming steel
21	0.128	0.22	1.3	0.018	0.003	0.04	0.0037	0.057	-	-	-	-	-	Conforming steel
22	0.120	0.17	1.1	0.020	0.002	0.04	0.0045	0.108	-	-	-	-	-	Conforming steel
23	0.129	0.03	1.5	0.011	0.002	0.04	0.0037	0.065	-	-	-	0.0015	-	Conforming steel
24	0.144	0.24	1.4	0.017	0.002	0.03	0.0054	0.109	0.092	-	-	-	As: 0.003	Conforming steel

(continued)

Steel No.	Chemical composition (mass%)													Remarks
	C	Si	Mn	P	S	Al	N	Ti	V	Nb	Mo	B	Others	
25	0.173	0.35	1.8	0.012	0.001	0.03	0.0051	0.113	-	0.043	-	-	Ca: 0.004, Sn: 0.002	Conforming steel
26	0.048	0.34	1.8	0.012	0.001	0.06	0.0040	0.130	-	-	0.15	-	Se: 0.005, Cr:0.23	Conforming steel
27	0.174	0.37	1.9	0.011	0.001	0.03	0.0046	0.125	-	-	-	0.0019	Sb: 0.004, Cu: 0.09, Ni: 0.18	Conforming steel
28	0.139	0.32	1.7	0.014	0.002	0.04	0.0057	0.074	-	-	-	0.0013	Co: 0.028, Cs:0.004	Conformity steel
29	0.102	0.09	0.8	0.024	0.002	0.04	0.0055	0.099	-	-	-	0.0018	Mg: 0.006, Ta: 0.03	Conforming steel
30	0.090	0.26	1.5	0.017	0.002	0.05	0.0050	0.082	-	-	-	0.0017	Pb: 0.004, W: 0.12	Conforming steel
31	0.169	0.13	1.0	0.022	0.002	0.03	0.0044	0.079	0.046	-	0.10	-	Zr: 0.04	Conforming steel
32	0.109	0.33	1.7	0.013	0.003	0.04	0.0056	0.065	-	-	-	-	Zn: 0.0012	Conforming steel
33	0.189	0.37	1.9	0.011	0.002	0.02	0.0030	0.075	-	-	-	-	REM: 0.06	Conforming steel

Table 2

Steel sheet No	Steel No.	Hot rolling conditions				Continuous annealing conditions							Albying treatment	Remarks
		Slab heating temperature (°C)	Finishes delivery temperature (°C)	Average cooling rate*1 (°C/s)	Coiling temperature (°C)	Maximum heating temperature (°C)	A ₁ *2 (°C)	A ₃ *3 (°C)	(A ₃ +A ₁)/2 (°C)	Holding time 1*4 (s)	Holding time 2*5 (s)	Average cooling rate*6 (°C/s)	Holding time 3*7 (s)	
1	1	1220	920	63	380	720	730	927	829	120	20	12	110	Comparative Example
2	2	1220	910	81	390	800	731	910	820	130	70	12	110	Example
3	3	1220	860	86	460	710	732	867	800	190	20	15	110	Example
4	4	1220	950	46	390	830	736	921	829	150	100	13	80	Comparative Example
5	5	1220	890	64	410	730	725	874	800	170	30	14	60	Example
6	6	1220	950	78	360	790	721	860	790	210	70	16	110	Example
7	1	1220	950	68	460	760	708	822	765	40	50	9	70	Comparative Example
8	8	1220	850	47	340	710	730	881	805	110	20	12	90	Comparative Example
9	9	1220	920	52	460	780	724	960	792	170	60	14	50	Example
10	10	1200	900	83	330	750	724	862	793	190	40	15	70	Example
11	11	1200	860	52	360	710	715	923	819	60	20	9	50	Comparative Example
12	12	1200	900	60	470	780	728	876	802	130	60	13	80	Comparative Example
13	13	1200	830	60	470		731	854	792	190	20	15	100	Comparative Example
14	14	1200	870	17	480	710	731	844	787	220	20	16	90	
15	15	1220	900	80	560	770	724	914	819	50	50	9	80	Comparative Example
16	16	1220	900	76	300	640	722	868	795	180	0	14	60	Comparative Example

(continued)

Steel sheet No	Steel No.	Hot rolling conditions				Continuous annealing conditions							Conting or Plating treatment	Albying treatment	Remarks
		Slab heating temperature (°C)	Finishes delivery temperature (°C)	Average cooling rate ^{*1} (°C/s)	Coiling temperature (°C)	Maximum heating temperature (°C)	A ₁ ^{*2} (°C)	A ₃ ^{*3} (°C)	(A ₃ +A ₁)/2 (°C)	Holding time 1 ^{*4} (s)	Holding time 2 ^{*5} (s)	Average cooling rate ^{*6} (°C/s)	Holding time 3 ^{*7} (s)		
17	17	1220	890	59	450	850	728	854	791	190	80	15	80		Comparative Example
18	18	1220	890	76	430	730	727	881	804	15	30	12	80		Comparative Example
19	19	1220	880	70	390	720	725	910	817	1150	20	10	90		Comparative Example
20	20	1220	890	49	320	730	722	887	804	130	230	12	70		Comparator Example
21	21	1220	890	84	460	770	726	870	798	150	20	4	70		Comparative Example
22	22	1220	900	69	350	760	725	881	803	140	30	13	5		Comparative Example
23	23	1180	920	58	450	740	719	861	790	110	50	13	70	Performed	Example
24	24	1180	900	58	350	770	720	880	800	170	50	14	70	Performed	Example
25	25	1220	900	45	340	760	731	853	792	210	40	15	40	Performed	Example
26	26	1220	890	49	310	750	721	922	821	50	50	9	40	Performed	Example
27	27	1220	900	85	320	750	718	861	789	210	50	15	40	Performed	Example
28	28	1220	910	57	420	780	722	869	796	160	40	14	50	Performed	Example
29	29	1220	900	79	370	780	725	896	811	120	40	12	100	Performed	Example
30	30	1220	900	76	410	770	722	896	809	100	60	11	60		Example
31	31	1220	890	78	410	750	729	865	797	450	60	15	90		Example
32	32	1220	910	60	440	780	723	878	801	190	50	12	50		Example

(continued)

Steel sheet No	Steel No.	Hot rolling conditions				Continuous annealing conditions							Coating or Plating treatment	Alloying treatment	Remarks
		Slab heating temperature (°C)	Finishes delivery temperature (°C)	Average cooling rate*1 (°C/s)	Coiling temperature (°C)	Maximum heating temperature (°C)	A ₁ *2 (°C)	A ₃ *3 (°C)	(A ₃ +A ₁)/2 (°C)	Holding time 1*4 (s)	Holding time 2*5 (s)	Average cooling rate*6 (°C/s)	Holding time 3*7 (s)		
33	33	1220	870	75	420	710	722	841	782	540	40	16	40		Example

*1 average cooling rate until steel sheet temperature reaches 500°C after finish rolling
 *2 A₁ point = $751 - 26.6 \times [\%C] + 17.6 \times [\%Si] - 11.6 \times [\%Mn] + 22.5 \times [\%Mo] + 233 \times [\%Nb] - 39.7 \times [\%V] - 57 \times [\%Ti] - 895 \times [\%B] - 169 \times [\%Al]$ ([%X] is content of X element in steel (mass%))
 *3 A₃ point = $937 - 476.55 \times [\%C] + 56 \times [\%Si] - 19.7 \times [\%Mn] + 38.1 \times [\%Mo] + 124.8 \times [\%V] + 136.3 \times [\%Ti] - 19 \times [\%Nb] + 3315 \times [\%B]$ ([%X] is content of X element in steel (mass%))
 *4 holding time of steel sheet temperature in temperature range of 600°C or more and 700°C or less
 *5 holding time of steel sheet temperature in temperature range of more than 700°C
 *6 average cooling rate from maximum heating temperature to 530°C
 *7 holding time in temperature range of 470°C or more and 530°C or less after cooling stops

[0077] A test piece was collected from each obtained hot rolled steel sheet, and subjected to microstructure observation, average dislocation density measurement, a tensile test, a hole expansion test, a blanking test, and manufacturing stability evaluation. The evaluation results are listed in Table 3. The test methods are as follows.

(i) Microstructure observation

[0078] A test piece was collected from each obtained hot rolled steel sheet, and polished in a cross-section (L cross-section) parallel to the rolling direction of the test piece and etched by nital. A micrograph taken with a scanning electron microscope (1000, 3000, 5000 magnifications) was used to determine the total area ratio of tempered bainite phase and tempered martensite phase, the area ratio of coarse pearlite phase, the total area ratio of martensite phase and retained austenite phase (MA), and the area ratio of phase other than these, through the use of an image analyzer. It is difficult to distinguish martensite phase and retained austenite phase from each other with a scanning electron micrograph. In this example, however, the total area ratio of coarse pearlite phase, martensite phase, and retained austenite phase is important, and accordingly the total area ratio of martensite phase and retained austenite phase (MA) was determined without distinguishing martensite phase and retained austenite phase from each other.

[0079] Moreover, a thin film made from each hot rolled steel sheet was observed using a transmission electron microscope (TEM), to measure the lath width in tempered bainite and tempered martensite and determine the proportion of Fe-based carbides with an aspect ratio of 5 or less in Fe-based carbides precipitated inside and at the boundaries of laths and the average particle size of MC-type carbides precipitated inside and at the boundaries of laths.

[0080] The lath width in tempered bainite and tempered martensite was measured as follows. In a transmission electron micrograph of 120 mm × 80 mm in size taken for 10 observation fields at 30000 magnifications, five straight lines orthogonal to the major axes of three or more consecutively aligned laths were drawn at intervals of 10 mm, the length of each line segment where the corresponding straight line intersects with the lath boundaries was measured, and the average length of the line segments was set as the average lath width.

[0081] The proportion of Fe-based carbides with an aspect ratio of 5 or less in Fe-based carbides precipitated inside and at the boundaries of laths was determined as follows. In a micrograph taken at 165000 magnifications, the major axis length and the minor axis length were measured for at least 100 particles of Fe-based carbides precipitated inside and at the boundaries of laths for 5 observation fields in total, to calculate the aspect ratio. The proportion of Fe-based carbides with an aspect ratio of 5 or less was thus determined.

[0082] The average particle size of MC-type carbides was determined as follows. In a micrograph taken at 300000 magnifications, the diameter was measured for at least 100 particles of MC-type carbides such as TiC for 5 observation fields in total, and an arithmetic average (average particle size d_{def}) was calculated. The lower limit of the measured particle size was 2 nm.

(ii) Average dislocation density measurement

[0083] A test piece was collected from each obtained hot rolled steel sheet, and the dislocation density of a 1/4 portion in sheet thickness was measured. Assuming that the dislocation density of a 1/4 portion in sheet thickness represents the average dislocation density of the steel sheet, the measurement was set as average dislocation density. The collected test piece was subjected to mechanical grinding and also polishing with oxalic acid for 0.1 mm, to adjust the sample so that the 1/4 portion in sheet thickness was exposed to the surface. Polishing with oxalic acid was intended to remove the layer worked by grinding.

[0084] For the sample adjusted in this way, the strain of the steel sheet was measured by an X-ray diffractometer. With an X-ray diffractometer, the diffraction intensity of (110) plane, (211) plane, and (220) plane of α -iron in the 1/4 portion in sheet thickness was measured using $\text{CoK}\alpha$ rays. The half-value breadth of the peak value of the reflection intensity of each crystal plane was calculated from the obtained measurement chart, and the local strain ϵ' applied to the steel sheet was determined according to the following Expressions (1) and (2).

$$\beta \cos \theta / \lambda = 0.9 / D + 2 \epsilon' \sin \theta / \lambda \quad \dots (1)$$

where β is the half-value breadth of the peak value (the value corrected according to Expression (2) was used), θ is the diffraction angle, λ is the wavelength of $\text{CoK}\alpha$ rays (0.1790 nm), D is the crystallite size (dislocation cell, crystal grain size), and ϵ' is the local strain.

$$\beta^2 = \beta_m^2 - \beta_0^2 \quad \dots (2)$$

[0085] where β_m is the half-value breadth of the peak of the sample subjected to dislocation density measurement, and β_0 is the half-value breadth of the peak of a strain-free sample.

[0086] Here, $\beta \cos \theta / \lambda$ was plotted against $\sin \theta / \lambda$, and ε' and D were calculated from the slope and the intercept. From the obtained local strain ε' , the dislocation density ρ was determined according to the following Expression 3.

$$\rho = 14.4 \varepsilon'^2 / b^2 \quad \dots (3)$$

where b is the Burgers vector (0.248 nm).

(iii) Tensile test

[0087] A JIS No. 5 tensile test piece (JIS Z 2001) was collected from each obtained hot rolled steel sheet so that the direction (C direction) orthogonal to the rolling direction was the tensile direction, and subjected to a tensile test in conformity with JIS Z 2241 to measure yield strength (YS), tensile strength (TS), and elongation (EI).

(iv) Hole expansion test

[0088] A test piece (size: 100 mm $\times$ 100 mm) was collected from each obtained hot rolled steel sheet, and blanked with a hole of 10 mm ϕ in initial diameter d_0 (clearance: 12.5% of the test piece sheet thickness). A hole expansion test was conducted using the test piece. In detail, a conical punch with a vertex angle of 60° was inserted into the hole of 10 mm ϕ in initial diameter d_0 from the punch side at the time of blanking, to expand the hole. The diameter d (mm) of the hole when a crack ran through the steel sheet (test piece) was measured, and the hole expansion ratio λ (%) was calculated according to the following expression.

$$\text{Hole expansion ratio } \lambda (\%) = \{(d - d_0) / d_0\} \times 100.$$

[0089] The stretch flangeability was evaluated as favorable in the case where tensile strength (TS) $\times$ {hole expansion ratio (λ)}^{0.5} was 6200·MPa%^{0.5} or more.

(v) Blanking test

[0090] A test piece (size: 30 mm $\times$ 30 mm) was collected from each obtained hot rolled steel sheet, and blanked with a hole of 10 mm ϕ in diameter d_0 (clearance: 20%, 30% of the test piece sheet thickness). After the blanking, the fracture state of the punched end surface was observed by a microscope (50 magnifications) on the whole circumference of the punch hole, to observe whether or not any crack, chip, or brittle fracture occurred. The blanking workability was evaluated as "pass" if there was no crack, chip, or brittle fracture, and "fail" otherwise.

Table 3

Steel sheet No.	Steel No.	Metallic microstructure						Mechanical properties						Remarks			
		Area ratio ^{*1} (%)				Average width of laths in TB and TM (μm)	Proportion of Fe-based carbides with aspect ratio of 5 or less ^{*2} (%)	Average particle size of MC-type carbides (nm)	Average dislocation density (×10 ⁴ m ⁻²)	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)	Hole expansion ratio λ (%)		TS × λ ^{0.5} (MPa × % ^{0.5})	Blanking workability	
		TB+TM	Coarse P	MA	Coarse P + MA												Balance
1	1	54	0	0	0	46	1.4	96	2	10	674	723	20	92	6935	Pass	Comparative Example
2	2	75	0	2	2	23	0.3	97	10	4	679	803	17	83	7316	Pass	Example
3	3	100	0	0	0	0	0.3	99	2	2	773	823	17	71	6935	Pass	Example
4	4	45	0	0	0	55	1.1	85	17	16	543	638	23	134	7385	Pass	Comparative Example
5	5	100	0	0	0	0	0.6	91	3	10	816	887	18	61	6928	Pass	Example
6	6	72	0	3	3	25	0.3	96	8	5	733	894	16	62	7039	Pass	Example
7	7	71	3	14	17	12	0.5	93	5	8	711	997	15	18	4230	Pass	Comparative Example
8	8	100	0	0	0	0	1.8	86	Not observed	15	592	622	24	116	6753	Pass	Comparative Example
9	9	74	0	3	3	23	0.6	87	3	14	707	846	19	78	7472	Pass	Example
10	10	87	0	1	1	12	0.5	98	8	3	875	992	15	46	6728	Pass	Example
11	11	100	0	0	0	0	0.6	87	5	14	993	1056	15	26	5385	Fail	Comparative Example
12	12	77	0	2	2	21	0.5	90	2	11	668	785	19	93	7570	Pass	Example
13	13	64	0	0	0	36	0.4	90	7	11	678	724	20	93	6982	Pass	Comparative Example
14	14	58	0	0	0	42	0.4	91	4	10	629	673	22	107	6962	Pass	Comparative Example
15	15	0	2	10	12	88	No lath observed	97	2	4	577	730	20	102	7373	Pass	Comparative Example
16	16	100	0	0	0	0	0.6	55	7	56	916	1035	15	43	6787	Fail	Comparative Example
17	17	32	0	2	2	66	1.9	90	28	11	658	722	21	92	6973	Pass	Comparative Example

*1 TB: tempered bainite, TM: tempered martensite, P: pearlite, MA: martensite-retained austenite

*2 all observed Fe-based carbides are cementite

Table 3 (cont'd)

Steel sheet No.	Steel No.	Metallic microstructure							Mechanical properties							Remarks	
		Area ratio ^{*1} (%)				Average width of laths in TB and TM (μm)	Proportion of Fe-based carbides with aspect ratio of 5 or less ^{**} (%)	Average particle size of MC-type carbides (nm)	Average dislocation density (×10 ¹¹ ·m ⁻²)	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)	Hot expansion ratio λ (%)	TS × λ ^{0.5} (MPa × % ^{0.5})	Blanking workability		
		TB+TM	Coarse P	MA	Coarse P + MA												Balance
18	18	97	0	0	0	3	1.1	95	2	6	653	713	21	100	7130	Pass	Comparative Example
19	19	100	0	0	0	0	0.4	No θ inside and between laths	6	8	742	796	18	81	7164	Fail	Comparative Example
20	20	95	0	0	0	5	1.6	No θ aside and between laths	22	15	681	745	20	83	6787	Fail	Comparative Example
21	21	80	10	2	12	8	0.4	98	8	3	730	810	19	53	5897	Pass	Comparative Example
22	22	85	0	13	13	2	0.3	93	3	8	714	928	16	32	5250	Pass	Comparative Example
23	23	89	0	1	1	10	0.6	93	2	12	710	831	19	74	7149	Pass	Example
24	24	80	0	2	2	18	0.4	89	3	12	823	960	16	53	6989	Pass	Example
25	25	84	0	2	2	14	0.6	85	6	16	877	1008	16	46	6837	Pass	Example
26	26	90	1	1	2	8	0.6	86	5	15	802	889	18	62	7000	Pass	Example
27	27	85	0	2	2	13	0.6	98	3	3	917	1039	16	41	6653	Pass	Example
28	28	75	0	3	3	22	0.5	89	6	12	732	867	18	71	7305	Pass	Example
29	29	79	0	2	2	19	0.2	96	5	5	751	879	16	68	7248	Pass	Example
30	30	82	0	2	2	16	0.4	95	4	6	714	821	19	80	7343	Pass	Example
31	31	89	0	1	1	10	0.3	96	4	5	817	917	16	58	6984	Pass	Example
32	32	76	0	2	2	22	0.5	90	7	11	683	805	20	86	7465	Pass	Example
33	33	100	0	0	0	0	0.6	95	7	6	873	933	17	52	6728	Pass	Example

*1 TB: tempered bainite, TM: tempered martensite, P: pearlite, MA: martensite+retained austenite

*2 all observed Fe-based carbides are cementite

[0091] As can be seen from Table 3, in all Examples, a hot rolled steel sheet having high strength such as a tensile

strength (TS) of 780 MPa or more and excellent stretch flangeability and blanking workability was obtained.

[0092] To evaluate the variations of the mechanical properties of each steel sheet, 100 JIS No. 5 tensile test pieces (JIS Z 2001) were optionally collected from the whole length and whole width of the hot rolled steel sheets of Examples so that the orthogonal direction (C direction) was the tensile direction. A tensile test was conducted in conformity with JIS Z 2241 to measure the tensile strength (TS), and their standard deviation σ was calculated. In all Examples, the standard deviation of the tensile strength (TS) was 10 MPa or less.

[0093] Thus, in all Examples, the mechanical properties of the steel sheet such as tensile strength (TS) had little variations, exhibiting excellent manufacturing stability.

Claims

1. A hot rolled steel sheet comprising:

a composition containing, in mass%,
 C: 0.03% or more and 0.20% or less,
 Si: 0.4% or less,
 Mn: 0.5% or more and 2.0% or less,
 P: 0.03% or less,
 S: 0.03% or less,
 Al: 0.1% or less,
 N: 0.01% or less, and
 Ti: 0.03% or more and 0.15% or less,
 optionally one or more of
 V: 0.01% or more and 0.3% or less,
 Nb: 0.01% or more and 0.1% or less,
 Mo: 0.01% or more and 0.3% or less,
 B: 0.0002% or more and 0.010% or less,
 REM, Zr, As, Cu, Ni, Sn, Pb, Ta, W, Cr, Sb, Mg, Ca, Co, Se, Zn, and Cs: 1.0% or less in total,
 with a balance being Fe and incidental impurities;
 a microstructure in which a total area ratio of a tempered bainite phase and a tempered martensite phase is 70% or more, a total area ratio of a coarse pearlite phase, a martensite phase, and a retained austenite phase is 10% or less, the tempered bainite phase and the tempered martensite phase have laths with an average width of 1.0 μm or less as a substructure, a proportion of Fe-based carbides with an aspect ratio of 5 or less in Fe-based carbides precipitated inside and at boundaries of the laths is 80% or more, and MC-type carbides with an average particle size of 20 nm or less are dispersed and precipitated inside and at the boundaries of the laths; and
 an average dislocation density of $1.0 \times 10^{14} \text{ m}^{-2}$ or more and $5.0 \times 10^{15} \text{ m}^{-2}$ or less.

2. The hot rolled steel sheet according to claim 1, comprising a coated or plated layer on a surface thereof.

3. A method of manufacturing a hot rolled steel sheet, comprising:

hot rolling a steel raw material having the composition according to claim 1, whereby the steel raw material is heated to an austenite single phase region at a temperature of 1150 °C or more and subjected to rough rolling and finish rolling to obtain a steel sheet, and the steel sheet is cooled and coiled after the finish rolling;
 pickling the steel sheet after the hot rolling; and
 then continuous annealing the steel sheet,
 wherein in the hot rolling, a finisher delivery temperature is 850 °C or more and 1000 °C or less, an average cooling rate to 500 °C after the finish rolling is 30 °C/s or more, and a coiling temperature is 500 °C or less, and in the continuous annealing, a maximum heating temperature of the steel sheet is 700 °C or more and $(A_3 \text{ point} + A_1 \text{ point})/2$ or less, a time during which a temperature of the steel sheet is 600 °C or more and 700 °C or less in heating the steel sheet to the maximum heating temperature is 20 s or more and 1000 s or less, a time during which the temperature of the steel sheet is more than 700 °C is 10 s or more and 200 s or less, an average cooling rate to 530 °C when cooling the steel sheet from the maximum heating temperature is 8 °C/s or more and 25 °C/s or less, and a time of holding the steel sheet in a temperature range of 470 °C or more and 530 °C or less after the cooling stops is 10 s or more.

4. The method of manufacturing a hot rolled steel sheet according to claim 3, further comprising performing a coating or plating treatment on the steel sheet, after the continuous annealing.

5 Patentansprüche

1. Ein warmgewalztes Stahlblech, umfassend:

eine Zusammensetzung in Massen-% enthaltend

C: 0,03% oder mehr und 0,20% oder weniger,

Si: 0,4% oder weniger,

Mn: 0,5% oder mehr und 2,0% oder weniger,

P: 0,03% oder weniger,

S: 0,03% oder weniger,

Al: 0,1% oder weniger,

N: 0,01% oder weniger, und

Ti: 0,03% oder mehr und 0,15% oder weniger,

optional eines oder mehrere von

V: 0,01% oder mehr und 0,3% oder weniger,

Nb: 0,01% oder mehr und 0,1% oder weniger,

Mo: 0,01% oder mehr und 0,3% oder weniger,

B: 0,0002% oder mehr und 0,010% oder weniger,

REM, Zr, As, Cu, Ni, Sn, Pb, Ta, W, Cr, Sb, Mg, Ca, Co, Se, Zn und Cs: insgesamt 1,0% oder weniger,

mit einem Gleichgewicht enthaltend Fe und beiläufige Verunreinigungen;

eine Mikrostruktur, in der ein Gesamtflächenverhältnis einer angelassenen Bainitphase und einer angelassenen

Martensitphase 70% oder mehr beträgt, ein Gesamtflächenverhältnis einer groben Perlitphase, einer Marten-

sitphase und einer zurückgehaltenen Austenitphase 10% oder weniger beträgt, die angelassene Bainitphase

und die angelassene Martensitphase Latten mit einer durchschnittlichen Breite von 1,0 µm oder weniger als

Unterstruktur haben, einen Anteil von Fe-basierten Karbiden mit einem Aspektverhältnis von 5 oder weniger in

Fe-basierten Karbiden, der innerhalb und an den Grenzen der Latten abgeschieden wird bei 80% oder mehr

liegt, und Karbide vom MC-Typ mit einer durchschnittlichen Teilchengröße von 20 nm oder weniger innerhalb

und an den Grenzen der Latten dispergiert und abgeschieden werden; und

eine durchschnittliche Versetzungsdichte von $1,0 \times 10^{14} \text{ m}^{-2}$ oder mehr und $5,0 \times 10^{15} \text{ m}^{-2}$ oder weniger.

2. Das warmgewalzte Stahlblech nach Anspruch 1, umfassend eine beschichtete oder plattierte Schicht auf einer Oberfläche davon.

3. Ein Verfahren zur Herstellung eines warmgewalzten Stahlblechs, umfassend:

Warmwalzen eines Stahlrohrmaterials mit der Zusammensetzung nach Anspruch 1, wobei das Stahlrohrmaterial bei einer Temperatur von 1150 °C oder mehr auf einen Austenit-Einphasenbereich erhitzt und einem Vorwalzen und Feinwalzen unterzogen wird, um ein Stahlblech zu erhalten, und das Stahlblech nach dem Feinwalzen abzukühlen und aufzuwickeln;

Beizen des Stahlblechs nach dem Warmwalzen; und

dann kontinuierliches Glühen des Stahlblechs,

wobei beim Warmwalzen eine Fertigungs-Auslegungstemperatur 850 °C oder mehr und 1000 °C oder weniger

beträgt, eine durchschnittliche Abkühlgeschwindigkeit auf 500 °C nach dem Feinwalzen 30 °C/s oder mehr

beträgt und eine Aufwickeltemperatur 500 °C oder weniger beträgt, und

beim kontinuierlichen Glühen eine maximale Erwärmungstemperatur des Stahlblechs 700 °C oder mehr und

(A3-Punkt + A1-Punkt)/2 oder weniger beträgt, eine Zeit, während der eine Temperatur des Stahlblechs 600

°C oder mehr und 700 °C oder weniger vorliegt, beim Erwärmen des Stahlblechs auf die maximale Erwär-

mungstemperatur 20 s oder mehr und 1000 s oder weniger beträgt, eine Zeit, während der die Temperatur des

Stahlblechs bei mehr als 700 °C liegt, 10 s oder mehr und 200 s oder mehr beträgt, eine durchschnittliche

Abkühlgeschwindigkeit auf 530 °C, wenn das Stahlblech von der maximalen Erwärmungstemperatur abgekühlt

wird, 8 °C/s oder mehr und 25 °C/s oder weniger beträgt, und eine Zeit des Haltens des Stahlblechs in einem

Temperaturbereich von 470 °C oder mehr und 530 °C oder weniger nach dem Abkühlen beträgt 10 s oder mehr

beträgt.

4. Das Verfahren zur Herstellung eines warmgewalzten Stahlblechs nach Anspruch 3, ferner umfassend das Durchführen einer Beschichtung oder Plattierungsbehandlung auf dem Stahlblech nach dem kontinuierlichen Glühen.

Revendications

1. Tôle d'acier laminée à chaud comprenant :

une composition contenant, en % en masse,

C : 0,03 % ou plus et 0,20 % ou moins,

Si : 0,4 % ou moins,

Mn : 0,5 % ou plus et 2,0 % ou moins,

P : 0,03 % ou moins,

S : 0,03 % ou moins,

Al : 0,1 % ou moins,

N : 0,01 % ou moins, et

Ti : 0,03 % ou plus et 0,15 % ou moins,

éventuellement l'un ou plus de

V : 0,01 % ou plus et 0,3 % ou moins,

Nb : 0,01 % ou plus et 0,1 % ou moins,

Mo : 0,01 % ou plus et 0,3 % ou moins,

B : 0,0002 % ou plus et 0,010 % ou moins,

terres rares, Zr, As, Cu, Ni, Sn, Pb, Ta, W, Cr, Sb, Mg, Ca, Co, Se, Zn et

Cs : 1,0 % ou moins au total,

le reste étant du Fe et des impuretés inévitables ;

une microstructure dans laquelle un rapport surfacique total de phase de bainite revenue et d'une phase de martensite revenue est supérieur ou égal à 70 %, un rapport surfacique total d'une phase de perlite grossière,

d'une phase de martensite et d'une phase d'austénite résiduelle est inférieur ou égal à 10 %, la phase de martensite revenue et la phase de martensite revenue ont des lattes avec une largeur moyenne de 1,0 μm ou

moins comme sous-structure, une proportion de carbures à base de Fe avec un rapport d'aspect inférieur ou égal à 5 dans des carbures à base de Fe précipités à l'intérieur et au niveau des joints des lattes est supérieur

ou égal à 80 %, et des carbures de type MC avec une taille moyenne de particule de 20 nm ou moins sont dispersés et précipités à l'intérieur et au niveau des joints des lattes ;

et une densité moyenne de dislocation de $1,0 \times 10^{14} \text{ m}^{-2}$ ou plus et $5,0 \times 10^{15} \text{ m}^{-2}$ ou moins.

2. Tôle d'acier laminée à chaud selon la revendication 1, comprenant une couche revêtue ou plaquée sur l'une de ses surfaces.

3. Procédé de fabrication d'une tôle d'acier laminée à chaud, comprenant :

le laminage à chaud d'une matière première d'acier ayant la composition selon la revendication 1, moyennant quoi la matière première d'acier est chauffée jusqu'à une région d'austénite monophasique à une température de 1150 °C ou plus et soumise à un laminage de dégrossissage et un laminage de finissage pour obtenir une tôle d'acier, et la tôle d'acier est refroidie et enroulée après le laminage de finissage ;

le décapage de la tôle d'acier après le laminage à chaud ; et

le recuit continu de la tôle d'acier,

où lors du laminage à chaud, une température de sortie de finissage est de 850 °C ou plus et 1000 °C ou moins,

une vitesse de refroidissement moyenne jusqu'à 500 °C après le laminage de finissage est de 30 °C/s ou plus

et une température d'enroulement est de 500 °C ou moins, et

lors du recuit continu, une température de chauffage maximale de la tôle d'acier est de 700 °C ou plus et de (point A_3 + point A_1)/2 ou moins, une durée pendant laquelle une température de la tôle d'acier est de 600 °C

ou plus et de 700 °C ou moins lors du chauffage de la tôle d'acier à la température de chauffage maximale est de 20 s ou plus et 1000 s ou moins, une durée pendant laquelle la température de la tôle d'acier est supérieure

à 700 °C est de 10 s ou plus et 200 s ou moins, une vitesse de refroidissement moyenne jusqu'à 530 °C lors du refroidissement de la tôle d'acier depuis la température de chauffage maximale est de 8 °C/s ou plus et 25

°C/s ou moins, et une durée de maintien de la tôle d'acier dans une plage de température de 470 °C ou plus et 530 °C ou moins après l'arrêt du refroidissement est de 10 s ou plus.

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4. Procédé de fabrication d'une tôle d'acier laminée à chaud selon la revendication 3, comprenant en outre la réalisation d'un traitement de revêtement ou de placage sur la tôle d'acier, après le recuit continu.

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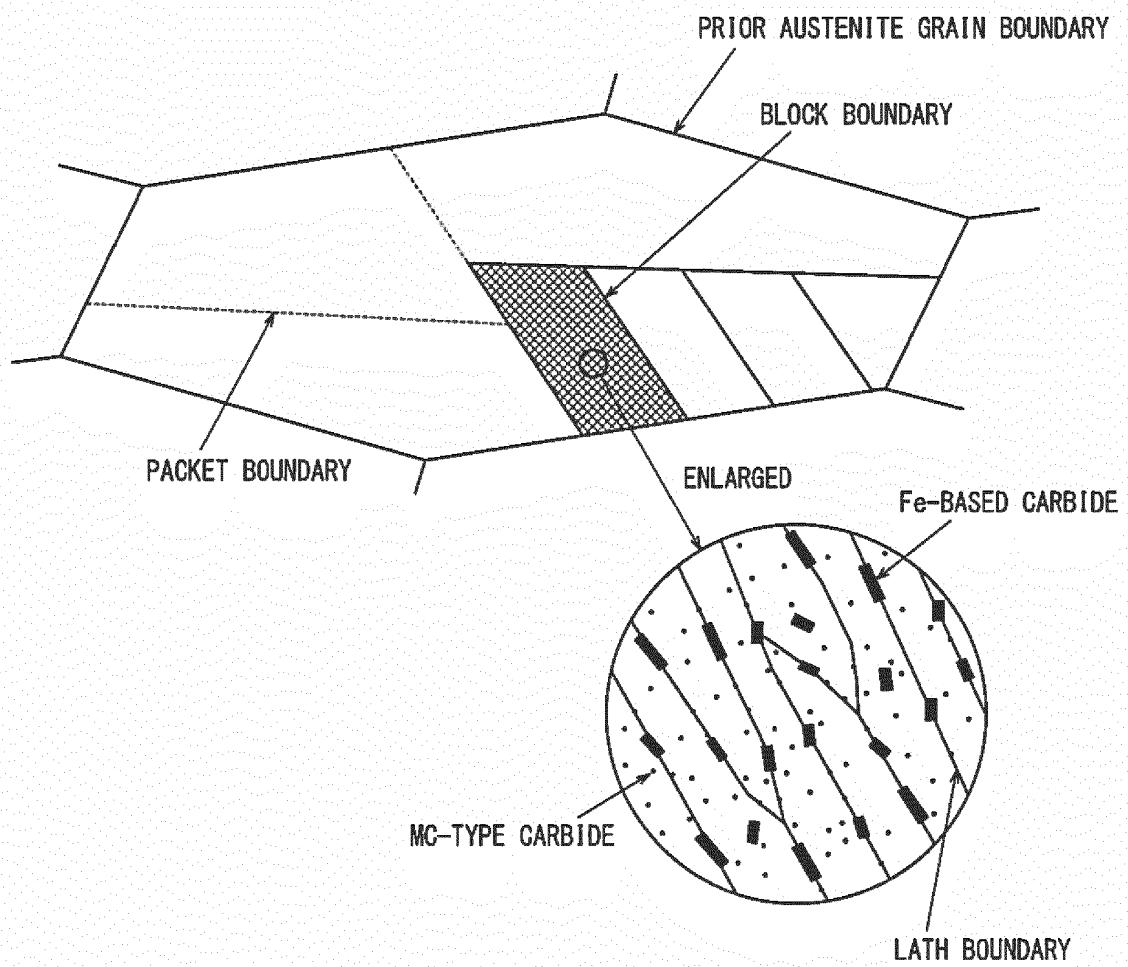
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FIG. 1



REFERENCES CITED IN THE DESCRIPTION

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