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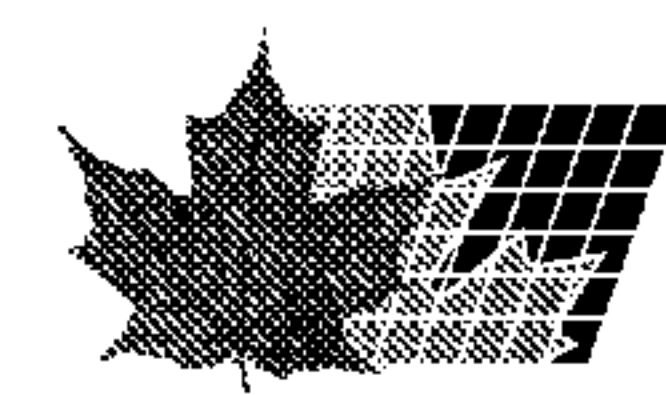
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(54) Titre : TOLE D'ACIER LAMINEE A FROID, A HAUTE RESISTANCE, DOTE E D'UNE OUVRABILITE ET D'UNE RESISTANCE AUX CHOCS REMARQUABLES, ET SON PROCEDE DE FABRICATION  
(54) Title: HIGH-STRENGTH COLD ROLLED STEEL SHEET HAVING EXCELLENT FORMABILITY AND CRASHWORTHINESS AND METHOD FOR MANUFACTURING THE SAME

(57) **Abrégé/Abstract:**

A high-strength cold-rolled steel sheet having excellent ductility and stretch-flangeability, and a method for manufacturing the same are provided. The high-strength cold-rolled steel sheet having excellent workability and impact resistance comprises, by mass, 0.05-0.3% C, 0.3-2.5% Si, 0.5-3.5% Mn, 0.003-0.100% P, 0.02% or less S, 0.010-0.5% Al, with the balance being Fe and inevitable impurities, and has a structure comprising, by surface area, 20% or more ferrite, 10-60% tempered martensite and 0-10% martensite and, by volume, 3-15% residual austenite.



## ABSTRACT

A high-strength cold rolled steel sheet having excellent ductility and stretch flangeability and a method for manufacturing the same are provided. A high-strength cold rolled steel sheet having excellent formability and crashworthiness includes, on a mass% basis, C: 0.05 to 0.3%, Si: 0.3 to 2.5%, Mn: 0.5 to 3.5%, P: 0.003 to 0.100%, S: 0.02% or less, Al: 0.010 to 0.5%, and balance being iron and unavoidable impurities, the high-strength cold rolled steel sheet having a microstructure including 20% or more of ferrite on an area fraction basis, 10 to 60% of tempered martensite on an area fraction basis, 0 to 10% of martensite on an area fraction basis, and 3 to 15% of retained austenite on a volume fraction basis.

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DESCRIPTION

Title of Invention

HIGH-STRENGTH COLD ROLLED SHEET HAVING EXCELLENT FORMABILITY  
AND CRASHWORTHINESS AND METHOD FOR MANUFACTURING THE SAME

Technical Field

[0001]

The present invention relates to a high-strength cold rolled steel sheet having excellent formability for use in structural parts and suspension parts mainly used in the automobile industry field, and to a method for manufacturing the same.

Background Art

[0002]

In recent years, from the viewpoint of preserving global environment, improving fuel efficiency of automobiles has become a key issue. There has been a trend toward increasing the strength of automobile body materials to achieve thickness reduction and decrease the weight of car bodies. However, increasing the strength of steel sheets leads to a decrease in ductility, i.e., a decrease in formability and workability. Thus, development of materials that have both high strength and high formability has been anticipated.

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[0003]

To fulfill such a need, various multi-phase cold rolled steel sheets including ferrite-martensite dual phase steel (hereinafter referred to as "DP steel") and TRIP steel that utilizes the transformation-induced plasticity of retained austenite have been developed.

[0004]

For example, Patent Literature 1 discloses a method for manufacturing a high-strength steel sheet having good formability, with which high ductility is achieved by adding large quantities of Si and thereby reliably obtaining retained austenite.

However, although DP steel and TRIP steel have good elongation properties, they have poor stretch flangeability. The stretch flangeability is an indicator of formability during flange-forming through expanding holes already made, and is a property as important as the elongation property required for high-strength steel sheets.

[0005]

Patent Literature 2 discloses a method for manufacturing a cold rolled steel sheet having good stretch flangeability with which the stretch flangeability is improved by forming a ferrite-tempered martensite multi-phase microstructure by conducting quenching and tempering after annealing and soaking. However, according to this

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technology, although high stretch flangeability is achieved, the elongation is low.

[0006]

According to existing technologies, cold-rolled steel sheets having good elongation property and stretch flangeability have not been obtained.

Citation List

Patent Literature

[0007]

PTL 1: Japanese Unexamined Patent Application  
Publication No. 2-101117

PTL 2: Japanese Unexamined Patent Application  
Publication No. 2004-256872

Summary of Invention

Technical Problem

[0008]

The present invention has been made by addressing the problems described above and an object of the present invention is to provide a high-strength cold rolled steel sheet having excellent ductility and stretch flangeability and a method for manufacturing the same.

Solution to Problem

[0009]

The inventors of the invention of the present application have conducted extensive studies from the points

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of view of steel sheet composition and microstructure so as to address the problems described above and manufacture a high-strength cold rolled steel sheet having excellent ductility and stretch flangeability. As a result, they have found that when the steel has alloy elements adequately controlled, intensively cooled to a 150 to 350°C temperature range during cooling from the soaking temperature in the annealing process, and reheated, a microstructure containing 20% or more ferrite and 10 to 60% tempered martensite in terms of area ratio and 3 to 15% retained austenite in terms of volume ratio can be obtained and high ductility and stretch flangeability can be achieved.

[0010]

Typically, when retained austenite is present, the ductility improves due to a TRIP effect of the retained austenite. However, it is known that the martensite generated by transformation of retained austenite under application of strain becomes very hard and as a result exhibits a hardness significantly different from that of the main phase ferrite, thereby degrading the stretch flangeability.

[0011]

However, according to the composition and microstructure of the present invention, both high ductility and high stretch flangeability are achieved simultaneously.



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Although the exact reason why high stretch flangeability is achieved despite the presence of retained austenite is unclear, it is presumed that co-existence of the retained austenite and tempered martensite reduces the adverse effects of the retained austenite on the stretch flangeability.

[0012]

It has also been found that when the average crystal grain diameter of the low-temperature transformation-forming phase constituted by martensite, tempered martensite, and retained austenite is 3  $\mu\text{m}$  or less, this steel sheet microstructure can exhibit high formability and improved crashworthiness.

The present invention has been made based on the findings described above and is summarized as follows.

[0013]

A first aspect of the present invention provides a high-strength cold rolled steel sheet having excellent formability and crashworthiness including, on a mass% basis, C: 0.05 to 0.3%, Si: 0.3 to 2.5%, Mn: 0.5 to 3.5%, P: 0.003 to 0.100%, S: 0.02% or less, Al: 0.010 to 0.5%, and balance being iron and unavoidable impurities, the high-strength cold rolled steel sheet having a microstructure including 20% or more of ferrite on an area fraction basis, 10 to 60% of tempered martensite on an area fraction basis, 0 to 10%

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of martensite on an area fraction basis, and 3 to 15% of retained austenite on a volume fraction basis.

[0014]

A second aspect of the present invention provides the high-strength cold rolled steel sheet having excellent formability and crashworthiness according to the first aspect, in which a low-temperature transformation-forming phase constituted by the martensite, the tempered martensite, and the retained austenite has an average crystal grain diameter of 3  $\mu\text{m}$  or less.

[0015]

A third aspect of the present invention provides the high-strength cold rolled steel sheet having excellent formability and crashworthiness according to the first or second aspect of the invention, further including, on a mass% basis, at least one element selected from Cr: 0.005 to 2.00%, Mo: 0.005 to 2.00%, V: 0.005 to 2.00%, Ni: 0.005 to 2.00%, and Cu: 0.005 to 2.00%.

[0016]

A fourth aspect of the present invention provides the high-strength cold rolled steel sheet having excellent formability and crashworthiness according to any one of the first to third aspects of the invention, further including, on a mass% basis, one or both of Ti: 0.01 to 0.20% and Nb: 0.01 to 0.20%.



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[0017]

A fifth aspect of the present invention provides the high-strength cold rolled steel sheet having excellent formability and crashworthiness according to any one of the first to fourth aspects of the invention, further including, on a mass% basis, B: 0.0002 to 0.005%.

[0018]

A sixth aspect of the present invention provides the high-strength cold rolled steel sheet having excellent formability and crashworthiness according to any one of the first to fifth aspects of the invention, further including, on a mass% basis, one or both of Ca: 0.001 to 0.005% and REM: 0.001 to 0.005%.

[0019]

A seventh aspect of the present invention provides a method for manufacturing a high-strength cold rolled steel sheet having excellent formability and crashworthiness, the method including hot-rolling and cold-rolling a slab having a composition described in any one of the first to sixth aspects of the invention to manufacture a cold rolled steel sheet and continuously annealing the cold rolled sheet, in which, during the continuous annealing, the steel sheet is held at a temperature of 750°C or more for 10 seconds or more, cooled from 750°C to a temperature in a temperature range of 150 to 350°C at a cooling rate of 10°C/s or more on

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average, heated to a temperature of 350 to 600°C, held thereat for 10 to 600 seconds, and cooled to room temperature.

[0020]

An eighth aspect of the present invention provides the method for manufacturing a high-strength cold rolled steel sheet having excellent formability and crashworthiness according to the seventh aspect of the invention, in which the average heating rate in the range of 500°C to  $Ac_1$  transformation point is 10°C/s or more.

Advantageous Effects of Invention

[0021]

According to the present invention a high-strength cold rolled steel sheet having excellent formability is obtained. The present invention achieves advantageous effects such as realizing both weight reduction and improved crash safety of automobiles and greatly contributing to improving performance of automobile bodies.

Description of Embodiments

[0022]

The present invention will now be described in detail.

[0023]

1. Regarding composition

The reasons for limiting the steel composition to those described above are first described. Note that the meaning

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of % regarding components is mass% unless otherwise noted.

[0024]

C: 0.05 to 0.3%

Carbon (C) is an element that stabilizes austenite and promotes generation of phases other than ferrite. Thus, carbon is needed to increase the steel sheet strength, generate a multiphase structure, and improve the TS-EL balance. At a C content less than 0.05%, it is difficult to reliably obtain phases other than ferrite even when the production conditions are optimized and TS × EL decreases as a result. At a C content exceeding 0.3%, hardening of welded portions and heat-affected zones is significant, and mechanical properties of the welded portions are deteriorated. Thus, the C content is within the range of 0.05 to 0.3% and preferably 0.08 to 0.15%.

[0025]

Si: 0.3 to 2.5%

Silicon (Si) is an element effective for strengthening the steel. Silicon is also a ferrite-generating element, suppresses C from becoming concentrated and forming carbides in the austenite, and thus serves to accelerate generation of retained austenite. When the Si content is less than 0.3%, the effects of addition are low. Thus, the lower limit is 0.3%. Excessive addition deteriorates the surface quality and weldability. Thus, the Si content is 2.5% or

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less. The Si content is preferably in the range of 0.7 to 2.0%.

[0026]

Mn: 0.5 to 3.5%

Manganese (Mn) is an element effective for strengthening the steel and accelerates generation of low-temperature transformation-forming phase such as tempered martensite. Such an effect is observed at a Mn content of 0.5% or more. However, when the Mn content exceeds 3.5%, the second phase fraction increases excessively, the ductility deterioration of ferrite due to solid solution strengthening becomes significant, and formability is degraded. Accordingly, the Mn content is within the range of 0.5 to 3.5% and preferably in the range of 1.5 to 3.0%.

[0027]

P: 0.003 to 0.100%

Phosphorus (P) is an element effective for strengthening the steel and this effect is achieved at a P content of 0.003% or more. When P is contained exceeding 0.100%, brittleness is induced by grain segregation and crashworthiness is deteriorated. Accordingly, the P content is in the range of 0.003% to 0.100%.

[0028]

S: 0.02% or less

Sulfur (S) forms inclusions such as MnS and causes

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deterioration of crashworthiness and cracking along the metal flow of the welded portion. Thus, the S content is preferably as low as possible but is limited to 0.02% or less from the production cost point of view.

[0029]

Al: 0.010 to 0.5%

Aluminum (Al) acts as a deoxidizing agent and is an element effective for cleanliness of the steel. Aluminum is preferably added in the deoxidizing process. When the Al content is less than 0.01%, the effect of addition is little and thus the lower limit is 0.01%. However, addition of large quantities of Al increases the risk of slab cracking during continuous casting and decreases the productivity. Thus, the upper limit of the Al content is 0.5%.

[0030]

The high-strength cold rolled steel sheet of the present invention contains the above-described components as the basic components and the balance is iron and unavoidable impurities. However, the following components can be adequately contained according to the desired properties.

[0031]

At least one selected from Cr: 0.005 to 2.00%, Mo: 0.005 to 2.00%, V: 0.005 to 2.00%, Ni: 0.005 to 2.00%, Cu: 0.005 to 2.00%

Chromium (Cr), molybdenum (Mo), vanadium (V), nickel



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(Ni), and copper (Cu) suppress generation of pearlite during cooling from the annealing temperature, promote generation of low-temperature transformation-forming phases, and effectively serves to strengthen the steel. Such effects are obtained when 0.005% or more of at least one of Cr, Mo, V, Ni, and Cu is contained. However, when the content of each of Cr, Mo, V, Ni, and Cu exceeds 2.00%, the effect is saturated and the cost will rise. Thus, the Cr, Mo, V, Ni, and Cu contents are each in the range of 0.005 to 2.00%.

[0032]

One or both of Ti: 0.01 to 0.20% and Nb: 0.01 to 0.20%

Titanium (Ti) and niobium (Nb) form carbon nitrides and have an effect of strengthening the steel by precipitation. Such effects are observed at a content of 0.01% or more for each element. In contrast, when Ti and Nb are contained in amounts exceeding 0.20%, excessive strengthening occurs and the ductility is decreased. Thus, the Ti and Nb contents are each within the range of 0.01 to 0.20%.

[0033]

B: 0.0002 to 0.005%

Boron (B) suppresses generation of ferrite from austenite grain boundaries and increases the strength. Such effects are obtained at a B content of 0.0002% or more. However, the effects saturate when the B content exceeds 0.005% and the cost will rise. Accordingly, the B content

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is within the range of 0.0002 to 0.005%.

[0034]

One or both of Ca: 0.001 to 0.005% and REM: 0.001 to 0.005%

Calcium (Ca) and a rare earth element (REM) improve the formability through sulfide morphology control. If needed, one or both of Ca and REM may be contained at an amount of 0.001% or more each. However, since excessive addition may adversely affect the cleanliness, the amount of each element is limited to 0.005% or less.

[0035]

## 2. Regarding microstructure

The microstructure of the steel will now be described.

[0036]

Area fraction of ferrite: 20% or more

When the area fraction of the ferrite is less than 20%, TS × EL decreases. Thus, the area fraction of ferrite is limited to 20% or more and preferably 50% or more.

[0037]

Area fraction of tempered martensite: 10 to 60%

Tempered martensite is a ferrite-cementite multiphase having a high dislocation density and is obtained by heating martensite to a temperature equal to or lower than  $A_{c1}$  transformation point and preferably to a temperature lower than  $A_{c1}$  transformation point. Tempered martensite effectively strengthens the steel. The microstructure

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obtained by heating martensite to a temperature exceeding  $A_{c1}$  transformation point is a microstructure that does not contain cementite in ferrite and is fundamentally different from the tempered martensite intended in the present invention.

[0038]

Compared to martensite, the tempered martensite has less adverse effects on stretch flangeability and is a phase effective for reliably obtaining the strength without significantly decreasing the stretch flangeability. When the area fraction of the tempered martensite is less than 10%, it becomes difficult to reliably obtain the strength. When the area fraction exceeds 60%, TS x EL is decreased. Thus, the area fraction of the martensite is limited to 10 to 60%.

[0039]

Area fraction of martensite: 0 to 10%

Martensite effectively increases the strength of the steel but significantly decreases the stretch flangeability once the area fraction of the martensite exceeds 10%. Thus, the area fraction of the martensite is limited to 0 to 10%.

[0040]

Volume fraction of retained austenite: 3 to 15%

Retained austenite not only contributes to strengthening of the steel but also effectively improves TS

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x EL of the steel. Such effects are achieved at a volume fraction of 3% or more. When the volume fraction of the retained austenite exceeds 15%, the stretch flangeability is decreased. Accordingly, the volume fraction of the retained austenite is limited to 3 to 15%.

[0041]

Average crystal grain diameter of low-temperature transformation-forming phases constituted by martensite, tempered martensite, and retained austenite: 3  $\mu\text{m}$  or less

Low-temperature transformation-forming phases constituted by martensite, tempered martensite, and retained austenite effectively improve the crashworthiness. In particular, finely dispersing the low-temperature transformation-forming phases improves the crashworthiness, and this effect becomes notable when the average crystal grain diameter of the low-temperature transformation-forming phases is 3  $\mu\text{m}$  or less. Accordingly, the average crystal grain diameter of the low-temperature transformation-forming phases is limited to 3  $\mu\text{m}$  or less.

[0042]

The phases other than ferrite, tempered martensite, martensite, and retained austenite may include pearlite and bainite but such phases do not present problem as long as the above-described phase structure is satisfied. However, the pearlite is preferably 3% or less from the view points

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of ductility and stretch flangeability.

[0043]

### 3. Regarding manufacturing conditions

A steel having a composition controlled as described above is melted in a converter or the like and formed into a slab by continuous casting or the like. This steel is hot-rolled, cold-rolled, and continuously annealed. The manufacturing methods regarding casting, hot-rolling, and cold-rolling are not particularly limited but preferable manufacturing methods are described below.

#### Casting conditions

The steel slab used is preferably manufactured by continuous casting in order to prevent macrosegregation of the components but an ingot casting technique or a thin slab casting technique may be employed. In addition to an existing method of cooling the manufactured steel slab to room temperature and then reheating the slab, an energy-saving process such as hot direct rolling or direct rolling which involves sending the hot slab to a heating furnace without cooling the slab to room temperature or which involves rolling the slab immediately after a short period of heat retention may be employed without any difficulty.

[0044]

#### Hot rolling conditions

Slab heating temperature: 1100°C or more



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The slab heating temperature is preferably low from the viewpoint of energy. At a heating temperature less than 1100°C, carbides cannot be sufficiently dissolved or the risks of troubles during hot-rolling increases due to an increased rolling load. In order to prevent the increase in scale loss attributable to oxidation weight gain, the slab heating temperature is preferably 1300°C or less.

[0045]

In order to avoid troubles during hot-rolling despite the decreased slab heating temperature, a sheet bar heater that heats the sheet bar may be employed.

[0046]

Finishing temperature:  $A_{r3}$  transformation point or more.

When the finishing temperature is less than the  $A_{r3}$  transformation point, ferrite and austenite are generated during rolling, and a band-like microstructure readily occurs in the steel sheet. Such a band-like microstructure remains after cold rolling and annealing, may generate anisotropy in the material properties, and may decrease the formability. Accordingly, the finishing temperature is preferably equal to or higher than  $A_{r3}$  transformation point.

[0047]

Coiling temperature: 450 to 700°C

When the coiling temperature is less than 450°C, the control of the coiling temperature is difficult and

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temperature nonuniformity may occur, thereby causing problems such as deterioration of cold-rolling properties. When the coiling temperature exceeds 700°C, problems such as decarburization in the base iron surface layer may occur. Thus, the coiling temperature is preferably in the range of 450 to 700°C.

[0048]

In the hot-rolling process of the present invention, in order to decrease the rolling load during hot rolling, part or all of the finish rolling may be conducted by lubrication rolling. Lubrication rolling is effective from the viewpoints of uniform steel sheet shape and material homogeneity. Note that the coefficient of friction during lubrication rolling is preferably in the range of 0.25 to 0.10. Preferable is a continuous rolling process of joining sheet bars next to each other and continuously finish-rolling the sheet bars. The continuous rolling process is also preferable from the viewpoint of operation stability of hot rolling.

[0049]

Next, the oxidized scales on the surface of the hot-rolled steel sheet are preferably removed by pickling and the steel sheet is cold-rolled to form a cold-rolled steel sheet having a particular thickness. The pickling conditions and the cold rolling conditions are not

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particularly limited and typical conditions may be used.  
The reduction of cold rolling is preferably 40% or more.

[0050]

Average heating rate from 500°C to Ac<sub>1</sub> transformation point:  
10°C/s or more

When the average heating rate in the recrystallization temperature zone, 500°C to Ac<sub>1</sub> transformation point, of the steel of the present invention is 10°C/s or more, recrystallization during heating is suppressed, austenite generated at Ac<sub>1</sub> transformation temperature or higher becomes finer, and the microstructure after annealing and cooling becomes finer. As a result, the average grain diameter of the low-temperature transformation-forming phase can be reduced to 3 μm or less.

[0051]

When the average heating rate is less than 10°C/s, α recrystallization occurs during heating and strain introduced into ferrite is released and thus the sufficient refining of grains cannot be achieved. Thus, the average heating rate from 500°C to Ac<sub>1</sub> transformation point is limited to 10°C/s or more and more preferably 20°C/s or more.

[0052]

Holding a temperature of 750°C or more for 10 seconds or more

When the heating temperature is less than 750°C or the

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holding time is less than 10 seconds, generation of austenite during annealing is insufficient and a sufficient amount of low-temperature transformation-forming phases cannot be reliably obtained after annealing and cooling. Although the upper limits of the holding temperature and the holding time are not particularly defined, the effects saturate and the cost will increase when the holding temperature is 900°C or more and the holding time is 600 seconds or more. Accordingly, the holding temperature is preferably less than 900°C and the holding time is preferably less than 600 seconds.

[0053]

Cooling from 750°C to a temperature range of 150 to 350°C at an average cooling rate of 10°C/s or more

When the cooling rate from 750°C is less than 10°C/s, pearlite is generated and TS × EL and stretch flangeability are degraded. Thus, the cooling rate from 750°C is limited to 10°C/s or more. The temperature condition of ending the cooling is one of the most crucial conditions of this technology. At the time cooling is stopped, part of austenite transforms into martensite and the rest forms untransformed austenite. When reheated, plated and alloyed, and cooled to room temperature, martensite turns into tempered martensite and untransformed austenite transforms into retained austenite or martensite. When the temperature



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of ending the cooling from annealing is low, the amount of martensite generated during cooling increases and the amount of the untransformed austenite decreases. Thus, controlling the temperature of ending the cooling determines the final area fractions of the martensite, the retained austenite, and the tempered martensite.

[0054]

When the temperature of ending the cooling is higher than 350°C, martensite transformation at the time cooling is stopped is insufficient and the amount of untransformed austenite is large, thereby ultimately generating excessive amounts of martensite or retained austenite and degrading the stretch flangeability. When the temperature of ending the cooling is lower than 150°C, most of austenite transforms into martensite during cooling, the amount of untransformed austenite decreases, and 3% or more of retained austenite is not obtained. Accordingly, the temperature of ending the cooling is set within the range of 150 to 350°C. As for the cooling method, any cooling method such as gas jet cooling, mist cooling, water cooling, or metal quenching, may be employed as long as the target cooling rate and cooling end temperature are achieved.

[0055]

Heating to 350 to 600°C and holding thereat for 10 to 600 seconds



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When the steel is held in the temperature range of 350 to 600°C for 10 seconds or more after being cooled to a temperature range of 150 to 350°C, the martensite generated during cooling is tempered and forms tempered martensite. As a result, the stretch flangeability is improved, the untransformed austenite that did not transform into martensite during cooling is stabilized, and 3% or more of retained austenite is obtained at the final stage, thereby improving the ductility.

[0056]

Although the detailed mechanism of stabilization of the untransformed austenite by reheating and holding is not clear, it is presumed that carbon diffuses from martensite, in which dissolved C is oversaturated, into untransformed austenite, thereby increasing the C concentration in the untransformed austenite and stabilizing the austenite. During this process, if the precipitation of cementite in the martensite occurs faster than diffusion of carbon, the concentration of C in the untransformed austenite becomes insufficient. Thus, it is important to delay the cementite precipitation and this requires addition of 0.3% or more of Si.

[0057]

If the reheating temperature is less than 350°C, the martensite is not sufficiently tempered and the austenite is

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not sufficiently stabilized, thereby degrading stretch flangeability and ductility. If the reheating temperature exceeds 600°C, untransformed austenite at the time cooling is stopped transforms into pearlite and 3% or more of retained austenite cannot be obtained at the final stage. Accordingly, the heating temperature is limited to 350 to 600°C.

[0058]

If the holding time is less than 10 seconds, the austenite is not sufficiently stabilized. If the holding time exceeds 600 seconds, untransformed austenite at the time the cooling is stopped transforms into bainite and 3% or more of retained austenite cannot be obtained at the final stage. Accordingly, the reheating temperature is set within the range of 350 to 600°C and the holding time within that temperature range is limited to 10 to 600 seconds.

[0059]

The annealed steel sheet may be subjected to temper rolling to correct shape, adjust surface roughness, etc. Moreover, treatment such as resin or oil/fat coating and various other coating may be performed.

Example 1

[0060]

A steel having the composition shown in Table 1 and balance being Fe and unavoidable impurities was melted in a

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converter and continuously casted into a slab. The slab is hot-rolled to a thickness of 3.0 mm. The hot rolling conditions were as follows: finishing temperature: 900°C, cooling rate after rolling: 10°C/s, and coiling temperature: 600°C. Then the hot-rolled steel sheet was pickled and cold-rolled to a thickness of 1.2 mm to manufacture a cold rolled steel sheet.

[0061]

The cold rolled steel sheet was annealed under the conditions described in Table 2 by using a continuous annealing line.

The cross-sectional microstructure, tensile properties, and stretch flangeability of the resulting steel sheet were investigated. The results are shown in Table 3.

[0062]

[Table 1]

(mass%)

Steel type	C	Si	Mn	P	S	Al	N	Cr	Mo	V	Ni	Cu	Ti	Nb	B	Ca	REM	
A	0.10	1.2	2.3	0.020	0.003	0.033	0.003											Example
B	0.07	1.7	2.0	0.025	0.003	0.036	0.004	0.30										Example
C	0.18	1.0	1.6	0.013	0.005	0.028	0.005		0.4									Example
D	0.25	1.5	1.4	0.008	0.006	0.031	0.003			0.05								Example
E	0.08	0.5	2.2	0.007	0.003	0.030	0.002				0.2	0.4						Example
F	0.12	1.1	1.9	0.007	0.002	0.400	0.001						0.05					Example
G	0.14	1.5	2.3	0.014	0.001	0.042	0.003							0.04				Example
H	0.10	0.9	1.9	0.021	0.005	0.015	0.004						0.02		0.001			Example
I	0.08	1.2	2.5	0.006	0.004	0.026	0.002									0.004		Example
J	0.09	2.0	1.8	0.012	0.003	0.028	0.005										0.002	Example
<u>K</u>	<u>0.04</u>	1.3	1.8	0.013	0.002	0.022	0.002											Comparative Example
<u>L</u>	0.17	0.6	<u>4.0</u>	0.022	0.001	0.036	0.002											Comparative Example
<u>M</u>	0.10	1.1	<u>0.3</u>	0.007	0.003	0.029	0.002											Comparative Example

Note: Underlined items are outside the range of the present invention.



[0063]

[Table 2]

No.	Steel type	AC1 transformation point °C	Average heating rate from 500°C to Ac <sub>1</sub> °C/s	Maximum temperature °C	Holding time Sec	Average cooling rate °C/s	Temperature after cooling °C	Reheating temperature °C	Holding time after reheating	
									Sec	Example
1	A	721	15	830	60	50	200	400	80	Example
2	A	721	15	810	60	50	100	420	80	Comparative Example
3	B	740	20	850	90	80	180	430	60	Example
4	B	740	20	720	60	80	250	430	60	Comparative Example
5	C	734	5	820	90	30	160	450	45	Example
6	C	734	5	820	5	30	120	450	45	Comparative Example
7	C	734	5	820	90	30	30	450	45	Comparative Example
8	D	735	30	780	150	70	150	450	60	Example
9	D	735	30	780	120	3	210	450	60	Comparative Example
10	D	735	30	780	120	100	380	450	50	Comparative Example
11	E	708	7	850	75	80	180	400	30	Example
12	E	708	7	850	60	80	200	250	60	Comparative Example
13	E	708	7	830	75	80	200	650	60	Comparative Example
14	E	708	7	850	75	80	40	400	30	Comparative Example
15	F	723	15	800	240	90	200	400	90	Example
16	F	723	15	820	240	90	220	400	0	Comparative Example
17	F	723	15	800	240	90	240	500	900	Comparative Example
18	G	725	15	850	60	100	200	500	30	Example
19	H	720	15	840	120	90	180	400	30	Example
20	I	718	15	830	75	150	220	500	45	Example
21	J	743	15	800	45	80	180	400	20	Example
22	K	730	15	800	200	100	210	550	10	Comparative Example
23	L	686	15	820	120	150	220	400	60	Comparative Example
24	M	745	15	840	90	150	160	400	20	Comparative Example

Note: Underlined items are outside the range of the present invention.



[0064]

[Table 3]

No.	Steel type	Ferrite		Martensite		Tempered martensite		Retained austenite		Average grain diameter of low-temperature transformation- $\mu\text{m}$	Other phases	TS		EL		TS $\times$ EL		Hole expansion ratio		Absorption energy up to 10% (AE)		AE/TS
		area%	area%	area%	area%	volume%	$\mu\text{m}$	MPa	%			MPa $\cdot$ %	%	MJ/m	MJ/m							
1	A	65	0	0	29	6	2.7		900	26	23400	85	57	0.063	Example							
2	A	63	0	0	35	2	2.8		915	20	18300	92	58	0.063	Comparative Example							
3	B	70	0	0	26	4	2.4		870	26	22620	88	56	0.064	Example							
4	B	73	0	0	8	0	1.9	P	835	21	17535	60	46	0.055	Comparative Example							
5	C	55	0	0	39	6	3.4		990	23	22770	75	57	0.058	Example							
6	C	62	0	0	9	1	2.6	P	935	20	18700	55	49	0.052	Comparative Example							
7	C	57	0	0	42	1	3.5		980	19	18620	92	56	0.057	Comparative Example							
8	D	57	0	0	31	12	1.7		975	26	25350	81	65	0.067	Example							
9	D	65	0	0	25	1	2.1	P	920	20	18400	63	50	0.054	Comparative Example							
10	D	58	20	0	0	14	1.8	B	970	25	24250	32	65	0.067	Comparative Example							
11	E	69	5	0	21	5	3.6		850	26	22100	89	47	0.055	Example							
12	E	70	13	0	15	2	3.6		859	22	18898	74	49	0.057	Comparative Example							
13	E	65	0	0	20	1	3.7	P	823	23	18929	88	40	0.049	Comparative Example							
14	E	75	0	0	24	1	3.5		820	22	18040	105	44	0.054	Comparative Example							
15	F	72	0	0	21	7	2.1		840	27	22680	74	54	0.064	Example							
16	F	70	12	0	17	1	2.0		865	21	18165	62	56	0.065	Comparative Example							
17	F	72	0	0	18	1	2.1	B	796	23	18308	82	50	0.063	Comparative Example							
18	G	53	0	0	37	10	1.8		1015	26	26390	76	72	0.071	Example							
19	H	65	0	0	30	5	2.2		900	25	22500	95	59	0.066	Example							
20	I	51	0	0	42	7	2.8		1068	23	24564	85	68	0.064	Example							
21	J	75	0	0	20	5	2.7		923	24	22152	92	60	0.065	Example							
22	K	91	0	0	8	1	1.8		611	28	17108	73	33	0.054	Comparative Example							
23	L	15	0	0	76	9	2.9		1325	14	18550	75	69	0.052	Comparative Example							
24	M	86	0	0	5	0	2.7	P	562	30	16860	65	31	0.055	Comparative Example							

Note: Underlined items are outside the range of the present invention.

\* : B represents bainite and P represents pearlite.

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[0065]

The cross-sectional microstructure of the steel sheet was observed by exposing the microstructure by using a 3% nital solution (3% nitric acid + ethanol), observing the position 1/4 of the thickness in the depth direction by using a scanning electron microscope, and conducting an image processing of a picture of the microstructure taken to determine the fraction of the ferrite phase (the image processing can be performed by using commercially available image processing software). The area fractions of the martensite and tempered martensite were determined by taking SEM photographs of adequate magnification, e.g., about 1000 to 3000 magnification, depending on the fineness of the microstructure and then determining the quantity by using image processing software. The average grain diameter of the low-temperature transformation-forming phase was determined by dividing the area of the low-temperature transformation-forming phases in the observed area by the number of the low-temperature transformation-forming phases, determining the average area therefrom, and raising the average to the power of 1/2.

[0066]

The volume ratio of the retained austenite was determined by polishing the steel sheet to a surface 1/4 in the thickness direction and measuring X-ray diffraction

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intensity of the 1/4 thickness surface. A MoK $\alpha$  line was used as the incident X ray, the intensity ratios were determined for all combinations of the integrated intensities of peaks of {111}, {200}, {220}, and {311} faces of the retained austenite phase and the {110}, {200}, and {211} faces of the ferrite phase, and the average value was assumed to be the volume fraction of the retained austenite.

[0067]

The tensile property was determined by using a JIS No. 5 specimen sampled from the steel sheet in such a manner that the tensile direction was orthogonal to the rolling direction, conducting a tensile test according to JIS Z2241 to measure TS (tensile strength) and EL (elongation), and determining the strength-elongation balance value represented by the product of the strength and elongation (TS  $\times$  EL).

[0068]

The hole expanding ratio  $\lambda$  was measured as an indicator for evaluating the stretch flangeability. The hole expanding ratio  $\lambda$  was determined by conducting a hole expanding test according to the Japan Iron and Steel Federation standard JFST1001 and determining the ratio from the initial diameter (10 mm $\phi$ ) of the hole upon punching and the diameter of hole at the time the crack at the hole edge penetrated the sheet upon hole expanding.

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[0069]

The shock absorption property was determined by using a specimen 5 mm in width and 7 mm in length sampled from the steel sheet in a direction orthogonal to the rolling direction, conducting a tensile test at a strain rate of 2000/s, and integrating the stress-true strain curve obtained by the tensile test within the range of 0 to 10% to calculate the absorption energy (refer to Tetsu-to-Hagane, 83 (1997) p. 748).

[0070]

The steel sheets of the examples of the present invention have excellent strength, ductility, and stretch flangeability, i.e.,  $TS \times EL$  of 22000 MPa·% or more and  $\lambda$  of 70% or more.

[0071]

In contrast, the steel sheets of comparative examples outside the range of the present invention did not achieve excellent strength, ductility, and stretch flangeability unlike the steel sheets of the examples of the present invention since  $TS \times EL$  was less than 22000 MPa·% and/or  $\lambda$  was less than 70%. Moreover, when the average grain diameter of the low-temperature transformation-forming phase is 3  $\mu\text{m}$  or less, the ratio of the absorption energy to  $TS$  ( $AE/TS$ ) is 0.063 or more, thereby achieving excellent crashworthiness.



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Industrial Applicability

[0072]

The present invention can contribute to weight reduction and decreasing the fuel consumption of automobiles by providing a high-strength cold rolled steel sheet having excellent formability and crashworthiness.



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## CLAIMS

[Claim 1]

A high-strength cold rolled steel sheet having excellent formability and crashworthiness comprising, on a mass% basis, C: 0.05 to 0.3%, Si: 0.3 to 2.5%, Mn: 0.5 to 3.5%, P: 0.003 to 0.100%, S: 0.02% or less, Al: 0.010 to 0.5%, and balance being iron and unavoidable impurities, the high-strength cold rolled steel sheet having a microstructure including 20% or more of ferrite on an area fraction basis, 10 to 60% of tempered martensite on an area fraction basis, 0 to 10% of martensite on an area fraction basis, and 3 to 15% of retained austenite on a volume fraction basis.

[Claim 2]

The high-strength cold rolled steel sheet having excellent formability and crashworthiness according to Claim 1, wherein a low-temperature transformation-forming phase constituted by the martensite, the tempered martensite, and the retained austenite has an average crystal grain diameter of 3  $\mu\text{m}$  or less.

[Claim 3]

The high-strength cold rolled steel sheet having excellent formability and crashworthiness according to Claim 1 or 2, further comprising, on a mass% basis, at least one element selected from Cr: 0.005 to 2.00%, Mo: 0.005 to 2.00%,

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V: 0.005 to 2.00%, Ni: 0.005 to 2.00%, and Cu: 0.005 to 2.00%.

[Claim 4]

The high-strength cold rolled steel sheet having excellent formability and crashworthiness according to any one of Claims 1 to 3, further comprising, on a mass% basis, one or both of Ti: 0.01 to 0.20% and Nb: 0.01 to 0.20%.

[Claim 5]

The high-strength cold rolled steel sheet having excellent formability and crashworthiness according to any one of Claims 1 to 4, further comprising, on a mass% basis, B: 0.0002 to 0.005%.

[Claim 6]

The high-strength cold rolled steel sheet having excellent formability and crashworthiness according to any one of Claims 1 to 5, further comprising, on a mass% basis, one or both of Ca: 0.001 to 0.005% and REM: 0.001 to 0.005%.

[Claim 7]

A method for manufacturing a high-strength cold rolled steel sheet having excellent formability and crashworthiness, the method comprising hot-rolling and cold-rolling a slab having a composition described in any one of Claims 1 to 6 to manufacture a cold rolled steel sheet and continuously annealing the cold rolled sheet, wherein, during the continuous annealing, the steel sheet is held at a

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temperature of 750°C or more for 10 seconds or more, cooled from 750°C to a temperature range of 150 to 350°C at a cooling rate of 10°C/s or more on average, heated to a temperature of 350 to 600°C, held thereat for 10 to 600 seconds, and cooled to room temperature.

[Claim 8]

The method for manufacturing a high-strength cold rolled steel sheet having excellent formability and crashworthiness according to Claim 7, wherein the average heating rate in the range of 500°C to  $A_{c1}$  transformation point is 10°C/s or more.