Title: HIGH STRENGTH API-X80 GRADE STEELS FOR SPIRAL PIPES WITH LESS STRENGTH CHANGES AND METHOD FOR MANUFACTURING THE SAME

Abstract: Disclosed are high strength API-X80 grade hot-rolled thick steels for spiral pipes with less strength variation after pipe forming, and a method for manufacturing the same. The steel is formed by suitably controlling contents of Nb, Ti, Mo and Ni, a ratio between these components, and carbon equivalent, while suitably adjusting heating temperature, reduction ratio, and cooling rate in hot rolling to have fine precipitates. The steel comprises 0.05-0.075% C, 1.65-1.85% Mn, 0.20-0.35% Si, 0.055-0.085% Nb, 0.02-0.06% Ti, 0.20-0.45% Ni, 0.30% or less Mo, 0.05% or less Al, 0.004-0.008 N, and the balance of Fe and unavoidable impurities, wherein an absolute value of Ti-4C-3.42N-1.5S is 0.15-0.31. Ceq represented by the following Equation is 0.45 or less, and microstructure of the steel includes 75% or more acicular ferrite and 3% or less pearlite in terms of area fraction. Ceq = C + Mn/6 + (Cr+Mo+V)/5 + (Ni+Cu)/15
Description
HIGH STRENGTH API-X80 GRADE STEELS FOR SPIRAL PIPES WITH LESS STRENGTH CHANGES AND METHOD FOR MANUFACTURING THE SAME

Technical Field
The present invention relates to high strength API-X80 grade hot-rolled thick steels for spiral pipes with less strength changes after pipe forming, and a method for manufacturing the same. More particularly, the present invention relates to steel for linepipes having a strength of API-X80 grade (yield strength of 552 MPa or more), which is formed by suitably controlling contents of niobium (Nb), titanium (Ti), molybdenum (Mo) and nickel (Ni) generally added for low carbon aluminum-killed steel, a ratio between these components, and carbon equivalent, while suitably adjusting heating temperature, reduction ratio, and cooling rate in hot rolling to have fine precipitates in the steel. The present invention also relates to a method for manufacturing the same.

Background Art
Conventional API-X80 grade hot-rolled steel for linepipes requires addition of a great amount alloying elements to ensure the strength of standard pipes.

As represented by carbon equivalent $\text{Ceq} = (C + \frac{Mn}{6} + \frac{(Cr+Mo+V)}{5} + \frac{(Ni+Cu)}{15})$, the added amount of alloy elements for the API-X80 grade steel is typically 0.47 or more, which generates many defects during steel pipe welding and disadvantageously leads to undesired strength of a pipe through decrease in strength of hot-rolled steel after pipe forming. Further, for the conventional API-X80 grade steel, hot working is performed by low temperature rolling to achieve low temperature toughness, which leads to an increase in rolling load and thus overall load of equipment.

$\text{Ceq} = C + \frac{Mn}{6} + \frac{(Cr+Mo+V)}{5} + \frac{(Ni+Cu)}{15}$

In manufacture of steel for pipes, a higher strength of steel pipes can be obtained through solid solution strengthening or precipitation strengthening by adding a higher amount of alloying elements. However, as the total added amount of alloying elements increases, the possibility of defect formation also increases during steel pipe welding.

One of representative defects is a cold crack, which is formed on a welded zone of the
steel pipe during a cooling process performed after melting and solidifying the welded zone, and which can be more easily formed with an increase in carbon equivalent and provides adverse effect to the steel pipe in view of stability. This is, when the alloying elements are added to steel, they lower transformation temperature, and thus facilitate transformation to microstructure that has high strength and causes brittleness cracks, thereby increasing hardenability of the steel. Therefore, it is necessary to manufacture high strength steel through control of carbon equivalent so as to be used as steel for linepipes.

On the other hand, hot-rolled steel is formed into a spiral pipe generally by a roll forming process. In this process, the steel undergoes deformation, which is accumulated in the steel and inevitably changes mechanical properties of the steel. If the mechanical properties of the steel are deteriorated, the final pipe fails to satisfy strength standard even though the steel satisfies the strength standard prior to pipe forming. Therefore, the steel for pipes must have a low difference in mechanical properties from the pipe, and have optimized microstructure and precipitates that determine the mechanical properties.

When a great amount of pearlite exists in microstructure of steel, the steel for pipes generally undergoes a rapid decrease in strength after pipe processing due to the Bauschinger effect as shown in Fig 1. Conversely, as is reported in the art, the steel having acicular ferrite not only undergoes low variation in strength after pipe forming, but also exhibits strength increase according to a pipe forming manner.

The aforementioned steel for linepipes is manufactured by steel manufacturing, continuous casting, and hot rolling, in which the hot rolling is generally performed in such a way that a slab prepared by continuous casting and having a thickness of 230~250mm is homogenized at a temperature of 1100~1250°C, hot-rolled in the order of rough rolling and finish rolling to have a desired thickness, and cooled to have desired microstructure. Here, the steel for linepipes can be distinguished from general hot-rolled steel by the rough rolling. Namely, compared to the rough rolling for the general hot-rolled steel performed in consideration of simple thickness reduction, the rough rolling for the linepipe steel should be performed via suitable control of deformation amount, temperature of rough rolling, and cooling conditions, since precipitation of added alloying components begins in this process. In rough rolling of the steel for linepipes, such a control is given for the purpose of finally obtaining fine structure, by which desired low temperature toughness can be ensured.

In other words, during the rough rolling of the linepipe steel, Nb-based precipitates
are generally formed in the steel and suppress austenite recrystallization, which occurs in high temperature rough rolling of austenite structure, and which will be defined as non-recrystallization rolling. Here, a total rolling reduction at a precipitation generating temperature or less, that is, a non-recrystallization reduction ratio, becomes one of important operation factors. If the Nb-based precipitates are not formed due to non-addition of Nb, strain energy accumulated in the steel by meaningful rolling deformation of the steel can be released by the recrystallization. However, for the linepipe steel, the austenite structure becomes denser by generation of precipitation during the rough rolling, and thus provides many nucleation sites for ferrite transformation in a cooling process, so that the final steel can have finer microstructure and improved low temperature toughness. Since the microstructure of the linepipe steel manufactured as described above has a great amount of grain boundaries capable of obstructing propagation of cracks, it suppresses propagation of cracks, which can occur during use of the linepipe, improving stability of the linepipe used for carrying gas or petroleum.

To form such precipitates in the steel, it is necessary to cool the steel to a non-recrystallization temperature in the rough rolling process. Thus, rolling of the linepipe steel is generally performed at a lower temperature than that of the general hot-rolled steel. As a result, rolling load of the linepipe steel is excessively increased to exceed a limit value of rolling equipment, which makes it difficult to provide a suitable rolling reduction to the steel. In particular, a small number of rough rolling mills leads to a significantly extended period of the rough rolling process, and a lower rolling load capacity of the rough rolling mill fails to impart a desired rolling reduction ratio, making it difficult to have desired microstructure.

Disclosure of Invention

Technical Problem

The present invention has been made to solve the foregoing problems of the prior art and therefore an aspect of the present invention is to provide a method of manufacturing API-X80 grade steel for spiral linepipes, which can sufficiently suppress welding cracks and strength reduction after pipe forming, and which ensures the spiral linepipes have a yield strength of 552 MPa or more.

The steel for spiral linepipes according to the present invention is formed by suitably
controlling contents of alloying components and ratio therebetween in the steel, thereby lowering frequency of welding defects and strength reduction after pipe forming, while ensuring desired low temperature toughness at a rolling load, which does not burden equipment.

[16] Technical Solution

[17] According to an aspect of the present invention, the present invention provides a high strength API-X80 grade hot-rolled steel plate, comprising, by weight %: 0.05-0.075% C; 1.65-1.85% Mn; 0.20-0.35% Si; 0.055-0.085% Nb; 0.02-0.06% Ti; 0.20-0.45% Ni; 0.30% or less Mo; 0.05% or less Al; 0.004-0.008% N; and the balance of Fe and unavoidable impurities, wherein an absolute value of Ti-4C-3.42N-1.5S is controlled in the range of 0.15-0.31, Ceq represented by the following Equation is 0.45 or less, and microstructure of the steel plate includes 75% or more acicular ferrite and 3% or less pearlite in terms of area fraction.

[18] Ceq = C + Mn/6 + (Cr+Mo+V)/5 + (Ni+Cu)/15

[19] Preferably, the steel plate further comprises 0.01% or less V.

[20] Preferably, the unavoidable impurities comprise 0.0015 wt% S and 0.01 wt% or less P.

[21] According to another aspect of the present invention, the present invention provides a method of manufacturing a high strength hot-rolled steel plate for linepipes, comprising: reheating a steel slab to 1,100~1,250°C, the slab comprising, by weight %(wt%), 0.05-0.075% C, 1.65-1.85% Mn, 0.20-0.35% Si, 0.055-0.085% Nb, 0.02-0.06% Ti, 0.20-0.45% Ni, 0.30% or less Mo, 0.05% or less Al, 0.004-0.008% N, and the balance of Fe and unavoidable impurities, wherein an absolute value of Ti-4C-3.42N-1.5S is in the range of 0.15-0.31, and Ceq represented by the following Equation is 0.45 or less; rough rolling the reheated steel slab at 1,100~950°C for recrystallization reduction and non-recrystallization reduction with a uniform distribution reduction ratio of 14-18%, followed by finish rolling at 760~850°C to prepare a hot-rolled steel plate; and cooling the hot-rolled steel plate to 600°C or less at a cooling rate of 10~20°C/sec, followed by coiling the cooled steel plate.

[23] Ceq = C + Mn/6 + (Cr+Mo+V)/5 + (Ni+Cu)/15

[24] Preferably, the steel slab further comprises 0.01% or less V.

[25] Preferably, the unavoidable impurities comprise 0.0015 wt% S and 0.01 wt% or less P.
[26] **Brief Description of the Drawings**

[27] The above and other aspects, features, and other advantages of the present invention will be more clearly understood from the following detailed description taken in conjunction with the accompanying drawings, in which:

[28] Fig. 1 is a graph explaining the Bauschinger effect which is a mechanism of stress-strain characteristics change of material when the material is formed to a pipe by pipe mill;

[29] Fig. 2 is a graph of Nb-precipitation behavior related to change in temperature of a steel plate;

[30] Fig. 3 is microphotographs of microstructure of inventive steel depending on a cooling rate;

[31] Fig. 4 is a microphotograph of microstructure of Inventive Steel 3;

[32] Fig. 5 is a graph depicting a change (drop) in yield strength of steel between before and after pipe forming;

[33] Fig. 6 is microphotographs showing microstructure of steel depending on a pearlite fraction, and drop of yield strength thereof; and

[34] Fig. 7 is a graph of percent ductile fracture of inventive steels depending on temperature by drop weight tear test.

[35] **Best Mode for Carrying Out the Invention**

[36] Exemplary embodiments of the present invention will now be described in detail with reference to the accompanying drawings.

[37] The present invention relates to a steel plate for spiral steel pipes.

[38] In order to prevent the steel plate for linepipes from suffering from deterioration in welding properties, it is advantageous that the composition of the steel plate is controlled as described below along with restriction for Ceq, an indicator of the welding properties, represented by the following Equation 2, to be 0.45 or less.

\[
\text{Ceq} = C + \frac{\text{Mn}}{6} + \frac{\text{(Cr+Mo+V)}}{5} + \frac{\text{(Ni+Cu)}}{15}
\]

[39] As mentioned above, Ceq is the indicator of determining the welding properties of a steel plate. An excessive increase in added amount of alloying elements for ensuring the strength of steel plate causes an increase in Ceq, which leads to deterioration of the welding properties. As a result, the steel plate is likely to experience undesired defects such as a cold crack. Therefore, it is desirable that the steel plate have Ceq of 0.45 or
Next, the composition of the steel plate is controlled as follows in order to ensure the strength of steel plate and to make it easy to regulate Ceq in the range of the present invention. Hereinafter, the present invention will be described in detail in view of composition, structure and manufacturing process.

(Composition of Steel)

C: 0.05-0.75 % by weight (hereinafter, %)

Carbon is a fundamental element that is combined with alloying elements to form precipitates in the steel. Such carbon-based precipitation is referred to as "Carbides", and provides different influence on the strength and toughness of steel depending on size and distribution thereof. Typically, finer precipitations make a higher contribution to improvement in strength of steel and have a lower influence on deterioration of low temperature toughness.

The content of carbon less than 0.05% becomes a cause of load in the steel manufacturing process and has little effect in strength improvement. On the other hand, an excessive content of carbon more than 0.75% leads to carbide coarsening and an increase in amount of pearlite, which are ineffective in strength improvement and provide adverse effects to the low temperature toughness. Additionally, an increase in content of carbon leads to an increase in solid solution temperature of the added alloying elements during reheating, which requires an excessive increase of the reheating temperature.

Therefore, the content of carbon is preferably in the range of 0.05-0.75%.

Mn: 1.65-1.85%

Manganese (Mn) generally combines with sulfur (S) during the steel manufacturing process to prevent red brittleness, and provides a solid solution strengthening effect when added in a predetermined amount or more. Further, Mn serves to retard initiation of transformation into ferrite or pearlite. Therefore, the content of Mn is set in the range of 1.65-1.85% to obtain desired strength of the steel according to the present invention. Here, an excessive added amount of Mn generates large amounts of central segregation in a slab, which leads to deterioration in low temperature toughness and can be advanced to central cracks during pipe working. Thus, the upper limit of Mn is set to 1.85%.
Si: 0.20-0.35%

Silicon (Si) serves to increase the strength of steel by solid solution strengthening. However, an excessive content of Si deteriorates the welding properties and can cause surface defects such as red scales during a process of manufacturing hot rolled steel. Therefore, the content of Si is preferably in the range of 0.20-0.35% in consideration of the strength and quality of steel.

Nb: 0.055-0.085%

Niobium (Nb) is an element that provides a more important influence on hot rolling properties of linepipes and mechanical properties of final steel than any of other alloying elements added to the steel. This is attributable to the fact that Nb is dissolved in the steel at a reheating temperature and forms carbonitride precipitates in the region of hot rolling temperature. As can be appreciated from Fig. 2 that shows a representative solid solution curve, the alloying element can be sufficiently utilized when temperature and time are controlled to allow the alloying elements to be sufficiently dissolved during the reheating process, and almost of all Nb exists in a solid solution state in the steel at about 1,200°C irrespective of steel kinds. On the other hand, although addition of Nb leads to a highly effective solid solution strengthening effect, it generates the precipitates during rough rolling, and thereby significantly changes austenite non-recrystallization temperature and hot rolling deformation resistance, which affect the hot rolling properties of the steel. Therefore, it is necessary to add Nb in consideration of metallurgical influence.

According to the present invention, the content of Nb is preferably in the range of 0.055-0.085%, which is a condition capable of providing optimal effects through addition of alloying iron in the compositional range of inventive steel in consideration of the hot rolling process and metallurgical characteristics of inventive steel. If the content of Nb is less than 0.055%, the non-recrystallization temperature is lowered. Thus, to obtain desired non-recrystallization rolling reduction, low temperature rolling is required. However, the low temperature causes a significant increase in deformation resistance of the steel, which increases rolling load above the equipment limit. As a result, the non-recrystallization rolling reduction cannot be sufficiently applied to the steel at the low temperature, making it difficult to ensure the strength and toughness of the steel. On the other hand, the content of Nb exceeding 0.085% can cause brittleness cracks at corners of the steel, making it difficult to process the hot rolled steel into a normal pipe.
Titanium (Ti) is a very intensive compound forming element and is precipitated in the sequence of sulfide (TiS), nitride (TiN), and carbide (TiC) with change in temperature of the steel from high temperature to low temperature. Thus, Ti is an element that highly influences precipitation hardening of the steel, and is preferably added in the range of 0.02-0.06\%. A predetermined amount of Ti combines with manganese sulfide (MnS), which is formed at a very high temperature, to increase the strength of MnS, and a more amount of Ti must combine with sulfides and nitrides precipitated at a high temperature, than what is actually necessary. Therefore, according to the present invention, Ti is preferably added to the steel in the lower limit of 0.02\% to obtain the precipitation hardening effect. This is for the purpose of obtaining the precipitation hardening effect by titanium carbide (TiC), which enables precipitation in a relatively wide temperature range from an austenite region to a ferrite region, compared to other alloying iron. Further, Ti precipitates have a complex shape like titanium carbonitride (Ti(C,\text{N})), and preferably have a fine size in a ferrite phase to ensure an effective increase in strength. Meanwhile, Ti also serves to increase the non-recrystallization initiation temperature with an increase in added amount of Ti as in the case of Nb. However, if the content of Ti exceeds 0.06\%, coarsening of TiN occurs to provide adverse effects in view of low temperature toughness. Thus, the content of Ti is preferably in the range of 0.02-0.06\%.

Although nickel (Ni) is known as an element for improving the low temperature toughness, it is very expensive and increases manufacturing costs. Therefore, according to the present invention, the content of Ni is preferably in the range of 0.20-0.45\% to ensure the desired low temperature toughness while preventing an increase in manufacturing costs.

Molybdenum (Mo) provides high influence during cooling of the steel, and serves to increase a fraction of acicular ferrite in a polygonal ferrite phase of the steel in the same conditions of hot rolling.

The acicular ferrite or the pseudo polygonal ferrite exhibits superior low temperature toughness to the polygonal ferrite, and can be produced via quenching. Thus, with
equipment that provides insufficient cooling conditions, Mo is added to surmount such insufficient cooling conditions by promoting generation of the acicular ferrite.

However, a highly added amount of Mo causes an increase in manufacturing costs and generation of bainite, which leads to deterioration in low temperature toughness. Compared to conventional 500MPa grade linepipe steel for low temperature toughness that comprises 0.3- 0.4% Mo, the steel according to the invention preferably comprises 0.30% or less Mo in order to suppress shape defects such as upward or downward bending of the hot rolled steel, which can occur during the hot rolling.

Al: 0.05% or less

Aluminum (Al) is an element used as an oxidizing agent for deoxidation of the steel and is generally added in an Al-killing process. Typically, Al exists in the form of an oxide in the steel and provides a structure refining effect. However, since a highly added amount of Al degrades cleanliness of the steel, Al is preferably added up to 0.05%.

N: 0.004-0.008%

Nitrogen (N) is an element that suppresses growth of austenite grains and forms TiN precipitates during slab heating, thereby suppressing the austenite grain growth in a welding heat affected zone. Therefore, N is preferably added in an amount of 0.004% or more. However, since an excessive content of N promotes surface defects on the slab while deteriorating the toughness of the welding heat affected zone, N is preferably added up to 0.008% or less.

Accordingly, the steel plate according to the present invention preferably comprises 0.05-0.075% C, 1.65-1.85% Mn, 0.20-0.35% Si, 0.055-0.085% Nb, 0.02-0.06% Ti, 0.20-0.45% Ni, 0.30% or less Mo, 0.05% or less Al, and 0.004-0.008% N in terms of wt%.

For further improvement in properties of the steel plate, the steel plate preferably further comprises 0.01% or more V. This is because V is an element that provides a similar function to Nb and effectively increases the strength of steel through formation of precipitates at low temperature. Here, since such effects of V are slightly less than that of Nb, it is desirable that Nb is preferentially added to the steel. However, since the effect as described above can be further improved by addition of Nb along with V, V can be further added to the steel. Nb and V are added for the purpose of increasing the strength of steel through formation of precipitates in the form of carbides.
considering addition of V leads to low temperature precipitation, it is necessary to
suitably control coiling temperature. However, since an excessively added amount of
V can cause deterioration in toughness and welding properties of the welding heat
affected zone, the content of V preferably has an upper limit of 0.1%.

In addition to the alloying elements described above, the steel plate may comprise
unavoidable impurities, of which kinds are too numerous to be individually referred to.
However, it is desirable that, as elements causing severely detrimental influence to the
properties of steel, sulfur (S) and phosphorus (P) be restricted as much as possible.
Therefore, it is more effective in obtaining the properties of steel according to the
invention to set preferable added amounts of S and P. Hereinafter, the preferable added
amount of each component and the reason thereof will be described.

S: 0.0015% or less

Sulfur (S) is a detrimental element in view of mechanical properties and is required
to be suppressed in content as much as possible in the steel. However, since it is
difficult to completely remove S from the steel due to current technical limit and load
in removal of sulfur, S can be allowed up to 0.0015%, which is considered not
problematic in mass production of the steel.

P: 0.01% or less

Phosphorus (P) is an element that causes central or grain boundary segregation in the
steel. In particular, for thick steel, since P facilitates brittleness fracture at the center of
the steel, it is also necessary to suppress the content of P as much as possible in the
steel. However, since load of the steel manufacturing process also increases as in the
case of S to manage the content of P to an excessively low content, P is allowed up to
0.01% in the present invention.

In addition to the advantageous composition of the steel plate as described above, it
is preferable that an absolute value of Ti-4C-3.42N-1.5S is in the range of 0.15-0.31.
Here, Ti, C, N, and S means wt% of respective elements. The relation of Ti-
4C-3.42N-1.5S can be used as an index indicating that except for the amount of Ti
contained in TiS and TiN precipitates at high temperature, Ti dissolved in the steel
during the hot rolling is formed to TiC precipitated at a relatively low temperature.
Since the Ti-based precipitates formed in the steel at the low temperature are fine pre-
cipitates, they are very effective in improving the strength of steel without reducing the
low temperature toughness. An index of Ti amount to remain in the steel for formation of the fine precipitates is advantageously controlled to 0.15 or more. However, since an excessively high index of Ti amount is likely to cause coarsening of TiN and TiS, which are stable at high temperature, and deteriorates the low temperature toughness, the index of Ti amount is preferably 0.31 or less.

[85] (Structure of steel plate)

[86] With Ceq, the composition, and relation between the respective components as described above, the steel plate for linepipes according to the invention has suitable conditions for realizing high strength and good welding properties. Additionally, to ensure high strength of API-X80 grade or more, it is desirable that the steel plate have microstructure as follows.

[87] In other words, when changing rolling and cooling conditions of the steel plate, the steel plate can be formed with various microstructure phases. In this regard, inventors of the present invention found that, when the steel plate comprises acicular ferrite as a main microstructure phase, it is preferable to ensure the strength and toughness of the steel and a pipe after pipe forming.

[88] Specifically, the microstructure of the steel plate preferably comprises acicular ferrite in an area fraction of 75% or more. Since hard microstructure such as bainite or martensite significantly reduces the toughness despite an increase in strength of the steel plate, they are unfavorable for the steel plate of the present invention. The acicular ferrite is formed as the main phase of the microstructure in the steel plate and pearlite as a secondary phase is preferably suppressed to minimize decrease in strength after the pipe forming. At this time, the fraction of pearlite is preferably 3% or less.

[89] When controlling the microstructure in the condition as described above, it is possible to ensure the strength of X80 grade or more without deteriorating the toughness of the steel for linepipes. However, this strength is strength of a steel plate prior to the pipe forming, and can be changed after the pipe forming. Fig. 1 shows stress-strain curves of a UOE steel pipe and a spiral steel pipe when performing the pipe forming. As can be seen from Fig. 1, the UOE steel pipe is subjected to a series of processes such as U-shape working, pipe forming, pipe expansion, etc., in which the UOE steel pipe undergoes a temporary reduction in yield strength during the pipe forming and an increase in yield strength during the pipe expansion to a desired diameter. The reduction in yield strength during the pipe forming is based on the so-called Bauschinger effect, by which although dislocations are piled up to act as
obstacles against deformation in the steel by application of stress during the pipe forming, they are converted into a cause of encouraging the deformation when stress is applied to the steel opposite the previous stress. Therefore, the yield strength of a steel pipe is generally lowered after the pipe forming. However, in manufacture of the UOE steel pipe, the pipe expansion is performed after the pipe forming to have a desired diameter, and increases the strength through work hardening caused thereby.

On the other hand, for the spiral steel pipe, since steel is formed to a pipe having a desired diameter during the pipe forming, the pipe expansion is not needed. Namely, the spiral steel pipe cannot be expected to have an increase in yield strength through the pipe expansion, so the yield strength decreased during the pipe forming becomes the yield strength of the pipe. As a result, there is possibility of a significant difference between the steel plate and the pipe in view of yield strength, which makes it difficult to satisfy the yield strength of the pipe demanded by the customer.

Accordingly, in order to minimize the difference in yield strength between the steel pipe after pipe forming and the final pipe, there is a need of minimizing the reduction in yield strength caused by the pipe forming.

The inventors of the present invention found that pearlite structure in a steel plate is a main cause of lowering the yield strength after the pipe forming. In other words, when the steel plate has pearlite, a strength reduction degree related to the Bauschinger effect increases substantially in proportion to an area fraction of pearlite in the steel. Thus, the area fraction of pearlite structure in the steel must be restricted to 3% or less to suppress the strength reduction degree so as not to cause a severe problem.

With the conditions as described above, not only does the steel plate have desired high strength and high toughness, but can also prevent the strength reduction even after spiral pipe forming. Although those skilled in the art can prepare the steel plate having the conditions as described above in various steel manufacturing methods, the inventors of the present invention suggest a new preferable method for manufacturing such a steel plate. Hereinafter, the method for manufacturing the steel plate according to the invention will be described in detail.

(Manufacturing method)

The method of manufacturing the steel plate according to the present invention comprises reheating a steel slab, hot rolling the reheated steel slab to provide a hot-rolled steel plate, and cooling the steel plate to a coiling temperature. Suitable conditions for the respective processes will be described as follows.
Slab reheating temperature: 1,100-1,250 °C

First, a steel slab having the composition as described above is reheated to change columnar structure, formed during continuous casting, into structure suitable for hot rolling while dissolving precipitates again in the steel slab. The reheating temperature is preferably in the range of 1,100-1,250 °C. If the reheating temperature is less than 1,100 °C, the alloying elements precipitated during the continuous casting cannot be sufficiently dissolved, thereby making it difficult to obtain desired metallurgical effects such as strength increase and the like. If the reheating temperature is above 1,200 °C, the slab is likely to have coarsened austenite grains, some of which remain in a final steel plate and deteriorate the properties of steel plate. Therefore, according to the invention, the steel slab is heated for 180 minutes or more in the temperature range as suggested above so as to prevent deterioration of hot rolling properties due to insufficient heating in the interior of the slab.

Rough rolling : reduction ratio of 14-18% per pass at 1,100-950 °C

For hot-rolling the reheated steel slab, the steel slab is subjected to rough rolling at 1,100-950 °C. If the rough rolling is performed at too high temperatures, compounds of elements such as Nb are not sufficiently precipitated, thereby making it difficult to achieve sufficient suppression of austenite grain growth. Conversely, if the rough rolling is performed at too low temperatures, a rolling load is increased above design load for equipment. Further, since austenite recrystallization and non-recrystallization are sequentially occurred in accordance with temperature decrease in this process, the reduction ratio per pass is set in the range of 14-18%. If the reduction ratio is 14% or less, deformation does not sufficiently propagate into the center of the steel plate, causing non-uniform properties in the thickness direction. On the other hand, if the reduction ratio is 18% or more in the rough rolling, the rolling load becomes excessively increased to provide a problem in view of equipment load or to provide an unfavorable shape to the steel plate, which makes it difficult to apply continuous rolling deformation to the steel plate.

Finish rolling temperature: 760-850 °C

Finish rolling is preferably performed at 760-850 °C. As in the rough rolling, if the finish rolling is performed at too low temperatures, finish rolling mills are likely to experience an increased load. Conversely, if the finish rolling is performed above 850 °C,
it is difficult to obtain desired toughness and strength.

[107]

[108] Cooling rate: 10-20 °C/sec

[109] The hot-rolled steel plate is cooled to 600 °C or less at a cooling rate of 10-20 °C/sec. A suitable cooling rate is one of important factors that determine final microstructure of steel. As can be appreciated from Fig. 3 showing a relationship between cooling rate and microstructure, an excessively low cooling rate leads to a soft phase, whereas an excessively high cooling rates leads to a decrease in ductility. Therefore, according to the invention, the cooling rate is set in the range as described above. During such a cooling process, if a coiling start temperature is maintained at 600 °C or more, the steel plate is in the pearlite transformation nose position, and formed with local pearlite. Thus, it is necessary to maintain the coiling temperature below this temperature in order to obtain a desired phase. After being cooled as above, the steel plate is coiled. In these operation conditions, the cooling rate and the coiling temperature are set based on the content of Mo set to 0.30%, and if the content of Mo exceeds 0.30%, acicular ferrite can be easily formed at a relatively slow cooling rate.

[110] When manufactured in the conditions as described above, the steel plate according to the invention satisfies desired properties of final hot-roll thick steel products and is characterized in that it has pearlite structure in an area fraction of 3% or less. Therefore, the present invention can suppress a strength reduction degree after the pipe forming to be 40 MPa or less, and thus ensures API-X80 grade strength of pipes.

[II]

Mode for the Invention

[112] Next, the present invention will be described more specifically with reference to examples. It should be noted that the following examples are given by way of illustration and do not limit the scope of the present invention. The scope of the present invention is limited only by the following claims and equivalents thereof.

[113]

[114] (Examples)

[115] Table 1 shows compositions of various steel samples, (unit : wt%)

[116]

[117] Table 1
After preparing steel slabs having the compositions of Table 1 by vacuum melting, each of the steel slabs was subjected to hot-rolling. Specifically, after preparing the steel slab by vacuum melting, the steel slab was reheated at 1,180 °C for 180 minutes in nitrogen atmosphere, followed by hot-rolling in conditions as shown in the following Table 2. The hot-rolling conditions shown in Table 2 were conditions that can be applied to conventional linepipe steel for low temperature toughness, in which all of the steel samples were rolled to the same thickness of 16.9 mm and cooling rates were obtained by dividing a difference between rolling finish temperature and coiling temperature by a cooling period.
Table 3 shows an average of mechanical properties measured twice in a direction of 30 degrees from a rolling direction of the respective steel samples that had the compositions in Table 1 and were manufactured in the conditions for hot rolling in Table 2. Since a hot-rolled steel plate is generally used to provide a spiral steel pipe, the mechanical properties of the pipe in a circumferential direction are important for such a spiral steel pipe and correspond to values in the direction of 30 degrees from a rolling direction of a steel plate prior to pipe processing. Since the hot-rolled steel for linepipes is hot-rolled at a lower temperature than that of other steel, the steel for linepipe is likely to be bent upward, that is, upward bending, which makes it difficult to perform the hot-rolling and deteriorates workability. Other steel samples except for conventional steel (Steel No. 4) did not exhibit upward bending during the hot-rolling.
As listed in Table 3, since Comparative Steels 5 and 7 had insufficient yield strengths and failed to obtain desired yield strength of 550 MPa for a pipe, they are not suitable for high strength linepipes. This is because alloying elements for ensuring the sufficient strength are not suitably added thereto. For Steel No. 4 (Prior Steel) prepared as a conventional steel for high strength linepipe, although it had desired mechanical properties, not only did it have Ceq exceeding 0.45 due to excessive addition of Mo, Ni, and Mn, but it also had a low commercial worth due to high manufacturing costs. Additionally, it experienced upward bending during the hot rolling, which made it difficult to perform the hot rolling. Fig. 4 shows microstructure of Inventive Steel 3.

As can be seen from Fig. 4, the suitable mechanical properties can be obtained by providing a high area fraction of acicular ferrite while suppressing a fraction of secondary phase such as pearlite as much as possible. The strength reduction after pipe forming is determined by the fraction of secondary phase, and changes in mechanical properties for Inventive Steel 3 and Comparative Steel 8 after the pipe forming are shown in Fig. 5. As shown in Table 3, results of Fig. 5 are closely related to generation of the secondary phase, that is, pearlite, and show that, when steel has 75% or more acicular ferrite and 3% or less pearlite, it is possible to prevent a significant reduction in strength of the steel.
Fig. 6 shows a fraction of pearlite structure determining strength reduction after pipe forming, and a strength reduction degree. The fraction of secondary phase was measured using an image analysis apparatus after picral etching for observing only the secondary phase, that is, pearlite. Although the conventional steel had a high Ceq and a low fraction of secondary phase leading to a lower strength reduction degree than any other steels after the pipe forming, it was inferior to the inventive steels in view of operability and easiness in manufacture. Accordingly, it could be concluded that the inventive steels with the strength reduction degree of 40 MPa or less after the pipe forming can be more advantageously used to manufacture the linepipes in view of operability and easiness in manufacture than any other steels.

Fig. 7 is a graph depicting low temperature of inventive steels by drop weight tear test. As can be seen from the graph, the inventive steels satisfy a condition for percent ductile fracture of 85% or more to -20 °C, as is required for a pipe.

As apparent from the above description, the present invention can provide API-X80 grade steel for linepipes, of which strength reduction degree after pipe forming is 40 MPa or less, and which ensures high strength and low temperature toughness at the same time.
Claims

1. A high strength API-X80 grade hot-rolled steel plate, comprising, by weight %:
   0.05-0.075% C, 1.65-1.85% Mn, 0.20-0.35% Si, 0.055-0.085% Nb,
   0.02-0.06% Ti, 0.20-0.45% Ni, 0.30% or less Mo, 0.05% or less Al,
   0.004-0.008% N, and the balance of Fe and unavoidable impurities, wherein an
   absolute value of Ti-4C-3.42N-1.5S is controlled in the range of 0.15-0.31, Ceq
   represented by the following Equation is 0.45 or less, and microstructure of the
   steel plate includes 75% or more acicular ferrite and 3% or less pearlite in terms
   of area fraction.

   \[ C_{eq} = C + \frac{Mn}{6} + \frac{(Cr+Mo+V)}{5} + \frac{(Ni+Cu)}{15} \]

2. The steel plate according to claim 1, further comprising: 0.01% or less V.

3. The steel plate according to claim 1, wherein the unavoidable impurities
   comprise 0.0015 wt% S and 0.01 wt% or less P.

4. A method of manufacturing a high strength API-X80 grade hot-rolled steel plate
   for linepipes, comprising:
   reheating a steel slab to 1,100-1,250 °C, the slab comprising, by weight %(wt%),
   0.05-0.075% C, 1.65-1.85% Mn, 0.20-0.35% Si, 0.055-0.085% Nb,
   0.02-0.06% Ti, 0.20-0.45% Ni, 0.30% or less Mo, 0.05% or less Al,
   0.004-0.008% N, and the balance of Fe and unavoidable impurities, wherein an
   absolute value of Ti-4C-3.42N-1.5S is in the range of 0.15-0.31, and Ceq rep-
   resented by the following Equation is 0.45 or less;
   rough rolling the reheated steel slab at 1,100-950°C with a reduction ratio of
   14-18% per pass;
   finish rolling at 760-850°C to prepare a hot-rolled steel plate; and
   cooling the hot-rolled steel plate to 600°C or less at a cooling rate of 10-20
   °C/sec, followed by coiling the cooled steel plate.

   \[ C_{eq} = C + \frac{Mn}{6} + \frac{(Cr+Mo+V)}{5} + \frac{(Ni+Cu)}{15} \]

5. The method according to claim 4, wherein the steel slab further comprises 0.01% or
   less V.

6. The method according to claim 4, wherein the unavoidable impurities comprise
   0.0015 wt% S and 0.01 wt% or less P.
FIG. 1
FIG. 2
FIG. 4
Comparative Steel

Inventive Steel

FIG. 5

Large strength reduction degree in high pearlite fraction → Pipe strength below standard (Comparative Steel)
INTERNATIONAL SEARCH REPORT

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/00(2006.01)i

According to International Patent Classification (IPC) or to both national classification and IPC

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)

IPC 8 C22C 38/14, C22C 38/16, C22C 38/50, C22C 38/58, C21D 9/06 C21D 9/46

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

Korean utility models and applications for utility models since 1975

Japanese utility models and applications for utility models since 1975

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

eKIPASS (KIPO internal) & keywords: high strength API-X80 grade hot-rolled steel plate, spiral pipes, and line pipe

C. DOCUMENTS CONSIDERED TO BE RELEVANT

<table>
<thead>
<tr>
<th>Category</th>
<th>Citation of document, with indication, where appropriate, of the relevant passages</th>
<th>Relevant to claim No</th>
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<tbody>
<tr>
<td>A</td>
<td>JP 08-209287 A (NIPPON STEEL CORPORATION) 13 August 1996 See the abstract, and claims 1-4 and examples</td>
<td>1 - 6</td>
</tr>
<tr>
<td>A</td>
<td>JP 07-173536 A (NIPPON STEEL CORPORATION) 11 July 1995 See the abstract, and claims 1-2 and examples</td>
<td>1 - 6</td>
</tr>
<tr>
<td>A</td>
<td>JP 2000-355729 A (NKK CORPORATION) 26 December 2000 See the abstract, and claims 1-2 and examples</td>
<td>1 - 6</td>
</tr>
<tr>
<td>A</td>
<td>JP 2004-04391 1A (JFE STEEL KK) 12 February 2004 See the abstract, claims 1-4, and examples</td>
<td>1 - 6</td>
</tr>
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</table>

Further documents are listed in the continuation of Box C

See patent family annex

* Special categories of cited documents
"A" document defining the general state of the art which is not considered to be of particular relevance
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"O" document referring to an oral disclosure, use, exhibition or other means
"P" document published prior to the international filing date but later than the priority date claimed

"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention
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Date of the actual completion of the international search
24 MARCH 2008 (24 03 2008)

Date of mailing of the international search report
24 MARCH 2008 (24.03.2008)

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<th>Patent family member(s)</th>
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</tr>
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<tr>
<td></td>
<td></td>
<td>CA2187028AA, CA2187028C</td>
<td>08.08.1996</td>
</tr>
<tr>
<td></td>
<td></td>
<td>CA2187028C, CN1148416A</td>
<td>31.07.2001</td>
</tr>
<tr>
<td></td>
<td></td>
<td>DE69607702C0, DE69607702T2</td>
<td>23.11.2000</td>
</tr>
<tr>
<td></td>
<td></td>
<td>KR100222302B1, N0964182A</td>
<td>01.10.1999</td>
</tr>
<tr>
<td></td>
<td></td>
<td>RU2136776C1, US5755895A</td>
<td>02.12.1999</td>
</tr>
<tr>
<td></td>
<td></td>
<td>W09623909A1</td>
<td>08.08.1996</td>
</tr>
<tr>
<td>JP07173536A</td>
<td>11.07.1995</td>
<td>NONE</td>
<td>NONE</td>
</tr>
</tbody>
</table>