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(54) STEEL PRODUCT FOR HIGH HEAT INPUT WELDING AND METHOD FOR PRODUCTION THEREOF

STAHLPRODUKT ZUM SCHWEISSEN MIT HOHEM WÄRMEEINTRAG UND VERFAHREN ZU SEINER HERSTELLUNG

ACIER POUR SOUDURES A FORT APPORT THERMIQUE ET SON PROCEDE DE PRODUCTION

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Description

BACKGROUND OF THE INVENTION

1. Field of the Invention

[0001] The present invention relates to steels used for various structures such as those in the fields of shipbuilding, construction, and civil engineering. More specifically, the present invention relates to steels suitably usable for high heat input welding with a heat input exceeding 400 kJ/cm.

2. Description of the Related Art

[0002] Generally, steels used in the fields of shipbuilding, construction, and civil engineering are processed by welding to fabricate structures having desired shapes. For these structures, the individual steels are required not only to exhibit a high parent-metal toughness as a matter of course but also to exhibit high weld-zone toughness. In recent years, there is a trend to fabricate larger structures for, for example, ships, by using steels having increased strengths and thicknesses. For weld-fabrication of such large structures, high-efficiency high heat input welding techniques, such as a submerged arc welding, an electrogas welding, and an electroslag welding, are employed. As such, steels exhibiting high weld-zone toughness need to be used for the weld-fabrication using the high heat input welding.

[0003] The JP-A-2 267241 discloses a steel for a line pipe having excellent hydrogen induced cracking resistance and sulfide stress corrosion cracking resistance. The composition of the steel for the line pipes is constitute of, by weight, <0.05% C, <=0.5% Si, 0.5 to 1.7% Mn, <=0.0070% P, <=0.0010% S, 0.01 to 0.10% Al, <=0.0050% N, <0.0010% O, 0.0010 to 0.0040% Ca and the balance Fe with inevitable impurities. To obtain such characteristic, the Japanese prior art document teaches to regulate ACR within the limits of 0.5<= ACR <=2.8, wherein ACR = (Ca (%) - (0.18 + 0.013 x Ca (%)) x O(%))/(1.25 x S(%)).

[0004] As is commonly known, with an increased welding heat input being applied, the microstructure of a weld heat affected zone (HAZ) is coarsened, and the toughness of the weld HAZ is thereby deteriorated. To overcome the problem of the reduction in toughness due to the high heat input welding, various countermeasures have been proposed to date.

[0005] For example, a technique that has already been put into practical use employs a method of suppressing the coarsening of austenite grains according to fine dispersion of TiN and actions of TiN serving as a ferrite transformation nucleus. In addition, techniques have been developed for dispersing oxides of Ti (Japanese Unexamined Patent Application Publication No. 57-51243) and for combining ferrite nucleation abilities BN (Japanese Unexamined Patent Application Publication No. 62-170459). Further, known techniques include a technique in which a high toughness is obtained by adding Ca and controlling the sulfide form (Japanese Unexamined Patent Application Publication No. 60-204863) and a technique in which a high toughness is obtained by adding REM and controlling the sulfide form (Japanese Unexamined Patent Application Publication No. 62-260041).

[0006] However, these conventional techniques arise problems as described hereunder. As described above, the techniques disclosed in the Japanese Unexamined Patent Application Publications No. 62-170459 and No. 60-204863 employ the method in which TiN is precipitated and the microstructure is refined to improve the toughness. However, in the weld HAZ heated in a range of temperatures causing TiN to be dissolved, the actions as described above do not occur. As such, the structure thereof is embrittled by solute Ti (titanium) and solute N (nitrogen), and the toughness is thereby significantly deteriorated. However, while the addition of B is effective for improving the HAZ toughness in a region where solid solution of TiN takes place, a problem occurs in a region where solid solution of TiN does not take place (where the heating temperature is 1,350°C or lower). In this region, the solute B that is non-connectable with N acts to significantly increase hardenability upon cooling after welding, and the HAZ microstructure is transformed into a hard bainite-based structure. Thereby, the toughness is significantly deteriorated.

[0007] In the technique disclosed in Japanese Unexamined Patent Application Publication No. 62-170459, the amount of addition of Al is reduced to prevent adverse effects of B. However, sufficient deoxidizing cannot be caused in steel making of the steel unless the amount of addition of Al exceeds 0.010%. In this case, since inclusions in the steel are increased, a sufficient toughness cannot be obtained.

[0008] In the technique disclosed in Japanese Unexamined Patent Application Publication No. 62-260041, REM is added to a region where dissolution of TiN takes place, and the region is refined into a fine structure according to actions of REM sulfides/oxides. However, there occurs a problem in that it is very difficult to cause sufficiently fine and uniform dispersion of the REM in steel making of the steel. This makes it difficult to secure a sufficient toughness in the weld HAZ heated up to a high temperature.

[0009] In the technique disclosed in Japanese Unexamined Patent Application Publication No. 57-51243, unlike the case of ordinary Al deoxidation, Ti deoxidation is performed, and Ti oxides or Ti compound oxides are dispersed into the steel to suppress the growth of austenite grains. As a result, while the oxide dispersion into the steel can be implemented to suppress the growth of austenite grains, it is very difficult to cause the oxides to disperse finely and uniformly into the steel. In addition, a critical problem occurs in that compared to TiN, the Ti oxides are coarser, and Charpy absorbed energy is thereby decreased. For these reasons, it is difficult to sufficiently suppress the growth of austenite grains in welding performed with a high heat input exceeding 400 kJ/cm, consequently making it difficult to secure a high toughness of the weld HAZ.

[0010] Moreover, problems remain pending resolution in the technique disclosed in the Japanese Unexamined Patent Application Publication No. 60-204863 in which Ca is added and the technique disclosed in the Japanese Unexamined Patent Application Publication No. 62-260041 in which REM is added. Even in these techniques, while a high toughness can be secured with a heat input of up to about 300 kJ/cm, it is difficult to secure a high toughness at the same level of the parent-metal toughness in welding with a high heat input exceeding 400 kJ/cm. Further JP 2002235114 discloses a method for producing a steel where ACR value is between 0 and 1 by which sufficient toughness can be obtained in a high heat input weld zone of >350 kJ/cm. In view of the conventional situations described above, an object of the present invention is to solve the above-described problems experienced with the conventional arts and to thereby provide a steel that enables a high weld HAZ toughness at the same level as that of a parent metal of the steel even after welding is performed with a high heat input exceeding 400 kJ/cm.

DISCLOSURE OF THE INVENTION

[0011] To achieve the object, the inventors carried out exhaustive researches. As a result, the inventors discovered that appropriate inclusion of Ca that is necessary for controlling the sulfide form is essential to improve the toughness of a weld heat affected zone (HAZ) even after welding performed with a high heat input exceeding 400 kJ/cm. More specifically, in order to improve the toughness of a high heat input weld HAZ, the inventors discovered that it is essential to suppress the coarsening of austenite grains in the high temperature region and to cause fine dispersion of the ferrite transformation nucleus that is necessary to accelerate ferrite transformation in the subsequent cooling stage. The conventional arts were insufficient in capability of achieving these essential factors.

[0012] Based on the discoveries of the researches, in the present invention, CaS is crystallized in the stage of solidification during formation of a steel plate from molten steel. In comparison with the oxide, since CaS is crystallized at a lower temperature, CaS can be finely dispersed. Noticeable one of the discoveries is that MnS is precipitated over the surface of CaS when a sufficient amount of solute S after crystallization of CaS is secured by controlling the contents of Ca and S and the amount of oxygen dissolved in the steel. MnS itself has the ferrite nucleation ability and is effective to accelerate the ferrite transformation by forming Mn depleted zones. In addition, the inventors discovered that the ferrite transformation is further accelerated by causing ferrite nucleation nuclei of TiN, AlN, and the like that are precipitated over MnS. These countermeasures described above enable fine dispersion of the ferrite-transformation generation nucleus that does not dissolve even at the high temperature during the high heat input welding. Consequently, the weld HAZ structure can be transformed into a ferrite and pearlite microstructure having a high toughness.

[0013] The present invention provides a steel for high heat input welding as defined in claim 1.

DESCRIPTION OF THE PREFERRED EMBODIMENT

[0014] Hereinbelow, a steel of an embodiment according to the present invention will be described. Specifically, the reasons for limiting individual compositions of the steel will be described.

C (Carbon): 0.03 to 0.15 Mass%

[0015] The lower limit of the C content is set to 0.03 mass% to secure a strength necessary for using the steel as a structural steel. Concurrently, the upper limit of the C content is set to 0.15 mass% in consideration of deterioration in weld-crack resistance. More suitably, the C content is preferably limited to a range from 0.05 to 0.10 mass%.

Si (Silicon): 0.05 to 0.25 Mass%

[0016] The Si content is required to be at least 0.05 mass% for steelmaking. However, with an Si content exceeding 0.25 mass%, the parent-metal toughness is deteriorated, and M-A (Martensite-Austenite) constituent is formed in a high heat input weld HAZ, whereby the toughness of the HAZ is deteriorated.

[0017] More suitably, the Si content is preferably limited to a range of from 0.13 to 0.22 mass%.

Mn (Manganese): 0.5 to 2.0 Mass%

[0018] An Mn content of 0.5 mass% Mn or higher is required to secure a sufficient parent-metal strength. However, with an Mn content exceeding 2.0 mass%, the weld-zone toughness is significantly deteriorated. More suitably, the Mn content is preferably limited to a range of from 0.8 to 1.6 mass%.

P (Phosphorus): 0,01 Mass% Maximum

[0019] The P content is limited to 0.01 mass% or lower.

S (Sulfur): 0.0015 to 0.0030 Mass%

[0020] An S content exceeding 0.0030 mass% acts to deteriorate the parent-metal toughness. The S content is limited to a range of from 0.0015 to 0.0025 mass% to generate CaS and MnS.

Al (Aluminum): 0.015 to 0.1 Mass%

[0021] An Al content of 0.015 mass% or higher is required to generate CaS and MnS. However, an Al content exceeding 0.1 mass% acts to deteriorate the parent-metal toughness and the weld metal toughness. More suitably, the S content is preferably limited to a range of from 0.02 to 0.06 mass%.

Ti (Titanium): 0.004 to 0.03 Mass%

[0022] Ti is precipitated in the form of TiN upon solidification, thereby contributing to suppression of coarsening of austenite grains in the weld HAZ, and contributes, as a ferrite transformation nucleus, to improvement in the toughness of the weld HAZ. With a Ti content lower than 0.004 mass%, the above-described effect is low; and with a Ti content exceeding 0.03 mass%, TiN grains are coarsened, and a desired effect cannot be obtained. More suitably, the Ti content is preferably limited to a range of from 0.008 to 0.02 mass%.

N (Nitrogen): 0.0020 to 0.0070 Mass%

[0023] N is an element necessary to secure a required amount of TiN. With an N content of lower than 0.0020 mass%, a sufficient amount of TiN cannot be secured. With an N content exceeding 0.0070 mass%, the toughness is significantly deteriorated due to an increase in the amount of solute N in a region where TiN is dissolved by a weld-heating cycle. More suitably, the N content is preferably limited to a range of from 0.0030 to 0.0055 mass%.

Ca (Calcium): 0.0005 to 0.0030 Mass%

[0024] Ca has a toughness-improving effect with S being fixed. Preferably, the Ca content is at least 0.0005 mass% to cause this effect to exhibit. However, the effect is saturated eveb with a Ca content exceeding 0.0030 mass%. For this reason, the content is limited to the range of from 0.0005 to 0.0030 mass%. More suitably, the Ca content is preferably limited to a range of from 0.0010 to 0.0020 mass%.

O (Oxygen): 0.0045 Mass% Maximum

[0025] With an O content exceeding 0.0045 mass%, since the amount of the inclusion is increased, the cleanliness of the steel is deteriorated. Consequently, the toughness is deteriorated.

$$0.3 \leq \mathrm{ACR} \leq 0.8$$

(wherein, ACR = (Ca - (0.18 + 130 × Ca) ×O)/1.25/S; and Ca, O, and S individually represent the contents (mass%) of the elements)

[0026] Ca and S need to be added to satisfy the relation 0.3 ≤ ACR ≤ 0.8. Fig. 1 shows the results of synthetic HAZ tests performed under two simulated hea-input conditions in which Ca was diversely added into a fundamental composition of the steel of the present invention. As can be seen from the figure, the toughness is significantly increased according to the relation 0.3 ≤ ACR ≤ 0.8 in either case where the time of 800-500°C cooling is 153 seconds or 270 seconds (improved by about 30°C in terms of vTrs). As shown in a microphotograph of Fig. 2, in the range of 0.3 ≤ ACR ≤ 0.8, the composition appears in the form of a compound sulfide with either MnS or MnS and TiN precipitated over CaS.

[0027] Unless a value representing ACR reaches 0.3, since CaS is not crystallized, S is precipitated in the form of only MnS. MnS is expanded by a rolling operation during steel-plate manufacture and causes reduction in the parent-metal toughness. Concurrently, since MnS is dissolved in the weld HAZ, which is a primary subject of the present invention, finely dispersion is not implemented. On the other hand, however, with the value of the ACR exceeding 0.8, since S is substantially fixed by the action of Ca, and MnS acting as the ferrite-generating nucleus is not precipitated over CaS, a sufficient function is not exhibited. Fig. 3 is a schematic view showing the relationship between ACR and the sulfide to be precipitated. With the optimum range of ACR according to the present invention, products formed by simultaneous precipitation of the compound sulfides of CaS and MnS and TiN exist. The quantity of the products is in a range of from 5×10^2 to 1×10^4 pieces/mm^2, and the average grain size thereof is in a range of from 0.1 to 5 μm. Thereby, ferrite-pearlite transformation in the weld HAZ is accelerated, and the toughness of the weld HAZ can be improved due to the microstructure refinement.

[0028] The present invention allows a steel of an embodiment to contain at least one or two selected elements from the elements V, Nb, Cu, Ni, Cr, and Mo that have a strength-improving function, as described hereunder.

B (Boron): 0.0004 to 0.0010 Mass%

[0029] B is effective in increasing hardenability during steel-plate manufacture. To secure this effect, the B content needs to be 0.0004 mass% or higher. However, addition of B exceeding 0.0010 mass% increases the hardenability, thereby decreasing the toughness of the weld HAZ.

V (Vanadium): 0.2 Mass% Maximum

[0030] V has the effect of improving the parent-metal strength and toughness. This effect can be secured with a V content of 0.01 mass% or higher. However, addition of V exceeding 0.2 mass% causes deterioration in the toughness.

Nb (Niobium): 0.05 Mass% Maximum

[0031] Nb has the effect of enabling the parent-metal strength and toughness and the weld-joint strength to be secured. This effect can be secured with an Ni content of 0.007 mass% or higher. Addition Nb exceeding 0.05 mass% causes deterioration in the weld-HAZ toughness.

Ni (Nickel): 1.5 Mass% Maximum

[0032] Ni has the effect of maintaining high parent-metal toughness and concurrently increasing the strength thereof. This effect can be secured with an Nb content of 0.10 mass% or higher. With a Ni content exceeding 1.5 mass%, since the effect is saturated, the content is specified to be the upper limit.

Cu (Copper): 1.0 Mass% Maximum

[0033] Cu exhibits an effect similar to that of Ni. This effect can be secured by including a Cu content of 0.10 mass% or higher. However, a Cu content exceeding 1.0 mass% causes hot embrittlement, thereby deteriorating the steel surface condition.

Cr (Chromium): 0.7 Mass% Maximum

[0034] Cr has the effect of increasing the parent-metal strength. This effect is secured with a Cr content of 0.05 mass% or higher. However, since addition of an excessive amount causes adverse effects on the toughness, the upper limit is set to 0.7 mass%.

Mo (Molybdenum): 0.7 Mass% Maximum

[0035] Mo has the effect of increasing the parent-metal strength. This effect is secured with a Cr content of 0.05 mass% or higher. However, since addition of an excessive amount causes adverse effects on the toughness, the upper limit is set to 0.7 mass%.

[0036] As described above, in the present invention, the compositions, specifically, Ca and S, are each regulated in the content to the limited range. Therefore, the steel exhibiting a high toughness in the weld HAZ in the high heat input welding can be provided.

[0037] The steel of the present invention is manufactured in, for example, a procedure as described hereunder. First, molten steel is refined using a convertor into steel. Then, an RH (Ruhrstahl-Heraeus) degassing process is performed, and the steel is formed into slabs through continuous casting or ingot-casting-blooming steps. Subsequently, each of the slabs is reheated to a temperature of 1, 250°C or lower, and is then hot-rolled to a predetermined thickness in a temperature range of from a heating temperature to 650°C. Thereafter, the hot-rolled steel is subjected to either an air-cooling process or an accelerated cooling process at a cooling rate of from 1 to 40°C/sec. Then, the cooling process is terminated at a temperature range of from 200 to 600°C, and air cooling is performed.

[0038] Alternatively, after the hot-rolling process, the hot-rolled steel is directly hardened from a temperature range of 650°C or higher, and is then tempered to a temperature of 500°C ±150°C. Still alternatively, the steel can be also manufactured according to a method selected from the steps wherein the hot-rolled steel is subjected to quenching after reheating in a temperature range of 850°C to 950°C and then, tempering to a temperature of 500°C±150°C; the hot-rolled steel is reheated to a temperature of 1,000°C or lower and normalized; or the hot-rolled steel is reheated to a temperature of 1,000°C or lower and normalized and subsequently, tempered to a temperature of 650°C or lower. Still alternatively, the manufacture can be achieved even under manufacturing conditions ordinarily used in hot rolling with a tandem roller being used. The steel plate according to the present invention is either a thick steel plate having a thickness of 6 mm or larger or a hot-rolled steel plate.

[0039] A welding method to be used for the steel plate of the present invention is not limited to a specific one. The welding method may be an arc welding method, a submerged arc welding method, an electroslag welding method, an electrogas welding method, or any one of other heating-source welding techniques.

(First Example)

[0040] Hereinbelow, the present invention will be described with reference to examples.

[0041] Steels having compositions shown in Tables 1 and 2 given below were produced using a 100-kg high frequency melting furnace and were cast into slabs each having a thickness of 100 mm. Subsequently, the slabs were heated for one hour up to 1,150°C and were then rolled by 50% of an overall draft in a temperature range of from 900 to 700°C into steel plates having a thickness of 20 mm. Subsequently, the steel plates were cooled in the manner of accelerated cooling at a cooling rate of 10°C/sec. The examples 33, 37, 38, 41 and 44 are of Tables 1 and 2 are Examples according to the present invention. All other examples are comparative.

[0042] From the steel plates, test specimens having a size of 80 mm (width) x 80 mm (length) x 15 mm (thickness) were prepared to measure properties after being subjected to welding thermal cycles. These test specimens were subjected to a welding thermal cycle set such that the rate of cooling from 800°C to 500°C after heating to 1,400°C was set to 1°C/sec (equivalent to a weld HAZ in electrogas welding with a heat input of 450 kJ/cm). Then, the test specimens were each evaluated for the weld-HAZ toughness according to the result of a 2-mm V-notch Charpy impact test. Table 3 shows thus-obtained weld-HAZ toughnesses together with parent-metal strengths and toughnesses. The parent-metal strengths were each obtained such that two JIS-Z2201 based test specimens were prepared from 1/2t-thick portions in the rolled direction of each of the rolled plates. The two test specimens were each tested in conformity with

[0043] the JIS-Z2241 requirements, and the average value was obtained from the test results. The toughnesses were each measured such that three JIS-Z2201 based V-notch test specimens were prepared from 1/2t thick portions in the direction perpendicular to the rolled direction of the rolled plate. The three test specimens were each tested in conformity to JIS-Z2242 to measure a brittle-ductile fracture transition temperature (vTrs). The toughness (represented by the fracture transition temperature) of each of the parent metals and the weld HAZs was determined to be excellent in accordance with a criterion set to a vTrs of -40°C or lower.

[0044] As can be seen from Table 3, in any one of inventive examples, a high weld-HAZ toughness satisfying vTrs ≤ -40°C was obtained. However, comparative examples were found to include those individually having low weld-HAZ toughnesses and even those individually having low parent-metal toughnesses. In these comparative examples, at least one of the value of (Ca - (0.18 + 130 x Ca) x O)/1.25/S and the contents of the compositions such as Ca, Ti, C, Mn, Si, S, N, Cu, Cr, Mo, V, and B was found to be out of the range specified in the present invention. For each comparative example steel 16 and 23, a steel plate having a thickness of 60 mm was produced by hot rolling. For each of these steel plates, a weld joint was produced by electrogas welding with a heat input of 450 kJ/cm, and a microstructure of a representative weld HAZ of a 1/4t thick portion was observed. Fig. 4 shows a microstructure taken of the inventive example steel 16, and Fig. 5 shows a microstructure taken of the comparative example steel 23. From these microstructures, grain-coarsening in the weld HAZ was found to appear conspicuously in the comparative example steel 23 shown in Fig. 5. In comparison, however, the microstructure of the weld HAZ in the inventive example steel 16 shown in Fig. 4 was found to have been refined to the same level as that of the microstructure of the parent metal. Thus, these results verify that the toughness of the high heat input weld HAZ is at the same level as that of the parent metal in the comparative example steel 16.

TABLE 1

		C	Si	Mn	P	S	Al	Ti	N	Ca	O	Expression (1)
	1	0.06	0.20	1.50	0.008	0.0015	0.030	0.012	0.0060	0.0020	0.0020	0.60
	2	0.07	0.15	1.45	0.007	0.0010	0.044	0.011	0.0055	0.0017	0.0018	0.78
	3	0.08	0.18	1.55	0.010	0.0015	0.040	0.010	0.0050	0.0010	0.0012	0.33
	4	0.10	0.11	1.10	0.012	0.0005	0.030	0.010	0.0050	0.0008	0.0020	0.37
	5	0.09	0.05	0.80	0.020	0.0025	0.031	0.024	0.0060	0.0022	0.0020	0.41
	6	0.05	0.10	1.80	0.008	0.0030	0.018	0.020	0.0065	0.0030	0.0015	0.57
	7	0.13	0.08	1.60	0.009	0.0010	0.015	0.004	0.0020	0.0020	0.0026	0.68
	8	0.15	0.14	1.50	0.004	0.0012	0.018	0.008	0.0035	0.0020	0.0019	0.78
	9	0.03	0.16	1.30	0.004	0.0005	0.030	0.012	0.0055	0.0028	0.0043	0.74
	10	0.07	0.18	1.40	0.010	0.0006	0.025	0.016	0.0060	0.0009	0.0016	0.57
	11	0.06	0.15	1.76	0.018	0.0017	0.020	0.028	0.0068	0.0020	0.0025	0.42
	12	0.08	0.19	1.48	0.013	0.0020	0.015	0.11	0.0044	0.0012	0.0010	0.35
	13	0.11	0.08	1.57	0.008	0.0015	0.015	0.009	0.0036	0.0014	0.0015	0.46
	14	0.07	0.24	0.95	0.007	0.0020	0.060	0.005	0.0025	0.0025	0.0016	0.68
	15	0.08	0.20	1.24	0.008	0.0005	0.055	0.012	0.0052	0.0016	0.0031	0.64
	16	0.08	0.15	1.50	0.012	0.0009	0.033	0.022	0.0057	0.0014	0.0019	0.63
	17	0.04	0.10	1.94	0.010	0.0015	0.025	0.018	0.0068	0.0020	0.0025	0.48
	18	0.15	0.12	0.58	0.008	0.0013	0.041	0.012	0.0054	0.0009	0.0012	0.33
	19	0.06	0.19	1.54	0.007	0.0020	0.030	0.012	0.0040	0.0002	0.0018	-0.07
	20	0.08	0.15	1.50	0.006	0.0015	0.035	0.018	0.0065	0.0028	0.0012	1.15
	21	0.08	0.10	1.45	0.013	0.0005	0.045	0.009	0.0043	0.0015	0.0013	1.62
	22	0.08	0.08	0.95	0.008	0.0029	0.007	0.010	0.0035	0.0045	0.0022	0.78
	23	0.14	0.16	1.75	0.015	0.0006	0.025	0.009	0.0025	0.0001	0.0015	-0.25
	24	0.05	0.30	1.50	0.008	0.0016	0.030	0.038	0.0075	0.0025	0.0025	0.62
	25	0.18	0.20	1.50	0.008	0.0063	0.025	0.011	0.0036	0.0015	0.0025	0.07
	26	0.08	0.12	2.45	0.008	0.0015	0.043	0.002	0.0085	0.0008	0.0016	0.18

Expression (1): (Ca - (0.18 + 130 x Ca) x 0)/1.25/S

TABLE 2

		C	Si	Mn	P	S	Al	Cu	Ni	Cr	Mo	Ti	Nb	V	B	N	Ca	O	Expression (1)
	27	0.05	0.15	1.55	0.007	0.0015	0.033	0.22				0.11				0.0059	0.0022	0.0018	0.73
	28	0.07	0.15	1.42	0.004	0.0011	0.044		0.41			0.025				0.0054	0.0015	0.0017	0.63
	29	0.05	0.13	1.58	0.006	0.0020	0.041			0.35		0.012				0.0030	0.0012	0.0012	0.32
	30	0.11	0.14	1.55	0.015	0.0025	0,039				0.33	0.012				0.0039	0.0015	0.0010	0.36
	31	0.14	0.08	1.50	0.022	0.0006	0.025					0.013	0.015			0.0057	0.0016	0.0027	0.74
	32	0.04	0.12	1.44	0.008	0.0018	0.020					0.010		0.035		0.0063	0.0025	0.0015	0.77
	33	0.07	0.17	1.12	0.007	0.0016	0.018					0.009			0.0009	0.0044	0.0021	0.0025	0.48
	34	0.06	0.09	1.55	0.006	0.0012	0.019					0.011	0.013		0.0010	0.0042	0.0020	0.0020	0.75
	35	0.08	0.16	1.30	0.007	0.0019	0.031					0.012	0.015			0.0056	0.0030	0.0035	0.42
	36	0.07	0.15	1.56	0.013	0.0017	0.042					0.014	0.018		0.0004	0.0057	0.0021	0.0016	0.65
	37	0.07	0.18	1.52	0.008	0.0017	0.035					0.024		0.030	0.0005	0.0055	0.0021	0.0019	0.58
	38	0.08	0.13	1.58	0.009	0.0016	0.040	0.20	0.40			0.013	0.011		0.0005	0.0055	0.0016	0.0015	0.51
	39	0.12	0.08	1.22	0.015	0.0016	0.019			0.20	0.20	0.008	0.012			0.0030	0.0016	0.0015	0.51
	40	0.15	0.05	0.95	0.018	0.0024	0.055		0.25			0.005	0.015	0.020	0.0007	0.0025	0.0024	0.0015	0.55
	41	0.09	0.20	1.85	0.006	0.0027	0.048		1.25			0.013	0.009	0.015	0.0006	0.0032	0.0014	0.0010	0.31
	42	0.03	0.24	1.75	0.017	0.0008	0.035	0.70	0.70	0.10	0.10	0.028				0.0068	0.0015	0.0019	0.79
	43	0.07	0.16	1.59	0.015	0.0030	0.020	0.20	0.40	0.10	0.10	0.022			0.0005	0.0060	0.0016	0.0011	0.31
	44	0.08	0.14	0.77	0.007	0.0016	0.046	0.15	0.30	0.10	0.20	0.015	0.008	0.010	0.0005	0.0053	0.0010	0.0009	0.36
	45	0.07	0.19	1.54	0.009	0.0021	0.031	0.25	0.30			0.013				0.0033	0.0001	0.0025	-0.15
	46	0.06	0.15	1.50	0.015	0.0016	0.040					0.014	0.013		0.0008	0.0045	0.0025	0.0012	0.95
	47	0.05	0.10	1.45	0.011	0.0008	0.055			0.20	0.20	0.008		0.050	0.0005	0.0062	0.0016	0.0012	1.13
	48	0.10	0.08	0.95	0.007	0.0028	0.010	0.20	0.20			0.012	0.009	0.020		0.0063	0.0044	0.0017	0.89
	49	0.13	0.16	1.75	0.016	0.0015	0.040	0.15	0.15	0.10	0.10	0.012	0.013	0.015	0.0007	0.0030	0.0002	0.0015	-0.06
	50	0.08	0.20	1.55	0.015	0.0019	0.044				1.30	0.015			0.0035	0.0060	0.0026	0.0022	0.61
	51	0.12	0.15	1.45	0.008	0.0022	0.050	0.30	1.50	1.10		0.015	0.055			0.0040	0.0016	0.0015	0.37

(continued)

	C	Si	Mn	P	S	Al	Cu	Ni	Cr	Mo	Ti	Nb	V	B	N	Ca	O	Expression (1)
52	0.05	0.16	1.50	0.009	0.0011	0.030	1.20				0.012		0.250		0.0030	0.0015	0.0020	0.55

Expression (1): (Ca - (0.18 + 130 x Ca) x 0)/1.25/S

TABLE 3

		YS	TS	BM (vTrs)	HAZ (vTrs)
	1	414	520	-68	-59
	2	413	515	-69	-59
	3	450	553	-65	-55
	4	374	495	-71	-63
	5	293	450	-75	-71
	6	460	567	-63	-54
	7	520	630	-57	-48
	8	528	614	-59	-47
	9	329	443	-76	-67
	10	404	497	-70	-60
	11	467	575	-62	-53
	12	435	535	-66	-57
	13	488	600	-60	-51
	14	303	450	-75	-70
	15	377	464	-74	-62
	16	431	530	-67	-57
	17	478	588	-61	-52
	18	327	435	-77	-67
	19	423	519	-68	-20
	20	431	530	-67	-10
	21	424	521	-68	-15
	22	308	379	-82	-27
	23	570	701	-50	-16
	24	406	500	-10	-5
	25	570	701	0	10
	26	635	781	-15	0
	27	409	503	-70	-59
	28	420	517	-68	-58
	29	505	622	-58	-49
	30	594	731	-47	-41
	31	511	629	-57	-49
	32	374	460	-74	-63
	33	343	422	-78	-66
	34	419	515	-68	-58
	35	394	485	-71	-61
	36	437	538	-66	-56
	37	433	533	-67	-57

(continued)

		YS	TS	BM (vTrs)	HAZ (vTrs)
	38	466	574	-63	-53
	39	542	667	-53	-46
	40	414	509	-69	-59
	41	571	702	-50	-43
	42	510	627	-57	-49
	43	516	635	-57	-48
	44	380	468	-73	-62
	45	445	547	-40	-15
	46	411	506	-30	-9
	47	506	623	-35	0
	48	349	429	-46	-3
	49	622	765	-29	-12
	50	625	725	10	15
	51	655	715	13	20
	52	422	585	-5	0
BM: Parent-metal toughness HAZ: Weld-HAZ toughness					

INDUSTRIAL APPLICABILITY

[0045] As described above, according to the present invention, a steel having a weld-HAZ toughness even after welding is performed with a high heat input of 400 kJ/cm or higher can be obtained. As such, the present invention greatly contributes to improvement in the quality of a large structure that is fabricated by high heat input welding, such as submerged arc welding, electrogas welding, and/or electroslag welding. Needless to say, the steel has a high weld-HAZ toughness in a heat-input range of 400 kJ/cm or lower.

Claims

1. A steel for high heat input welding, **characterized in that** a composition of the steel comprises:

C (carbon)	0.03 to 0.15 mass percent (mass%);
Si (silicon)	0.05 to 0.25 mass%;
Mn (manganese)	0.5 to 2.0 mass%;
P (phosphorus)	≤ 0.01 mass%;
S (sulfur)	0.0015 to 0.0030 mass%;
Al (aluminium)	0.015 to 0.1 mass%;
Ti (titanium)	0.004 to 0.03 mass%;
N (nitrogen)	0.0020 to 0.0070 mass%;
Ca (calcium)	0.0005 to 0.0030 mass%,
B (boron)	0.0004 to 0.0010 mass %; and
O (oxygen)	≤ 0.0045 mass %;

wherein:

individual contents of Ca, O (oxygen), and S satisfy the following expression (1), and the balance of the composition Fe (iron) and unavoidable impurities.

$$0.3 \leq ACR \leq 0.8 \qquad (1)$$

where ACR = (Ca - (0.18 + 130 x Ca) x O)/1.25/S, where Ca, O, and S each represent the content (mass%) thereof; and optionally comprising one or two or more selected from.

V (vanadium)	0.2 mass% maximum;
Cu (copper)	1.0 mass% maximum;
Ni (nickel)	1.5 mass% maximum;
Cr (chromium)	0.7 mass% maximum;
Nb (niobium)	0.05 mass % maximum; and
Mo (molybdenum)	0.7 mass% maximum.

Patentansprüche

1. Stahl zum Schweißen mit hohem Wärmeeintrag, **dadurch gekennzeichnet, dass** eine Zusammensetzung des Stalles umfasst:

C (Kohlenstoff)	0,03 bis 0,15 Masseprozent (Masse-%);
Si (Silizium)	0,05 bis 0,25 Masse-%;
Mn (Mangan)	0,5 bis 2,0 Masse-%;
P (Phosphor)	s 0,01 Masse-%;
S (Schwefel)	0,0015 bis 0,0030 Masse-%;
Al (Aluminium)	0,015 bis 0,1 Masse-%;
Ti (Titan)	0,004 bis 0,03 Masse-%;
N (Stickstoff)	0,0020 bis 0,0070 Masse-%;
Ca (Calcium)	0,0005 bis 0,0030 Masse-%,
B (Bor)	0,0004 bis 0,0010 Masse-%; und
O (Sauerstoff)	≤ 0,0045 Masse-%;

wobei:

die Einzelgehalte von Ca, O (Sauerstoff) und S den folgenden Ausdruck (1) erfüllen und der Rest der Zusammensetzung Fe (Eisen) und unvermeidbare Verunreinigungen umfasst:

$$0{,}3 \leq ACR \leq 0{,}8 \qquad (1)$$

wobei ACR = (Ca - (0,18 + 130 x Ca) x O)/1,25/S, wobei Ca, O und S jeweils den Gehalt (Masse-Prozent) dieser darstellen; und optional umfassend ein oder zwei oder mehr gewählt aus:

V (Vanadium)	maximal 0,2 Masse-%;
Cu (Kupfer)	maximal 1,0 Masse-%;
Ni (Nickel)	maximal 1,5 Masse-%;
Cr (Chrom)	maximal 0,7 Masse-%;
Nb (Niob)	maximal 0,05 Masse-% und
Mo (Molybdän)	maximal 0,7 Masse-%.

Revendications

1. Acier pour soudure à fort apport thermique, **caractérisé en ce qu'**une composition de l'acier comprend :

C (carbone)	0,03 à 0,15 pour cent en masse (% en masse) ;
Si (silicium)	0,05 à 0,25 % en masse ;
Mn (manganèse)	0,5 à 2,0 % en masse;
P (phosphore)	≤ 0,01 % en masse;
S (soufre)	0,0015 à 0,0030 % en masse ;
Al (aluminium)	0,015 à 0,1 % en masse ;
Ti (titane)	0,004 à 0,03 % en masse;
N (azote)	0,0020 à 0,0070 % en masse ;
Ca (calcium)	0,0005 à 0,0030 % en masse,
B (bore)	0,0004 à 0,0010 % en masse ; et
O (oxygène)	≤ 0,0045 % en masse;

où :

les teneurs individuelles en Ca, O (oxygène), et S satisfont l'expression (1) suivante, et
le reste de la composition Fe (fer) et impuretés inévitables,

$$0,3 \leq ACR \leq 0,8 \qquad (1)$$

où ACR = (Ca - (0,18 + 130 x Ca) x O)/1,25/S, où Ca, O, et S représentent chacun leur teneur respective (% en masse) ; et comprenant éventuellement un ou deux ou plusieurs choisis parmi

V (vanadium)	0,2 % en masse maximum ;
Cu (cuivre)	1,0 % en masse maximum ;
Ni (nickel)	1,5 % en masse maximum ;
Cr (chrome)	0,7 % en masse maximum ;
Nb (niobium)	0,05 % en masse maximum ; et
Mo (molybdène)	0,7 % en masse maximum.

FIG. 1

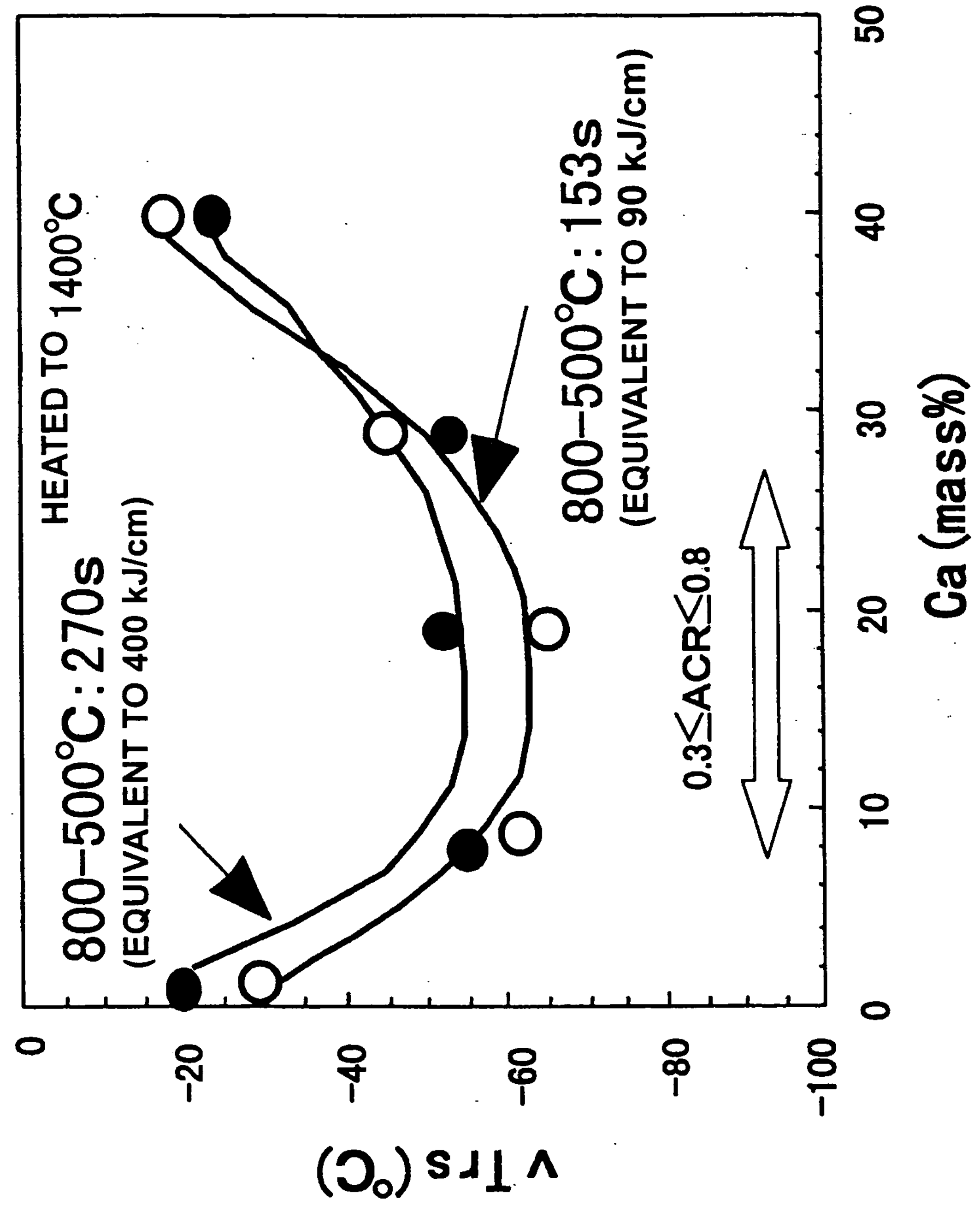

HEATED TO 1400°C
800−500°C : 270s
(EQUIVALENT TO 400 kJ/cm)
800−500°C : 153s
(EQUIVALENT TO 90 kJ/cm)
0.3≦ACR≦0.8
Ca (mass%)
v Trs (°C)
0
-20
-40
-60
-80
-100
10
20
30
40
50

FIG. 2

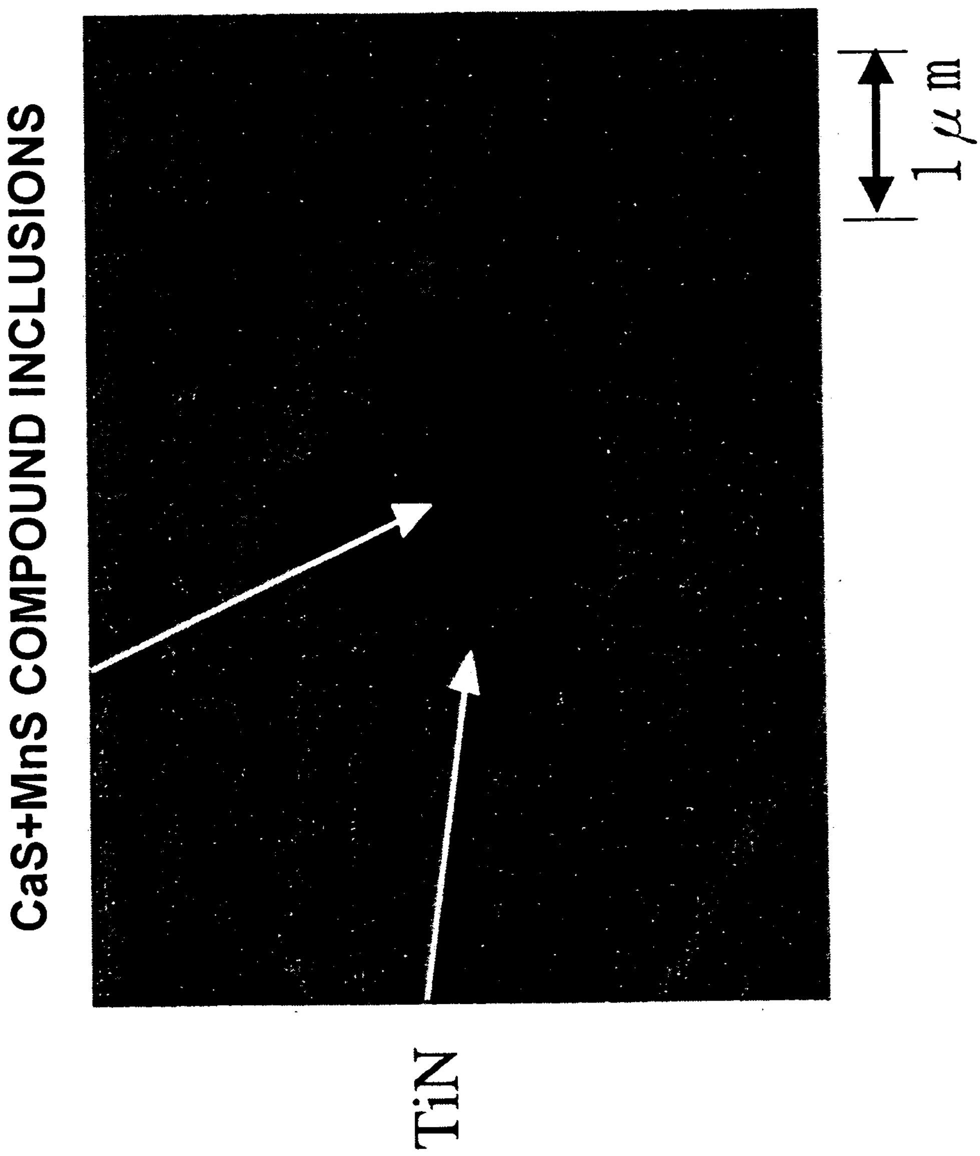


FIG. 3

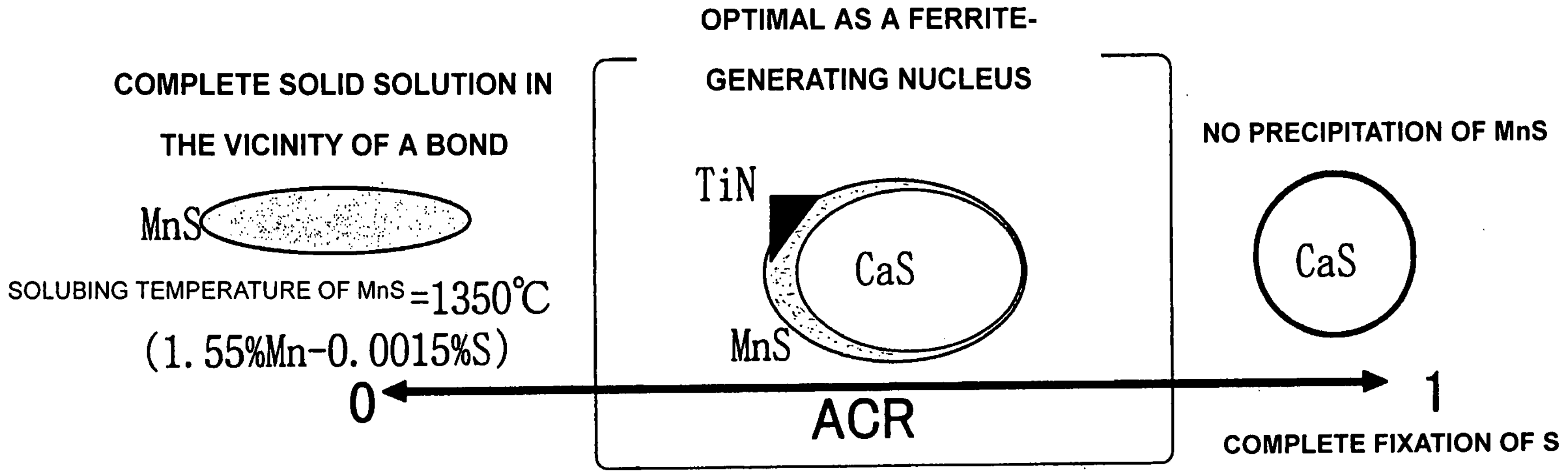

COMPLETE SOLID SOLUTION IN
THE VICINITY OF A BOND
MnS
SOLUBING TEMPERATURE OF MnS =1350℃
(1.55%Mn-0.0015%S)
OPTIMAL AS A FERRITE-
GENERATING NUCLEUS
TiN
CaS
MnS
NO PRECIPITATION OF MnS
CaS
0
ACR
1
COMPLETE FIXATION OF S

FIG. 4

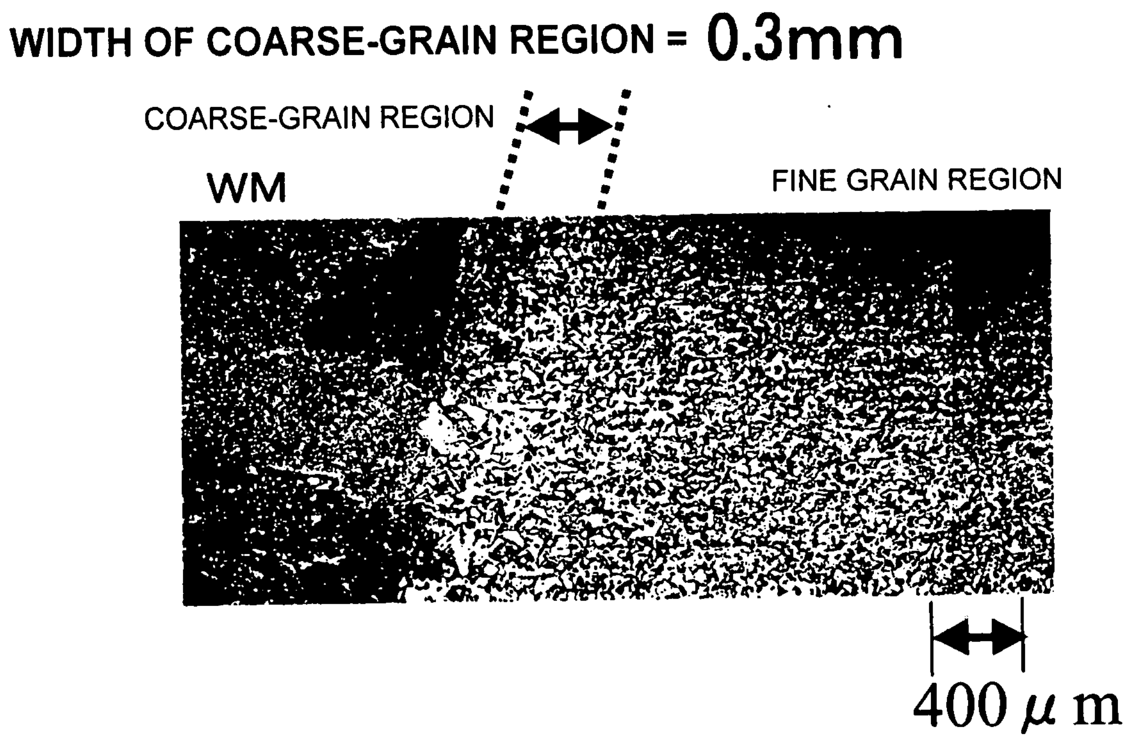
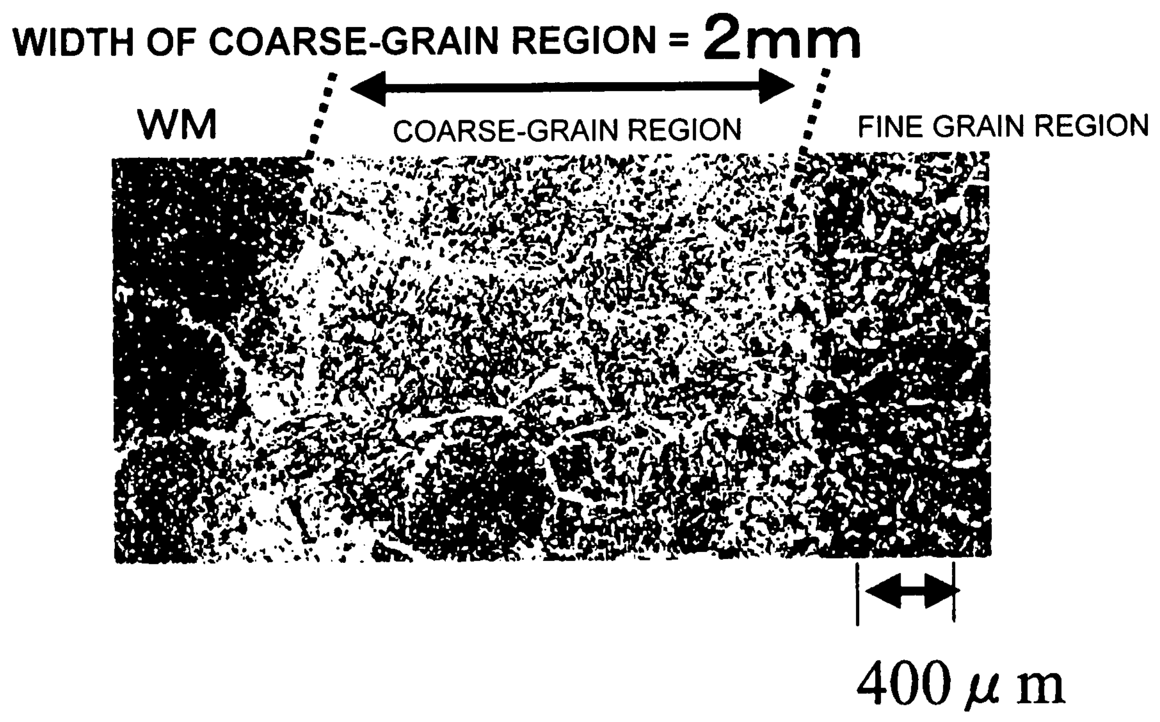

WIDTH OF COARSE-GRAIN REGION = 0.3mm
COARSE-GRAIN REGION
WM
FINE GRAIN REGION
400 μ m

FIG. 5


WIDTH OF COARSE-GRAIN REGION = 2mm
WM
COARSE-GRAIN REGION
FINE GRAIN REGION
400 μ m

REFERENCES CITED IN THE DESCRIPTION

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