UDC621.791.01:539.4

Development of High-Performance Welding Technology for Steel Plates and Pipe for Structural Purposes

Yukihiko Horii*1 Shigeru Ohkita*1
Kouichi Shinada*1 Kunio Koyama*1

Abstract:

Oil development is proceeding into arctic and deep-sea regions, and efficiency of the operation of oil pipelines and refining facilities is being improved by a high pressure process. The properties required of steel structures in the energy field are higher than ever low-temperature toughness and strength as well as greater weldability, these include lower preheat temperature and higher welding efficiency. Sulfide stress cracking (SSC) resistance and localized seawater corrosion resistance must also be satisfied in some cases. The construction field calls for stronger steels, such as TS780 and TS590, and now demands fire resistance as well. The compatibility between steels and welding materials assumes increasing importance with these stricter property requirements. From the standpoint of oxide metallurgy, technologies for microstructural control and weld metal design have been developed for specific steels, and welding technology is now available offshore structures, SSC-resistant steel pipe, and fire-resistant steels, among other things.

1. Introduction

Development of oil and gas resources is expanding into the North Sea, the Arctic Ocean and other frigid regions, and pipeline and other oil and gas production and refining facilities are operating at increasing pressure for higher efficiency. The changes in steel property requirements accompanying these worldwide moves in the energy development field are summarized as the relationship between the tensile strength and Charpy test temperature of steels in **Fig. 1**.

As offshore oil drilling structures increase in water depth and size, the substructure must be built of steel of greater thickness to avoid buckling, and the steel of the superstructure must be stronger to reduce weight. Cold and other arctic regions call for maximum Charpy impact and CTOD (crack tip opening displacement) test temperatures of −60°C and −40°C, respectively, which are lower than those for conventional regions[1,2].

The high pressure transmission through pipelines is accelerating the use of steel pipe of higher strength: X-70 and X-80[3] in place of X-55. Technical trends in the energy development field are clearly toward greater strength and toughness. Furthermore, there is a call for steels weldable with high heat input which can be fabricated on existing equipment efficiently.

New properties are being demanded in addition to strength and low-temperature toughness. In icy sea regions, welds in the splash zone must be protected against localized seawater corrosion[4]. Sulfide stress cracking (SSC) is a problem with pipeline and tanks for gas and oil containing hydrogen sulfide (H_2S)[5]. To prevent SSC, Vickers hardness of welds is often specified at 248 or less. Environmental disruption also has drawn attention, and

*1 Technical Development Bureau

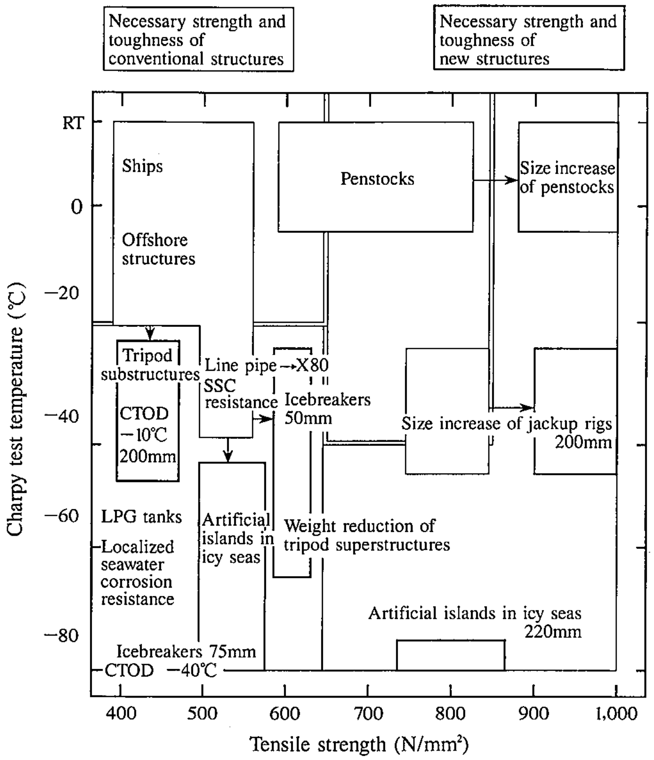


Fig. 1 Changes in strength and toughness required in the energy development field

oil tankers must now be constructed with double-hulls to prevent oil from spilling even if a tanker is grounded. This double-hull requirement has increased the length of one-side welding and fillet welding by about 1.5 times, calling for improvement in welding efficiency[6].

Emphasis in Japan is now on augmentation of social capital and expansion of domestic demand. In the construction field, earthquake resistance is accomplished by reducing the yield ratio of welded joints, and rigid joints where the weld metal does not yield before the base metal are considered essential. The Yokohama Landmark Tower is built of new SM580 steel in combination with conventional SM490 steel[7]. TS780 steel is required for high-rise building, and it is necessary to develop steels and welding techniques with weldability in mind[8]. The development of fire-resistant steels[9] has called for fire-resistant welding materials.

The energy and construction fields demand material of higher strength, toughness, and weldability (lower preheat temperature and greater efficiency), and have also added those that are resistant to corrosion and fire. Various types of steels to meet these requirements have been developed utilizing the thermomechanical control process (TMCP). To make the best use of the properties of these new steels requires appropriate welding technology, and the development of such technology which meets recent requirements is reviewed here.

2. Problems with Welding Materials for Low-Temperature Service

Welding materials for low-temperature service can be categorized into three main types by their toughening mechanism: the low-nickel type where 11% or less nickel is added to improve toughness by the solid solution of nickel; the titanium-boron type where the grain size of the microstructure is refined by adding titanium and boron[10]; and the high-nickel type where austenite are fully formed to prevent brittle fracture. The low-nickel type and high-nickel type are not suited for welding with high heat input, but the titanium-boron type permits high-heat input welding. It is anticipated that welding technology will be developed to guarantee the specified impact strength at −60°C or less for steels with a tensile strength of 490 to 590 N/mm².

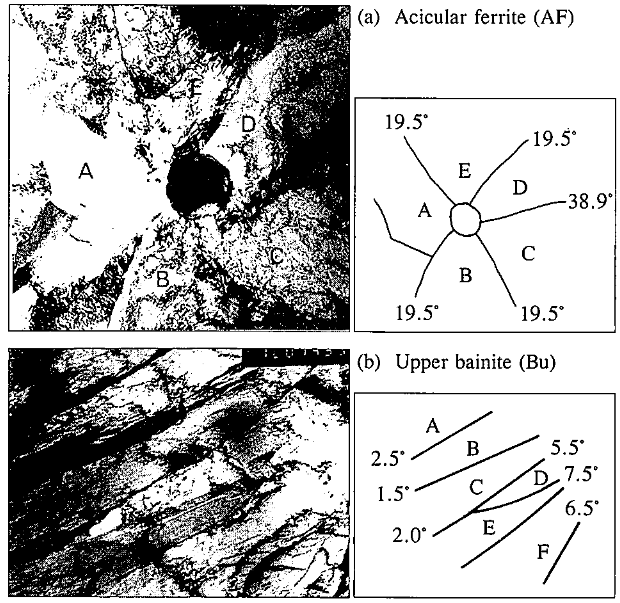


Fig. 2 TEM observation examples of low-alloy steel weld metal

The titanium-boron type produces ferrite called acicular ferrite (AF) and radially transforms from the nucleus of a nonmetallic inclusion as shown by the dark area in **Fig. 2**. Each acicular ferrite grain acts as a fracture facet; this assures high toughness. With the non-titanium-boron types, adjacent ferrite grains transform into almost parallel upper bainite (Bu) grains. However, two or more adjoining Bu grains usualy fail at the same time, resulting in larger facets and lower toughness.

Fig. 3 shows the toughness of welds made from steels with different aluminum and other strong deoxidizer content using some welding materials of type. It is found that the toughness of the weld metal depends on its aluminum content, and there is an optimum aluminum content for toughness. When the aluminum content is optimal, the weld metal consists of acicular ferrite. With higher aluminum content, the weld metal is upper bainite and toughness is reduced, and with lower alminum content the weld metal is grain-boundary ferrite (GBF). Aluminum content changes with dilution from the base metal, and the aluminum content of the base metal is seen to influence the toughness of the weld metal.

High-heat input weldable steels based on different concepts of toughness improvement, such as TiB steel[11], TiN steel[12] and TiO steel[13], were developed as TMCP steels for low-temperature service. These steels differ greatly in aluminum content. When the service temperature is low and the dilution rate is high, welding materials appropriate for specific steels are required. The transformation nucleus of **Fig. 2** was regarded as oxide inclusion containing TiO[10]. There have been examples of the formation of acic-

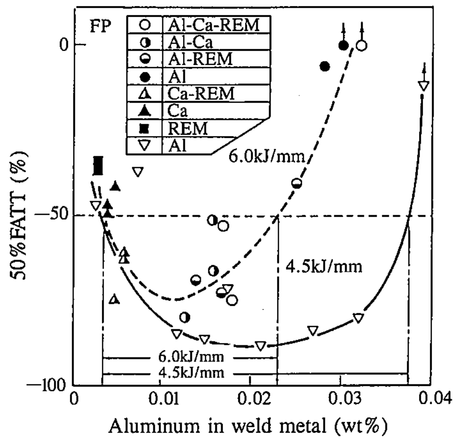


Fig. 3 Effect of aluminum on toughness of Ti-B weld metal

ular ferrite in the weld metal being governed by the aluminum content of the base metal, and acicular ferrite being obtained without appreciable titanium addition[14,15]. Stable acicular ferrite formation technology must be analyzed from the standpoint of oxide metallurgy.

3. Stable Formation of Acicular Ferrite by Oxide Metallurgy

3.1 Caracteristic of nuclei and effective way to get acicular ferrite

A weld metal with an AF + GBF microstructure (A), one with a Bu microstructure (B), and one with an AF + GBF microstructure (C) (**Table 1**) were deposited using titanium-boron-bearing, non-titanium bearing, and trace titanium-bearing electrodes, respectively, and analyzed for nonmetallic inclusions.

The X-ray diffraction results of the electrolytic extraction residues of the three samples are shown in **Fig. 4**, and TEM observations of the nonmetallic inclusions in **Fig. 5**. The weld metal B with the upper bainite microstructure has no particular peaks in the X-ray diffraction spectrum of **Fig. 4**, and its electron beam analysis image shows a halo pattern; this means that the nonmetallic inclusions in this metal are amorphous. The weld metals A and C, each with acicular ferrite, have almost the same diffraction peaks (**Fig. 4**), meaning that they contain the same nonmetallic inclusions. Considering the X-ray diffraction results in combination with the electron beam analysis results of **Fig. 5**, these nonmetallic inclusions were identified as spinel oxide ($MnO \cdot Al_2O_3$). The lattice misfit of the spinel oxide with alpha iron is 1.8% and is better than 3.2% for titanium monoxide (TiO) which has reported as nucleation site. Acicular ferrite can easily grow epitaxially from spinel nuclei.

The formation of acicular ferrite from an amorphous interface requires a large amount of energy. The microstructural differences between samples A and B depend on whether or not the oxides contained are coherent. Manganese sulfide (MnS)[16], claimed to be effective for transformation in steel, was similarly deposited on both samples and was not effective in a weld metal. The results of analysis by energy-dispersive X-ray spectroscopy (EDS) show that the three samples differ in the composition of their nonmetallic inclusions, only in that samples A and C contain titanium. This shows that titanium plays an important role in forming spinel. It is also known to be a glass crystallization catalyst, and its presence in the oxide is considered to have promoted the formation of spinel. The titanium detected in both samples A

Table 1 Chemical compositions of weld metals for analysis of nonmetallic inclusions (wt%)

Symbol	C	Si	Mn	P	S	T.Al	Ins.Al	T.Ti	Ins.Ti	B*	N*	O*	Microstructure
A	0.04	0.17	1.62	0.006	0.003	0.009	0.009	0.011	0.009	2	26	269	AF+GBF
B	0.11	0.18	1.68	0.016	0.005	0.007	0.005	<0.002	<0.001	<1	30	134	Bu+GBF
C	0.07	0.24	1.55	0.012	0.003	0.009	0.009	0.002	0.002	<1	67	302	AF+GBF

*:ppm

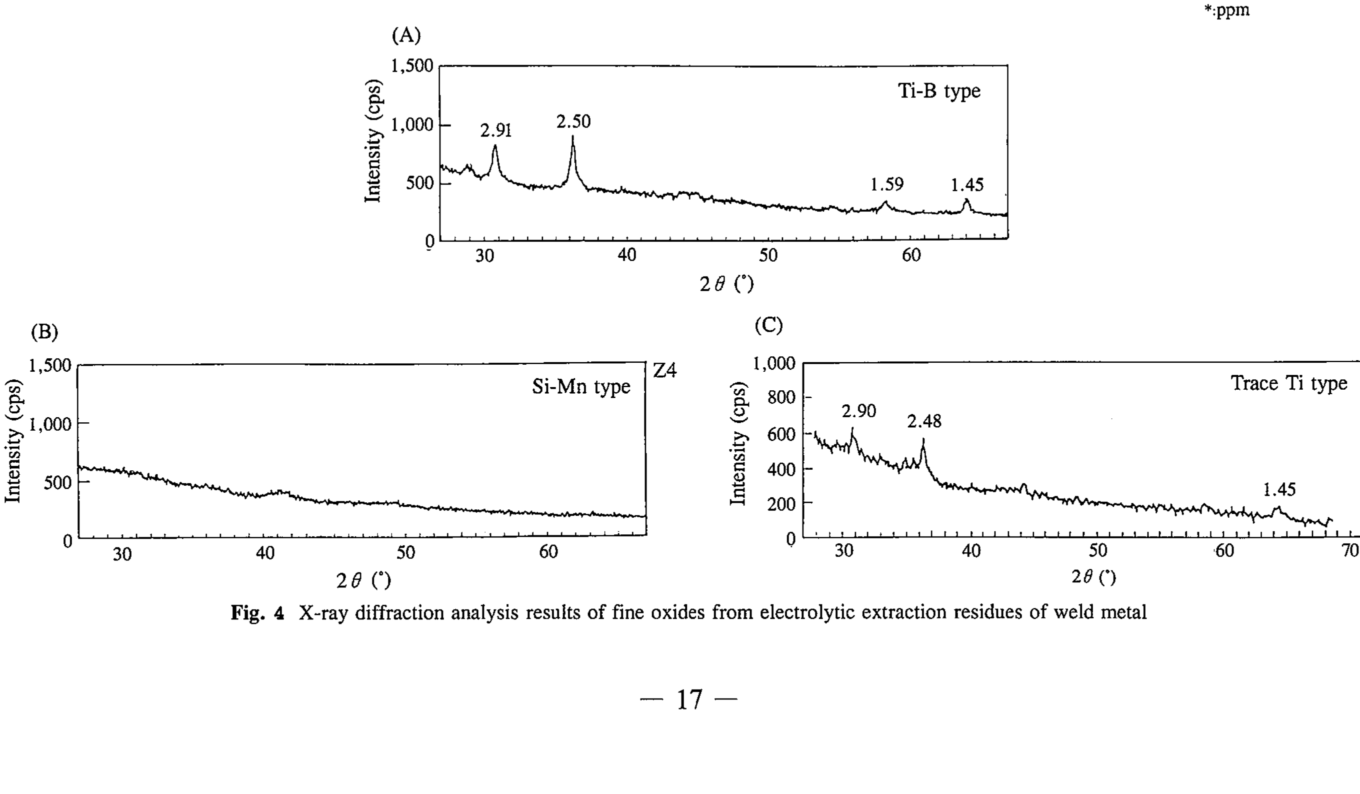
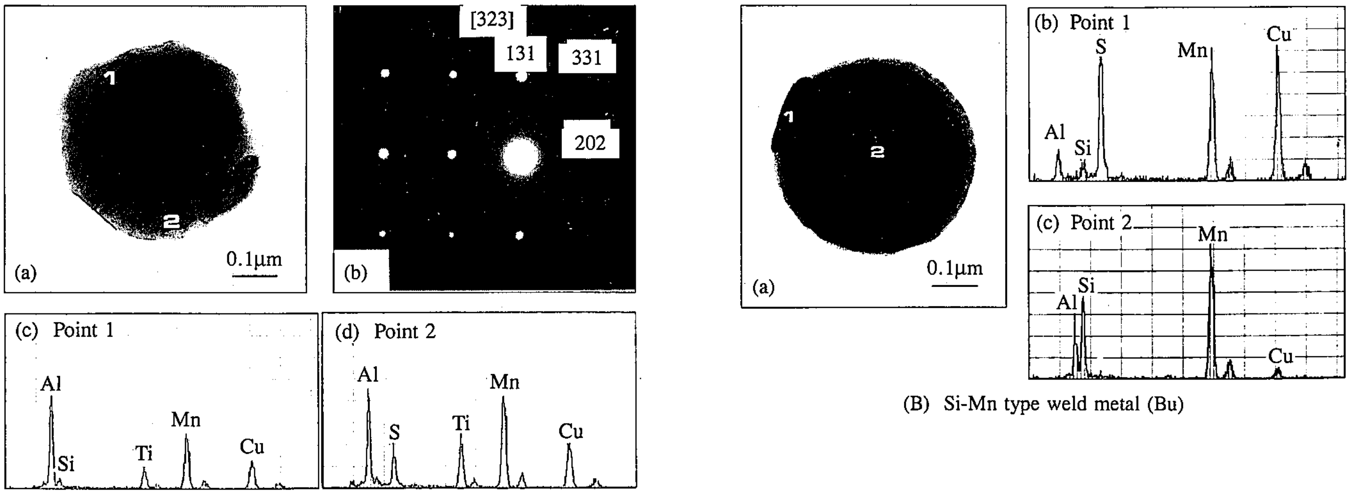


Fig. 4 X-ray diffraction analysis results of fine oxides from electrolytic extraction residues of weld metal



(A) Ti-B type weld metal (AF + GBF)

(B) Si-Mn type weld metal (Bu)

Fig. 5 TEM observations of nonmetallic inclusions

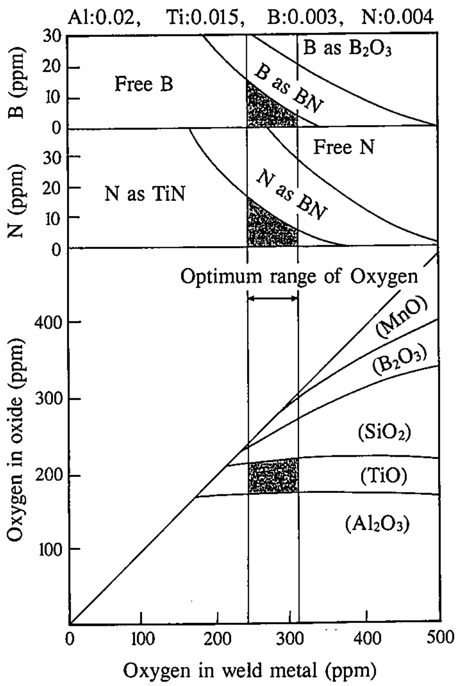


Fig. 6 Optimum ranges of Al, Ti, B, N, and O in Ti-B type weld metal

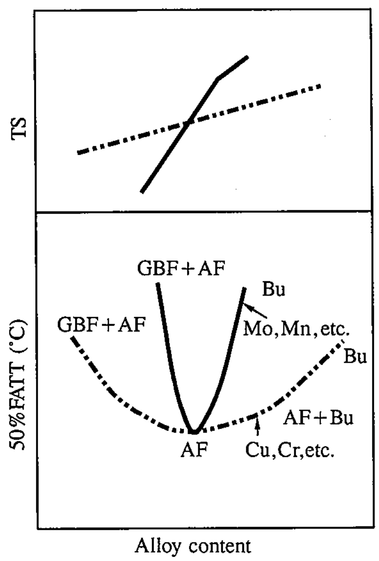


Fig. 7 Schematic showing effects of alloying elements on strength and toughness of Ti-B type weld metal

and C is thought to have replaced some of the manganese and aluminum to form (Mn,Ti)O·$(Al,Ti)_2O_3$. Since the oxidation reaction of deoxidation elements in the weld pool proceeds in an increasing order of affinity for oxygen[17], all oxygen reacts with aluminum at the high end of the aluminum range where the aluminum/oxygen ratio of 1.125 in Al_2O_3 is exceeded. This prevents the oxidation of titanium and hence the formation of spinel. Both aluminum and titanium must be oxidized for the formation of spinel. An important condition is the oxygen potential (Al/O < 1.125) after cessation of aluminum deoxidation. In addition the consumption of free boron, which is necessary to inhibit the formation of grain-boundary ferrite must be prevented from oxygen and nitrogen. It is important to strike a proper balance among aluminum, titanium, boron, oxygen, and nitrogen by considering the amounts of these elements introduced from the base metal. The optimum ranges of these elements are schematically shown in **Fig. 6**.

3.2 Increase in toughness by adjustment of other elements

The effect of other elements due to dilution from the base metal has been determined for the titanium-boron weld metal where acicular ferrite is obtained by the optimization of aluminum, titanium, boron, oxygen, and nitrogen[17]. This effect is schematically shown in **Fig. 7**, together with the optimum strength and alloy content for the toughness of the weld metal. Grain-boundary ferrite and upper bainite increase more than acicular ferrite, respectively, when the actual alloy content is smaller and greater than the optimum level. Toughness declines in each

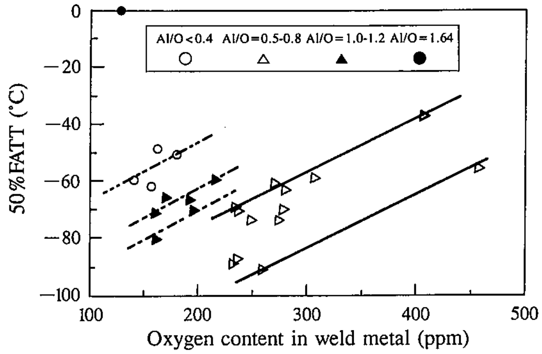


Fig. 8 Effect of oxygen content on toughness of weld metal

case. Copper and nickel are relatively wide in the optimum range, while manganese and molybdenum are narrower in the optimum range. The effect of dilution from the base metal must be carefully controlled. If the Al/O ratio is held constant, the lower the oxygen content, the higher will be the toughness as shown in **Fig. 8**. The oxygen content should be keep as low as possible.

4. Increase in Toughness of Reheated Weld Metal

For welding in multiple passes with a titanium-boron bearing system and high heat input, the heat-affected zone (HAZ) in the weld metal reheated by the welding heat of the successive passes is embrittled like the HAZ in the base metal. Since such brittle zones are a governing factor for the CTOD value of the weld metal, measures should be taken to avoid embrittlement of the reheated zone as well as the as-welded zone described in the previous chapter.

Heating temperature of the reheated zone continuously changes, making it difficult to analyze. Like the base metal, the reheated zone was carefully analyzed using a weld thermal cycle simulator. The results of thermal cycle simulation are given in **Fig. 9**, and microstructures obtained at typical heating temperatures are shown in **Photo 1**. Cause of the formation of the brittle zone varies with the peak heating temperature. In the brittlest grain-refined zone (at about 950°C), ferrite grains are larger than AF grains, and the martensite-austenite constituent (MAC) is present. In the grain-coarsened zone (at or above 1,000°C), GBF is more than in the as-welded zone. In the dual-phase zone (near 750°C), the MAC is formed in the solidified segregation region. Lowering the carbon content reduces the MAC content and improves the toughness of the grain-refined zone, but increases GBF in the grain-coarsened zone and embrittles this zone. There

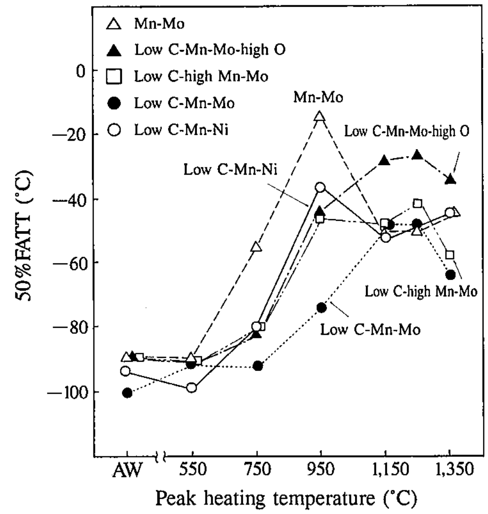


Fig. 9 Toughness of reheated zone in various weld metal compositions

is no other measure than a reduction in oxygen content to improve toughness that can be commonly applied to these zones. In the case of low carbon-nickel type weld metal, retained austenite formed in place of the MAC, is less embrittled than the molybdenum weld metal (see **Fig. 9**), and low carbon-nickel type weld metal is suitable for use when the toughness of the reheated zone is critical.

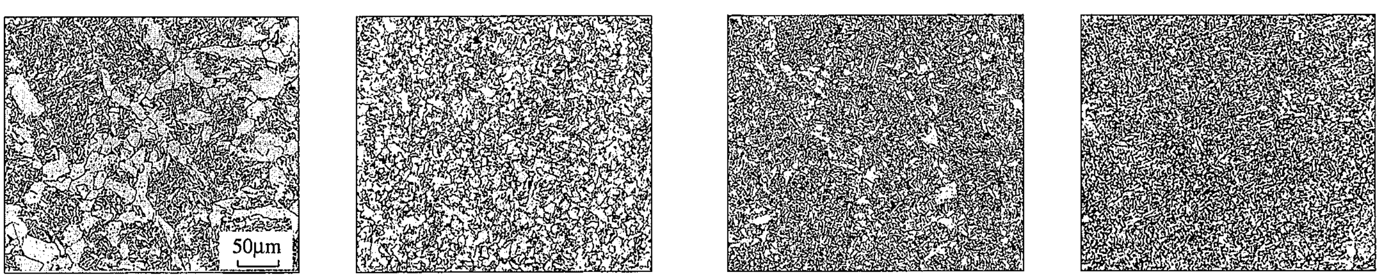
5. Development of Welding Materials for Low-Temperature Service

5.1 Development of low-temperature high-heat input welding materials for specific steels

Submerged arc welding (SAW) materials for steels with tensile strengths of 490 and 590 MPa were developed, based on the findings discussed above. Typical examples are given in **Table 2**. Localized seawater corrosion is reported not to occur when the absolute value of the parameter ΔEC due to the differences in copper and nickel content between the base metal and the weld metal is less than unity[18]. Welding materials marked by an asterisk are designed taking this into account.

5.2 Development of low-hardenability high-toughness weld ing materials for SSC-resistant steel pipe

Sulfide stress cracking (SSC) is likely to occur in pipeline handling large amounts of hydrogen sulfide (H_2S) and sour gas. To prevent SSC, hardness is often specified at Hv ≤ 248. Since girth welds of pipes are made with a low heat input of 1.0 kJ/mm or less, the HAZ of the seam weld metal by girth welding is



(a) Peak heating temperature: 1,350°C (b) Peak heating temperature: 950°C (c) Peak heating temperature: 750°C (d) As welded

Photo 1 Reheating temperatures and microstructural changes

Table 2 Typical mechanical properties of SA weld metals for low-temperature service

Welding method	Steel	Thickness (mm)	Heat input (kJ/mm)	AW PWHT	YP (N/mm²)	TS (N/mm²)	$_vE_{-80°C}$ (J)	$_vE_{-60°C}$ (J)	50%FATT (°C)	Minimum CTOD (mm)	Composition	Flux × Wire
Both-side multipass	TiN	50	4.5	AW	499	565	90/50	159/151	−73	−30°C/2.68	Mn	55L×YD
				PWHT	503	599	43/38	159/143	−74	−30°C/2.78		
	TiN	50	5.0	AW	549	656	122/118	171/165	−85	−50°C/0.46	0.5Ni-0.5Cu	*55L×YDCN
	TiN	32	4.5	AW	558	643	58/53	136/134	−73	−30°C/2.30	Mn-0.18Mo	60L×YDM3
				PWHT	503	620	92/54	165/156	−79	−30°C/2.56		
	TiO	38	4.5	AW	552	662	143/113	184/175	—	−30°C/2.19	Mn-0.18Mo	60L×YDM3
				PWHT	587	663	—	61/39	—	−30°C/2.80		
	TiN	75	10.0	AW	545	662	44/44	176/93	−70	−40°C/0.2	Mn	55E-B×YE
			10.0	AW	—	593	83/44	128/102	—	−40°C/0.33	Mn-2.5Ni	*55E-B×Y8NI
Both-side single-pass	TiB	32	11.0	AW	—	639	59/20	124/59	—	−40°C/0.4	Mn-0.2Mo	55E-A×YDM
			11.0	AW	—	673	93/49	136/69	—	−60°C/0.4	Mn-2.5Ni	*55E-Am×Y8NI
	TiN	38	13.1	AW	—	656	45/20	137/69	—	−30°C/1.17	Mn-0.2Mo	55E-B×YDM
	TiO	32	11.0	AW	537	666	85/63	172/168	−80	—	Mn-0.2Mo	55E-T×YDM
	TiO	32	9.2	AW	515	673	144/98	157/149	−93	−60°C/0.25	Mn-2.7Ni	*55E-Tm×Y6NI
One-side two-pass	TiN	38	14.8	AW	539	666	—	66/61	−56	−50°C/0.37	Mn-1.6Ni	*H52-Bm×Y3NI
	TiB	32	14.4	AW	—	653	—	61/39	—	—	Mn-Ni-Mo	*H52-Am×Y3NT
One-side single-pass	TiB	30	19.0	AW	481	664	45/34	57/39	−55	—	Mn-2.8Ni	*H55-Am×Y6NI
	TiO	30	19.0	AW	481	696	40/40	74/71	−41	—	Mn-1.6Ni	*H55-Tm×Y6NI

*: $-1 \leq \Delta Ec = 3.3 \times \Delta Cu + 1.1 \times \Delta Ni + 0.3 \leq 1$, where ΔCu = Cu (base metal) —Cu (weld metal) and ΔNi = Ni (base metal) —Ni (weld metal)

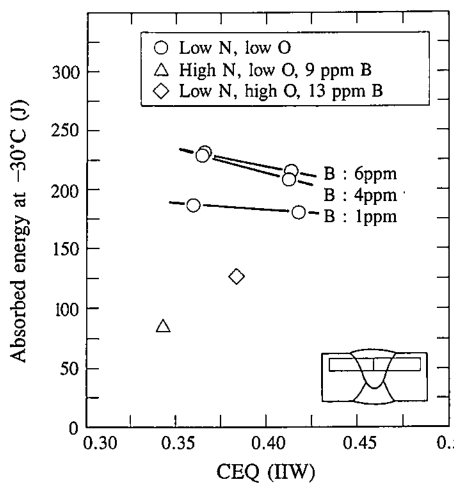


Fig. 10 CEQ and absorbed energy of low-nitrogen low-oxygen weld metal for SSC-resistant line pipe

hardened most. Therefore, the seam weld metals must have high toughness as well as adequate strength at low hardenability. Based on the findings described in the previous chapter, welding materials were developed for SSC-resistant steel pipe.

The basic concept is to add titanium and achieve Al/O ≤ 1 (0.8), in order to produce a microstructure composed largely of acicular ferrite by both-side single-pass welding. As discussed in Chapters 3 and 4, reduction of carbon and oxygen content is an effective means of increasing toughness without increasing hardenability, and the carbon equivalent (CEQ) is made as low as possible to meet the Charpy impact energy requirement of $_vE_{-30°C}$. A high-basicity fused flux was developed that accommodates the high-speed welding of UO pipe and assures low oxygen and nitrogen contents. Using this new flux, weld metals with different CEQ values were prepared[19]. The Charpy test results of the low-oxygen low-nitrogen weld metal as well as high-nitrogen

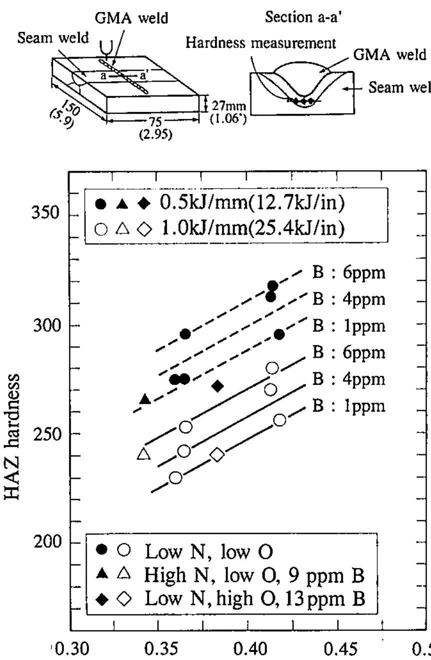
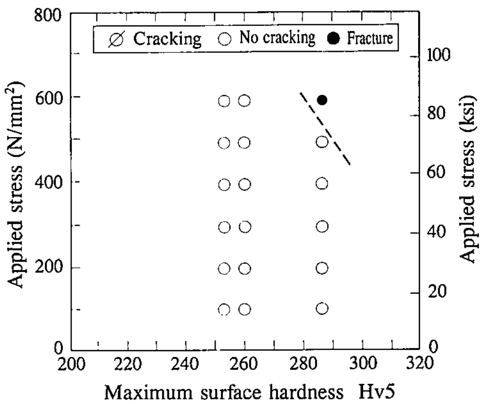
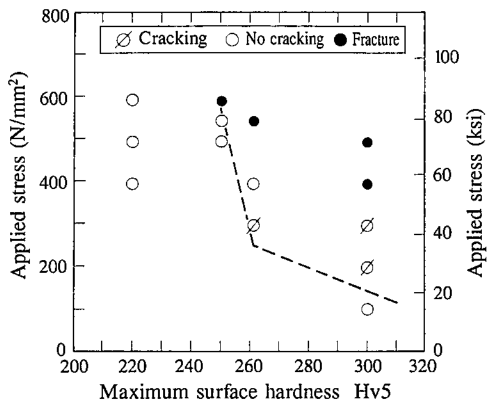


Fig. 11 Hardenability test results of weld metal for SSC-resistant line pipe

weld metal and the high-oxygen weld metal used are given in **Fig. 10**. The low-oxygen low-nitrogen type has higher toughness than the high-nitrogen type and the high-oxygen type, despite its lower boron content. Using the method shown in the upper part of **Fig. 11**, the welds were reheated to the same degree as reached during girth welding. The hardenability of the reheated welds is shown in the lower part of **Fig. 11**. All welds exceeded the HAZ hardness of 250 when the heat input was 0.5 kJ/mm. When the heat input was 1.0 kJ/mm, the region where



(a) New weld metal (low N, low O, 9 ppm B)



(b) Comparative weld metal (low N, high O, 20 ppm B)

Fig. 12 SSC characteristics of weld metals for SSC- resistant line pipe

the hardness requirement of Hv ≤ 248 can be met was near the CEQ value of 0.36%.

Both the high-nitrogen type and the high-oxygen type show almost the same degree of hardness as the low-oxygen low-nitrogen weld metal with the boron content of 1 ppm, except for the high-nitrogen weld metal with the heat input of 1.0 kJ/mm. This is because the boron in the high-nitrogen type and in the high-oxygen type forms boron nitride (BN) and boron oxide (B_2O_3), respectively, and there is no free boron to increase the hardenability. Since boron nitride is unstable, reheating at 1.0 kJ/mm decomposes some of it in the high-nitrogen weld metal and free boron is formed. This free boron is probably responsible for the slight hardening observed in the high-nitrogen weld metal. In other words, more boron should be allowed in the weld metal because it contains more nitrogen and oxygen.

The SSC resistance of the weld metal thus developed was evaluated by a four-point bending test method under the NACE solution. The results are shown in **Fig. 12**. Compared with the high-oxygen weld metal, the newly developed weld metal cracked only when the hardness and applied stress were both high; its very good SSC resistance was thus confirmed. With the establishment of technology to assure the desired toughness of low-alloy weld metal, low-hardenability and high-toughness weld metals can now be designed. The new welding material has already been applied to the production of more than 100,000 tons of SSC-resistant steel pipe.

6. Development of Welding Technology for High-Functionality Steels in Construction Field

The Ministry of Construction of Japan conducted a project to develop technology for a fire-resistant design system for use in buildings to assure their safety. The new system resulted in designs complying with the high-temperature strength of steel and led to the development of fire-resistant construction steels with a guaranteed resistance up to 600°C[9]. Fire-resistant steels have many advantages, including the elimination or reduction of fire-resistant coverings, reduction in construction cost, and effective utilization of space. Welding materials currently in use for heat-resistant steels are of a better quality than needed for fire-resistant steels, and new materials appropriate for the strength and price of fire-resistant steels must be developed.

The percentage of the room-temperature yield point that

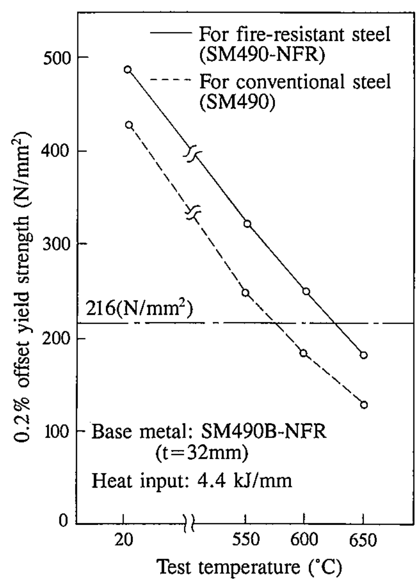


Fig. 13 Temperature dependence of yield strength of SA weld metal deposited by welding materials for fire-resistant steel for conventional steel

remains at 600°C depends on the molybdenum content, irrespective of the welding process employed. The optimum composition of each fire-resistant steel was determined by considering the dilution rate for each welding process. The yield strength of the weld metal with high-temperature properties of the weld metal are given in **Fig. 13**.

High-heat input SAW is applied to box column corner joints. Formation of acicular ferrite by adding titanium and boron is requisite in terms of toughness. When the titanium and boron contents are high, the high-temperature ductility of the weld metal declines as shown in **Fig. 14**. The high boron content weld metal fractured at austenite grain boundaries and short-time creep strength was low at 600°C[20]. This phenomenon was observed in high-heat input welds. Since these welds are large in austenite grain size and likely to increase in grain-boundary boron concentration, only the smallest necessary amounts of titanium and boron were added.

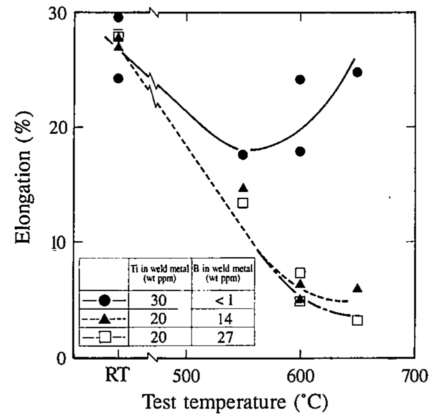


Fig. 14 Effects of titanium and boron on high-temperature ductility of weld metal

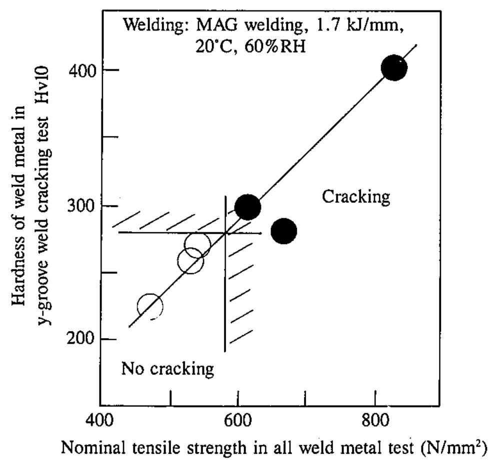


Fig. 15 Relationship among tensile strength, hardness, and cracking of weld metal

6.2 Manufacturing technology for HT780 steel box columns for superhigh-rise buildings

Superhigh-rise buildings must be constructed of box columns with a tensile strength of 780 N/mm^2. The primary problems involved in welding of such box columns are prevention of cold cracking and reduction in preheat temperature. TS780 steel with lower preheat temperature was developed as reported elsewhere[8,21].

The two-electrode AC-MIG technique was established to make the best use of new steel. Use of large-diameter wire increased the welding current, improved welding efficiency, reduced hydrogen content to an extremely low level, and enabled the welding of 100-mm thick plates with a preheat of only 50°C. If a welding material with a nominal strength of 490 N/mm^2 is used for jig installation, cold cracking does not occur. As the heat input is low, strength of weld metal is over 780 N/mm^2, and field weldability does not suffer, as shown in **Fig. 15**.

7. Development of High-Productivity Technology

Preventing oil spillage which is one cause of environmental pollution necessitates the double-hull construction of oil tankers and increases the weld length by about 1.5 times. Increase in welding speed is expected to improve the construction efficiency

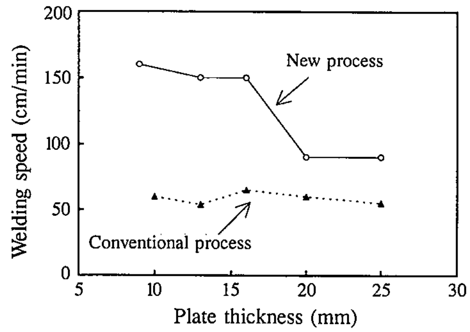


Fig. 16 Welding speed comparison of new one-side welding process with four-electrodes and conventional one-side welding process

of double-hulled ships over that of conventional ships. By one-side single pass SAW with four electrodes (two electrodes for uranami formation and two electrodes for surface bead formation) and modification in flux, the welding speed was increased as shown in **Fig.16**[6]. The heat input is reduced, rotational deformation that causes toe cracks, a problem in one-side welding, is halved, and repair time and labor is reduced.

Fillet welding for a leg length of 5 mm can be performed at a high speed of 1.5 m/min by the combination of a flux-cored arc welding (FCAW) technique and two-electrode fillet welding equipment[22,23].

8. Conclusions

Greater strength and higher low-temperature toughness are demanded for welds in new steel structures fabricated in the energy and building construction fields. In addition welding efficiency calls for higher heat input, lower preheat temperature, and higher speed. Based on the principles of oxide metallurgy, there is now technology to assure the desired toughness of the weld metal, and it is possible to design weld metals that resist corrosion and SSC and have other properties and which are suitable for specific steels.

A new flux copper backing process (FCuB) and fillet welding process have more than doubled welding speed. Although not described here, a new steel with good electron beam weldability has also been developed, making it possible to weld heavy-gage plates at higher speed.

Technology making it possible to impart higher strength and toughness to a weld metal and enhancing welding productivity has made great progress and will continue to do so.

References

1) Murata, S. et al.: Journal of the Society of Naval Architects of Japan. 156, 489 (1984)
2) Imura, Y. et al.: Journal of the Society of Naval Architects of Japan. 157, 359 (1985)
3) Nakasugi, H. et al.: WELDING-90. July 1990, Hamburg, Germany
4) Lingnau, D.G. et al.: Corrosion. Paper 395, 1986
5) Takahashi, A. et al.: CAMP-ISIJ. 6 (3), 644 (1994)
6) Miyazaki, T. et al.: Hitachi Zosen Giho. 54 (1), 71 (1993)
7) Ohashi, M. et al.: CAMP-ISIJ. 6 (2), 406 (1994)
8) Yamashita, T. et al.: Tekkotsu Gijutsu. 7 (8), 63 (1994)
9) Chijiiwa, R. et al.: Shinnittetsu Giho. (348), 55 (1993)
10) Mori, N. et al.: Journal of the Japan Welding Society. 50 (8), 74 (1981)
11) Ohno, T. et al.: Tetsu-to-Haganè. 73, 94 (1987)

12) Kanazawa, S. et al.: Tetsu-to-Hagané. 61, 2589 (1975)
13) Ohkita, S. et al.: Seitetsu Kenkyu. (327), 9 (1987)
14) Bhatti, A.R. et al.: Welding Journal. (7), 224S (1984)
15) Jang, J. et al.: Material Science. 22, 689 (1987)
16) Okumura, M. et al.: Quarterly Journal of the Japan Welding Society. 6, 144 (1988)
17) Horii, Y. et al.: Seitetsu Kenkyu. (327), 3 (1987)
18) Koseki, T. et al.: CAMP-ISIJ. 4, 809 (1991)
19) Haze, T. et al.: Seitetsu Kenkyu. (326), 36 (1987)
20) Ichikawa, K. et al.: Pre-Prints of the National Meeting of JWS, 1993, p. 10
21) Watabe, Y. et al.: Shinnitettsu Giho. (348), 17 (1993)
22) Kamada, M. et al.: Welding Journal. 72 (3), 49 (1993)
23) Maki, N.: Nittetsu Bead. 100, 17 (1992)