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Evaluation of Cold Crack Susceptibility in Weld Metal of High Strength Steels Using LB-TRC Test†

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Abstract

The lower critical stress of weld metal of 60, 70, 80, 90 and 100 kg/mm² steels in strength levels was measured by the LB TRC test, previously reported, with SMA and GTA argon and argon-hydrogen mixed shielding gases welding. Moreover the susceptibility of cold cracking was evaluated by "Embrittlement Index", I, etc.

Consequently, it was seen that the lower critical stress of weld metal decreases rapidly with an increase in diffusible hydrogen content in the range 0 to 3 ml/100 g and the weld metal of 100 kg/mm² class steel in strength level was most susceptible to cold cracking.

KEY WORDS: (Cold Cracking) (Weld Metal) (High Strength) (TIG Welding) (Weldability Test)

1. Introduction

It has been reported[1] by the authors that the weld metal cracking was easy to occur in HY-130 type steel and generally it remains problem that a transverse cold cracking in weld metal occurred in multipassed weld metal of HT80 in 80 kg/mm² strength level and Cr-Mo steel. Accordingly in the previous report[2], for the purpose of investigation on susceptibility for weld metal cold cracking of high strength steel, a new test which was named the LB-TRC (Longitudinal Bead-Tensile Restraint Cracking) test was developed by the authors. Consequently it was concluded that the LB-TRC test was available to evaluate the susceptibility of weld metal cold cracking.

In this work, the lower critical stress of various high strength weld metals, which were from 50 to 100 kg/mm² in strength levels, was studied with the LB-TRC test with shielded metal-arc (SMA) and gas tungsten-arc (GTA) welding, in which the diffusible hydroben content was changed by changing hydrogen gas content in Ar shielding gas.

Consequently, the lower critical stress of various high strength weld metal was quantitatively obtained and the cold crack susceptibility of weld metal was evaluated using "Embrittlement Index", I and the ratio of the lower critical stress to the yield stress of deposited metal.

2. Materials Used and Experimental Procedures

2.1 *Materials used*

Base metals used were HT60, HT70, HT80, HT90 and HT100, which were weldable heat-treated high strength steels. The figures of these designations indicate the nominal ultimate tensile strength in kg/mm². Among these, HT70, HT90 and HT100 were HY-type high-toughness steels corresponding to HY-90, HY-110 and HY-130, respectively. Designations, type of chemical compositions and nominal ultimate tensile strength of materials used were shown in **Table 1.** The yield stress of deposited metal was measured using deposited metal test specimen with gas metal-arc (GMA) argon-oxygen mixed shielding gas. Chemical compositions and mechanical properties of these materials were described in detail elsewhere.[1-2] These base metals were welded by SMA welding using covered electrodes (4 mm dia.) of low-hydrogen type and by GTA welding using filler wire (1.6 mm dia.) which well matched base metals in strength levels. In Table 1, W60 and W80 are commercial covered electrodes and the others are tentative ones, and F50, F60, F70 and F80 are commercial filler wires and the others are tentative ones.

2.2 *The LB-TRC test*

The LB-TRC test has been developed to evaluate the susceptibility of weld metal cold cracking. Testing method has been described in detail elsewhere.[2] Briefly describing, it essentially consists of two plates butted together to provide a slit across which a test bead was laid as shown in **Fig. 1**, and then a constant tesile load was applied parrallel to the weld line after welding.

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Table 1 Designation and type of chemical compositions of materials, and strength of deposited metals used

Base metal	Covered Electrode	Filler wire	Yield stress of deposited metal (kg/mm^2)	Nominal ultimate stress of deposited metal (kg/mm^2)
HT60(0.2Mo-V) [35mm thickness]	—	F50 [1.6mm dia.]	42	50
	W60(0.6Ni-Mo) [4mm dia.]	F60(0.3Mo) [1.6mm dia.]	56	60
HT70(2.9Ni-Cr-Mo-V) (HY-90 type) [35mm thickness]	W70(2.7Ni-Mo) [4mm dia.]	F70(2.0Ni-Mo) [1.6mm dia.]	67	70
HT80(1.1Ni-Cr-Mo-V-B) [25mm thickness]	W80(1.7Ni-Cr-Mo) [4mm dia.]	F80(2.6Ni-Mo) [1.6mm dia.]	76	80
HT90(3.9Ni-Cr-Mo-V) (HY-110 type) [35mm thickness]	W90(2.7Ni-Cr-Mo) [4mm dia.]	F90(2.4Ni-Cr-Mo) [1.6mm dia.]	86	90
HT100(5.2Ni-Cr-Mo-V) (HY-130 type) [35mm thickness]	W100(2.6Ni-Cr-Mo) [4mm dia.]	F100(2.5Ni-Cr-Mo) [1.6mm dia.]	93	100

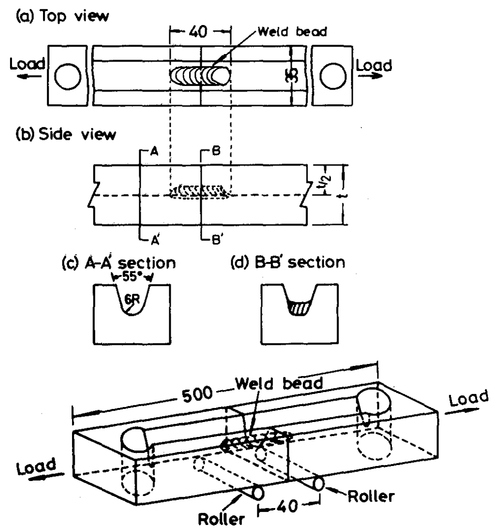


Fig. 1 LB-TRC test specimen and testing method

In this study, test welds were deposited with SMA welding utilizing 170A, 25V and 150 mm/min, which corresponded to a heat input of 17 kJ/cm, and GTA welding with argon-hydrogen ($Ar+H_2$) mixed shielding gas and constant wire feed system utilizing 300A, 14V and 120 mm/min, which corresponded to a heat input of 21 kJ/cm. Hydrogen gas content was changed 0, 0.35, 1.0, 2.0 and 3.2% in volume. Total gas flow rate was 20 l/min. These mixing gases were obtained with a gas mixing chamber. Cooling time from 800 to 500°C with GTA and SMA welding without preheating condition was about 3 and 4 sec, respectively. The preheating temperature was selected as shown in **Table 2**. The method of preheating was that the specimen had been heated for 5 to 10 min in an electric furnace at about 350°C and then was welded after the temperature on the butted surface of the specimen cooling down to the set preheating temperature. Cooling time from 800 to 500°C with GTA and

Table 2 Preheating temperature in LB-TRC test

Base metal	Filler wire, Electrode	Welding process	Preheating temperature(°C)		
HT60	F50	GTA	R.T*	75	—
HT60	F60	GTA	R.T	75	—
	W60	SMA	R.T	—	—
HT70	F70	GTA	R.T	75	125
	W70	SMA	R.T	75	—
HT80	F80	GTA	R.T	75	125
	W80	SMA	R.T	75	—
HT90	F90	GTA	R.T	75	125
	W90	SMA	R.T	75	—
HT100	F100	GTA	R.T	75	125
	W100	SMA	R.T	75	125

R.T*: Room temperature

SMA welding with 125°C preheating condition was about 5 and 7 sec, respectively at 35 mm thickness.

In order to obtain the fracture stress which could be regarded as the fracture stress in hydrogen free condition, the welded specimen had been released for 10 days after welding and then was stretched to fracture with a cross-head speed of 1.2 mm/min. Therefore this fracture stress is regarded as the lower critical stress without hydrogen. Besides, it is seemed that this fracture stress indicates the strength and toughness of materials, because the LB-TRC test is regarded as the notch tensile test.

The content of diffusible hydrogen was measured by modified JIS method in which mercury at 45°C was used as a confining liquid instead of glycerine because of low level in diffusible hydrogen. The diffusible hydrogen content was represented in ml/100g deposited metal.

3. Results and Discussions

3-1 *Effect of diffusible hydrogen content, preheating temperature and yield stress of deposited metal on lower critical stress*

Relationship between hydrogen gas content in Ar shielding gas and diffusible hydrogen content ($[H_D]$) in the weld deposited metal is shown in **Fig. 2**. An increase in hydrogen gas content in Ar shielding gas increases $[H_D]$ in all materials used and $[H_D]$ nearly depends on only hydrogen gas content in Ar shielding gas irrespective of materials.

On the other hand, **Table 3** shows $[H_D]$ in SMA welding. Comparing $[H_D]$ in Fig. 2 with in Table 3, it is considered that $[H_D]$ in GTA welding with Ar+0.35%H_2 shielding gas is nearly equal to those in SMA welding.

Figure 3 (a) and (b) shows the effects of hydrogen gas content in Ar shielding gas and preheating temperature on the lower critical stress (σ_{cr}) of weld metal of HT70 and HT100. In both steels, an increase in hydrogen gas content in Ar shielding gas decreases σ_{cr}. Then, **Fig. 4** which was obtained by substituting $[H_D]$ for hydrogen gas content in Ar shielding gas in abscissa of Fig. 3 shows the effect of $[H_D]$ and preheating temperature on σ_{cr}. For the comparison, σ_{cr} in SMA welding is also plotted. In Figs. 3 and 4, the point with an asterisk on the ordinate shows the fracture stress (σ_F^*) at 10 days after welding described in 2.2. In Fig. 4 (a), σ_{cr} in HT70 decreases precipitously with $[H_D]$ in the range 0 to 3 ml/100g and then gently in more than 3 ml/100g in all preheating temperature. Besides, σ_{cr} increases with an increase in the preheating temperature and exceeds the yield stress of deposited metal ($(\sigma_{depo})_y$) under 125°C preheating condition.

On the other hand, σ_{cr} in HT100 also decreases with an increase in $[H_D]$ similar to HT70, however it doesn't exceed $(\sigma_{depo})_y$ in more than about 4 ml/100g even with 125°C preheating condition. By the way, it is noticed that σ_{cr} in HT70 with SMA welding is approxi-

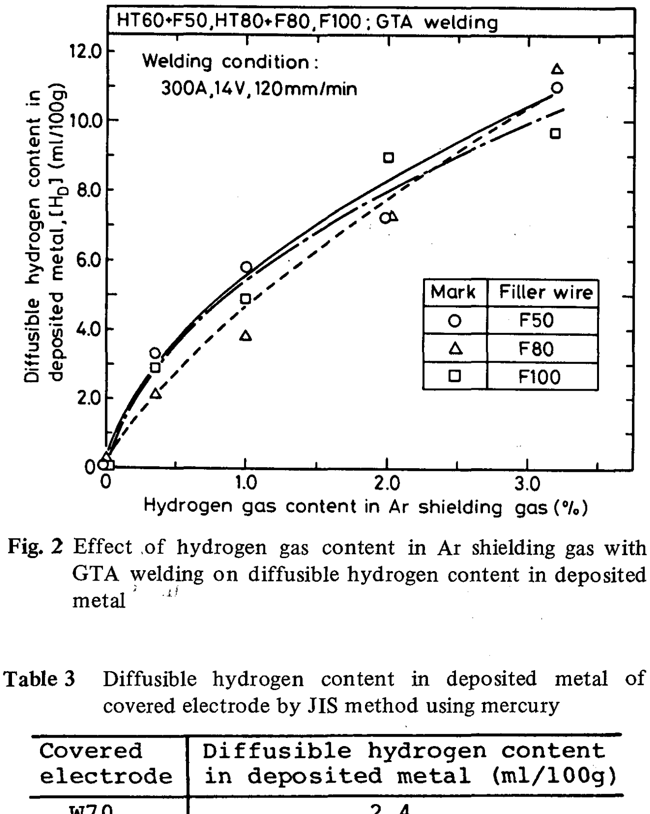
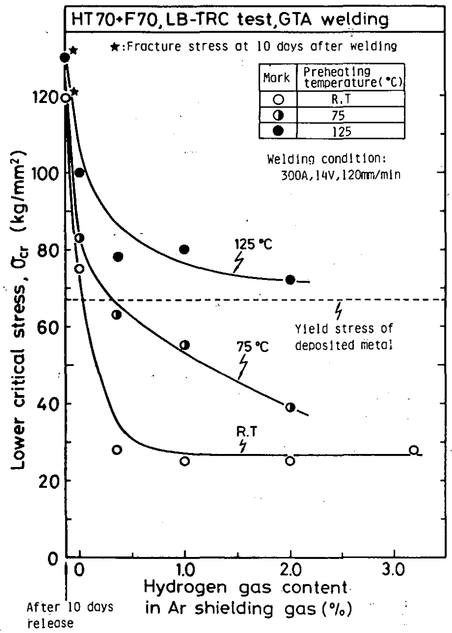


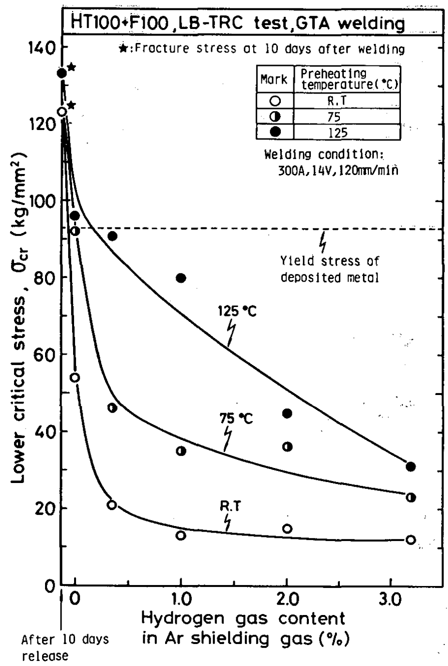
Fig. 2 Effect of hydrogen gas content in Ar shielding gas with GTA welding on diffusible hydrogen content in deposited metal

Table 3 Diffusible hydrogen content in deposited metal of covered electrode by JIS method using mercury

Covered electrode	Diffusible hydrogen content in deposited metal (ml/100g)
W70	2.4
W80	3.5
W90	1.0
W100	1.8

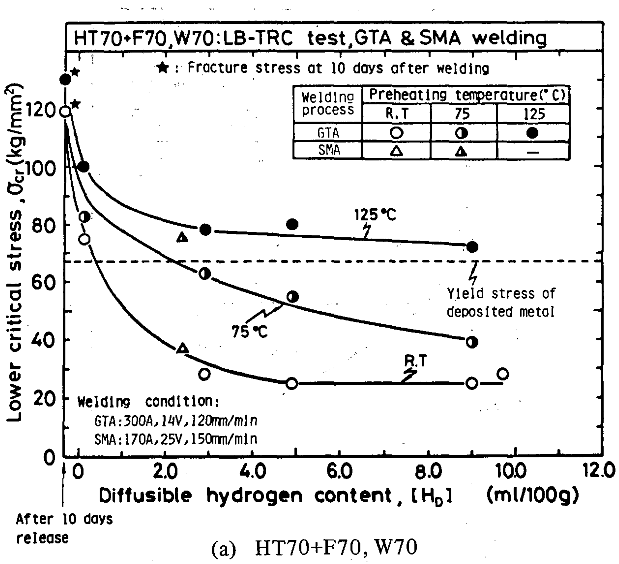


(a) HT70+F70

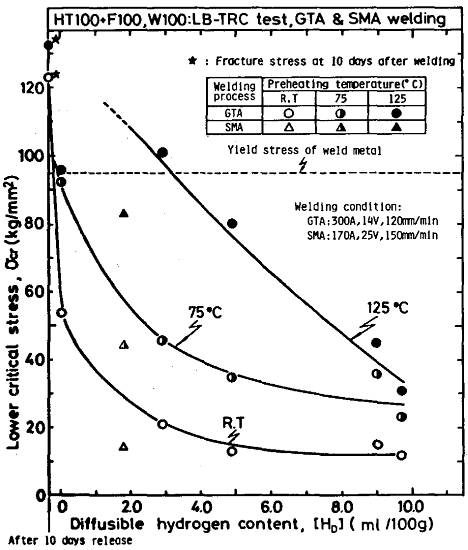


(b) HT100+F100

Fig. 3 Relationship between hydrogen gas content in Ar shielding gas and lower critical stress of weld metal



(a) HT70+F70, W70



(b) HT100+F100, W100

Fig. 4 Effect of diffusible hydrogen content in deposited metal on lower critical stress of weld metal

mately equal to that with GTA welding under the same $[H_D]$ condition, but σ_{cr} in HT100 with SMA welding is lower than that with GTA welding.

Judging from the above, the weld metal of HT100 is more susceptible to cold cracking than that in HT70. **Figure 5** shows the relation between $(\sigma_{depo})_y$ and σ_{cr}, σ_F^* in the LB-TRC test. Essentially in the abscissa of Fig. 5, the yield stress of weld metal in the LB-TRC test specimen has to be used instead of $(\sigma_{depo})_y$, however in this work conveniently $(\sigma_{depo})_y$ is substituted for the yield stress of weld metal. In Fig. 5, σ_{cr} with Ar shielding gas without preheating is gradually decreases with an increase in $(\sigma_{depo})_y$ except for the range 70 to 80 kg/mm^2 and is lowered about 50 kg/mm^2 in HT100 class steel. The same tendency are observed in Ar+0.35%H_2 and Ar+2.0%H_2 shielding gas, whereas both σ_{cr} in these

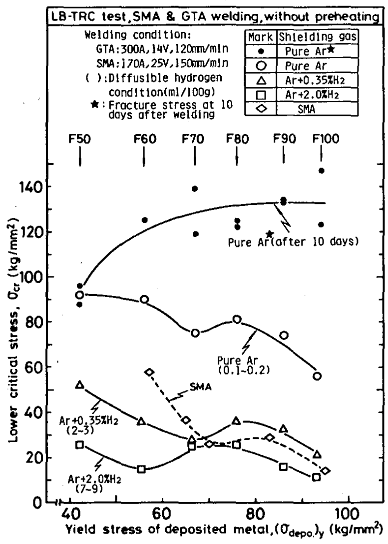


Fig. 5 Effect of yield stress of deposited metal on lower critical stress of weld metal

shielding gases are decreases more gentle than σ_{cr} in Ar shielding gas with an increase in $(\sigma_{depo})_y$. Now it can't be understood that σ_{cr} is increased with an increase in $(\sigma_{depo})_y$ in the range 70 to 80 kg/mm^2, however this is a interesting result, so it will be investigated in the future. Comparing σ_{cr} in GTA weld with Ar+0.35%H_2 shielding gas with σ_{cr} in SMA weld, the latter is higher within $(\sigma_{depo})_y$ up to about 70 kg/mm^2 and then is lower in more than 70 kg/mm^2 than the former.

3.2 Evaluation of susceptibility to cold cracking of weld metal in high strength steel

Now, the authors have defined the critical diffusible hydrogen content ($[H_D]_{cr}$), which is maximum $[H_D]$ required to prevent cold cracking. Assuming that the restraint stress can't exceed the yield stress of deposited metal, $[H_D]_{cr}$ can be easily obtained from the intersection point of the experimental curve and the broken line of $(\sigma_{depo})_y$ respectively in Fig. 4. **Figure 6** summarized the relation between $(\sigma_{depo})_y$ and $[H_D]_{cr}$. $[H_D]_{cr}$ decreases with an increase in $(\sigma_{depo})_y$ and increases with an increase in preheating temperature in the same stress level. This result also means that the welding of HT100 is difficult under 75°C preheating condition, because $[H_D]_{cr}$ is practically zero. **Figure 7** shows the relation between $(\sigma_{depo})_y$ and $\sigma_{cr}/(\sigma_{depo})_y$. Fracture stress at 10 days after welding (σ_F^*) of each material was larger than its $(\sigma_{depo})_y$ and moreover, was approximately constant in the range of $(\sigma_{depo})_y$ more than 70 kg/mm^2 as shown in Fig. 5. In Fig. 7, $\sigma_F^*/(\sigma_{depo})_y$ decreases linearly with an increase in $(\sigma_{depo})_y$ and $\sigma_{cr}/(\sigma_{depo})_y$ in various shielding gases decreases rapidly with an increase in $(\sigma_{depo})_y$ in the range 40 to 70 kg/mm^2. According

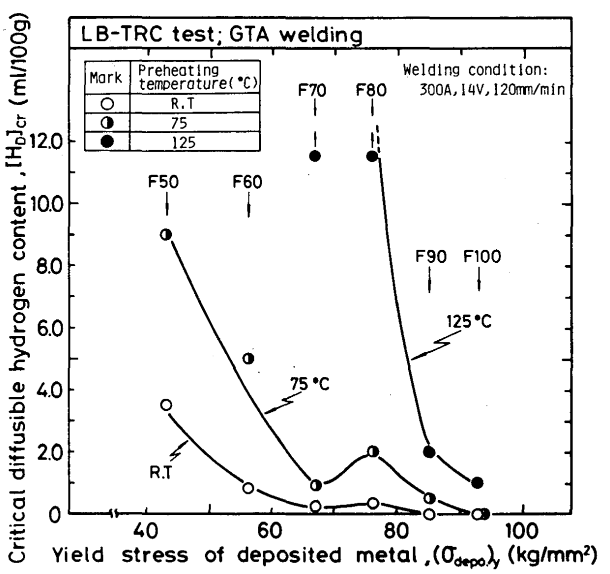


Fig. 6 Relationship between yield stress of deposited metal and critical diffusible hydrogen content

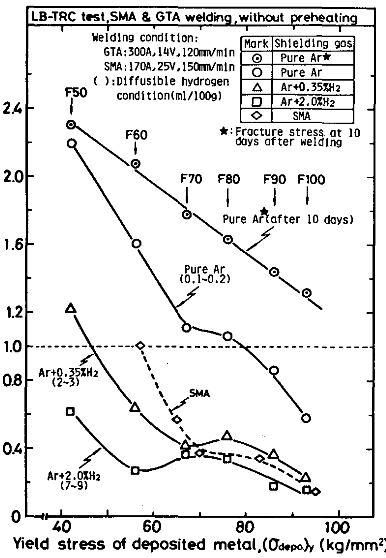


Fig. 7 Relationship between yield stress of deposited metal and ratio of lower critical stress of weld metal to yield stress of deposited metal

to the above assumption, the ratio of unity is regarded as the threshold. Consequently from this figure, $(\sigma_{depo})_y$ which is allowable maximum stress to exceed the yield stress of deposited metal is estimated in various shielding gases.

On the other hand, Boniszewski[3)], Kikuta[4)] and Savage[5)] et al evaluated the hydrogen embrittlement of steels using "Embrittlement Index", I which is often calculated from the relationship I = (NTS − LCS)/NTS where NTS = notch tensile strength and LCS = lower critical stress.

In this study, as the above mentioned, σ_F^* is regarded as the notch tensile strength. **Figure 8** shows the relation

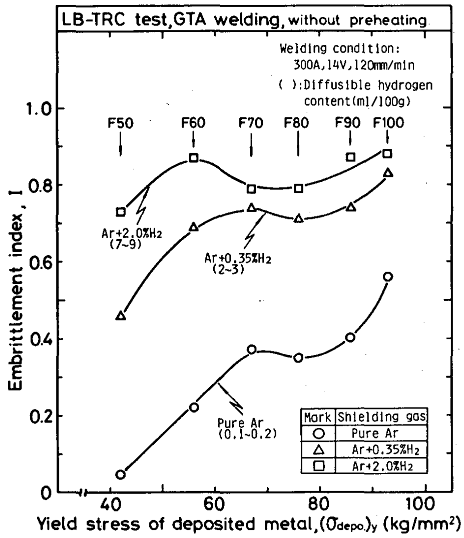


Fig. 8 Effect of yield stress of deposited metal on susceptibility to hydrogen embrittlement of weld metal

between $(\sigma_{depo})_y$ and the index I, and **Fig. 9** shows the relation between $[H_D]$ and the index I in F50, F70 and F100 of filler wire. From these figure, it is noted that the index I increases with an increase in $(\sigma_{depo})_y$ and is highest in HT100 class steel in all shielding gases. This tendency is increased with a decrease in hydrogen gas content in Ar shielding gas, that is to say, in diffusible hydrogen content in the deposited metal. Therefore, in another word, it is considered the more the yield stress of deposited metal increases, the higher the susceptibility to hydrogen embrittlement is in the lower hydrogen level.

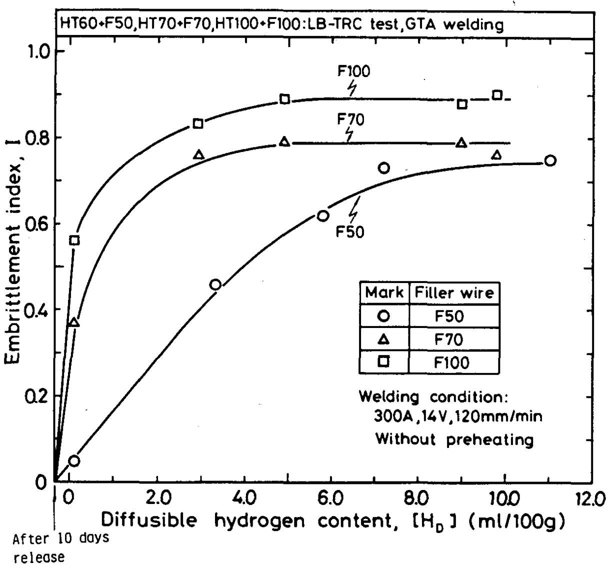


Fig. 9 Effect of diffusible hydrogen content in deposited metal on susceptibility to hydrogen embrittlement of weld metal.

4. Conclusions

The lower critical stresses of weld metal of various high strength steels in the range of HT60 to HT100 were obtained by the LB-TRC test in SMA and GTA welding where hydrogen gas contents in Ar shielding gas were changed.

Main conclusions are summarized as follows:

(1) The lower critical stress (σ_{cr}) of weld metal of all materials used decreases rapidly with an increase in diffusible hydrogen content ($[H_D]$) in the range 0 to 3 ml/100g and gently in more than 3 ml/100g. Besides σ_{cr} increases with an increase in preheating temperature.

(2) Cracking susceptibility of all weld metals were discussed using the critical hydrogen content ($[H_D]_{cr}$) to prevent the cold cracking, the ratio of σ_{cr} to the yield stress of deposited metals ($(\sigma_{depo})_y$), and that of σ_{cr} to the fracture stress at 10 days after welding (σ_F^*).

(3) The weld metal of HT100 class steel is most susceptible to cold cracking of all weld metal investigated. In the welding of HT100 class steel, hydrogen level in deposited metal should be kept as low as possible and is practically zero at 75°C preheating and about less than 1.0 ml/100g at 125°C preheating in order to maintain the lower critical stress in excess of the yield stress of the deposited metal.

(4) Moreover, it is considered that the more the yield stress of deposited metal increases, the higher the susceptibility to hydrogen embrittlement is in the lower hydrogen level.

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