

INCREASING THE FRACTURE TOUGHNESS OF 18 NI-MARAGING
STEEL WELD METAL

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INTRODUCTION

At ICF3 the reasons for the low fracture toughness of TIG-weld metal in maraging steels were reported [1]. Microsegregation occurring between the dendrites in the weld led to local enrichment of alloying elements Ni, Mo and Ti, so that at these spots the re-austenitizing temperature is lowered. During the post-weld heat treatment necessary to produce precipitates - usually performed for 3 hr at 480° C - at these zones with higher contents of the above mentioned alloying elements, austenite spots with a micro-hardness of about 100 HV 0.005 form, whilst the adjacent matrix, formed from nickel-martensite containing precipitates from the types Fe Ti, Ni₃Mo and Fe₂Mo [2], shows hardness values above 530 HV 0.005. Under applied load, the austenite will first deform plastically, then initiate cracks, whilst the martensite is deformed only elastically. The micro-cracks then may propagate into the matrix and cause the low fracture toughness of the welded joint. In the meantime similar observations have been made on welded joints of maraging steel, welded with different techniques [3]. A predominantly large decrease of fracture toughness occurs when welding with plasma beam, Figure 1. Within this type of weld metal, especially large austenitic sites are formed while the EB-welding process promotes relatively small pools, Figure 2. In order to improve the fracture toughness in the weld metal of these types of steels, the formation of austenite has to be prevented or restricted. Two approaches are possible:

- a) eliminating the micro-segregations by high-temperature heat treatment, above annealing temperature;
- b) suppressing the formation of local austenite by selecting a lower aging temperature, or shorter times during the precipitation heat treatment.

The results of these procedures applied to TIG- and EB-weldments of a maraging steel containing 0.006% C, 18% Ni, 8% Co, 5% Mo and 0.42% Ti will be reported.

TEST PROGRAMME

For the preparation of suitable test specimens, 12 mm thick plates of this type of steel in the annealed condition (1 hr at 820° C/AC) were TIG- and EB-welded. From these plates, three-point bend specimens corresponding to ASTM-Standards, as well as flat tensile specimens were machined. One series of specimens were then heat treated for 1 hr in the temperature range from 820 to 950° C, air cooled and finally aged for 3 hr at 480° C. This type of specimen was marked with the letter "H" and the corresponding annealing temperature in ° C. For the other series of specimens only a

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post-weld aging at 480° C was applied, the aging time being varied from 30 to 180 min. These specimens were marked with the letter "A" and the corresponding aging time in minutes. All testing of specimens was performed at room temperature under standardized conditions.

RESULTS OF THE METALLOGRAPHIC EXAMINATIONS

The cross-sections of TIG- and EB-weldments are compared in Figure 3. The TIG-weldment, made from 24 weld passes, shows recrystallized as well as dendritic structure within the weld metal, surrounded by a relatively broad heat affected zone. The EB-weld fusion on the other hand is restricted to about 0.8 mm width, and solidifies completely to fine dendrites. On both sides of the fusion zone, about 1.0 mm wide heat affected zones are to be found. A 1 hr annealing at temperatures of 820, 850 and 900° C yields within the TIG-weld metal a globular structure, where indeed the amount of austenite has diminished, but were especially the long-formed austenite particles are preserved. Within the EB-weld metal, the dendritic structure also has been eliminated completely and a globular structure has been formed after heat treatment at 820, 850, 900 and 950° C. At the highest annealing temperature a considerable grain-coarsening was observable, thus leading to the conclusion that additional precipitation effects take place at these temperatures. Within the EB-weld metal austenite particles are still detectable, which in comparison with those in the TIG-weld metal, have smaller and more globular dimensions, Figure 4.

At light-optical magnification the short-time aging at 480° C produces nearly no visible differences within the structure of the weld metal. The formation of precipitates during these aging treatments is restricted to the sub-microscopical range [2, 3, 4]. Within the TIG- as well as in the EB-weld metal, different etching of the matrix and the interdendritic parts was observed, but this different etching behaviour may be explained by variations of the chemical composition. There is no evidence as to whether austenite has already reformed or not. There is only an indication from the appearance of the structure, that after aging for three hours the lighter etched particles are distinctly bounded by the matrix, whereas after one hour aging treatment this effect is not apparent, Figure 5.

HARDNESS MEASUREMENTS

A comparison of micro-hardness measurements (HVO.3) within base metal TIG- and EB-weld metal yielded indications to the metallographic condition. Already after two hours of aging the base metal reached its final hardness. Shorter aging times did not considerably affect the base metal hardness: after one hour, 88%, and after 30 minutes already 81% of the final hardness is attained, Figure 6. The annealing heat treatment did not influence the base metal hardness.

Concerning the short time aging, within the TIG- and EB-weld metal similar trends comparable to the base metal were observed. In all cases the hardness values of the weld metal were more or less lower than those of the base metal, Figure 6. From this, the conclusion may be drawn, that the precipitation process within the rolled structure of the base metal occurs faster than within the cast metal structure of the weldments. After three hours of aging, the EB-weld metal reached the final hardness of the base metal, whereas the hardness of the TIG-weld metal remained about 80 HVO.3

below the base metal value. The reason for this may be the lower Ti-content of the TIG-weldment - only 0.37% Ti - so that even after annealing at high temperatures, the TIG-weld metal did not reach the hardness of the base metal.

RESULTS OF TENSION TESTS

Stress-strain curves for base metal, TIG- and EB-weld metal have been determined separately by means of strain gauges with 1.5 and 0.6 mm gauges length. Tensile characteristics are shown in Figure 7 for specimens, which have been heat treated in the usual manner for 3 hr at 480° C. Using these curves, the corresponding values for the yield point (0.2% offset) and the UTS could be determined, and these results are summarized in Figure 8 for the case of short-time aging.

The results of the tension tests extensively confirmed the conclusions drawn already from the hardness measurements: when an annealing heat treatment is performed, UTS and yield point of base metal and EB-weld metal will not further increase. All EB-weldments of this type ruptured within the base metal. Annealed TIG-weldments on the other hand, failed within the weld metal, the UTS of which was merely equivalent to the yield point of the base metal. The yield point of the TIG-weld metal decreased to 90 to 95% of the base metal value.

With a short aging time, the yield point of the base metal decreases only slightly (by about 12% after a heat treatment of 30 minutes in comparison to a heat treatment of 3 hr). EB-weldments, which after 3 hr aging still ruptured in the base metal, failed, after shorter aging times, within the weld metal. The decrease of the yield point of the EB-weld metal after 30 minutes aging time reaches 20% of the base metal after 3 hr aging. TIG-weldments attained a similar value after a corresponding aging time although all these weldments always ruptured within the weld metal, since the yield point of this type, even after three hours of aging, still remained 14% below the base metal value.

FRACTURE TOUGHNESS BEHAVIOUR

The test procedure for evaluating K_{IC} values was in agreement with the ASTM Standards. The fatigue crack was always initiated after full heat treatment. The specimen-thickness of 10 to 12 mm was sufficient for the determination of valid K_{IC} values.

In comparison to a K_{IC} value of 3190 N/mm^{3/2} (101 MPa·m^{1/2}) for the base metal, after aging for 3 hr (condition A 180), the fracture toughness of the TIG- as well as of the EB-weld metal decreased by about 40%. An additional annealing heat treatment in the temperature range from 820 to 900° C produced no increase in fracture toughness for the TIG-weld metal. For the EB-weld metal also no marked increase of K_{IC} value occurred as a result of annealing at temperatures from 850 to 950° C. Only the heat treatment at 820° C caused a distinct improvement.

On the contrary, as a result of short time aging, a considerable improvement of weld fracture toughness was attained for TIG- and EB-weldments. Most effect was to be observed after 30 minutes aging time. These specimens now attained nearly the same K_{IC} values as measured for the base metal after aging for three hours, Figure 9.

CONCLUSIONS

On comparing the results of metallographic examinations with the fracture toughness values of the annealed weldments, the insignificance of the total elimination of the dendritic structure on the fracture toughness of the weld metal becomes apparent. Although this expensive additional heat treatment succeeds in diminishing part of the micro-segregations, no worthwhile increase in fracture toughness could be achieved, except in the case of EB-weldments annealed at 820° C. Annealing temperatures about 820° C even cause a deterioration of the material toughness. In this case the explanation given by Kalish and Rack [5] seems to be most likely, that within this temperature range titanium carbides as well as titanium carbonitrides are precipitated along the grain boundaries.

A definite possibility to improve the toughness of the weld metal, is given by short time aging at 480° C. It seems to be remarkable that the considerable increase in fracture toughness is not connected with gross losses in yield strength, UTS or hardness. For practical applications it is necessary to decide from case to case, whether this small loss in strength is balanced by the considerable increase in fracture toughness.

REFERENCES

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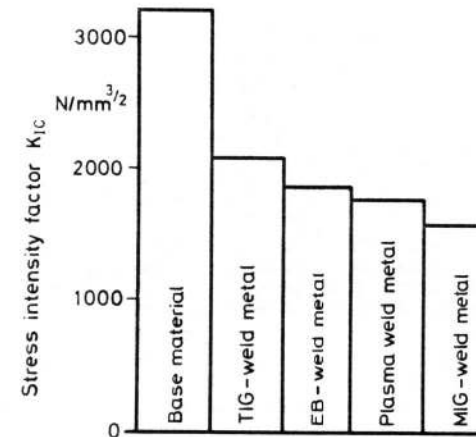


Figure 1 Fracture Toughness K_{IC} of Base Metal and Different Weld Metal of 18 Ni-Maraging Steel. (1000 N/mm^{3/2} = 31.6 MPa·m^{1/2})

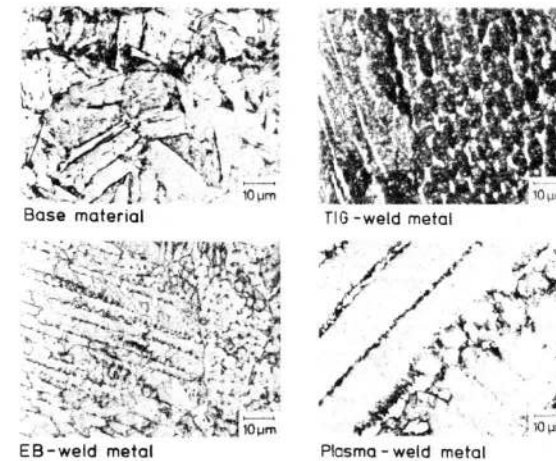


Figure 2 Micro-Structure of Base Metal and Different Weld Metal of 18 Ni-Maraging Steel, Etchant: 10 g FeCl₃ + 1 ml HCL + 186 ml Methanol

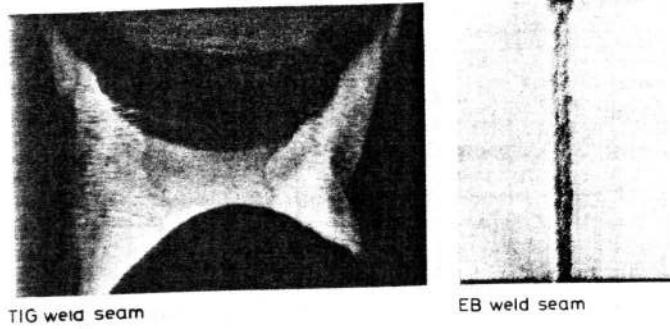


Figure 3 Cross-Sections of TIG- and EB-Weldments in 12 mm Thick 18 Ni-Maaging Steel; Etchng as in Figure 2

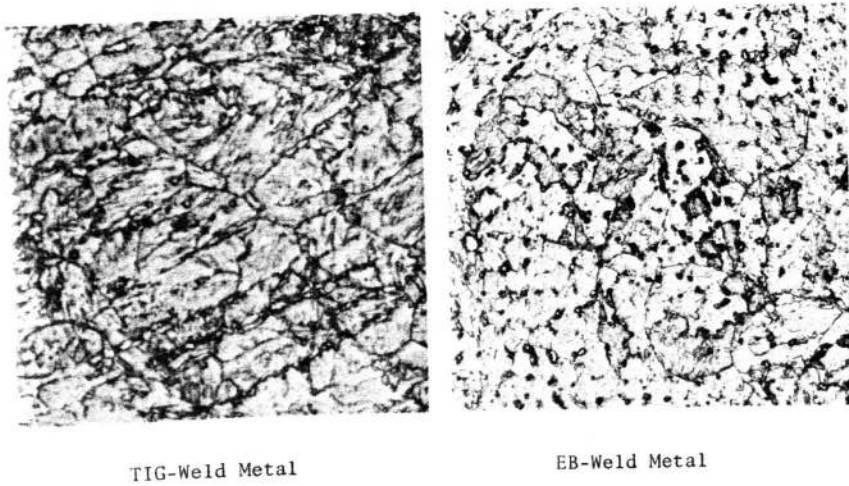


Figure 4 TIG- and EB-Weld Metal Microstructure After Annealing 1 hr at 850° C/ Air and Additional Aging 3 hr at 480° C; Etchant as in Figure 2

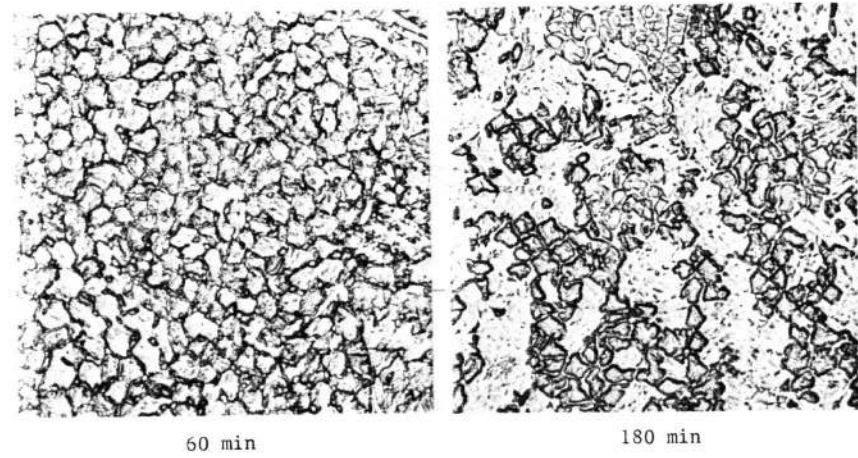


Figure 5 EB-Weld Metal Microstructure after Aging at 480°C for Different Aging Times; Etchant as in Figure 2

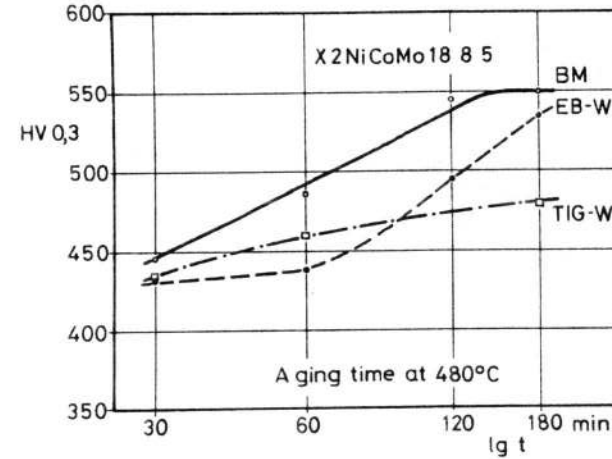


Figure 6 Micro-Hardness of Base Metal, TIG- and EB-Weld Metal after Aging at 480° C for Different Aging Times

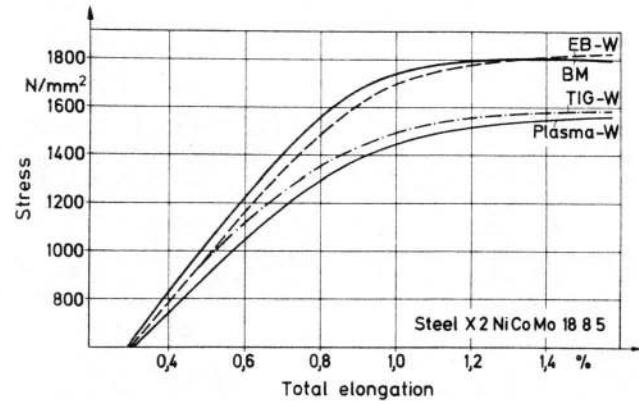


Figure 7 Stress Strain Curves of Base Metal and Different Weld Metal (Aged 3 hr at 480° C) (1 N/mm² = 1 MPa)

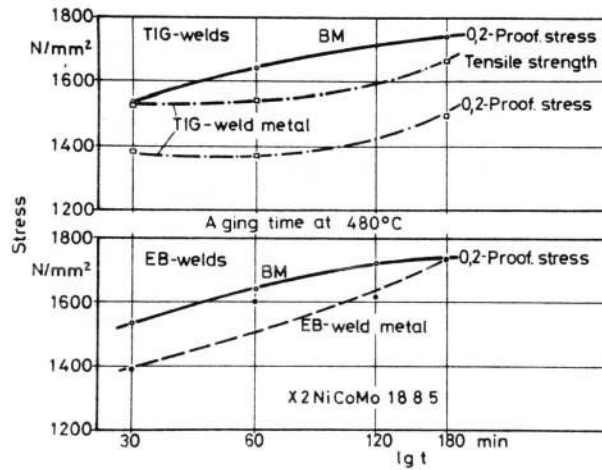


Figure 8 Yield Strength and UTS of Base Metal and TIG- and EB-Weld Metal as a Function of Aging Time at 430° C (1 N/mm² = 1 MPa)

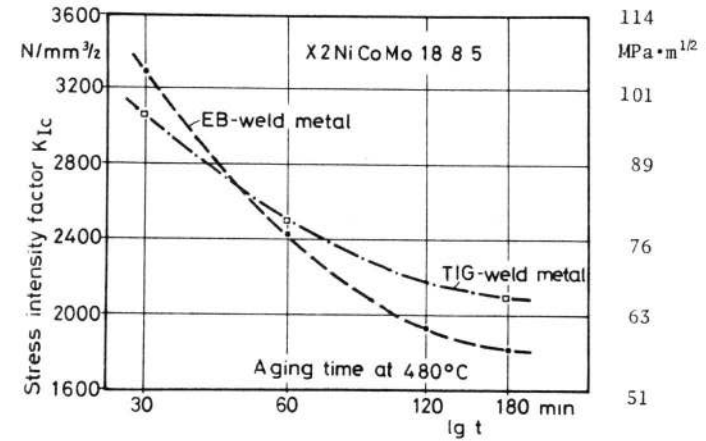


Figure 9 Fracture Toughness K_{IC} of TIG- and EB-Weld Metal as a Function of Aging Time at 480° C (1 N/mm^{3/2} = 0.0316 MPa·m^{1/2})