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Influence of Chemical Composition on the Mechanical Properties of High Strength Steel Weld Metals for Application in Mooring Components

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Abstract— **The present work is part of a research program for the development of welding procedures for chains and accessories for application in mooring systems of oil platforms. In the present work, the effect of changes in the chemical composition of the weld metal is discussed. These changes aimed to obtain high mechanical strength of 690 MPa and impact toughness of 40 joules at –20°C. Weld metals were deposited by FCAW process for evaluation of the mechanical properties, using a preheat of 200°C, direct current, flat position and nominal heat input of 1.2 kJ/mm. After welding, tensile, impact Charpy-V, hardness tests and metallographic examination were carried out in samples of all weld metal, both in as welded and heat treated conditions. The post weld heat treatment (PWHT) was performed at 580ºC for 2 hours. The results show that the obtained weld metals have mechanical properties higher than the minimum required for the welding of an IACS W22 R3 Grade steel, and particularly good impact properties. These results indicate that a particularly Mn-Ni well-balanced chemical composition allows to achieve an adequate strength/toughness relationship for high strength steel weld metals, where the PWHT is mandatory**.

Index Terms— **High Strength Steel, Mechanical Properties, Post Weld Heat Treatment, Weld Metal.**

I. INTRODUCTION

Mooring lines of offshore oil exploitation platforms are built with long lengths of steel chain links, steel wire and polyester ropes, anchors and other accessories [1]. Usually, these lines are designed for an operational life of about 20 years and have to attend to the requirements of the Classification Societies Rules [2]-[4]. The mooring lines chains must be removed and periodically inspected according to the requirements of API RP 2I [5], in order to guarantee safety performance of the mooring lines. A failure of a single element in a mooring line can promote incalculable environmental damage and severe economic losses. Based on this scenario, manual welding repair can be an interesting alternative to avoid unnecessary costs of replacement of a single component.

Nevertheless, to obtain the welding procedures qualification according to the stringent requirements for steels used in mooring components (Table 1), the development of welding consumables adequate to this application become necessary. This development is however a more complex task, since the execution of post welding heat treatment (PWHT) procedure is mandatory to reduce residual stresses [6]-[10]. This PWHT generates an additional problem to be solved, since the qualification standards of welding consumables [11]-[13] only have requirements for the as welded condition. Consequently, the viability of using commercial consumables must be considered and evaluated on a case by case basis, and the approval of specific heats of the welding consumables are the only possible alternative.

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Considering that previous works involving shielded metal arc welding process (SMAW) [6],[14],[15], obtained mechanical properties superior to the minimum required for R3 Grade steel, the present work investigates the correlation between microstructure and mechanical properties of a high strength steel weld metal obtained by Flux Cored Arc Welding process (FCAW) in order to promote an improvement on the productivity for welding of this Grade.

Table 1. Mechanical properties requirements for offshore mooring chain and accessories [4].

Grade	YS (MPa)	UTS (MPa)	El (%)	RA (%)	Charpy-V energy at -20ºC (joules)
R3	410	690	17	50	40
R3S	490	770	15	50	45
R4	580	860	12	50	50
R4S(1)	700	960	12	50	56
R5(1)	760	1000	12	50	58

YS – Yield Strength, UTS – Ultimate Tensile Strength, El – Elongation, RA-Reduction of Area

II. EXPERIMENTAL PROCEDURE

A. Consumables

Three different heats of commercial metal cored wires welding consumable with 1.2mm diameter according to the AWS 5.28 class 110C-G were studied. The chemical composition of the welding consumable supplied by the manufacturer is showed in Table 2.

Table 2. Chemical composition of the welding consumable according to the manufacturer (wt,%).

Element, Heat	C	Si	P	S	Mn	Mo	Ni
A	0.03	0.51	0.018	0.019	1.26	0.58	2.23
B	0.03	0.65	0.018	0.010	1.73	0.64	2.55
C	0.01	0.31	0.018	0.009	1.32	0.57	1.98

B. Welding

Welding was performed by FCAW process, in the flat position, with a mix of 25%CO_2-75%Ar as shielding gas, preheat of 200ºC, nominal heat input of 1.2 kJ/mm. The weld joint geometry is illustrated in Fig.1. The calculated cooling times between 800 and 500^{0}C are shown in Table 3. For these calculations it was taken into consideration the sampling position of the mechanical test pieces at the filling passes and at top beads, according to EN 1011-2 Annex D [16]. Although, the top bead microstructure was submitted to different cooling rates (Table 3), this region will also be evaluated due to its importance for the understanding of the microstructure of the reheated region.

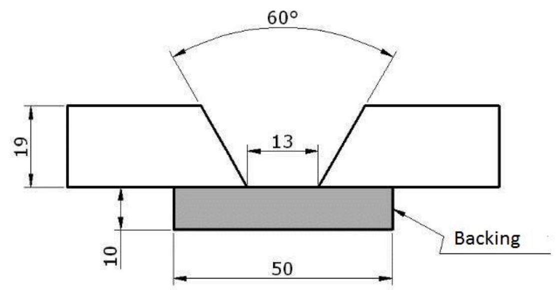


Figure 1. Weld joint geometry (mm).

Table 3. Calculated cooling times.

Weld Metal	A	B	C
$\Delta t_{8/5}$(fill pass), s	9.6	7.7	7.6
$\Delta t_{8/5}$(top beads), s	5.4	7.7	3.9

The weld metal was tested for both conditions: as-welded and after post welding heat treatment (PWHT) performed at 580°C for 2 hours followed by air-cooling.

C. Metallographic Examination

Traditional metallographic sample preparation was performed using Nital 2% etchant. Metallographic analysis was carried out by Optical (OM) and Scanning Electron Microscopy (SEM) at the top bead and at regions related to positioning of the Charpy-V notch. It should be noted that all samples were removed transversally to the weld bead. Microstructural quantitative analysis of columnar and reheated regions was performed at the same position where Charpy-V impact notch was placed. It is known that the volume fraction of regions crossed by the crack path can contribute for the impact toughness value of the weld metal [17].

D. Mechanical Tests

Tensile, impact Charpy-V and hardness tests, were performed in accordance to ASTM A-370 standard [18]. All weld metals tensile specimens were sampled in the longitudinal direction of the weld metal bead and tests of were performed at room temperature.

Charpy-V impact tests at –20ºC temperature were also performed on standard test pieces (10x10x55mm) removed transversally to the weld bead. The notch was positioned at the weld metal center line within the thickness section. The Vickers 1 kgf hardness test was performed transverse to the weld bead.

III. RESULTS AND DISCUSSION

Tensile tests results obtained from the weld deposits, Table 4, shows higher values for weld metal C, whereas the lowest values are for weld metal A.

Table 4. Results of tensile tests.

Weld metal	Condition	YS (MPa)	UTS (MPa)	El (%)	RA (%)
A	AW	631	690	22	61
	PWHT	580	673	25	63
B	AW	659	744	26	65
	PWHT	640	742	23	66
C	AW	717	775	12	31
	PWHT	668	759	22	58
Spec.W22 [4]		580	690	17	50

AW – As welded, PWHT – Post Weld Heat Treated.
YS – Yield Strength, UTS – Ultimate Tensile Strength,
El – Elongation, RA-Reduction of Area.

Ramirez [19], Surian et al. [20] and Talas [21], stated that the weld metal strength increases with increasing carbon equivalent number. Although the carbon equivalent was originally developed with the objective of evaluating the base metal cold cracking susceptibility, these general empirical equations can also be useful in understanding the complex relationship between the high-strength steel weld metal hardenability as controlled by the alloying content, the resulting microstructural transformation behavior of the weld deposit and associated tensile properties [19].
The results obtained in the present work as shown Table 5 and Fig. 2, are in agreement with this statement. It should be noted that these results are also in good correlation with other published experimental data [15],[19], [20], [22]-[29].

Table 5. Chemical composition of the deposited weld metals (wt, %).

Weld Metal	C	Si	Mn	Mo	Ni	Cr	Cu	V	Ceq (*)
A	0.05	0.31	1.09	0.51	2.45	0.03	0.02	0.01	0.51
B	0.05	0.41	1.32	0.52	2.48	0.02	0.02	0.01	0.55
C	0.03	0.45	1.37	0.54	2.66	0.03	0.03	0.01	0.55

$$Ceq_{IIW} = C + Mn/6 + (Cr+Mo+V)/5 + (Cu+Ni)/15$$

The effect of PWHT on UTS shows no meaningful variation for all weld deposits (Table 4) being a maximum reduction of only 3% observed for weld A, taking the UTS requirements of the IACS standard as a reference.

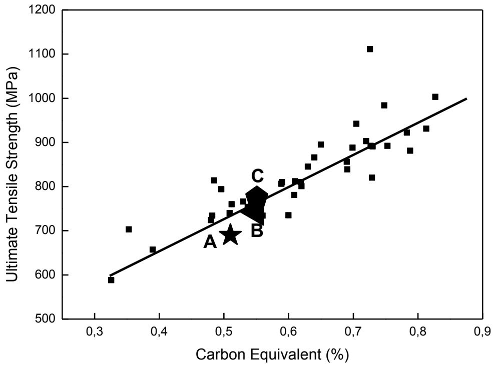


Figure 2. Influence of carbon equivalent on the ultimate tensile strength. A, B and C weld metals of the present work in comparison with other authors [15],[19],[20],[22]-[29].

According to Surian et al. [21], it is well known that when selecting a C-Mn steel as base alloy in order to increase tensile strength of the weld metal, it is necessary to add alloying elements such as Ni, Mo, and/or Cr, which in turn will modify other properties such as impact toughness. It is also known that an increase in UTS is frequently accompanied by a loss in toughness, particularly at low temperatures. For this reason, when designing an electrode formulation starting with C-Mn consumables, the main concern is devoted to maintaining the toughness requirements and the achievement of adequate tensile properties is seldom considered. Consequently, and mainly when PWHT is applied, there are problems to achieve the requirements for this property when working with high and extra high strength steel weld metals.

Weld A, presented the lowest Mn content (Table 5) and did not reach the UTS minimum value required for R3 grade steel, although all other properties are within the Specification W22 [4] as shown in Table 4 and Fig.3. A possible solution to this problem, is to increase Mn content to an adequate level since Ni is considerably less potent strengthener than Mn [30]-[32]. The results obtained for weld B containing 1.32%Mn (Table 5) achieving UTS superior to 690 MPa (Table 4), are an indication that this is an effective solution.

Charpy-V impact tests (Fig. 3) results are higher than the minimum required [4] for all conditions studied. Although, a reduction of the impact toughness for the welds B and C after PWHT is observed, these results confirm that impact toughness is not a problem for these high strength steel weld metals [20]. Similar behavior has already been observed previously [22],[23],[33].

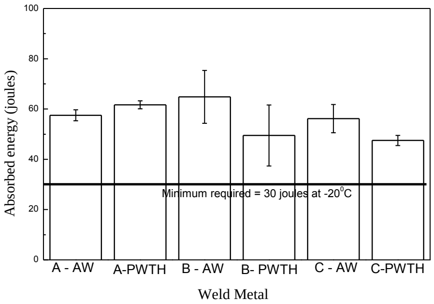


Figure 3. Results of Charpy-V impact tests.

Vickers hardness (Fig. 4) of the welds showed the same tendency of UTS, as expected. Weld metal C showed the highest hardness values while weld metal A showed the lowest ones.

In general, the region near to the top bead has higher hardness, since this region was not submitted to the effect of multiple passes.

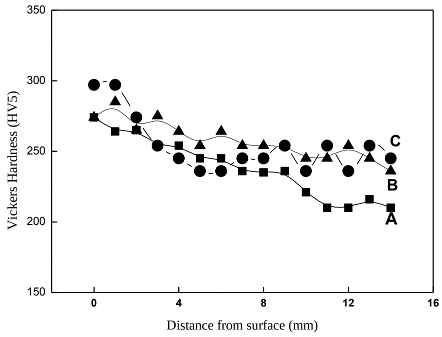
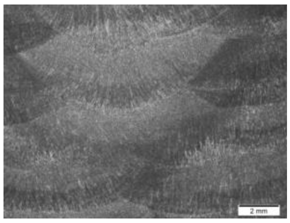


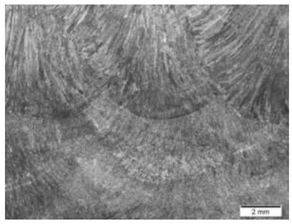
Figure 4. Hardness (HV5) tests results.

A detailed image of the region where the Charpy-V notch was positioned (Fig. 5), shows several beads. These regions were subjected to multiple, complex and varying thermal cycles containing both columnar and reheated zones in an alternated pattern [29]. The influence of the distribution of these regions, originated from multipass welding procedure, ahead of the Charpy-V crack path affecting both mechanical and microstructural properties of weld metals is a well-known fact.

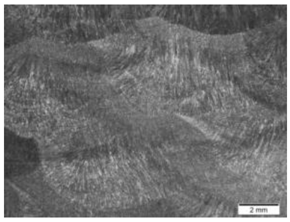
According to Narayanan et al. [34], the classification of welding microstructures using optical microscopy proposed by the International Institute of Welding [35] is sufficient and very useful for traditional C–Mn and most low alloy weld metals. However, for higher strength steel weld metals, this classification system is inadequate to show detailed differences of the weld microstructures, as those obtained in the present work (Fig. 6).



Weld A
62% of Columnar

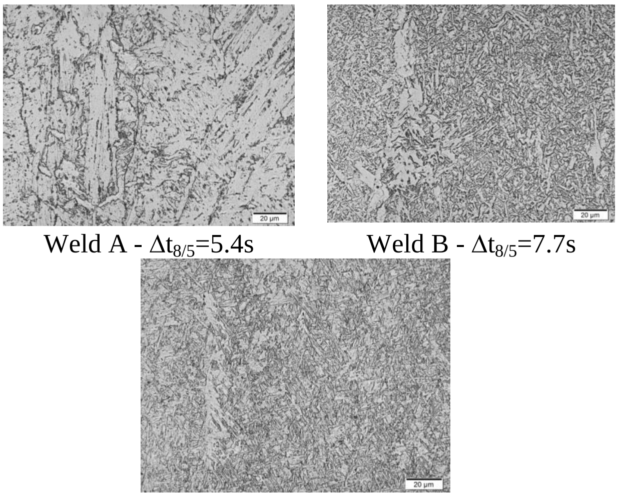


Weld B
69% of Columnar Region



Weld C
62% of Columnar Region

Figure 5. Optical micrographs (OM) of the region from where specimens for mechanical testing were sampled.



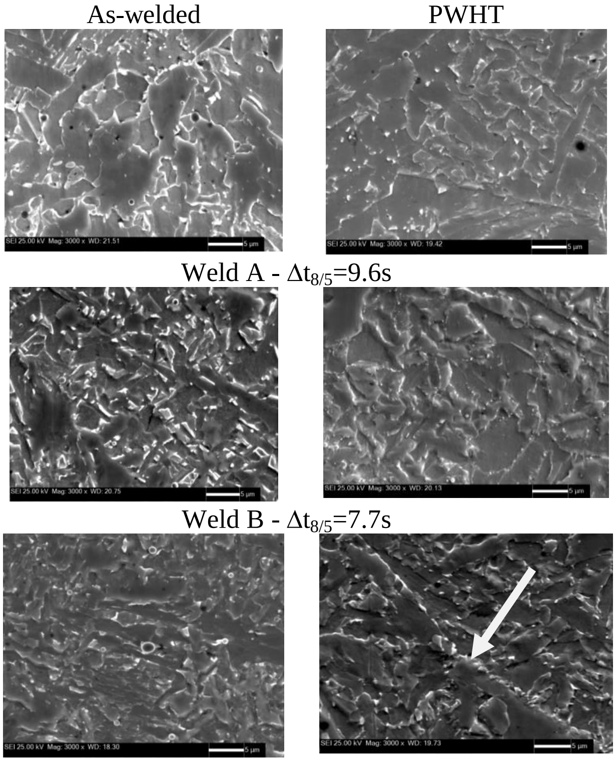
Weld A - $\Delta t_{8/5}$=5.4s Weld B - $\Delta t_{8/5}$=7.7s

Weld C - $\Delta t_{8/5}$=3.9s

Figure 6. Top bead microstructures of the weld metals observed by Optical Microscopy (OM).

The cooling rate and chemical composition resulted microstructures can be observed in Figs. 6 and 7. Ferrite with second phase (FS) microstructure was observed for weld A with low Mn content. The same microstructure was observed even at the top bead where the cooling time was slower (Fig. 6). For weld B with higher Mn content a mixture of acicular ferrite (AF), primary ferrite (PF) and ferrite with second phase (FS) was predominant. The increase of Ni in Weld C, in comparison with Weld B, promoted a mixture of Ferrite with second phase (FS) and martensite (M) whose proportions are dependent on the cooling rate (Figs. 6 and 7). On top bead (Fig. 6), it can be observed that the microstructure is martensite.

As-welded PWHT



Weld A - $\Delta t_{8/5}$=9.6s

Weld B - $\Delta t_{8/5}$=7.7s

Weld C - $\Delta t_{8/5}$=7.6s

Figure 7. Weld metal microstructures of the regions correspondent to the position of the Charpy-V notch. Etching: nital 2%.

Therefore it can be suggested, that small variations at cooling rates can promote greater changes on microstructures of these high strength steel weld metals as previously shown [26],[28],[29],[36].

After PWHT an improvement on impact toughness for weld A due to the tempering of microstructure can be observed.

The superior impact toughness of the Weld B was due to the presence of acicular ferrite and observed only for the as-welded condition. In this case, the apparent decrease of this property after PWHT can be consequence of the interdendritic segregation in some selected areas of the low temperature reheated zones (Fig. 8), being this phenomenon also reported [31],[38] as deleterious to the impact toughness of weld metals. The segregation associated with some selected areas where the Charpy-V notch were positioned also contributed to the great dispersion on Charpy-V test results for this condition (Fig. 3).

For weld C, a small decrease on impact toughness in comparison to the as-welded condition can be attributed to the occurrence of a continuous precipitation at the grain boundaries (Fig. 7) and to the variation which normally occurs in Charpy-V tests [19],[26].

IV. FINAL COMMENTS

It has been established that some high-strength weld metals exhibit a high degree of variability in mechanical property test results [38]. The variability of the properties of a weld metal can arise from various sources including the consumable lot-to-lot variation. Chemical composition variations sometimes may explain the differences, but in many cases, it is not clear.

In this respect, more systematic investigations involving the relationship between nickel and manganese have been reported [30], [37], [39], [40]. According to Svensson [41], two ways of achieving a good combination of high strength and high impact toughness has been reported, one with medium content levels of both manganese and nickel and another one with very low manganese and high nickel content.

In the present work, to reach the requirements of an IACS W22 Grade R3 steel, it was observed that it is necessary to use at least, medium manganese. Low manganese content (weld A) do not allow to achieve the UTS value required, which is the critical property to be reached, mainly after PWHT. Weld metals B and C achieved the requirements for application in mooring components, being interesting to mention that PWHT did not promoted substantial impairment on mechanical properties, allowing reliable strength-toughness relationships for both weld metals.

V. CONCLUSIONS

Based on the aspects discussed in the present work, the main conclusions are:

a) High strength steel all weld metals obtained by FCAW can reach high mechanical strength and impact toughness after PWHT;
b) PWHT did not promote significant changes on the mechanical properties and;
c) Strict control of the Mn-Ni relationship is necessary to reach the minimum requirement for R3 grade steel after PWHT.

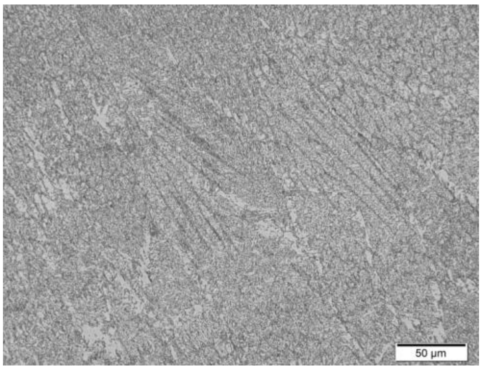


Figure 8. Occurrence of segregation associated with M-A constituents. Etching: Nital 2%.

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