INFLUENCE OF THE SOLUTION TREATMENT TEMPERATURE UPON THE CAVITATION EROSION RESISTANCE FOR 17-4 P.H. STAINLESS STEELS

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Abstract

The 17-4P.H. stainless steels also called "half austenitic" are used especially in manufacturing high speed aircrafts and missiles and because those devices are running at temperatures over maximum value allowed for the high resistance alloys with aluminium and manganese (over 200 \degree C).

The same stainless steels can also be used in manufacturing hydraulic machineries subjected to intense cavitation erosion. They combine the most attractive properties of austenitic steels (weldability and deformability) and martensitic steels (high mechanical and cavitation erosion resistance). Their modification can be obtained through a solution quenching followed by aging.

The present research examines the effect of various solutions quenching temperatures (950 \degree C respectively 1050 \degree C) upon the cavitation erosion behaviour of steels with 17 \% Cr, 4 \% Ni, 2 \% Mo, 2 \% Co, 2 \% Mn.

The cavitation erosion tests were carried out in a vibratory facility with piezoceramic crystals. The eroded surfaces were analyzed, after different testing times, through optical and scanning electron microscopy.

Keywords: 17-4 P.H. stainless steels, cavitation erosion, microstructure

1. REQUIREMENTS IMPOSED FOR 17-4 P.H. STAINLESS STEELS

The stainless steels with controlled transformation from the 17-4 P.H. group, also called "half austenitic" stainless steels combine the most attractive properties of austenitic steels (weldability and deformability) and martensitic steels (high mechanical resistance). Their modification can be obtained through a heat treatment after technological processing.

The main requirements imposed for such a steel are the following [5]:

- a softening treatment (hardness up to 200 HV) to be possible by forming a predominantly austenitic microstructure technology to facilitate technological processing;
- martensitic transformation to be achieved by a relatively simple treatment that does not involve heating to high temperatures with adverse consequences on the phenomena of oxidation and deformation of parts;
- the steel must have a limited amount of ferrite to allow a positive response to the primary aging heat treatment;
- to be able to obtain a wide range of mechanical resistance by applying relatively simple heat treatment at a moderate temperature;
- good ductility and toughness characteristics both in terms of the softening and hardening treatment;
- a good metallurgical behaviour at welding, the weld metal and heat affected zone to respond in a manner similar to the base metal;
good resistance to intercrystalline corrosion, tenso-cracking and cavitation erosion. Many of these demanding requirements are contradictory and therefore in practice are required some compromises [3].

The main objective to be pursued in this paper aims at optimizing the heating temperature for implementing such a solution which has the starting point martensitic transformation Ms situated near room temperature and at which the ferrite δ content is limited on first, to obtain mechanical properties as high and good resistance to erosion by cavitation, and on the other hand, for a prompt and favorable response to the primary aging heat treatment.

2. EXPERIMENTAL PROCEDURE

The selection of the heating temperatures for implementing the solution was based on the determination of the critical temperature of solid state transformation and the results of sclerometric examinations. The cavitation erosion tests were conducted by a vibratory machine with piezoceramic crystal [6], built after the regulatory requirements of ASTM G32-2010, similar to the machine model from the University of Michigan [1]. As liquid environment, drinking water from the public network was used. During the researches, the water temperature was maintained at the value of 22° ± 1° C. For each heating temperature for releasing in solution, were tried 3 samples, whose surfaces were polished at a roughness Ra = 0.2 - 0.8 μm. The total duration of testing each sample was 165 minutes, which is divided in 12 periods (one each for 5 and 10 minutes, and 10 each for 15 minutes). At the end of each test period, the mass of material was determined by weighing the mass loss by erosion cavitation, respective the average depth of erosion (MDE) and corresponding speed (MDER), and the cavitation eroded surfaces were examined by the optical microscope and photographed. Also, after the completion of the testing procedures by erosion cavitation (165 minutes), the eroded surfaces were examined with scanning by the electronic microscope.

3. EVALUATION AND INTERPRETATION OF EXPERIMENTAL RESULTS

The chemical composition of the steel studied is shown in Tab. 1.

Tab. 1 Chemical composition of the steel

<table>
<thead>
<tr>
<th>Accompanying and allying elements, %</th>
</tr>
</thead>
<tbody>
<tr>
<td>C</td>
</tr>
<tr>
<td>-----</td>
</tr>
<tr>
<td>0.10</td>
</tr>
</tbody>
</table>

For a preliminary outline of the influence of alloying elements on the constitution of the steel, it has started from the Schäffler diagram [2].

The calculus relationships of chromium and equivalent nickel used to define the nominal characteristic point are:

\[
\begin{align*}
Cr_e &= \% Cr + 1.5x \% Si + \% Mo + \% Al + 0.5x \% (Ta + Nb) + 2x \% Ti + \% W + \% V = 19.01 \% \\
Ni_e &= \% Ni + 30x \% C + 0.5x \% Mn + 0.5x \% Co = 9.92 \% 
\end{align*}
\] (1) (2)

Transposing these values on the Schäffler diagram, it resulted that the microstructure of the cast steel consists of austenite + martensite + 10 % δ ferrite.

For the experimental determination of the critical points of transformation in solid state, were made samples having the sizes 15 x 3 x 3 mm, which were heated in high vacuum (10⁻⁵ mm Hg) to 1050 ° C.

For the tests it was used an Ernst-Leitz Dilatometer which enabled photographic record of the expansion curves.
In Figure 1 is rendered such a recording, from which are the following observations:

- $A_c_1$ critical point is lowered to approx. 536 °C especially due to the gamagene action of Ni and Mn;
- $A_c_3$ critical point is approx. 914 °C and as a result, the heating temperature for the release in solution must be higher then 920 °C, given the high degree of alloying steel;
- the martensitic transformation start temperature, $M_s$ is shifted to approx. 65 °C for all alloying elements except Co and Al.

The heating temperature in solution must ensure that all or most of the secondary phase particles (carbides, carbonitrides, etc.) and $\delta$ ferrite, and after cooling in air to obtain a specified minimum hardness of austenitic microstructure with a small proportion of martensite and ferrite $\delta$. After hardening in order to release in the solution, the steel must be readily processed by plastic deformations at hot and cold, or welding operations, and later it is needed that the austenitic structure to be able to be transformed in martensite at temperatures which would not produce deformations of the made products. The further increase of the mechanical strength will be made though a hardening treatment by precipitation performed at relatively low temperatures.

The results of the scleromatic examinations summarized in Tab. 2, show that the temperature at which the austenitification is made, influences significantly the microstructure of the steel and hence the behaviour at cavitation.

Micrographic images of Figure 2 and Figure 3 come to complete the sclerometric examination results.
### Tab. 2 Effect of the heating temperature on the hardness of the quenched samples

<table>
<thead>
<tr>
<th>Item No.</th>
<th>Heat treatment</th>
<th>Sample No.</th>
<th>Pieces No.</th>
<th>Vickers HV5 hardness</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>Experimental values</td>
</tr>
<tr>
<td>1</td>
<td>Quenching 900 °C, air</td>
<td>2.1</td>
<td>3</td>
<td>A: 340; 342; 341</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 354; 352; 353</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 347; 341; 343</td>
</tr>
<tr>
<td>2</td>
<td>Quenching 925 °C, air</td>
<td>2.2</td>
<td>3</td>
<td>A: 337; 341; 342</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 360; 379; 358</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 334; 330; 332</td>
</tr>
<tr>
<td>3</td>
<td>Quenching 950 °C, air</td>
<td>2.3</td>
<td>3</td>
<td>A: 353; 352; 354</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 320; 333; 325</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 314; 314; 315</td>
</tr>
<tr>
<td>4</td>
<td>Quenching 975 °C, air</td>
<td>2.4</td>
<td>3</td>
<td>A: 255; 259; 268</td>
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<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 284; 282; 285</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 308; 304; 304</td>
</tr>
<tr>
<td>5</td>
<td>Quenching 1000 °C, air</td>
<td>2.5</td>
<td>3</td>
<td>A: 251; 267; 267</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 230; 230; 239</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 278; 262; 257</td>
</tr>
<tr>
<td>6</td>
<td>Quenching 1000 °C, 1 h, air</td>
<td>2.6</td>
<td>3</td>
<td>A: 237; 232; 234</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 234; 232; 237</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 247; 242; 242</td>
</tr>
<tr>
<td>7</td>
<td>Quenching 1050 °C, 1 h, air</td>
<td>2.7</td>
<td>3</td>
<td>A: 254; 254; 254</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>B: 231; 232; 228</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td></td>
<td>C: 236; 233; 235</td>
</tr>
</tbody>
</table>

**Fig. 2** Quenching 950 °C / air (x500)  
**Fig. 3** Quenching 1050 °C / air (x500)

Thus, quenching temperatures of 900 °C – 950 °C manifest by a lower release in solution of carbides and other intermetallic phases, less chemical homogenization of austenite, hence a lifting of the martensitic transformation starting temperature; therefore, the microstructure obtained will consist of austenite, martensite, ferrite and undissolved carbides (Fig. 2).

Raising the heating temperature to 1040 °C – 1050 °C favors the release in solution of the secondary phases, dissolution of a ferrite σ part, an increase of stability of austenite transformation or the lowering of the Ms point; therefore, the microstructure will consist of austenite and a small proportion of ferrite σ (Fig. 3) and its hardness (231HV5 – 254 HV5) ensures good technological properties.

The application of the primary aging treatment to 700 °C favors the increase of the Vickers hardness (328 HV – 338 HV) only at the samples austenitized to 1050 °C; this is due to the transformation of the metastable
austenite part in the martensite; instead, samples quenched from 950 °C and subsequently subjected to primary aging to 700 °C, have practically hardness values (318 HV – 323 HV) similar to those specific quenched for releasing in solution; this is due to low carbon content of the steel, implicitly a limited proportion of carbide precipitates and a similar concentration of carbon in austenite chromium after releasing in the solution and after the primary aging treatment. The final recovery treatment at 450 °C is manifested by the appearance of a secondary hardening (335 HV – 345 HV at those austenitized at 1050 °C), which can be attributed to the transformation of an important part of austenite in martensite and the precipitations of secondary phases due to the presence of the chemical composition of alloying elements Mo, Cr, Al.

Firstly, it can be seen that the martensitic structure obtained after quenching from 1050 °C followed by primary aging treatment to 700 °C is hardly attacked by metallographic reagent. The quenching making from 950 °C and primary aging to 700 °C, although it promotes a pronounced martensitic transformation, keeps a certain amount of austenite and carbides in the microstructure.

Secondly, it should be noted that the primary aging treatment manifests by the phenomena precipitation of carbides mainly on the edges of the ferrite δ crystals. In this way, there is a decrease in carbon content in the alloying carburigene elements in austenite and so a raising of the martensitic transformation domain, which makes the air cooling until the room temperature to trigger the martensitic transformation [4]. The presence of a small amount of approx. 10% ferrite δ in the microstructure speeds the precipitation of the M23C6 carbide, and so the primary aging process is favored. This phenomenon is justified by increasing the separation areas of austenite and ferrite δ, they constitute the core of M23C6 carbides, and the transformation of austenite with low carbon content (temperate Ms will be higher) is relieved. Figure 4 and Figure 5 exemplify specific microstructures of the recovered steel at temperatures of 450 °C. They show increased sensitivity to metallographic attack with the emerge and development of the precipitation phenomena responsible for the secondary hardening of the steel.

![Fig. 4 Quenching 950 °C / air + Primary aging](image)

700 °C / air + Recovery 450 °C / air (x500)

![Fig. 5 Quenching 1050 °C / air + Primary aging](image)

700 °C / air + Recovery 450 °C / air (x500)

These structural modifications confirm the differences occurred in the behaviour of the analyzed steel at cavitation. In the diagrams shown in Figure 6 and Figure 7 are characteristic curves of cavitation erosion, indicating changes of the mean depth of penetration of erosion, respectively it’s speed with the duration of the attack of the cavitation generated in the T2 vibratory device. The OH12NDL steel was used as a standard, from which the hydraulic turbine blades from the Hydroelectric Power Plant Portile de Fier I and II (Romania), were executed.
The obtained data demonstrates that the quenching temperature for the releasing in the solution of 1050 °C provides the highest resistance to erosion by cavitation and that both heat treatment conditions based on intercalation of aging to 700 °C between quenching and tempering leads to better results than the standard steel.

**Fig. 6** Average penetration depth variation with cavitation time attack (MDE)
1 - Quenched 17-4 P.H. steel at 1050 °C
2 - Quenched 17-4 P.H. steel at 950 °C
3 - Standard OH12NDL steel

**Fig. 7** Mean depth of penetration speed variation with cavitation time attack (MDER)
1 - Quenched 17-4 P.H. steel at 1050 °C
2 - Quenched 17-4 P.H. steel at 950 °C
3 - Standard OH12NDL steel

4. **CONCLUSIONS**
1. The heating temperature for releasing in the solution of the researched 17-4 P.H. stainless steel influences the hardening level reached at the heat treatment and implicitly the cavitation erosion resistance.
2. The increase of the austenitizing temperature to 1050 °C favors the releasing in the solution of the secondary phases, dissolution of a ferrite δ part, an increased stability of austenite transformation, respectively a lowering of the Ms point.
3. The application of the primary aging treatment to 700 °C, causes the transformation of an austenite metastable part in martensite, and the final recovery treatment to 450 °C is manifested by the appearance of a secondary hardening, slightly more pronounced compared to the austenitized samples to 950 °C, with consequences of increasing resistance to cavitation.

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**REFERENCES**