HARDENING OF AUSTENITIC STEELS WITH HIGH Mn AND Al CONTENT

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Abstract: Estimates and experimental study of aluminum, carbon and nitrogen content effect for thermally and mechanically stable austenitic alloys based on Fe-(20–25)%Mn-(5–10)%Al-(0–10)%Ni-C-N, and hardening efficiency of these alloys by nitrogen addition are conducted. It is shown that the alloys based on the investigated system with C content of >1.2% for nickel-free system and ≥1.4 % for 5% and 10% of Ni can be proposed as high-strength non-magnetic cryogenic both in heat-treated and aged states. Nitrogen addition in the alloy containing 5% Al can be considered as micro-alloying, enhances dispersion of the cast structure. Ni enhances the dispersion hardening effect due to increase in carbide release rate resulted from the reduction of C solubility in austenite. Rational content of nickel equals to ~5%.

KEYWORDS: IRON-MANGANESE ALLOYS; TRIPLEX ALLOYS; NITROGEN MICROALLOYING; HIGH-STRENGTH ALLOYS; MULTIPHASE STRUCTURE.

1. Introduction

In the past few decades, the efforts of many researchers focused on the development of the advanced high-strength steel (AHSS) with adequately high ductility and toughness. It is often trying to get the heterogeneous structure with ultrafine particles (up to nanosizes) and one or several of such particles should be metastable to deformation under load [1]. It provides high mechanical hardening and enhanced ductility.

The alloys based on Fe-Mn-Al-C system, or so called TRIPLEX alloys with high Mn and Al content satisfy this requirement and can be applied as light high-strength steels.

This work is directed to calculation assessment and experimental research of the effects of Al, Ni and C content on the thermally and mechanically stable austenitic steels based on the Fe-(20–25)%Mn-(5–10)%Al-(0–10)%Ni-C systems.

2. Material and Experiment

Fe-20%Mn-(5–10)%Ni-10%Al-C system is chosen as the basic calculation system in order to provide the maximum strengthening due to the Al alloying.

Thermo-Calc software using thermodynamical database TCFE7 for the chosen basic content is applied to plot the polythermal section of phase diagrams and to determine the thermal-concentration regions of equilibrium phase existence such as austenite, carbides nitrides and intermetallides. Generalized Hall-Petch law is used to assess the influence of the chemical and phase-changing content of the alloy under the thermal treatment, more specifically, the solid-state and dispersion strengthening, as the effect of structure strengthening of the steels with different content is generally determined by technological parameters of metal production processes.

Experimental researches on the influence of the content on the mechanical properties and phase content in cast and deformed steel are made for certain alloys (see Table 1 for the chemical content).

Mechanical properties are tested for flexure under GOST 14019-2003 on the Instron 3300 test system under the normal temperature.

| Table 1. Chemical composition of investigated alloys |
| No. | Alloys | Mn | Al | Si | Mo | C | N |
| 1 | 35K | 21.2 | 6.2 | 0.50 | 0.20 | 1.10 | 0.006 |
| 2 | 33K | 19.1 | 9.0 | 0.50 | 0.03 | 2.18 | 0.001 |
| 3 | 34K | 23.9 | 4.0 | 0.54 | 0.01 | 1.80 | 0.032 |
| 4 | 44K | 24.1 | 5.3 | 0.32 | <0.001 | 0.04 | 0.024 |

* – base is Fe

3. Results and Discussion

Phase diagrams show that all Fe-20%Mn-10%Al-(0–10)%Ni-C alloys with low carbon content crystallize according to L+γ→L+α+γ→α+γ method and when C content exceeds 1.3 and 1.4 % the crystallization finishes with γ-phase formation.

Addition of 5 % of Ni does not influence the type and sequence of different carbides formation. The higher the C content, the higher the temperature of carbide formation in alloy structure. Addition of 10 % of Ni supports the extension of γ-region and the reduction of required concentration of C till 1.05 % under 1000 °C and 1.35 °C under 1200 °C.

For the sake of further assessment the C content of 1.4 % is taken as the basic content since it allows to create and treat the alloys in the fully austenitic state (without proeutectoid constituent) and their dispersion strengthening as the result of aging.

In Fe-Mn-Al-Ni-C-N nitrides of aluminum are formed at temperatures considerably exceeding the liquidus temperature, therefore, the nitrogen content in the aluminum is its solubility at this temperature. The solubility of nitrogen in the melt and in the solid δ- and γ-phases decreases with decreasing temperature and increasing content of aluminium [2].

Table 2 shows the assessed temperatures of the martensite transformation initiation for steels Ms as well as temperatures allowing the formation of 50 % of martensite under the deformation of 30 %, Md30 according to [3, 4].

It is clear, that thus received assessments are very rough, but in this case only evaluation of assessment changes due to the austenite content change is important.

As we see, the full solution hardening results in the stability of all alloys up to the temperatures close to 6 K. The reduction of Ms and Md30 due to addition of 10 % Ni is about 200–250 K. This allows significant dispersion hardening of such alloys in case the stable austenite matrix is persistent till the cryogenic temperatures.

The volume of released phases f and the change of austenite content under the aging calculated using the Thermo-Calc software are indicated in Table 2.
Table 2. Estimation of temperatures* $M_S$ and $M_{D30}$ estimated alloys Fe-20%Mn-10%Al-1.4% of C-x%Ni

<table>
<thead>
<tr>
<th>Ni, %</th>
<th>After treatment for solid solution</th>
<th>After aging</th>
<th>The content of elements in the austenite, % wt.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$M_S$, K</td>
<td>$M_{D30}$, K</td>
<td>$T$, °C $f$, % Ni Mn Al</td>
</tr>
<tr>
<td>0</td>
<td>17</td>
<td>&lt;4</td>
<td>800 1,17 1,34 0 19,77 10,06 40 &lt;4</td>
</tr>
<tr>
<td>5</td>
<td>&lt;4</td>
<td>&lt;4</td>
<td>790 5,67 1,20 5,3 20,75 9,90 &lt;4 &lt;4</td>
</tr>
<tr>
<td>10</td>
<td>&lt;4</td>
<td>&lt;4</td>
<td>827 (T1) 4,95 1,24 10,52 21,04 9,83 &lt;4 &lt;4</td>
</tr>
<tr>
<td></td>
<td>773</td>
<td>0,96</td>
<td>11,16 21,31 9,84 &lt;4 &lt;4</td>
</tr>
</tbody>
</table>

* for the calculation of temperatures $M_S$ and $M_{D30}$ for austenite solution heat treatment we take the content corresponding to the chemical compound of the alloy, for the condition after aging – considering the release of proeutectoid constituents from austenite

Results of calculation of $M_S$ and $M_{D30}$, possible number of proeutectoid constituents after aging for certain experimental alloys content and the actual phase content are shown in Table 3.

Calculated temperatures of $M_S$ and $M_{D30}$ for heat-treated high-carbon steels 33K, 34K and 35K are very low, i.e. austenite in such compounds is thermally and mechanically stable till cryogenic temperatures. Under the complete release of carbides from austenite $M_S$ in 34K alloy or $M_{D30}$ in 33K and 35K increase to the values above normal temperature. This proves that hardening of such alloys due to aging should be done in regimes maintaining the proper stability of austenite. In order to prevent the release of carbides in cast alloys higher cooling rates are needed after the casting and solidification.

Under the experiment the cast alloys 33K, 34K and 35K under cooling rate of ~$10^2$ K/s show the existence of austenite in the form of matrix phase (by X-ray method), 35K alloy shows c-carbide traces and 33K exhibit the significant amount of c-carbide (see Table 3).

High-carbon alloys 33K and 34K exposed to the cold deformation using three-point bending method remain austenitic under the normal temperature and 35K showed small amount of α-martensite. The total content of N in researched experimental alloys is very low (less than 0.01–0.03%) and is represented in the form of AlN nitrides. So, it can be expected that the influence of N on steel properties is similar to the influence of other proeutectoid constituents.

Estimation of the alloy flow limit is calculated by the following relation [5]:

$$YS = Y_{S0} + \Delta S_{S,S} + \Delta S_d + \Delta S_c + \Delta S_{S,G} + \Delta S_{p},$$

where $Y_{S0}$ = YS of pure iron (30–40 MPa); $\Delta S_{S,S}$ – solid-solution hardening; $\Delta S_d$ – dispersion hardening due to the second constituent particle; $\Delta S_c$ – grain boundary strengthening; $\Delta S_{S,G}$ – hardening due to formation of subgrain structure upon polygonization; $\Delta S_{p}$ – hardening due to increase in $p$ dislocation density.

Solid solution hardening may be estimated by equation

$$\Delta S_{S,S} = \Sigma K_i C_i,$$

where $K_i$ – hardening due to each dissolved element. Empirical expressions are used to assess the certain effects of austenite content on flow limit and ultimate strength, see [4].

Calculated assessment of the austenite content influence on mechanical properties after the treatment on solid solution and after the aging, as well as assessment of dispersion hardening due to the particles of released constituents upon aging (see Table 2, Figures 1 and 2) for estimated (model) and experimental steels.

Fig. 1. Comparison of calculated (solid lines) and experimental (dashed) values of strength UTS (a) and YS (b) for alloys 44K, 35K and 34K and traditional corrosion-resistant austenitic steel Cr18Ni10 [6]

Table 3. Phase composition and temperature rating* $M_S$ and $M_{D30}$ experimental Fe-Mn-Al-C alloys

<table>
<thead>
<tr>
<th>Alloys</th>
<th>The phase composition of the alloys in the cast state</th>
<th>After cold deformation</th>
<th>After treatment for solid solution</th>
<th>After aging</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$\gamma + [k]^{**}$</td>
<td>$\gamma + \alpha$</td>
<td>54 79</td>
<td>804 (T1) 7612 6 0,628 73 104</td>
</tr>
<tr>
<td>33K</td>
<td>$\gamma + \kappa$</td>
<td>$\gamma + \kappa$</td>
<td>&lt;4 &lt;4</td>
<td>971 (T1) 7572 29,430 153 122</td>
</tr>
<tr>
<td>34K</td>
<td>$\gamma$</td>
<td>$\gamma$</td>
<td>&lt;4 &lt;4</td>
<td>985 (T1) 6602 17,769 321 376</td>
</tr>
<tr>
<td>44K</td>
<td>$\gamma$</td>
<td>$\gamma$</td>
<td>234 542</td>
<td>– – – – – –</td>
</tr>
</tbody>
</table>

* for the calculation of temperatures $M_S$ and $M_{D30}$ for austenite solution heat treatment we take the content corresponding to the chemical compound of the alloy, for the condition after aging – considering the release of proeutectoid constituents from austenite;

** trace (less 1 %)
4. Conclusion

Alloys based on Fe-20%Mn-10%Al-(0–10)%Ni-C with C content of >1.2% for nickel-free system and ≥1.4% for 5% and 10% of Ni can be proposed as high-strength non-magnetic cryogenic steels both in heat-treated and aged states.

Nitrogen addition in the alloy containing 5% Al can be considered as micro-alloying, enhances dispersion of the cast structure.

Ni enhances the dispersion hardening effect due to increase in carbide release rate resulted from the reduction of C solubility in austenite. Rational content of nickel equals to ~5%.

Results of experiments confirmed the adequacy of phase content calculation as well as the assessments of solid-solution and dispersion hardening of alloys based on Fe-(20–25)%Mn-(5–10)%Al-(0–10)%Ni-C.

Cast high-carbon alloys Fe-(19.1–23.9)%Mn-(4.0–9.0)%Al-(1.1–2.18)%Ni-C allow to reach the high strength and adequate ductility under deformation conditions at normal temperature. Strength values: YS=1000–1200 MPa, UTS=1300–1600 MPa upon deformation of ε=10%.

5. References


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