EVALUATION OF MICROSTRUCTURAL STABILITY OF DISSIMILAR WELD JOINTS

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Resume

The microstructural changes occurring in the weld joint P92/316Ti during his long-term exposure at high temperature were studied. In parallel to experiments were carried out calculations of phase equilibria for the base materials and the weld joint using the ThermoCalc software. Based on the experimental results and computational modeling results were evaluated a microstructural stability and the application of the base materials and the weld joint.

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1. Introduction

At present time, large number of power generating facilities undergoes reconstruction and refitting. The most widely used materials in these applications are creep-resistant steels which are often connected by dissimilar weld joints [1], [2]. Generally, the necessary high creep-strength of these materials and their joints is ensured by a microstructural stability [3].

In this work a microstructural stability of creep-resistant steels P92 and 316Ti and their dissimilar welds is examined. As a suitable tool for evaluation of microstructural stability, computational modelling of phase composition at thermodynamical equilibrium was selected. This modelling approach forms nowadays one of the standard tools of material designing process [4]. The Thermo-Calc software, which uses the CALPHAD method [5, 6] presents a generally accepted standard software used for computational phase equilibria determination. Thermodynamic database STEEL 16 [7] was used in the calculations.

2. Experimental

P92 and 316Ti steels of standard purity were used as experimental material. The chemical composition of the used steels is in Table 1. The steels were supplied in heat-treated state.

Table 1

Chemical composition of used steels (wt. %)

Steel	C	Mn	Si	Cr	Ni	Mo	V	W	Ti	Nb	N
316Ti	0.02	1.83	0.6	17.1	11.8	2.25	0.14	-	0.19	0.02	0.06
P92	0.09	0.52	0.34	8.96	0.36	0.4	0.23	1.5	-	0.05	0.03

Table 2

Temperatures and times used for annealing of experimental samples

Series	A	B	C	D	E
Temperature (°C)	500	600	650	750	1000
Time (h)	1000	160	100	60	8

Cylindrical samples with one polished basis were machined out of the materials. The samples were resistance-welded to form the experimental weld of P92/316Ti. These were subsequently annealed in evacuated glass capsules at temperatures 500 – 1050 °C, for 8 – 1000 hours (see Table 2). After annealing the samples were rapidly cooled down in water. Samples were cut up from the heat treated samples perpendicularly to the weld interface. Metalographical evaluations of microstructure and microhardness measurements were performed on the samples across the weld interface.

3. Results

3.1. Base materials

The 316Ti steel belongs in the group of stabilized austenitic steels. According to computed diagram in Fig. 1., the material has austenitic matrix hardened with intermetallic phases and small amount of carbides and nitrides.

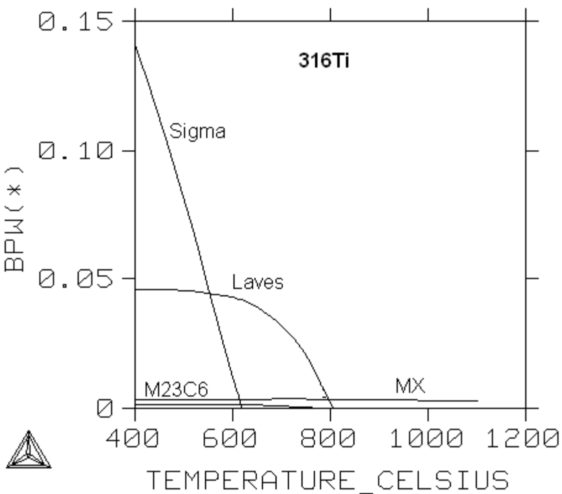

Fig. 1. Temperature dependence of weight fraction of minor phases in the steel 316Ti (ThermoCalc)

The default microstructure of samples was formed of austenitic matrix with a high amount of titanium nitrides. In the microstructure was also observed a low content of intermetallic phases streamlined in the direction of rolling of the initial stock. Significant changes in the microstructure of steel were observed only after annealing at 750 °C. At this temperature the intermetallic phases on grain boundaries was formed and the formerly created particles was coarsen. Titanium nitrides stayed stable at all temperatures. By a measuring of hardness was found a decrease from about 190 HV 0.1 at the unannealed sample to about 160 at all annealed samples.

The P92 steel belongs in the group of 9 – 12 % Cr steels. According to computed diagrams in Fig. 2., the material is hardened by $M_{23}C_6$ carbides and MX carbonitrides which are present in the martensitic matrix. Up to 700 °C the tungsten Laves phase is also stable.

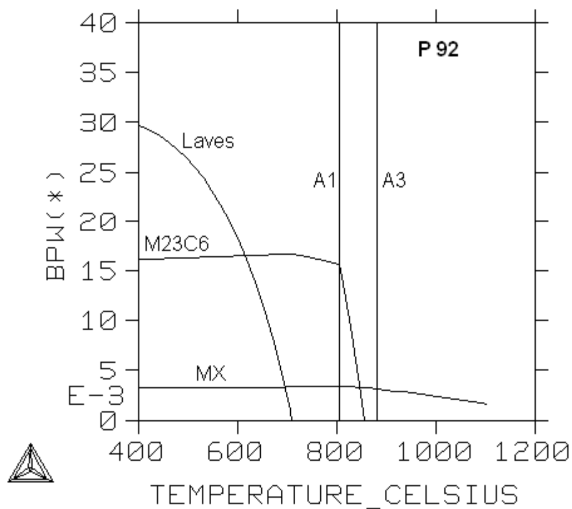

Fig. 2. Temperature dependence of weight fraction of minor phases in the steel P92 (ThermoCalc)

The microstructure of all samples of the P92 steel had fine-grained martensitic matrix. Up to 650 °C changes in microstructure were not significant. It was observable only very slow dissolution of particles of minority phases with increasing temperature. At 750 °C a sorbitic microstructure was observed. At temperature 1050 °C the steel P92 was fully austenitic, which correspond to the microstructure of the sample consisting of coarse martensite, which was created by the rapid cooling of the sample. Hardness decreased with increasing temperature from about 590 HV 0.1 at the unannealed sample to 215 HV 0.1 at 750 °C.

3.2. The weld joint

A typical heat affected zone was formed in the steel P92 during welding of both materials. This led to the creation of 30 μm wide belt of ferrite in the weld interface area and to grain coarsening in adjacent area. The grain coarsening was occurred to on the site of steel 316Ti in the weld interface area. This area was in all samples significantly more resistant against etching.

The microstructure of the steel 316Ti did not experience noticeable changes around the weld interface area up to temperature of 600 °C (Fig. 3.). At the temperature 650 °C was precipitated particles of intermetallic phases on the austenite grain boundaries.

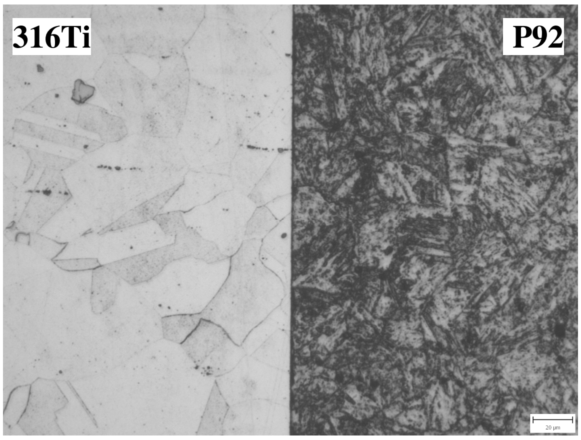

Fig. 3. Weld interface after annealing at 600°C/160h

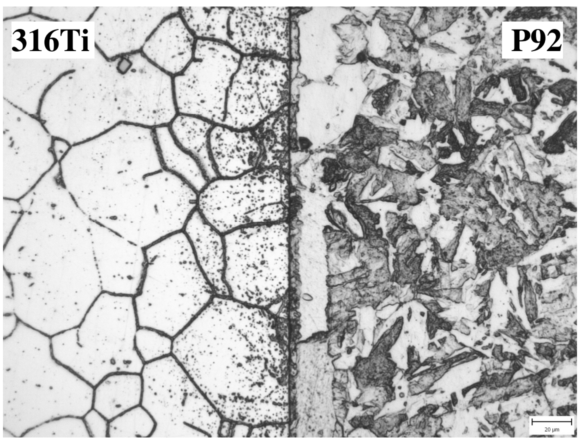

Fig. 4. Weld interface after annealing at 750°C/60h

At 750 °C a network of intermetallic phases was formed at grain boundaries, these phases were also excluded in the form of rows parallel to the direction of rolling of the original stock (Fig. 4.). Compared to the base material was near the weld interface several times higher amount of intermetallics. These intermetallics were identified as Laves phase on the basis of computational modeling. There was also a significant precipitation of carbide particles in area close to the weld interface. As a result of carburizing at the temperature 1050 °C a more significant precipitation of carbide particles was observed compared to the basic material to a distance of about 300 μm from the interface.

For P92 steel a ferrite has precipitated in the area around the weld interface on the grain boundaries at the temperature 500 °C. At the temperature 600 °C the ferrite formed on the interface during welding did not occurred. Microstructure around the weld interface did not show significant changes compared to the base material. At the temperature 650 °C was observed slight coarsening of carbides and growing of their inter-particle distance. At a temperature of 750 °C was dissolved large part of carbides precipitated inside the grains. Carbides excluded at grain boundaries was slightly coarsen. At the temperature 1050 °C was dissolved a most of a ferrite produced during welding.

Microhardness profiles measured across the weld interface was characterized by a continuous transition between the hardness of both basic materials in weld interface area (Fig. 5). Only in the case of 750 °C a sharp step change in the hardness was occurred on the interface (Fig. 6.). Hardness increased at the interface of about 80 HV 0.1 in the steel 316Ti, while in P92 steel hardness decreased by about 30 HV 0.1.

These results correspond with diffusion redistribution of carbon across the weld interface in direction from martensitic to austenitic steel. This is consistent with computed difference of thermodynamic activity between both basic materials (Fig. 7.).

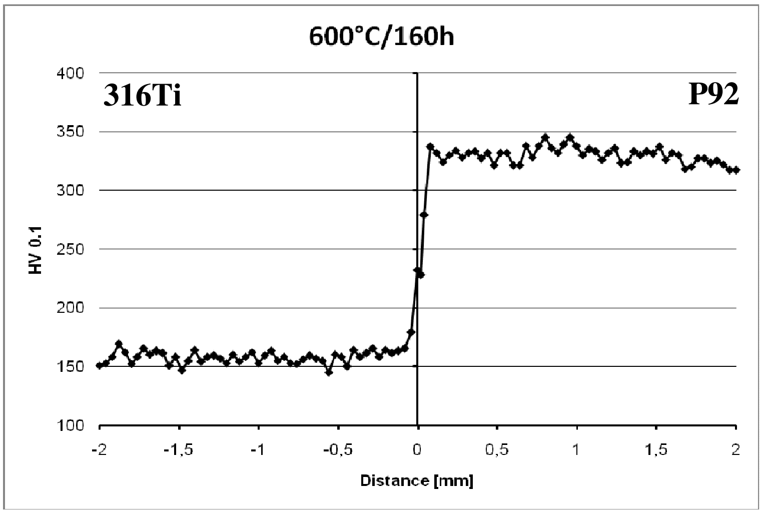

Fig. 5. Microhardnes values profile across the weld interface measured after annealing at 600°C/160h.

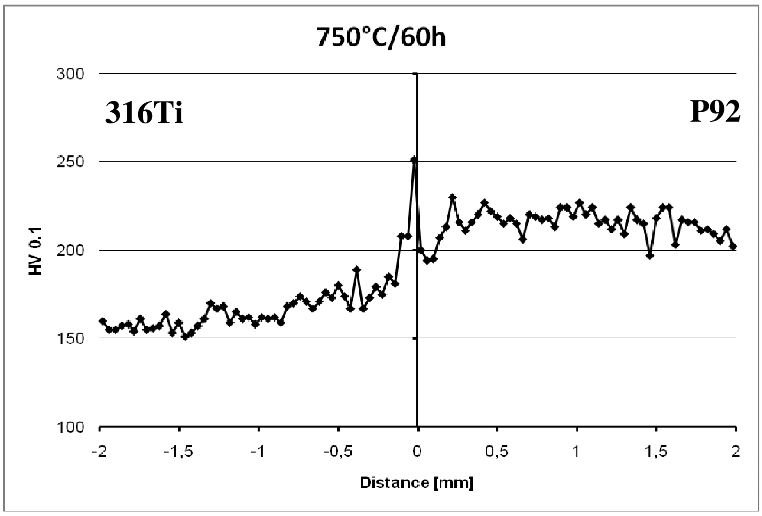

Fig. 6. Microhardnes values profile across the weld interface measured after annealing at 750°C/60h.

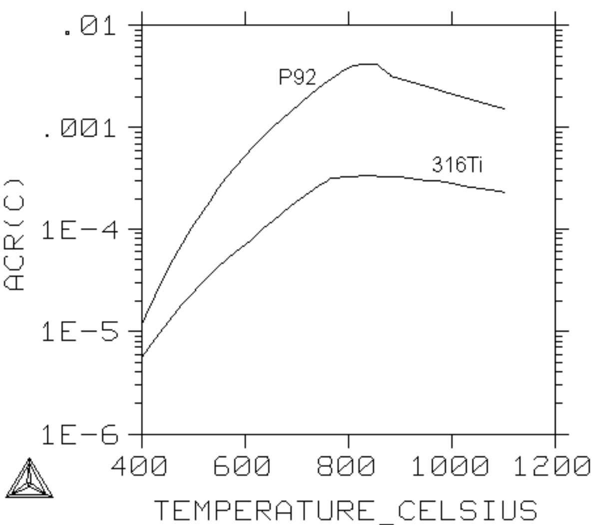

Fig. 7. Temperature dependence of activity for used steels (ThermoCalc calculation)

4. Discussion

The results show the instability of the studied weld joint in the range of examined temperatures. Welding of both materials had a negative impact in formation of ferrite in steel P92 caused by the redistribution of carbon across the weld interface. The absence of the ferrite on interface after annealing at a temperature 600 °C was probably caused by intensive precipitation at this temperature. By intensive precipitation in the ferrite the microstructure was changed to the microstructure visually similar to fine tempered martensite or bainite. Temperature of 1050 °C is in fact austenitization temperature for steel P92, at this temperature occurs transformation of ferrite produced during welding to austenite during homogenization. On the other hand, this temperature has no significant influence on the microstructure of steel 316Ti. Due to these facts can be expected a positive influence of post weld heat treatment on the microstructure and properties of evaluated weld joint.

Up to temperatures around 650 °C the diffusion effects influence the weld joint in terms of mechanical properties rather positively. Up to these temperatures there is a continuous change of mechanical properties in the weld interface area. As critical for the investigated joint a temperature of 750 °C was show. This temperature is practically the same as tempering temperature of steel P92, so there is a significant drop in hardness of the base material in the weld interface area, in addition, due to decarburization is there also significant recrystallization of matrix and the dissolution of carbides. Because this temperature is well below the solubility temperature of Laves phase, which is steel 316Ti about 800 °C, a rapid precipitation of intermetallic Laves phase in steel is occurred at this temperature. This phenomenon is near the interface also accelerated due to the redistribution of alloying elements. These effects result in significant step change of the hardness of the investigated weld joint in the weld interface area at temperatures around 750 °C.

5. Conclusions

The dissimilar weld joint P92/316Ti is unstable at temperature 500 – 1050 °C from microstructural point of view. During their high-temperature long-term exposition, carbon diffuses across the weld interface, which is reflected by changes of mechanical properties in the weld interface area. Due to the observed changes in microstructure caused by diffusion processes taking place across the weld interface during welding and subsequent high temperature exposure, it is proving effective use of post weld heat treatment. Based on current results is the studied weld joint safely useful up to temperatures around 650 °C, in the case of short operating times up to 700 °C. Due to the expected use of the studied materials and the weld joint for long-term high-temperature applications, a further comprehensive research in terms of influence of the overall metallurgical quality on microstructural stability and creep properties of the studied materials and the weld joint is needed.

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References

[1] R. Viswanathan, W. J. Bakker: J. Mater. Eng. Perform.10 (2001) 81-95.

[2] Y. Yamamoto, M. Takeyama, Z.P. Lu, C.T. Liu, N.D. Evans, P.J. Maziasz, M.P. Brady: Intermetallics 16 (2008) 453-462.

[3] V. Vodárek: Fyzikální metalurgie modifikovaných (9-12)%Cr ocelí. (*Physical metallurgy of modified (9-12)% Cr steels)* VŠB – Technická Univerzita Ostrava, Ostrava 2003 *(in Czech)*

[4] H. K. Danielsen, J. Hald: Comp. Coupl. Phase Diagrams and Thermochem. 31 (2007) 505–514.

[5] N. Saunders, A. P. Miodownik: CALPHAD, Elsevier Science, Amsterdam 1998.

[6] ThermoCalc User`s Guide, Div. of Comput. Thermodynamics, Dept. of Mater Science and Engineering, Royal Inst. of Technology, Stockholm 1998

[7] A. Kroupa, J. Havránková, M. Coufalová, M. Svoboda, J. Vřešťál: J. Phase Equilib. 22 (2001) 312-323.