

## Effect of destabilization treatment on microstructure, hardness and abrasive wear of high chromium hardfacing

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### Abstract

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Hardfacing metals are widely used in arc welding and plasma transfer arc technologies and industries. However, application of hardfacing in agriculture is limited due to low toughness after weld depositing. This study was focused on destabilization of high chromium hardfacing metal. The hardfacing was destabilized at 900 and 1,000°C in the different treatment time intervals. Destabilization treatment showed precipitation of secondary carbides leading to partial transformation of austenite to martensite phase in the matrix. The results show that increasing destabilization temperature increased volume of carbide phase in austenitic matrix which affects abrasive wear resistance.

**Keywords:** weld deposit; heat treatment; carbides; precipitation; microhardness

RICHARDSON (1967) tested several materials for agricultural technology focusing on low wear cutting soil tools. He found that high chromium cast iron can be used as a wear resistant material for agricultural applications but the use of this material is very limited as a result of its low fracture toughness. FOLEY et al. (1988) also tested ceramic protected agriculture subsoilers and his results showed that low fracture toughness is a limiting factor for its use in high wear resistant material for agriculture industry. High chromium cast iron and hardfacing are used in mining and earth industry (HOU et al. 2006; HAO et al. 2010). Their wear resistance properties can be improved by adding Ti, W, V, Mo, Nb elements for the formation of MC type carbides into matrix high chromium cast iron (BEDOLLA-JACUINDE et al. 2005; WANG et al. 2005; CHUNG et al. 2009; KAZEMIPOUR et al. 2010; SABET et al. 2011). Second effect of the added carbide forming elements is a change

in phase into matrix, for example, Vanadium in the range of 1.5–3.7% wt gave a ferrite matrix and fine eutectic carbide structure in the hypoeutectic high chromium cast iron (WIENGMON et al. 2005). Niobium in the hardfacing leads to forming NbC which are randomly deployed in hypoeutectic or eutectic matrix. Their presence in the hardfacing positively influenced three-body (CORREA et al. 2007) or two-body (CHOTĚBORSKÝ et al. 2008) abrasion resistance. Also Tungsten and molybdenum are added to white cast iron or hardfacing material with the aim to increase wear resistance properties (WANG et al. 2011) using MC and  $M_6C$  type's hard carbides. Actually, several researchers used rare earth oxides to modify morphology of carbides in white cast iron and hardfacing metal. Studies (HOU et al. 2009; HAO et al. 2011; HOU 2012) have shown that rare earth oxides lead to formation of smaller primary  $M_7C_3$  carbides in the hardfacing.

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High chromium hardfacing has been proven to be an effective material for applications in aggressive environments where abrasion and erosion resistance are required. The high wear resistance of high chromium hardfacing is attributed to the combination of hard primary and/or eutectic carbides of  $M_7C_3$  (M – metal: iron, chromium and other strong carbide former elements) and a relatively ductile ferrous matrix (Ferritic, Pearlitic, Martensitic, Austenitic) (TABRETT, SARE 2000; LIN et al. 2010). The hardness of  $M_7C_3$  is in a range of 1,200 HV, which may vary with the composition (up to 1,700 HV). The ferrous matrix binds the hard carbides and provides the material with certain toughness vital for handling impact. Solidification of high chromium hardfacing begins with the formation of the primary phase, which may be austenite in hypoeutectic alloys or  $M_7C_3$  carbides in hypereutectic ones, followed by simultaneous precipitation of eutectic mixture of both phases. However, the formation of coarse or large primary  $M_7C_3$  carbides is not desired since they reduce the toughness of the material being crucial to its impact resistance (BUCHELY et al. 2005; BUYTOZ 2006). Results of DOĞAN et al. (1997) show that chromium content in the high chromium white iron directly affected wear resistance because chromium content changed a size and number of primary carbides with a changed phase of the matrix, too. But in the hardfacing of metal welding conditions and heat transferred are determinants showing size of chromium carbides formed in hardfacing. Abrasive wear of hardfacing is influenced by abrasive particle size (CHOTĚBORSKÝ et al. 2009) which has a different effect on hypoeutectic, eutectic and hypereutectic hardfacing structure.

Studies of POLAK et al. (2008) and BADISCH et al. (2009) show that abrasive wear resistance of metal matrix composites depends on mean diameter of hard phase and their inter-particle distance and phase volume. In the hypoeutectic high chromium hardfacing a large austenitic phase exists which is relatively soft and due to abrasion small areas transform to martensite. But hardness ratio between abrasive particle and austenitic phase is higher than hardness ratio between abrasive particles and eutectics. This leads to higher wear rate of soft phase and cracked eutectics. One possible way to increase hardness of the matrix is to transform austenite (gamma phase) to martensite by cryogenic heat treatment. This structure transformation will result in higher abrasive resistance (LIU et al. 2008).

Next is the use of tempering of hardfaced deposits at 400 up to 650°C which leads to secondary hardening and higher wear resistance (KARANTZALIS et al. 2008). It will be seen differently by adding alloy elements that are influenced by volume of austenite in the matrix (INTHIDECH et al. 2006). Destabilization heat treatment up to solidus curve which depends on chemical composition (LI et al. 2009) should be used for a relatively ductile high chromium hypoeutectic white cast iron and hardfacing metals. Destabilization heat treatment leads to the formation of secondary carbide phase into austenite dendrites and austenite into eutectic cell. Incipient secondary phase was created soon by critical temperature but volume secondary phases in austenite phase depend on destabilization temperature and time (KARANTZALIS et al. 2009; ALBERTIN et al. 2011). Results of WANG et al. (2006) have shown that precipitation of the secondary carbides starts at 580°C (subcritical heat treatment) as cubic  $M_{23}C_6$  carbide type. Differences are seen in the secondary carbide precipitation and transformation, depending on the thermal treatment conditions. At the 1,000°C destabilization heat treatment, the initial  $M_{23}C_6$  carbides transform to  $M_7C_3$ . Destabilization heat treatment positively influences fracture toughness of the high chromium cast iron (KOOTSOKOS, GATES 2008) but can negatively influence corrosion resistance (WIENGMON et al. 2011).

In this present study the high chromium hardfacing was produced by the gas metal arc welding (GMAW) technique from OK Tubrodur 14.70 electrode. The coatings were deposited on low carbon steel substrate. The aim of the work was to characterize destabilized treated hardfacing metal by their abrasive wear resistance.

## MATERIAL AND METHODS

In this experiment, hardfacing OK Tubrodur 14.70 from the ESAB (Vamberk, Czech Republic) was used. The commercial hardfacing electrode was applied onto S235JR steel plates (one weld bead; ArcelorMittal Ostrava a.s., Ostrava, Czech Republic). The nominal chemical composition of S235JR steel and electrode (OK Tubrodur 14.70; ESAB, Vamberk, Czech Republic) is shown in Tables 1 and 2, respectively. The deposition was carried out in flat position using ESAB Mini 2A welding machine (ESAB, Vamberk, Czech Republic). The samples were cut after deposition (25 mm width, 40 mm

Table 1. Chemical composition of base material (% wt); (producer datasheet)

C	Mn	S	P	Fe
0.074	0.33	0.006	0.002 5	Rest

Table 2. Chemical composition of electrode (% wt); (producer datasheet)

C	Cr	Mo	V	Si	Mn	Fe
3.5	22	3.5	0.4	0.4	0.9	rest

length) and they were grinded. Minimal width of the grinded surface was eleven millimetres.

The hardness of the hardfacing deposits was measured by the Vicker's method using a load of 294 N (HV30) which was repeated eleven times per sample. Optical microscopy (Zeiss Jenavert; Carl Zeiss, Jena, Germany) was used to analyse the microstructure of the specimens. The surface was grinded, polished and etched with picric acid (2% solution) before analysing. The hardness of the matrix phases was measured by the Vicker's method for microhardness using a load of 0.049 N (HV0.05) and was also repeated eleven times per sample. For structure analysis Beraha's reagent (1 ml HCl, 99.4 ml water, 1 g  $K_2S_2O_5$ ) was used as is shown in Fig. 1. The structure of hardfacing was austenite (grey area in Fig. 1) and chromium rich carbides (light area in Fig. 1) after deposition. Samples were controlled by Vicker's hardness test before heat treatment. Hardness value showed no significant difference. For the wear test, samples were further controlled by the ratio of austenite microstructure on each sample surface. The *T*-test result shows that means of hardness and ratio of austenite were similar at 95% level of significance.

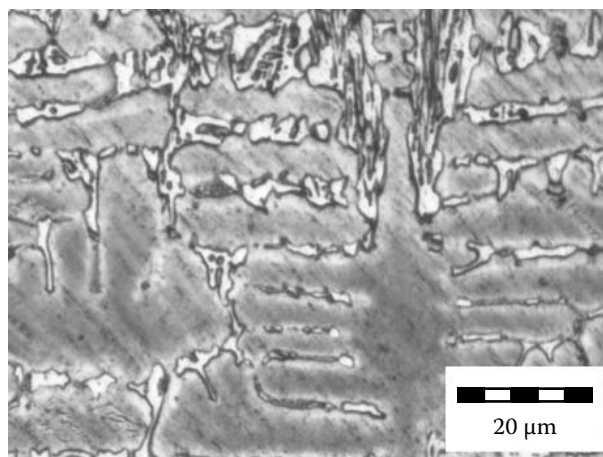


Fig. 1. Microstructure of the sample after weld deposited by Beraha's etching reagent

Samples for wear test were heat treated by a different thermal cycle as is shown in Table 3. Samples were heat treated in the furnace with an inert gas (argon).

Abrasive wear test (five times per samples) was carried out in a dry rubber wheel machine using sand particle 0.2–0.3 mm (CHOTĚBORSKÝ 2013). The load on specimen was 30 N, wear distance was 250 m. Diameter of rubber wheel was 130 mm and width was 10 mm. Before testing, all specimens were cleaned in ultrasonic bath and rinsed with warm air. The abrasive wear resistance was determined from the volume loss results, which was measured with 0.1 mg resolution.

The volume loss  $V$  ( $m^3$ ) was determined by mass loss using Eq. (1).

$$V = m/\rho \quad (1)$$

where:

$m$  – mass loss of material (kg)

$\rho$  – density of the tested material ( $kg/m^3$ )

Table 3. Hardness and austenite volume of the specimens after weld deposit and heat treatment

Sample No.	Hardness HV30*	Austenite (dendrites)**	Heating temperature/time at elevated temperature, cooled in air
1	557 ± 12	57.0 ± 1.3	900°C/0.5 h
2	573 ± 17	56.5 ± 1.9	900°C/2 h
3	585 ± 9	57.8 ± 2.2	900°C/8 h
4	592 ± 10	54.4 ± 1.4	1,000°C/0.5 h
5	584 ± 14	56.8 ± 2.3	1,000°C/2 h
6	575 ± 8	57.9 ± 1.5	1,000°C/8 h
7	562 ± 19	58.2 ± 2.0	–

\*mean values; \*\*mean (%)

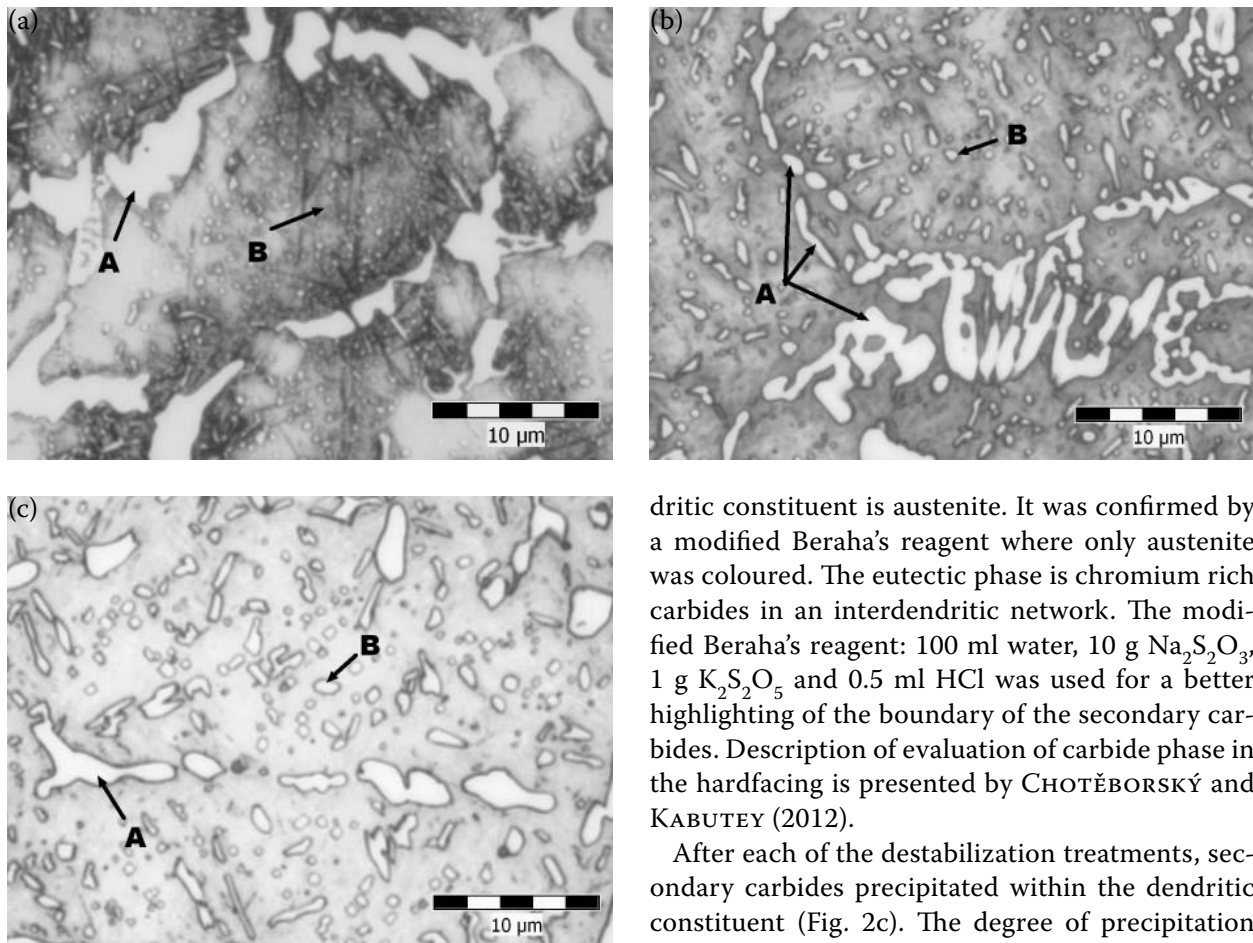


Fig. 2. Microstructure of the hardfacing destabilized at (a) 900°C/2 h, (b) 1,000°C/2 h and (c) 1,000°C/8 h (modified Beraha's reagent, immersion oil objective with blue optical filter)

A – eutectic carbide, B – secondary carbide in austenite dendrites

Density of hardfacing layer was  $7,580 \pm 21 \text{ kg/m}^3$ , it was determined according to hydrostatic method.

## RESULTS AND DISCUSSION

### Effect of heat treatment on microstructure of the hardfacing

Microstructure of the hardfacing was studied on samples No. 1–6 after destabilization heat treatment. Specimen No. 7 was used as reference specimen without heat treatment as a standard for wear test. Fig. 2 shows the microstructure of the deposit after heat treatment, 900°C/2 h, 1,000°C/2 h and 1,000°C/8 h. Given the composition and low hardness value of the deposited material, the den-

dritic constituent is austenite. It was confirmed by a modified Beraha's reagent where only austenite was coloured. The eutectic phase is chromium rich carbides in an interdendritic network. The modified Beraha's reagent: 100 ml water, 10 g  $\text{Na}_2\text{S}_2\text{O}_3$ , 1 g  $\text{K}_2\text{S}_2\text{O}_5$  and 0.5 ml HCl was used for a better highlighting of the boundary of the secondary carbides. Description of evaluation of carbide phase in the hardfacing is presented by CHOTĚBORSKÝ and KABUTEY (2012).

After each of the destabilization treatments, secondary carbides precipitated within the dendritic constituent (Fig. 2c). The degree of precipitation depends on the solubility of carbon and alloying elements (predominantly chromium) in the primary phase – austenite. Therefore the amount of precipitated carbide is a function of both alloy composition and destabilization temperature. For high chromium white irons in the as-cast state, the austenite dendrites are supersaturated in chromium, carbon and other alloying elements (LI et al. 2009). We assume that supersaturating in the hardfacing will be higher thanks to higher cooling rate from liquid to solid state. The higher temperatures of destabilization treatment of hardfacing alloys reduced the solubility of these elements in austenite more than in as-cast chromium white iron. A destabilized structure will therefore have precipitates of chromium carbides within its microstructure. However, as the destabilization temperature increases the solubility limits of chromium and carbon increase (KOOTSOKOS, GATE 2008). The result shows that as the destabilization temperature increases the amount of secondary carbide precipitation decreases. Destabilization at 1,000°C and longer time lead to destabilized eutectic carbides which changed their morphology from eutectic cell to carbide chains (Figs 2b and 2c). It seems

Table 4. Secondary carbide volume and size in the hardfacing after destabilization

Sample No.	Secondary carbide volume (%)	Size of secondary carbides ( $\mu\text{m}^2$ )		
	mean	median	low quartile	high quartile
1	8.1 ± 0.5	0.157	0.102	0.255
2	10.8 ± 1.9	0.158	0.102	0.264
3	12.4 ± 2.4	0.172	0.107	0.264
4	5 ± 0.9	0.212	0.132	0.286
5	4.6 ± 0.8	0.246	0.132	0.384
6	5.6 ± 1.2	0.331	0.188	0.418

that destabilization treatment has the same effect on eutectic carbides in hardfacing alloys as a destabilization of high boron steel, where the iron diborides are destabilized due to thermal input (XIANG, YANXIANG 2010; ZHANG et al. 2011).

Researchers WANG et al. (2006) and KOOT-SOOKOS and GATES (2008) have shown that for destabilization temperatures higher than 1,100°C, the precipitated carbides are of the  $M_7C_3$  form, for destabilization treatments lower than 1,100°C they showed  $M_7C_3$  carbides and  $M_{23}C_6$  carbides. Results of WANG et al. (2006) show that longer time of the destabilization treatments at temperatures lower than 1,100°C leads to transformation of fine  $M_{23}C_6$  cubic precipitates into  $M_7C_3$  rods, with an orientation relationship between  $M_{23}C_6$  and  $M_7C_3$ . The results of this present study show that the destabilization time significantly influenced the carbide size into austenitic matrix and the precipitation process caused a change of chemical content of austenite. This is seen on the transformation of the austenite to martensite around eutectics. This was observed at destabilization treatment at 900°C; destabilization at 1,000°C creates martensitic structure.

### Hardness and abrasive wear resistance

Destabilization heat treatment significantly influenced hardness of the hardfacing alloy as is shown in Tables 3–5. Microstructure analysis showed no change in eutectic cells and only secondary carbides could change hardness of hardfacing after destabilization heat treatment. Microhardness of austenite dendrites in hardfacing after weld deposit without destabilization treatment was  $725 \pm 55$  HV0.05 and destabilized treatment changed the microhardness of austenite due to precipitation of the secondary carbide. Results of microhardness are shown in Table 5; it can be seen that increasing the secondary carbide phase volume and size increased the hardness of hardfacing. Both variables (Table 6) have shown a significant effect of gamma phase (austenite/martensite) on microhardness. Regression analysis showed that precipitation of secondary carbides positively influenced the microhardness of gamma phase, but this effect decreased with increasing carbide size. Carbide size after destabilization treatment at 1,000°C depends on time and the microhardness of the gamma phase which decreased with

Table 5. Hardness, microhardness and volume loss destabilized hardfacing alloy

Sample No.	Hardness HV30*	Microhardness of gamma phase HV0.05	Volume loss ( $\text{mm}^3$ )
1	958 ± 28	1,414 ± 66	0.284 ± 0.075
2	960 ± 32	1,431 ± 42	0.281 ± 0.061
3	985 ± 17	1,451 ± 44	0.248 ± 0.055
4	925 ± 39	1,364 ± 62	0.286 ± 0.051
5	928 ± 31	1,344 ± 37	0.384 ± 0.061
6	949 ± 24	1,308 ± 54	0.418 ± 0.075
7	562 ± 19	725 ± 55	0.65 ± 0.120

\*mean values

Table 6. Results of regression analysis for microhardness and volume loss of the hardfacing alloy

Parameter	Coefficient	T-test	P-value
Intercept	1,420	82	< 0.001
Secondary carbide volume	8.6	8.3	0.004
Secondary carbide size	-480	-9.54	0.0025
Volume loss – regression analysis, adj. $R^2 = 0.79$ ; $F = 19.8$ ; $P = 0.011$			
Intercept	0.123	2.72	0.051
Secondary carbide size	0.911	4.46	0.011

microhardness of gamma phase – regression analysis;  $R^2 = 0.98$ ;  $F = 237$ ;  $P < 0.001$

increasing of the carbide size. Regression analysis of the carbide size and microhardness of the gamma phase as variables for destabilization treatment at 900°C showed no significant results. But results showed that secondary carbide volume leads to an increase of the gamma phase microhardness. In the microstructure of the samples destabilized at 900°C martensitic structure was observed, because martensite volume in gamma phase probably grew up with destabilization time and the secondary carbide volume increased. Results of the microhardness of the gamma phase at 900°C are not clear. While secondary carbide volume has a direct effect on microhardness, martensite phase has no significant effect. Some studies (INTHIDECH et al. 2006; KARANTZALIS et al. 2009) showed that the max. hardness value for high chromium cast iron was reached after destabilization treatment at 1,000°C but it can be seen that for hardfacing alloy the max. hardness was reached after destabilization at 900°C. This could be due to the fact that the precipitation kinetics of the second-

ary carbide phase could be affected by a different composition if the austenite phase was cooled with different cooling rate, usually higher than casting.

If the microstructure has an effect on microhardness of the gamma phase the same effect can be seen on abrasive wear resistance and wear loss in the test. Using a regression analysis only size of the secondary carbides (Table 6) was found as the significant variable which influences the wear loss. Therefore it can be seen that the results of wear loss (Fig. 3) show a generally decreasing trend if the secondary carbide size increases. Comparison of the samples No. 1, 2 and 4 in Table 5 shows that the volume loss was similar for both temperatures, although the sample No. 4 destabilized at 1,000°C showed a lower secondary carbide volume than samples No. 1 and 2 destabilized at 900°C.

## CONCLUSION

The present work investigated the effect of destabilization temperature and time on the microstructure, hardness and abrasive wear of high chromium hardfacing. Based on the results we can conclude that:

Destabilization heat treatment at temperatures 900°C and 1,000°C with longer time can lead to precipitation of secondary carbide particles. The austenitic matrix partially transformed to martensite at 900°C while at 1,000°C there was no transformation effect.

Destabilization heat treatment caused a change of the hardness of gamma phase in the hardfacing. Hardness was growing due to precipitation of the secondary carbide into austenite dendrites. Volume and size of the secondary carbide phase significantly influenced hardness of austenite dendrites.

Fine secondary carbide in austenite dendrites positively influenced the abrasive wear resistance

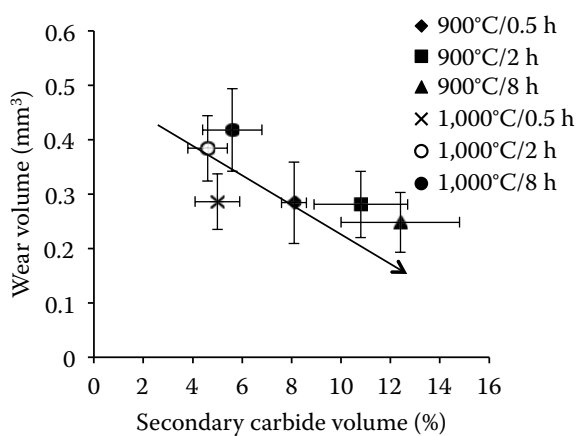


Fig. 3. Trend of the dependence between wear volume and secondary carbide volume

of the hardfacing after destabilization treatment. Effect of the secondary carbide volume was not established but the trend of this effect was obvious.

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