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Understanding the dissolution mechanism of fused tungsten carbides in *Ni*-based alloys: An experimental approach

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ABSTRACT

Ni-based alloys reinforced with fused tungsten carbides (FTCs) become more and more important as hardfacing material in the mining and oil drilling industry. The reason for the high abrasive wear resistance of FTCs lies in the microstructure, characterized by an eutectoid morphology of mono-tungsten carbide (*WC*) and di-tungsten carbide (W_2C). The benefit is a combination of high hardness and fracture toughness. However, FTCs are thermally instable and dissolute in case of high thermal loads during welding. An indicator for the dissolution of FTCs is the formation of a degradation seam along the interface to the surrounding *Ni*-based alloy. However, the formation mechanism of the degradation seam has not been fully understood and is diversely discussed. By means of an experimental approach, the formation of the degradation seam was reproduced experimentally and proven for the first time. The degradation is initiated by the selective dissolution of the phase W_2C. Due to *W* and *C* dissolved in the melt, *WC* precipitates in the *Ni*-based alloy. Fused tungsten carbides act as nucleus and promote the precipitation of *WC*. The precipitation of *WC* leads to the formation of the typical degradation seam.

1. Introduction

To improve the wear resistance of tools for mining and oil drilling applications, *Ni*-based alloys reinforced with fused tungsten carbides (FTCs) gain in importance as hardfacing material. The reason for the high abrasive wear resistance of FTCs lies in the microstructure, characterized by an eutectoid morphology of mono-tungsten carbide (*WC*) and di-tungsten carbide (W_2C) [1]. The benefit is a combination of high hardness and fracture toughness, exceeding the tribological properties of other commercial available hard phases [2]. However, high cooling rates are necessary during the manufacturing process in order to stabilise the high temperature phase W_2C below ~1200 °C and to ensure the presence of both phases at room temperature [1,3,4]. Due to the manufacturing process, FTCs are thermally instable and dissolute in case of high thermal loads during welding. Consequently, the hard phase content and wear resistance decrease [5]. Dissolution of FTCs in *Ni*-based alloys is clearly indicated by a degradation seam along the interface FTC-matrix [5,6]. However, the formation of the degradation seam has not yet been reliably understood and experimentally proven. According to *Schreiber* et al. the dissolution of FTCs is based on the diffusion of *C*, leading to a phase transformation from W_2C to *WC* [1]. *Choi* et al. postulates a selective dissolution of the instable phase W_2C [7]. According to *Vespa* et al. the degradation of FTCs is not based on the selective dissolution of W_2C [8]. Consequently, both phases W_2C and *WC* dissolve simultaneously due to a comparable ΔG [9]. As a result of dissolved *W* and *C* in the melt, [7,8] detected the precipitation of *WC* in the solidified *Ni*-based matrix and assumed the solidification of *WC* along the FTC interface to be the reason for the degradation seam.

The above mentioned assumptions concerning the formation of the degradation seam have never been proven. A novel approach was developed to understand and reproduce the degradation mechanism experimentally.

2. Methodology and materials

The experimental approach is based on gas metal arc welding (GMAW). As it is known that FTCs already start to dilute uncontrolled during the droplet transfer [5,7,8], GMAW does not enable a stepwise reconstruction of the degradation mechanism. In addition, the droplet transfer during GMAW influences the insertion of FTCs in the melt pool due to the non-wetting effect [10,11]. These features were solved by inserting an additional, resistance heated filler wire (hot wire) in the melt bead, designated in the scientific literature as hot wire assisted GMAW [10,12].

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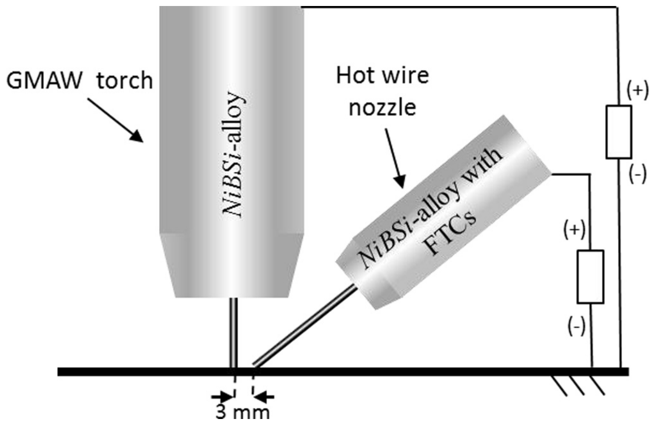

Fig. 1. Schematic drawing of the experimental set-up.

The experimental set-up is illustrated in Fig. 1. Only the hot wire was used to insert the FTCs in the melt, whereby the filler wire comprised the elements of the *Ni-B-Si* alloy, *C* (0.4 wt%), *Si* (1.2 wt%), *B* (2.05 wt%), *Ni* (40.85 wt%), and FTCs (55.50 wt%). The element composition of the consumable GMAW electrode was comparable with the *Ni-B-Si* matrix (without FTCs).

The experimental methodology is based on the assumption that the amount of *W* and *C* dissolved in the melt can be controlled by the amount of inserted FTC. Due to the restricted solubility of both elements in the solidified *Ni*-based alloy, the amount and size of *W*-rich precipitates can be controlled precisely. The amount of FTCs in the matrix was varied in dependence on the wire feed rates of the consumable GMAW electrode v_{GMAW} and the hot wire v_{HW}. Based on a total wire feed rate $v_{GMAW+HW}$ of 10 m/min, v_{HW} was systematically increased from 1 to 5 m/min.

The experiments were performed with a GMAW welding power source (Alpha Q 552). As the hot wire power supply, a Triton 260 TGD was used for resistance heating. The welding speed v_S was continuously set to 300 mm/min. *Ar* mixed with 30% *He* was used as the protective gas. Unalloyed steel S355 J2+N with the dimensions of 5 × 50 × 150 mm was the substrate for a defined weld seam length of 120 mm.

Cross-sections were prepared for metallographic analysis. Etching of the FTCs was performed with Murakami [13]. Metallurgical investigations were performed by means of light microscopy, scanning electron microscopy (SEM) and energy-dispersive X-ray analysis (EDX). The global element composition of the melt was determined with X-ray fluorescence (XRF) at a spot diameter of 1.0. These measurements were only performed in areas of the cross-section, where precipitations due to degradation were apparent. The elements *Ni*, *W*, and *Fe* were considered, whereby the main purpose was to estimate the solubility limit of *W* in the *Ni*-based alloy.

3. Results and discussion

Fig. 2 shows the typical appearance of FTCs embedded in the *Ni*-based alloy. The high thermal impact during GMAW leads to the dissolution of the hard phases. Typical indicators are the formation of the degradation seam and precipitates in the matrix material, see Fig. 2a. From literature it is known that the degradation seam and precipitates are *WC* [6,7,8]. A detailed observation of the degradation seam reveals a porous morphology, which is designated as the "HALO"-region [7], see Fig. 2a. Hereby, EDX mappings show an enrichment of *Ni* inside the porous morphology, see Fig. 2b. Up to now, the formation of the degradation seam is not fully understood. Based on hot wire assisted GMAW, the developed approach for reproducing the dissolution mechanism is presented in the following. The rough topography of the FTCs in Figs. 3 and 4 is due to the influence of Murakami etching solution. W_2C was

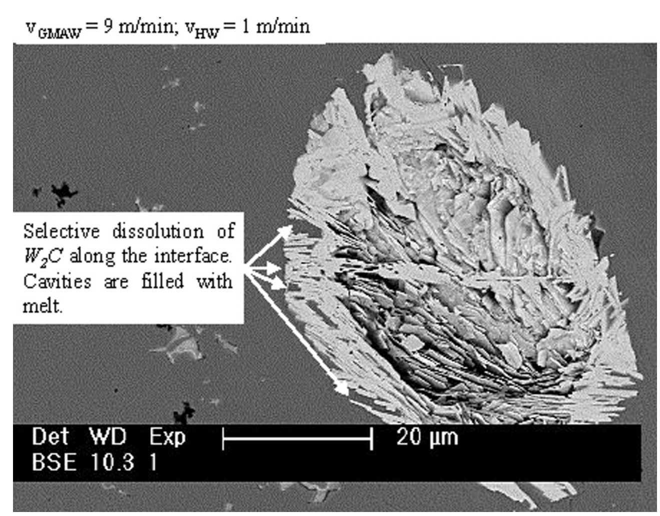

Fig. 3. Selective dissolution of W_2C, initiating the degradation of FTC.

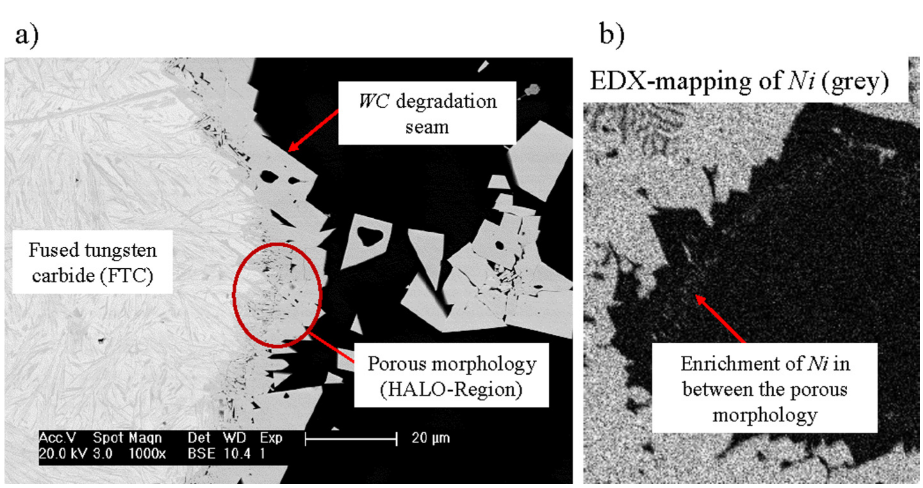

Fig. 2. EDX mapping of the elements *W* and *Ni*.