

The eta phases and mechanical properties of TIG welded joints of WC-Co cemented carbide and steel

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Abstract The tungsten-inert-gas (TIG) arc welding experiments of cemented carbide YG30 and steel 45 were carried out using the Ni-Fe-C filling alloys. The eta phases and mechanical properties of welded joints were analyzed by means of scanning electronic microscope (SEM), transmission electronic microscope (TEM) coupled with selected diffraction, electronic probe microanalysis and bending strength methods. The experimental results show that the chemical composition of the filling alloys affects eta phase formation. When the carbon and nickel contents in filling alloys are 0.61 wt% and 55.29 wt%, respectively, no eta phases form. And the joint bending strength is the highest to 1.352 GPa. But if they are 0.01 wt% and 55.38 wt%, the eta phases are formed at the boundaries of the cemented carbide and the weld, and the thickness of eta phase layer is about 110 micrometers. And the joint bending strength is low. Usually, these eta phases are anomalously granular, and easy to accumulate at the boundaries between cemented carbides and the weld. They are multiple M_6C rich in tungsten and iron.

Key words eta phase, TIG welding, cemented carbide, bending strength

0 Introduction

The demands for various kinds of tools made in WC-Co cemented carbides are increasing because of their excellent properties in hardness, wear-resistance, toughness, corrosion resistance and high heat durability. However, the fields of application are limited because of their high cost and difficulty in their large-scale products with complicated shape. Some of processes are used to joint cemented carbides and steels as the effective methods. Brazing, usually adopted at present, has low adhering strength and suitable for lower using temperature than 400°C^[1]. Diffusion welding requires expensive production facilities and has some limitations both in size and in shape for products. Meanwhile, the detrimental multiple carbides, called eta (η , hereafter) phases, to the bending strength of the welded joints are easily formed at the interface between cemented carbides and steels^[2,3]. The η phases here were normally pointed as the multicarbides with M_6C , $M_{12}C$ and other types, including Co_3W_3C ,

Co_2W_4C , Co_6W_6C , Co_3W_6C-F , $Co_3W_6C-104F$, $Co_3W_9C_4$, Co_2W_8C , $Co_3W_{10}C$, etc. in WC-Co cemented carbides^[4-6]. Co_3W_3C and Fe_3W_3C formed in sintering were called as η phases in author's preliminary experiments^[2]. When diffusion welding is used to weld the cemented carbide and steel, the carbon atoms in the cemented carbides would be thought to diffuse to the steel, which would result in that the carbon content decreased obviously and the η phases were formed at the cemented carbide side of the interface because the carbon range in the normal phase (WC + γ) zone in the phase diagrams of the WC-Co alloy and the WC-Fe alloy was all very narrow^[2].

Comparing with diffusion welding, the tungsten-inert-gas (TIG) arc welding has many advantages. For example, it is suitable for various kinds of positions of welding, apt to realize the automation, high in welding speed, and low in production cost, etc. Moreover, by changing the

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chemical compositions of the filling alloys, the formation of η phases could be restrained through decreasing the carbon diffusion. But the TIG welding of cemented carbide and steel has been little researched, and the forming mechanism of η phases in welding and their influence on the properties are not clear.

In this article, the formation and microstructure of the η phases and the mechanical properties of the joints were investigated by adjusting the chemical composition of the filling alloys.

1 Experimental materials and methods

Cemented carbide YG30 (WC-30wt% Co) and carbon steel 45 (0.45wt% C) were used in the experiments, and four kinds of Ni-Fe-C alloy as filling alloys in diameter 2 mm were used, which their essential compositions are shown in Table 1.

Table 1 Chemical composition of filling alloys (wt%)

No.	Ni	C	Fe
1	55.38	—	39.97
2	55.29	0.61	39.45
3	34.70	0.62	60.00
4	73.66	0.61	21.15

Cemented carbide YG30 ($\phi 50$ mm $\times$ 6 mm) and steel 45 ($\phi 50$ mm $\times$ 6 mm) were cut in semicircular form for one side welding of butt joints, respectively, and single V groove of slope in 30 degree was cut along the direct side. TIG welding for one-sided welding of butt joints in three layers was carried through with every filling alloy. The welding parameters are welding current of 90 A, pulse frequency of 80 Hz, and the welding speed of 40 mm/min.

The cross section samples of welded bead were cut vertically to prepare metallurgical analysis. The etchant with 20% KOH + 20% $K_3[Fe(CN)_6]$ + 80% H_2O composition was used to reveal η phases and WC. By means of JSM-6360LV scanning electron microscope (SEM), the microstructure, η phases, and the morphology of fracture surface were analyzed. Element distribution near the inter-

face between cemented carbides and welds, and the composition of carbides were determined using JXA-8800R electron probe micro-analyzer. Transmission electronic microscope (TEM) observations and analyses were carried out on a Hitachi-800 with operating voltage of 200 kV.

The four-point bending tests were carried out to measure the bending strength (σ_{bb}) of the welded joints in the way of Japanese Standard JISR1601-1996.

2 Experimental results and discussions

2.1 Microstructure of the welded joints

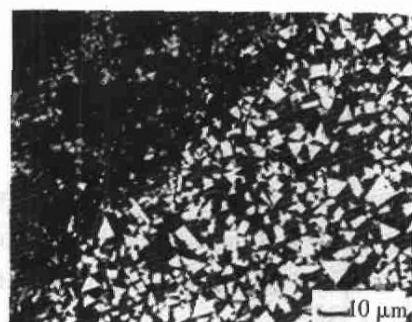
It was found that all the welded joints are well combined metallurgically and there are not welding defects at the interface in optical microscope. Fig. 1 shows the macro image of sample No. 1. In the microstructure of sample No. 1 shown in Fig. 2a, η phases (the black) are observed both at the heat-affected-zone (HAZ) of YG30 side and at the interface of YG30 and weld bead. The thickness of η phase layer is decreased gradually from the bottom to the top of the whole slope, and the biggest thickness of layer is about 110 μ m. Most η phases are anomalously granular and accumulated at the boundaries of cemented carbides and the weld. Some of fine η phases are formed both at the interface and in the HAZ of the YG30. In the most of bigger η phases and a few of tiny η phases there are obvious grain boundaries and WC (white). These distribution and morphology of η phases suggested that many η phases be formed by assembling some tiny η phases under the stirring condition in welding, and the small η phases which has similar size with WC in cemented carbides are transformed from WC by atomic diffusion.



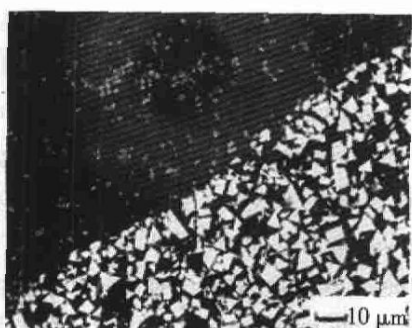
Fig. 1 Image of welded joint

When the carbon content is 0.61 wt % and other elements content in the filling alloy are constant (sample No. 2), no η phases are formed, as shown in Fig. 2b.

When the carbon content is same as sample No. 2, but Ni content decreases from 55.29 wt% to 34.70 wt% (sample No. 3), there are only big η phases formed at the interface. Sample No. 4 also contains the same carbon content with sample No. 2, while Ni content is increased from 55.29 wt% to 73.66wt%, no η phase is observed.



(a) No.1



(b) No.2

Fig. 2 SEM backscattered electron morphologies of the YG30/weld boundaries

2.2 Element redistribution near the interface

In the thermal cycle of welding process, the chemical element distribution in weld would change significantly whether in the liquid state or in the solid state. Because of the difference of chemical composition of filling alloys and YG30, the Fe and Ni atoms would diffuse from filling alloys to YG30, and C, Co and W atoms in YG30 would diffuse from YG30 to weld at same time. Especially, this redistribution of some elements in melted alloys would be promoted by the stirring action in the molten pool. The elemental distribution would affect the microstructure and properties of the welded joints apparently. By means of electron probe microanalysis (EPMA), the element distribution near the interface of YG30 and weld had been ana-

lyzed in the range of 500 μm . The thickness of W, C and Co diffusing layer is about 70 ~ 100 μm , 100 μm , and 140 ~ 170 μm , respectively. Meanwhile, obvious diffusion of the element Ni and Fe from the weld to YG30 was observed at these diffusion layers.

From the analysis in sample No. 1 and No. 2 shown in Fig. 3, it was found that, in the carbides of both sample No. 1 and No. 2, the content of W, C, Co far away from the interface were almost same. Moreover, the C content (51at%) is a bit higher than W (46 at%), and Co content in the WC is about 3at% ~ 4at%. Comparing with WC far away from the interface, the composition of η phase in sample No. 1 changes obviously. The content of W and C in η phase decreases while that of Fe, Co, Ni increases substantially. The decrease of C content is most remarkable, from 51at% to 13at%. Besides, W content reduces from 46at% to 41at%, and Fe, Co, Ni content increases to 28at%, 10at%, 8at%, respectively. Generally, the η phase of M_6C is the type of $\text{Co}_3\text{W}_3\text{C}$ and $\text{Fe}_3\text{W}_3\text{C}^{[2,5]}$. In theory, the carbon content of η phase in M_6C is 14at% and it is close to that of the η phase formed in this experiment, this suggests that the η phase formed in the experiment be multicarbides of M_6C rich in tungsten and iron. At the interface of sample No. 2 there is not η phase formed, but WC at the interface contains almost same content of Fe, Ni, about 2at% as sample No. 1, and the content of W, C decrease a little. These results indicate that when the content of Fe and Ni in the weld are suitable, η phase formation depend on the carbon content

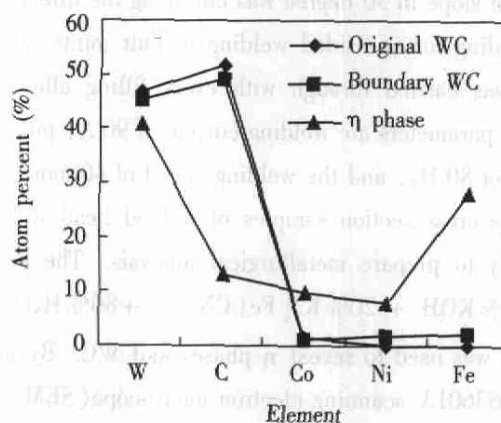


Fig. 3 Results of the composition analysis of carbides

of the weld.

2.3 Formation mechanism of η phases

According to the composition analysis, it might be thought that in the welding process Fe atoms in the weld diffuse to the base material YG30 over the interface between the cemented carbide and the weld. At same time some WC solutionized in Co phase. Therefore, it is possible to form Fe-W-C-Co system in some areas in Co phase. If the activity of carbon atom in Co phase were higher than that in the weld, C atoms would diffuse from YG30 to the weld. As a result, the C content in Co phase would decrease and WC would continuously decompose to complement the loss of C in Co phases. In the equilibrium diagram of system Co-W-C and Fe-W-C, the range of the normal phase ($WC + \gamma$) zone in system Fe-W-C is very narrow, so η phase is very easy to form when the system lost C slightly. From the analysis results of the element diffusing at the boundary between YG30 and the weld by means of EPMA, it was found that C atoms had diffused from YG30 to weld in all four samples (from No. 1 to No. 4), but η phase was formed only in sample No. 1 with lower C content (0.01 wt%) and No. 3 with lower Ni content (34.70 wt%). Although C had diffused from YG30 to weld in all four samples, they were in different cases. If the difference of C content in Co phase and in weld were small, the diffusion speed of C atoms would be slow and C content decrease little in YG30, therefore, WC would not transform to η phase. For example, in Fig. 3 the composition of WC at the boundary of sample No. 2 was shown, and C content decrease was not resulted in η phase formation. These indicate that C content in the weld is the main factor of the η phase formation. When Ni content is 55.29 wt% and C content is higher than 0.6 wt% in the filling alloy, η phase formation could be restrained. On the other hand, when C content is 0.6 wt%, η phase is easy to form if Ni content is too low, such as sample No. 3. In the system of Ni-W-C formed at the interface of the weld and YG30, the range of C content in the normal phase zone is wider than that in system Fe-W-C or Co-W-C according to the equilibrium diagram of system Ni-W-C^[7]. Therefore, C loss is not sensitive to η phase formation in system Ni-W-C.

2.4 Bending strength and fracture surface of the welded joints

2.4.1 Bending strength of the welded joint

Fig. 4 shows the result of the four-point bending test. The bending strength of the specimens in which no η phase exists are higher than that η phase does, they are 1.352 GPa and 0.96 GPa for sample No. 2 and No. 1, respectively.

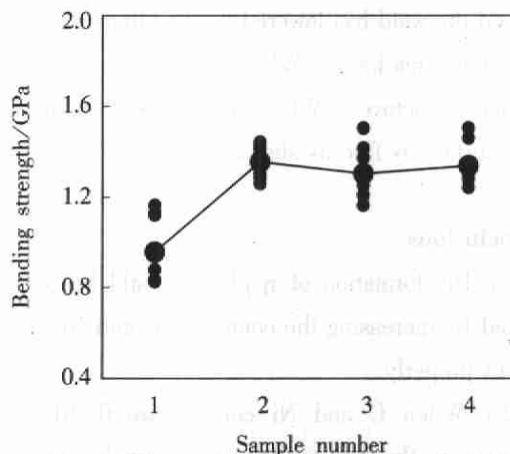
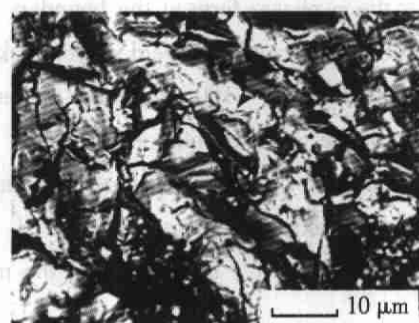
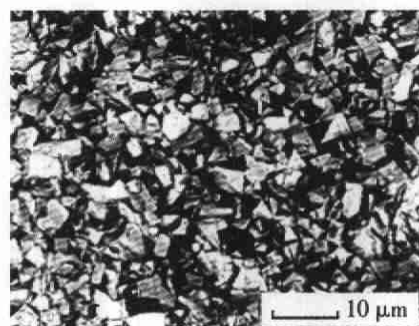


Fig. 4 Bending strength of welded joints



(a) Big granular η phases



(b) Cemented carbide YG30

Fig. 5 SEM backscattered electron images of fracture surface

2.4.2 Fracture surface analysis

Three types of the ruptures were observed, which are (1) rupturing along the interface between YG30 and weld, (2) originating from the interfaces and then spreading to YG30, and (3) destroying in weld.

SEM backscattered electron images of fracture surface are shown in Fig. 5. Coarse η phases are seen in the fracture surface of sample No. 1 in Fig. 5a. The existence of coarsely granular η phases at the boundaries between YG30 and the weld has lowered the bending strength of the welded joint remarkably. When the crack is in the YG30, which microstructure is WC and γ -Co solid solutions, the fracture surface is fine as showed in Fig. 5b.

3 Conclusions

(1) The formation of η phases could be reduced or restrained by increasing the content of C and Ni in the filling alloys properly.

(2) When C and Ni content are 0.61wt% and 55.29 wt% in the filling alloy, respectively, no η phase is observed. The bending strength of the welded joint is the highest, 1.352 GPa. When C is free and Ni content is 55.38wt%, the η phases form at the boundary of the cemented carbide YG30 and the weld. The thickness of η phases layer is about 70 ~ 100 μm which makes the bending strength of the welded joints decrease.

(3) If coarse η phases exist at the boundaries of the YG30 and the weld, the rupture occurred along the boundary at bending test, and these are the main factor

that decreases the bending strength of the welded joint seriously.

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