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Microstructural development of dissimilar weldments: case of MIG welding of Cu with Fe filler

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Microstructure development during MIG welding of copper with iron filler has been studied to gain insight to the process of dissimilar welding. The microstructure of the iron rich bids consist of martensitic bcc iron with cellular network of fcc copper. The scale of network depends on the processing conditions. However, the average composition remains fairly uniform with 20at% Cu excepting at the boundary regions of the bid and the copper plates. A characteristic banded structure could be observed in these regions whose width scales with traverse speed. Evidence of phase separated copper globule suggests access to the submerged miscibility gap and significant undercooling of the melt during welding.

1. Introduction

Development of dissimilar weldments represents major challenge in modern manufacturing processes. One of the main reasons for the poor progress in this area is the relative lack of basic understanding of the process. In particular, very little is known about the weldments both in terms of heat transfer, fluid flow and the microstructure development. All these are crucial in developing sound dissimilar weldments in future. We have embarked on a major programme of understanding dissimilar weldments from a basic point of view. Some of the recent results are given elsewhere primarily using laser as the heat source [1–6]. In order to understand the process that take place during MIG welding, work is initiated with model systems. In the present letter, we report some early work on a model system where two copper plates are welded with an iron filler rod. The choice of the system arises from two facts. Firstly, fusion welding of copper to iron, if successful, has great technological relevance. The more important reason for the choice of the system lies in the fact that they are immiscible in solid state while miscible in liquid state. However, there exist a liquid miscibility gap in undercooled temperature. Thus, heat transfer, fluid flow and mass transport will leave its imprint on the weld pool microstructure which will allow us to understand the conditions prevailed during the welding and subsequent microstructure development.

2. Experimental

Commercial pure copper blocks are MIG welded using the iron wire. The composition of the iron wire is 0.08% C, 0.9% Si, 1.45% Mn. The welding was done under different traverse speeds of the work piece. The input power was varied in addition to traverse speed to optimize the welding conditions. Table I shows the different power inputs and the traverse speed used in the present investigation. The microstructure development of the weldments was studied by optical and scanning electron microscopy. A JEOL JSM840 scanning microscope equipped with EDAX analyzer of OXFORD make was used for the electron microscope studies. The hardness profile of the weld pool was obtained using a Vickers microindentor tester.

3. Results

The x-ray diffraction patterns of the weldments indicate the presence of copper together with the bcc iron. Fig. 1 shows typical low magnification morphology of the weld bead. Fig. 1a shows a weld bead at a traverse speed of 325 mm/s while Fig. 1b shows the same with a traverse speed of 225 mm/s. In both the cases the energy input to the electrode was identical (7.7 kW). Fig. 2a

TABLE I Experimental parameters used in the present investigation

Set	Voltage	Current	Traverse speeds (mm/min)
1	30	250	270–350
2	32	235	225–325
3	28	275	225–325
4	32	200	225–325

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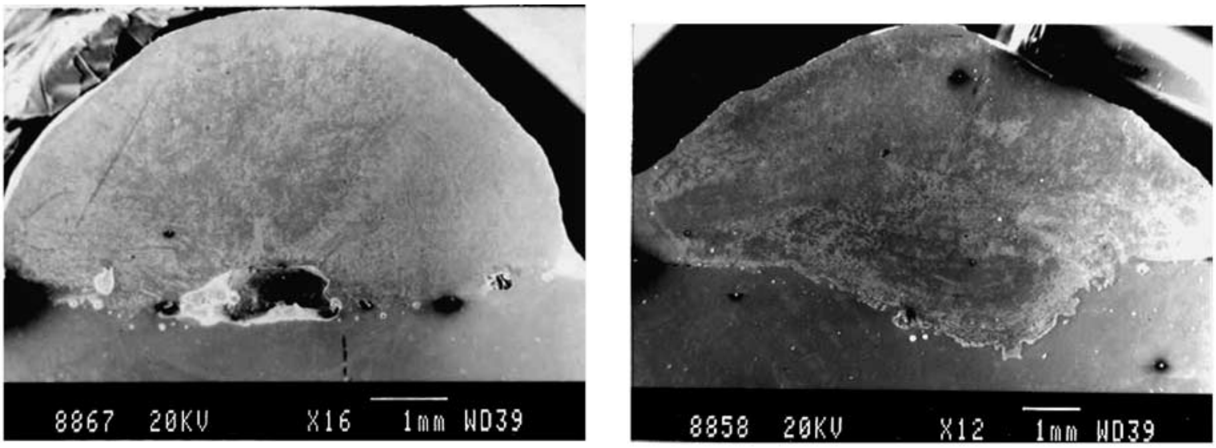
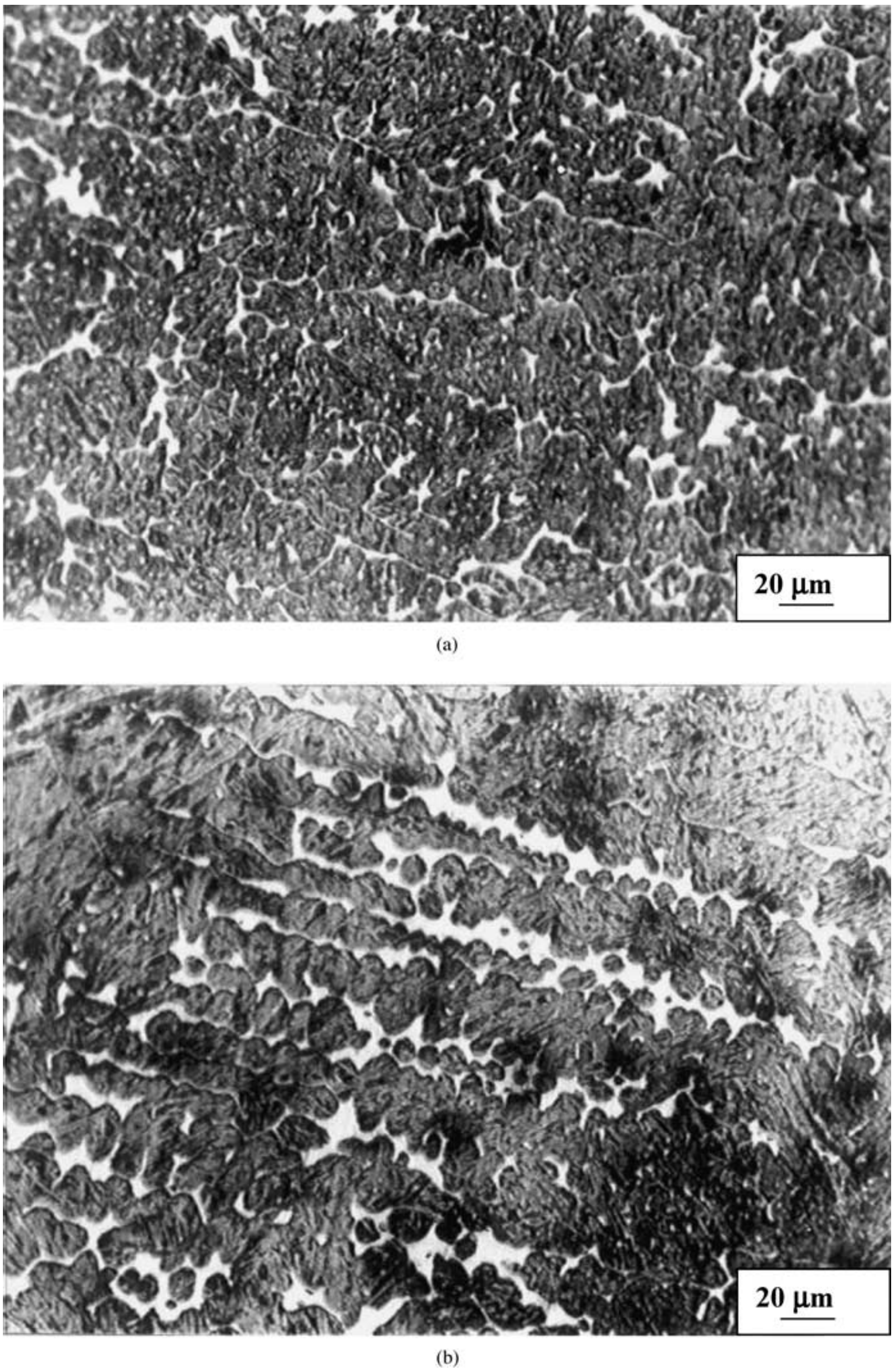


(a) (b)

Figure 1 Low magnification micrographs of weld bead at (a) low heat input (28 V, 275 A, 325 mm/min) and (b) high heat input (28 V, 275 A, 225 mm/min).



(a)

(b)

Figure 2 Microstructure revealing copper network in the weld processed at 32 V, 200 A, 225 mm/min.

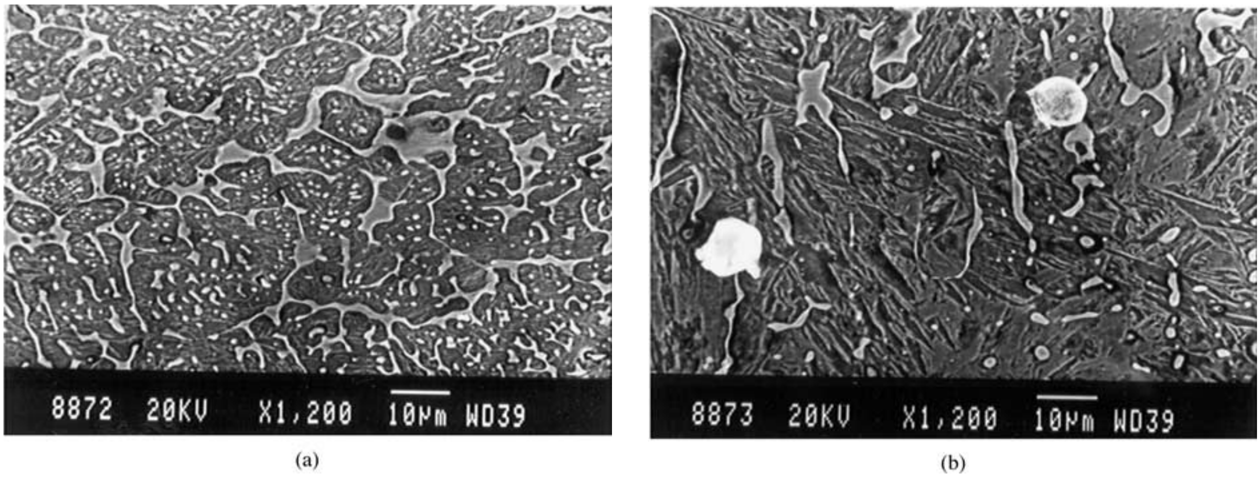


Figure 3 Microstructure showing copper globules in a martensite matrix for the weld processed at 30 V, 250 A, 270 mm/min.

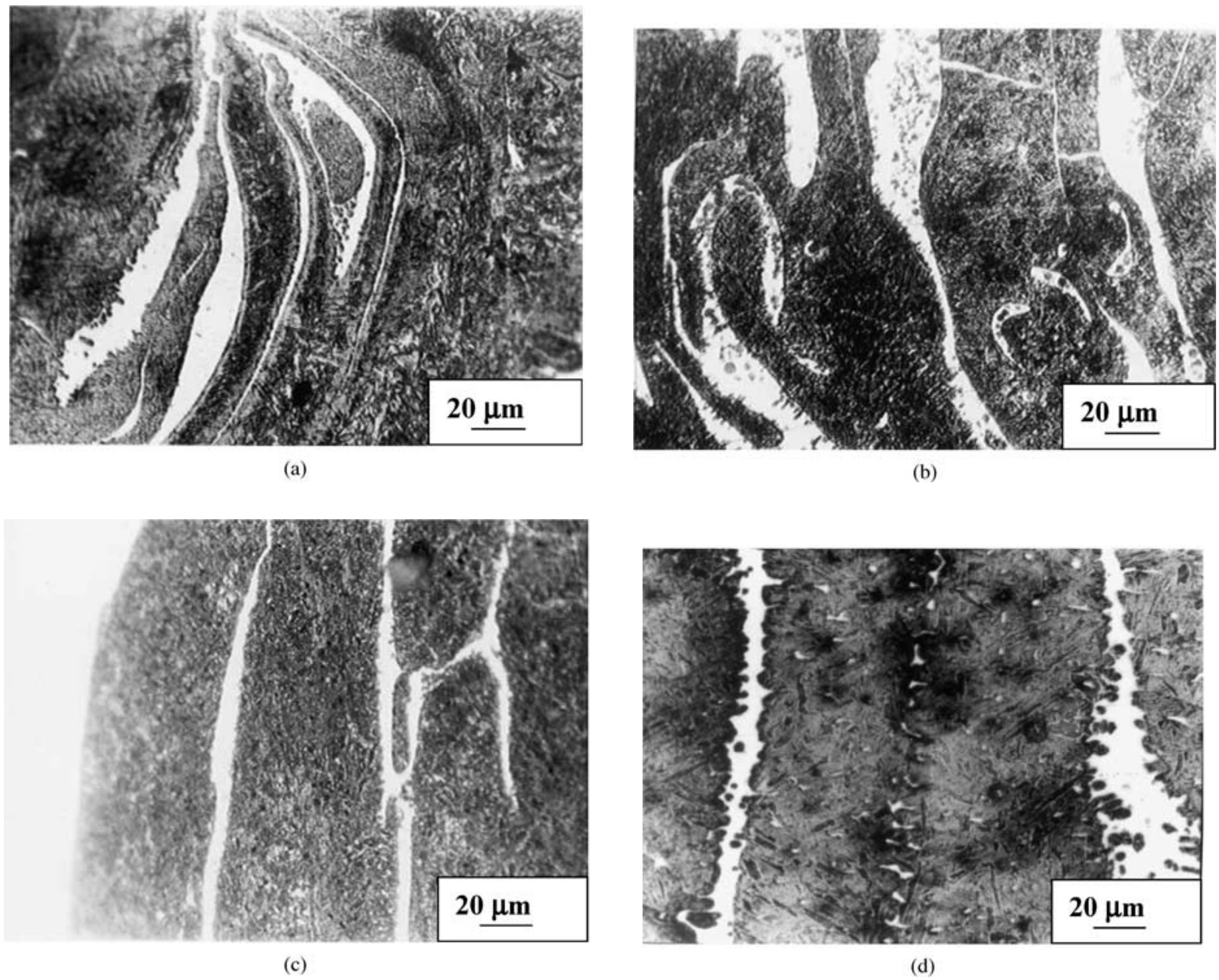


Figure 4 Banded morphology of weld bead-base metal interface at (a) 30 V, 215 A, 225 mm/min and (b) 28 V, 275 A, 325 mm/min (c) 32 V, 200 A, 225 mm/min, and (d) 34 V, 235 A, 325 mm/min.

shows the microstructure of the top region of the bead processed with the identical traverse speed but with slightly lower power to the electrode (6.4 kW). The copper network in the microstructure can be clearly seen. The microstructure of the weldment just below the top surface is shown in Fig. 2b. Besides the segregated network, one also notes that the cells are narrow and elongated, characteristic of dendrites at high growth rate. Microstructures of the samples also reveal the presence of the copper globules embedded in the iron matrix. This is clearer in the SEM micrograph of Fig. 3a and b. Careful observation also indicates the presence of needle shaped morphology of the iron matrix typical of martensite (Fig. 3b). The microstructure of copper-iron interface shows distinctly different morphology. These interfaces under two traverse speeds (power 6.45 kW) are shown in Fig. 4. One can clearly see a banded microstructure of copper and iron in the iron rich bead near the interface. The nature of the band is not sensitive to the input power. However, it is

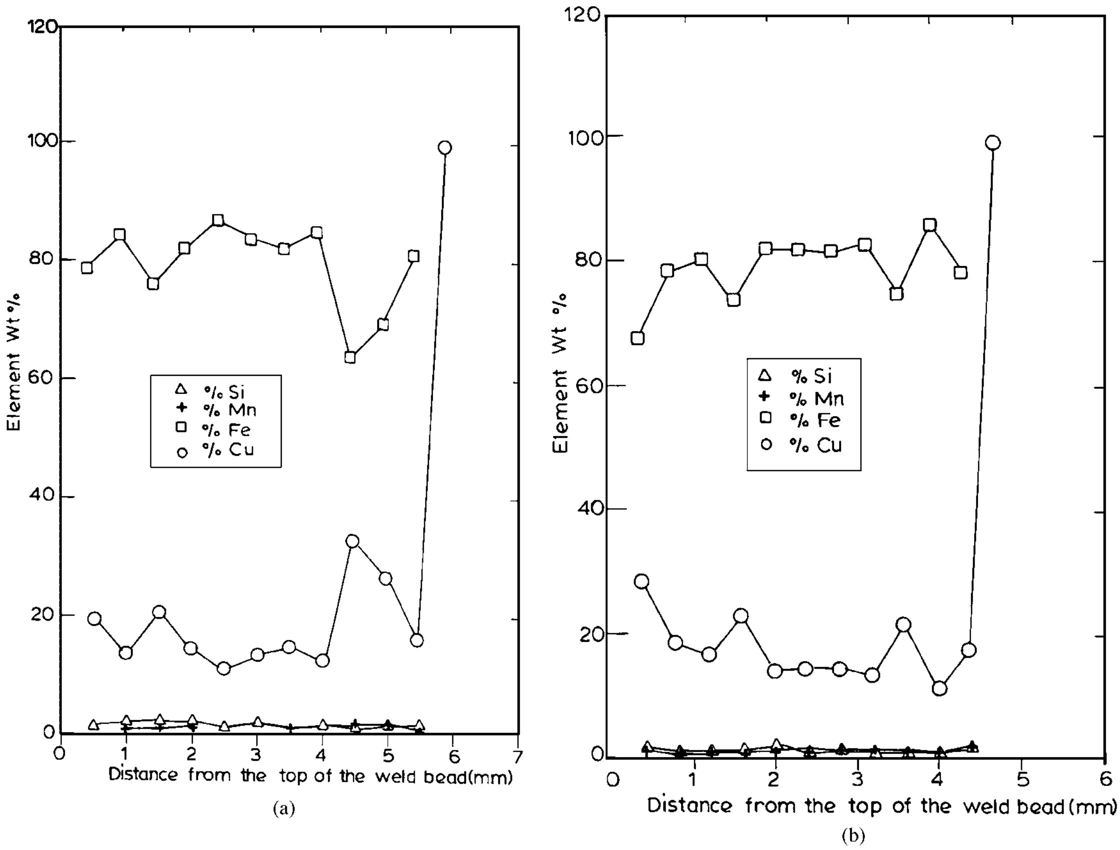


Figure 5 Composition profile along the depth of the weld bead at (a) 32 V, 235 A, 225 mm/min and (b) 34 V, 235 A, 325 mm/min.

influenced by the traverse speed. The bands are broader for the slower traverse speed (Fig. 4a and b) and narrower at higher traverse speed (Fig. 4c and d). Typical composition profiles from top to bottom of the central bead from samples processes with two scan speeds (power of 7.5 kW) are shown in Fig. 5. One can see that on an average the compositions are uniform except near the interface. Composition profiles obtained from samples processed under different intermediate processing conditions suggest that the composition does not change much and the profile remains fairly uniform in composition with an average of 20% copper. Hardness of bead is fairly high. Typical hardness profile from top of the bead up to the interface is shown for two scan speeds of 225 mm/min and 325 mm/min. (power 7.5 kW) in Fig. 6. From the profiles, it can be seen that in most of the cases the hardness is maximum slightly away from the interface and lower in the top of the bid. However, in most of the bid region it hovers around 300 to 350 VHN. The hardness profile is more uniform at higher scan speed than at lower scan speed. The small fluctuations that are observed in micro level near the interface are due to presence of soft copper zones. The maximum hardness depends on the processing conditions with the highest value observed being 430 VHN.

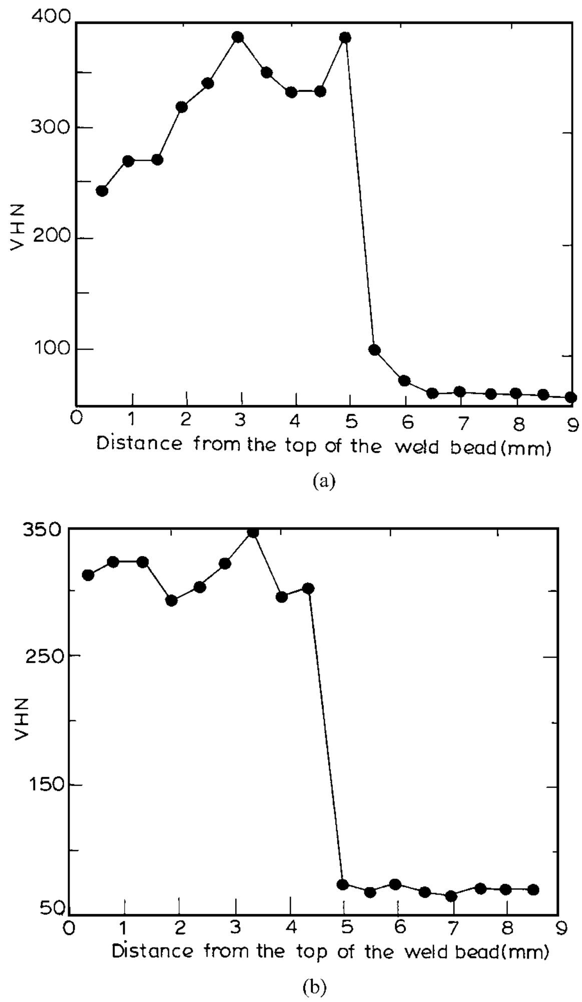


Figure 6 Hardness profile along the depth of the weld bead at (a) 32 V, 235 A, 225 mm/min and (b) 32 V, 235 A, 325 mm/min.

4. Discussion

One of the major problems in the dissimilar welding is the problem of mixing. The problem not only depends on the heat source and the melting point of the two constituent materials but also in the relative conductivity of heat. It is the latter which often determines the temperature profile and the mixing behaviour. This is clear from the present experiments. Although the melting point of the copper is much less compare to iron, our results clearly indicate that the amount of copper that has melted and mixed is much less. The microstructural analysis reveals several important results. First, the mixing is uniform and no significant composition gradient could be observed excepting near the interface. This indicates the presence of strong convection driven circulation of the melt. The mixing of iron and copper melt is primarily due to convection as can be evidenced from the banded layers near the interface. Clearly the circulation of iron melt near the two liquid interface drags the molten copper layer into the iron rich pool formed by the filler wire. Due to the very rapid heat transfer by the copper, the molten layer of copper is limited in extent and explains the roughly 20% composition under all the processing conditions utilized in this investigation. This result is very similar to our earlier results on the surface alloying of copper on iron [7]. However, in that case the difficulty of coupling of the laser beam with the copper was responsible for the observation of almost similar copper concentration on the iron surface under different processing condition. Since coupling is not an issue in the case of MIG welding, our observation could be explained in terms of the high heat transfer of copper, which makes the small difference in heat input insignificant in controlling the molten copper layer.

The microstructure exhibits a typical cellular structure away from the interface with copper network in the intercellular regions. A look at the phase diagram [8] clearly indicates such a microstructure will result due to the presence of wide freezing range. The earlier work on laser cladding with similar composition revealed similar feature. However, the observation of copper particulates in our microstructure requires different explanation. The Fe-Cu system exhibits submerged miscibility gap which can be accessed by undercooling the melt [9]. An earlier thermodynamic calculation by us reveals that the undercooling necessary to access this miscibility gap for Fe 20%Cu is roughly 180 K [10]. Such a high undercooling is generally achieved by rapid solidification. The observation of particulate copper, therefore, indicates a very high heat transfer from the molten iron rich bead by the copper workpiece.

The high heat transfer is also responsible for the rapid cooling even in the solid state leading to a martensitic transformation of the iron in the weld bead. Since the heat transfer is maximum near the interface, one observes a mild hardness gradient peaking near the interface. It is worth noting that such a martensitic transformation could not be observed in the case of cladding. We are currently attempting modeling the heat and fluid flow to understand the process further.

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