LASER CLADDING OF LOW ALLOY STEEL SUBSTRATES WITH CARBON-FREE TOOL STEELS Fe-Co-Mo/W

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Abstract
Cladding by laser deposition is an effective way to modify metallic surfaces, e.g. to improve the hardness and wear resistance. In the present work, carbon-free tool steel layers of the general composition Fe-Co-Mo(-W) were produced on structural steel surfaces by laser depositing gas atomized spherical powders. In a preliminary test series, a large variety of different compositions were prepared by pressing, sintering and laser remelting and tested with regard to their hardening behaviour. From two promising grades, prealloyed powders were manufactured by gas atomizing and deposited on Cr alloy structural steel, a CO\textsubscript{2} laser being used. Then the layers were heat treated by aging at different temperatures. It showed that these precipitation-hardenable tool steel grades are particularly well suited for laser deposition since they are fairly soft and ductile as-deposited, thus avoiding the cracking problems common with standard carbidic tool steels, which exhibit a brittle ledeburitic microstructure after laser deposition, and also enable soft machining. After a subsequent isothermal aging process at about 600°C, a full hardness of about 65 HRC is obtained, the isothermal treatment resulting in virtually distortion-free hardening.

Keywords: laser deposition, carbon-free tool steels, precipitation hardening, temper resistance

INTRODUCTION
For many tools and other wear loaded components, a typical feature is that the high hardness and wear resistance is required only at the – or at a specific – surface while the remaining material should possess sufficient toughness. One solution for this requirement is surface cladding, also termed hardfacing, i.e. deposition of a hard and wear resistant layer on a tough substrate [1]. This also offers the benefit that only fairly small amounts of the hard material are required, which is frequently quite expensive, and the remaining material may be a cost-effective solution such as e.g. a structural steel or in many cases even mild steel. To some extent, hardfacing can be counted among the various techniques termed “additive manufacturing” today.

There are numerous techniques for hardfacing, many of which are variants of welding. In recent years, laser cladding has become a widely used method to deposit hard phases onto a substrate. This is due to the availability of high energy lasers that enable the focusing of high energy on a small, well defined area. The material to be deposited is usually injected into the laser beam as powder, being transported by a carrier gas, and is...
rapidly melted in the laser beam but then also rapidly solidified after the beam has moved further on. Scanning a large surface with the laser enables hardfacing of the entire surface, but one of the main advantages is the focused hardfacing of selected areas, i.e. those for which wear resistance is the main requirement.

Hard surfaces can be obtained by cladding them with tool steels [2], such as e.g. cold work tool steels or high speed steels. These steel grades with high carbide contents offer excellent hardness and wear resistance. However, it should be remembered that if manufactured through the classical ingot metallurgy route, these steels exhibit a very brittle as-cast microstructure, due to the high-carbide eutectic network formed during solidification [3]. Only after considerable hot working which disintegrates the eutectic network into carbide stringers, sufficient toughness is attained. If laser cladding is done using such tool steel powders, the structure in the powder particle is very fine, but after laser melting and subsequent solidification the as-cast structure is present, martensitic hardening being afforded by the rapid cooling through heat transfer into the substrate. Both effects result in a very hard and brittle deposited layer that is prone to cracking under thermal stresses.

There are however tool steel grades which do not form such eutectic structures on solidification and which are also fairly soft after deposition. These are the carbon-free precipitation hardened steel grades of Fe-Co-Mo/W type described already in the early 1930s by Köster and Tonn [4-6] and have been later thoroughly investigated in the USSR by Geller et al. [7-9]; these alloys showed to be markedly more temper resistant than carbide tool steels, being thus particularly suited for the machining of stainless steels and Ti alloys. In both cases however, the ingot metallurgy route was chosen. Manufacturing by the PM route has been suggested by Köster in the late 1930s [10] but was performed much later by Karpov et al. [11] who started from a coprecipitated and coreduced Fe-Co-W-Mo powder. Danninger et al. [12] showed that the blended elemental approach can also be successfully used if suitable starting powders are used and sintering is done at sufficiently high temperatures. Today a Fe-Co-Mo steel grade is commercially available from Boehler Edelstahl, Austria, under the designation MC-90 Intermet [13]. One characteristic feature of this steel is that it is fairly soft after solution treatment and quenching, hardening being attained by a subsequent isothermal aging treatment at 550-600°C that forms fine precipitates of µ phase [12,14]; the exact mechanism of precipitate formation has been studied by Eidenberger et al. [14-16]. From this behaviour it was expected that the steel should be well suited for laser deposition, its as-deposited ductility avoiding the cracking typical for carbide tool steels.

**EXPERIMENTAL PROCEDURE**

For preliminary studies, to check the structure and heat treatment response of different compositions during laser remelting, compacts were prepared from elemental starting powders by mixing, uniaxial die compaction and sintering. The starting powders were Carbonyl Fe CN (BASF), Mo powder (Plansee), W powder (WOLFRAM), Co < 5 µm (UMEX). The powders were dry blended in a tumbling mixer, compacted at 600 MPa under die wall lubrication to bars 55 x 10 x approx. 7 mm3 and then sintered in flowing hydrogen at 1300°C for 1 hr. This was not a sintering proper that resulted in fully homogeneous microstructure [17], but it was sufficient to give the compacts adequate strength for the subsequent laser treatment.

This treatment was done using a Rofin-Sinar CO2 laser with 3500 W power. The laser beam was moved over the specimen surface with a speed of 0.3 m/min, the bars remaining fixed at the working table. After surface remelting of the track, the bars were
aged at temperatures ranging from 450 to 700°C in flowing N₂ (99.999%). Then cross sections were prepared metallographically and the microstructures were studied as well as hardness values measured.

For the laser cladding experiments, a different approach was selected. Here, prealloyed steel powders Fe-25%Co-15%Mo and Fe-25%Co-14%W-7.5%Mo, respectively, were prepared by inert gas atomization at Treibacher, Austria, and Nanoval, Germany, respectively; as a reference, tool steel powder AISI M42 (Carpenter, USA) was also used. The powders were laser deposited using the same CO₂ laser, but here the laser remained fixed and the substrate was moved. The laser power was varied from 2200 to 2600 W, and the speed was set at either 0.4 or 0.6 m/min. The gas feed rate for powder delivery was 40 to 70 l Ar/min. The substrate was a low Cr alloyed structural steel In principle simple mild steel could have been used, but it was decided to use an alloy steel so that any mixing effects, e.g. uptake of substrate material into the deposited layer, would stand out more clearly. Also here the laser-treated specimens were aged at varying conditions and then characterized.

**LASER REMELTING**

To assess the response of different types of carbon-free tool steels to laser melting, the resulting self-quenching and subsequent aging, different binary and ternary compositions were studied. The specimens were prepared and laser treated as described above. In Figure 1a the laser remelted specimens are shown; melted tracks are clearly visible. In Figure 1b, the cross section of such a track is shown; the virtually structureless melted area is clearly visible as compared to the surrounding sintered material in which the comparatively coarse white µ phases are clearly discernible. Figures 1c and d depict light optical images of two alloys with high Mo and W contents, respectively; it is evident that the Mo containing materials exhibit grain boundary phases while the W alloyed grade does not; here it must be considered that when comparing not the wt. % but the atomic %, the former material contains about double as much VIa element as the latter.

In the wear tracks the hardness was measured as-deposited as well as after aging. The results are shown as temper graphs in figures for some of the materials and are summarized in Table 1 and graphically in Fig.2. As can be seen, the hardness levels attainable after aging are quite high, up to 1100 HV1, which indicates that also hardening through intermetallic phases is highly effective. From the only moderate drop of the hardness from 600 to 650°C aging temperature can be seen that the high resistance to overaging of the intermetallic precipitates, which has already been described by Köster and Tonn [4-6], holds true also for the laser melted materials. Finally, what is also striking is the huge hardness difference between the as-deposited and the aged state, which is much larger than e.g. for carbidic tool steels; this underlines the specific hardening mechanism here, which is more closely related to maraging steels than to the classic carbidic tool steels.
Fig. 1a. Presintered bars Fe-25Co-30W, laser remelted

Fig. 1b. Cross section through the remelted track of specimen as in Fig. 1a

Fig. 1c. Laser track in Fe-25%Co-22.5%Mo

Fig. 1d. Laser track in Fe-25%Co-22.5%W

Fig. 1. Pressed and sintered specimens, laser remelted.

Tab. 1. Composition of experimental Fe-Co-Mo/W alloys and hardness as-deposited and after aging at varying temperatures.

<table>
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<tr>
<th>Fe mass%</th>
<th>Co mass%</th>
<th>Mo mass%</th>
<th>W mass%</th>
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<th>HV1 550°C</th>
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CLADDING OF STEEL SUBSTRATES

For these experiments, the compositions Fe-25%Co-15%Mo (A5) and Fe-25%Co-14%W-7.5%Mo (W5) were chosen, with carbide HSS M42 as a reference; prealloyed powders were available as described above. The morphology of the powders can be seen in Fig.3. After a thorough parametric study, the optimum cladding parameters could be identified which resulted in reproducibly clad surfaces. The excess material can be machined off; considering the moderate hardness of the as-deposited material this machining is much less effort than e.g. with carbide tool steels. In any case, deposition was possible without cracks for both carbon-free alloys while deposition of the carbide high speed steel M42 resulted in severe cracking of the beads.
In metallographic cross sections the microstructure was studied, and elemental profiles were taken by SEM-EDX. The results are shown in Fig.4a, b (please note that the Fe signal is shown with 10% of the real intensity); as can be seen, mixing effects, which are frequently a problem with cladding techniques, are marginal here, as indicated by the very steep gradients in particular for W and Mo. The Cr content of the deposited layer is negligible, indicating that the uptake of substrate material into the layer has been very low.

Furthermore, hardness profiles were taken. Such profiles are shown in Fig.4c, d for both materials. It stands out clearly that in the deposited zone the hardness is moderate, being in the range of 450 HV1, which for the W containing steel agrees well with the data given in Table 1, while for the Mo alloyed grade it is slightly higher, probably due to different cooling conditions during self-quenching. In any case however, the hardness is still sufficiently low to enable soft machining of the material.

The hardness of the substrate, the low alloy steel, is still markedly lower than that of the deposited material; typically, the highest hardness is recorded in the transition zone where the materials are mixed. This is not surprising since here the high contents of carbide-forming alloy elements – Mo and W – interact with the carbon from the substrate, resulting in low-carbon steel with very high hardenability. Nevertheless, the difference to the hardness of the bead is not too pronounced, the hardness in the transformation area being a 600 HV1 maximum. In any case however, this shows that the laser cladding parameters should be adjusted such to ensure a minimum mixing effect.
Fig.4c. Hardness profile, Fe-25%Co-15%Mo (A5)

Fig.4d. Hardness profile, Fe-25%Co-14%W-7.5%Mo (W5)

Fig.4. SEM-EDX Elemental and hardness profiles of structural steel laser clad with carbon-free tool steels, as-clad.

The hardness levels measured here should be regarded against those attained after the deposition of classical tool steel, e.g. AISI M42. This is shown in Fig.5. Here the large difference in as-clad hardness is evident: the M42 results in as-deposited hardness > 750 HV1 compared to about 450 HV1 for the carbon-free grades. Also for the M42 the transition zone shows the highest hardness; this is probably due to carbon transfer to the substrate, which results in locally lower carbon and thus less retained austenite in the transition zone.

Fig.5. Hardness profiles of low alloy structural steel laser clad with AISI M42 as compared to the carbon-free grades.

Carbon-free tool steels are heat treated by solution anneal, to bring the alloy elements Mo and W into the solution, followed by a quench and then a temper = aging treatment to form the secondary precipitates. Therefore, such heat treatment was also performed with the laser-clad specimens, and once more metallographic sections and hardness profiles were taken. In Fig.6, the profiles are shown for different states: as-quenched (2), aged at 550°C (3) and, for comparison, also the initial state as-clad (1). It is clearly evident that after quenching the carbon-free tool steel is still soft, exhibiting about the same hardness as after cladding, while the substrate shows about 700 HV1, as a consequence of martensite formation. After the aging treatment, this hardness drops to about 550 HV1, while that of the tool steel increases to about 900 HV1 for both carbon-free tool steel grades, due to the precipitation hardening effects.
Fig. 6. Hardness profile of structural steel laser clad with carbon-free tool steels, differently heat treated.

**MICROSTRUCTURES**

In Figures 7-9 a-d, typical microstructures of the tool steels are shown in different states. The extremely fine microstructures of the carbon-free tool steels are visible; after laser deposition hardly any features are discernible except a few grain boundaries. After solution treatment some μ phases can be seen which are in the low μm range; as shown in [12, 16], such μ phases are necessary to avoid excessive grain coarsening during solution treatment. After aging, the microstructure is etched dark, and the larger μ phases stand out as white, more or less spherical dots.

Fig. 7. Laser deposited Fe-25%Co-15%Mo, as-clad vs. differently heat treated.
Fig. 8. Laser deposited Fe-25%Co-14%W-7.5%Mo, as-clad vs. differently heat treated.

Fig. 9. Laser deposited AISI M42, as-clad vs. differently heat treated.
The microstructure of the M42, for comparison, looks much coarser, and regardless of the heat treatment the eutectic network remains there (Fig.9), indicating that the material is inherently brittle also after heat treatment.

CONCLUSIONS

Carbon-free tool steels of the type Fe-Co-Mo/W can be successfully deposited on low alloy steel by laser cladding with very low intermixing effects. Since these steels do not form hard martensite during rapid cooling, but remain relatively soft as-quenched – or, similarly, as-laser-deposited - they are not prone to cracking due to thermal stresses, which is a considerable problem with the classical carbidic tool steels. Rapid melting and cooling during laser remelting or laser deposition results in an almost featureless microstructure with a moderate hardness of 400-450 HV1, which enables conventional machining; after solution treatment, oil quenching and aging, hardness levels > 900 HV1 are attained, the microstructures being practically identical to those of the same material prepared in bulk by HIP. Conventional high speed steels processed in the same way, in contrast, retain the brittle eutectic solidification structure, the properties thus being markedly inferior to those of the same material manufactured through the HIP or even the classical ingot metallurgy (casting and hot working) route. The carbon-free grades thus showed to be much better suited to deposition by local melting than do the carbidic grades.

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REFERENCES