

SINTER-HARDENING RESPONSE OF SINTERED STEEL BASED ON ASTALOY Mo AND ITS DIFFUSION BONDED DERIVATIVES

S. Karamchedu, S. Hatami, L. Nyborg, M. Andersson

Abstract

This study addresses the sinter-hardening response of sintered steel based on Astaloy Mo grade (prealloyed with 1.5 wt.% Mo) and its derivatives Distaloy DH- diffusion bonded with 2 wt.% copper and Distaloy DC-diffusion bonded with 2 wt.% Ni, respectively. Compacts of these powder grades were sintered at 1120°C for 30 minutes and cooled at a rate of 2.5°C/s within the range of 800 to 300°C, followed by stress relieving at 200°C for 60 minutes. Tensile testing and metallographic investigations were performed in order to understand the relation between microstructure and the mechanical properties. It is shown that the distribution of the alloying elements, as expected, has a crucial impact on microstructural development. The regions lean in Cu or Ni become bainitic, while regions with Cu or Ni become martensitic. Also, despite the fact that Cu is more homogeneously distributed in the DH material, while Ni partly remains in Ni-rich regions in the DC material, both alloys show similar sinter-hardenability. The compensating effects of carbon redistribution upon bainite formation and Ni on C activity are proposed to be the key factors in this context.

Keywords: *tensile testing, Astaloy Mo, Distaloy DC, Distaloy DH, diffusion bonded powders*

INTRODUCTION

Sinter-hardening incorporates two processes: sintering and hardening into one operation making it economically attractive. This is achieved by cooling the sintered compact from the sintering temperature at a rate that is fast enough for a significant portion of the material matrix to transform into martensite. Commonly, cooling rates in the order of 1-10°C/s within the temperature range of 900 to 425°C are employed [1] and certain powder grades with suitable hardenability have been developed for this purpose. The objective of this study is to evaluate the impact of Ni and Cu on the sinter-hardening behaviour when they are introduced in diffusion-bonded form to a ferrous base powder prealloyed with 1.5 wt.% Mo, commonly known as Astaloy Mo. The resulting diffusion bonded grades are commonly known and referred to as Distaloy DC and Distaloy DH, respectively. Pre-alloying with molybdenum, which has a low solid solution strengthening effect, and introducing Cu and Ni through diffusion bonding, promote compressibility of these powders [2].

MATERIALS AND METHODS

The nominal chemical compositions of the materials are given in Tab.1. Distaloy DC and Distaloy DH will hereafter be referred to as DC and DH in the following text.

Tab.1. Nominal chemical composition (wt.%) of the alloys. ¹ Prealloyed in the base powder.

Material	Ni	Mo	Cu	Graphite	Fe
Astaloy Mo	-	¹ 1.5	-	0.55	Bal.
Distaloy DH	-	¹ 1.5	2	0.55 & 0.7	Bal.
Distaloy DC	2	¹ 1.5	-	0.55 & 0.7	Bal.

Sinter-hardened compacts of the three powder grades Astaloy Mo, DH and DC were produced by Höganäs AB. The compacts were prepared in the form of tensile specimens (dog bone: ISO 2740) by pressing powder at corresponding pressures to achieve a sintered density of 7.1 g/cm³. The specimens were sintered at 1120°C for 30 minutes in 90N₂/10H₂+0.2% CH₄ atmosphere. All samples were cooled at an average cooling rate of 2.5°C/s in the temperature range of 850 to 300°C. The samples were tempered after sinter-hardening at 200°C for 1 hour in air. Metallographic observations were carried out using light optical microscopy (LOM). The AxioVision software by Zeiss was used for determining the amount (vol.%) of each micro-constituent (averaged over 10 measurements). Scanning Electron Microscopy (SEM) and Energy Dispersive X-ray Spectroscopy (EDX) were used for more detailed microstructural investigations. Tensile tests were performed in accordance with the Swedish standard SS-EN-10002-1:2001 with a strain rate of 10⁻⁴ s⁻¹. Tensile strength and the strain to fracture were determined from the engineering stress-strain curve and the 0.2% proof stress- yield strength was determined graphically. For every material the tests were repeated 5 times. Micro-hardness measurements were performed with a Vickers indenter in accordance with MPIF standard 51: 2010 (results averaged from 15 indentations).

RESULTS AND DISCUSSION

Figure 1 (a-e) shows representative microstructures obtained by LOM. The Astaloy Mo material exhibited a homogenous bainitic microstructure due to the prealloyed nature of the powder, while the microstructures for DC and DH constituted both bainitic- marked 'B', and martensitic phases- marked 'M', for both carbon levels examined (see Fig.1).

These results are in agreement with the metallographic observations and CCT curves reported earlier [1,3,4], which also indicate that higher cooling rates (more than 6°C/s) would be necessary for obtaining fully martensitic microstructures for the materials used in the present study. Besides bainite and martensite, Ni-rich regions, marked 'N', were identified in the microstructure for the DC samples.

The volume fraction of the bainitic and martensitic phases in DC and DH were similar for both carbon levels and the increase in martensite volume fraction with an increase in carbon content was also similar (see Fig.1f). To study the distribution of alloying elements in the microstructure, SEM and EDX analysis was performed. Representative results from EDX mapping complemented with point analysis obtained from DH and DC samples are shown in Fig.2. Each figure has images numbered with locations from where EDX point analysis was obtained and the corresponding results are shown in the table at the bottom-right on the image.

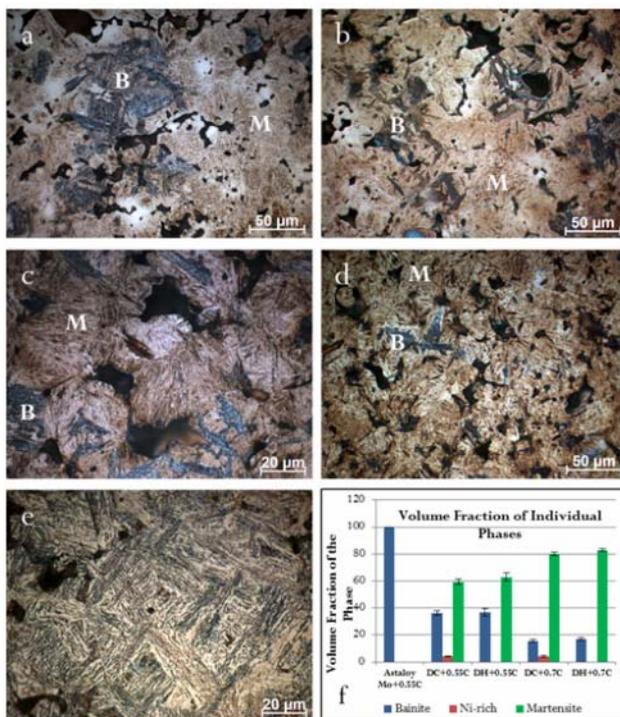


Fig.1. LOM micrographs of a) DC with 0.55 wt.% C b) DC with 0.7 wt.% C c) DH with 0.55 wt.% C d) DH with 0.7 wt.% C e) Astaloy Mo with 0.55 wt.% C; etchant: 10 ml HCl and 3 ml HNO₃ f) Volume fraction of individual phases.

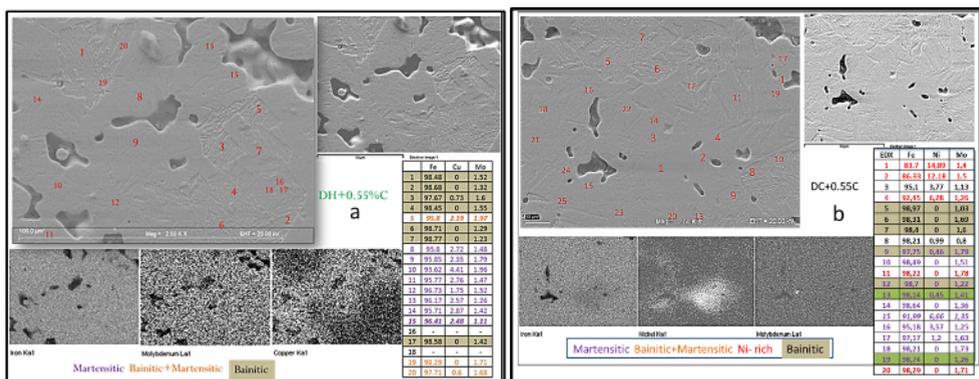


Fig.2. SEM and EDX analysis results from a) DH-0.55% C and b) DC-0.7% C.

Elemental maps obtained from the same region are shown in the top right corner with the resultant maps in the second row. EDX analysis shows that the presence of bainitic regions in the microstructure can be attributed to the absence of the alloying elements Cu or Ni in these regions, reducing the hardenability. This was also reported earlier [5]. This results from the diffusion-bonded nature of the DH and DC powder grades. The prealloyed Mo, in contrast, was uniformly distributed throughout the microstructure for all the samples. Nickel and Cu show different responses to sintering (mechanisms during

sintering). Copper melts at 1083°C i.e., below the sintering temperature, while nickel has a melting temperature (1455°C) well above the sintering temperature. EDX analysis shows that copper is distributed rather uniformly within the martensitic regions in the microstructure while nickel is not. For the DC samples, additionally, Ni-rich regions were identified from the EDX analysis (see Fig.2b); also observed from the metallographic investigation, see Fig.1 a-d. This is due to the slower inter-diffusion for the Fe-Ni with Fe diffusion into Ni being dominant, with the diffusion coefficient of Fe being one order of magnitude higher and resulting in gradient microstructures with varying concentration of Ni [6,7]. Since the total content of alloying elements is similar for both DC and DH, the average composition of nickel within the martensitic regions will be lower compared to copper. This is due to nickel concentration gradients resulting from the complex inter-diffusion effects referred to earlier [6,7]; and also, since nickel is arrested in the Ni-rich regions, its availability in the other phases (connected to the former matrix powder) is decreased. These effects are evident when comparing the EDX analysis results from Figures 2a and b. The amount of martensite, in spite of these effects, is similar for both the DC and DH alloys (see Fig.1f) for both carbon levels examined.

Examining the micro-hardness values, it can be seen that the hardness for the bainitic phase remains similar for all the alloys and the values are similar to those observed for Astaloy Mo. This is in agreement with the SEM and EDX analysis showing a composition similar to Astaloy Mo in the bainitic regions for all the materials. Martensite in DC shows similar values at both the carbon levels, which are similar to the values obtained for DH at the higher carbon level. Considering the case for a wrought steel with composition similar to DC and assuming uniform distribution of Ni, the hardness values for fully martensitic microstructures obtained at different carbon levels have been calculated using JMatPro Modelling software, see Table 2. Grain-size was varied in the calculations in order to obtain information about fully martensitic microstructure. The hardness values in the table concern un-tempered condition for martensite and the tempered values can be expected to be lower. Regarding the hardness values for the martensite phase, for the DH samples, an increase in the hardness is evident with an increase in carbon, as shown in Fig.3a. The average values measured for martensite for DC at lower and higher carbon levels were 650 and 676 HV, respectively, with standard deviations of 46 and 32 HV, respectively, resulting in comparable hardness ranges, see Fig.3a.

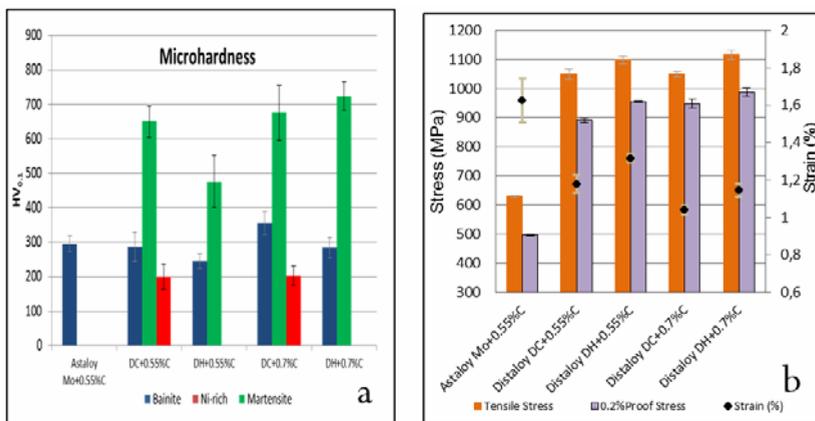


Fig.3. Micro-hardness measurements for the individual phases (a) and tensile test results showing yield strength, elongation to fracture and ultimate tensile strength values (b).

Considering the higher and lower limits from the standard deviation, the values correspond to un-tempered martensite with 0.5 to 0.65 and 0.6 to 0.7 wt.% according to the calculations in Table 2. Carbon redistribution can occur during bainitic transformation, as discussed in [8], in which case the lower bainite level for the higher carbon content could contribute to a compensating effect resulting in similar hardness ranges.

Tab.2. Hardness values (HV) for martensite phase calculated from JMatPro v.8, modelling software.

C [wt.%]	0.4	0.5	0.65	0.7
Astaloy Mo	608	668	710	747
DC	629	688	724	750

Also, considering the substantial effect of carbon on the hardness of martensite, it can be hypothesized that at both the carbon levels, DC has 'similar carbon content' but non-uniformly distributed within martensite. This hypothesis is supported by the previous investigations [5,9], according to which, a carbon repelling effect is observed around the Ni-rich areas, since nickel increases the chemical activity of carbon. This, along with the gradient microstructures resulting from Fe-Ni inter-diffusion mentioned earlier, would result in a range of carbon values within martensite which are close to the estimated values recorded in Table 2. It can also be expected that regions around the Ni-rich areas, even though leaner in Ni, can transform into martensite, due to the improved hardenability contributed by the diffused carbon as a result of this repulsion effect. This repelled carbon contributes both to a higher hardness, already at lower carbon level, and improved hardenability, which results in DC having similar martensite volume fraction as the DH materials.

Tensile testing results are shown in Fig.3b. Astaloy Mo has the lowest tensile and yield strengths and the highest elongation, due to the fully bainitic microstructure. For DC, with increased carbon content, an increase in yield strength was observed, but UTS remained similar and elongation decreased. A similar trend was observed for the DH samples with respect to increase in carbon. Comparing DC and DH samples, at the lower carbon level, UTS, yield strength and elongation were higher for DH. At the higher carbon level, DH still exhibited higher UTS, yield strength and elongation. One explanation for DH exhibiting better mechanical properties could be the presence of martensite with higher carbon content in DC -resulting from the repelling effect- which can contribute to brittleness. Previous studies also report that for the sintered compacts alloyed with Ni, primary cracks form and grow frequently at the weak nickel-rich carbon-lean areas, or at the inter-phase boundaries, with one phase being Ni-rich, occurring at the necks [10]. This could also contribute to inferior mechanical properties of DC.

CONCLUSIONS

Metallographic examinations showed that Astaloy Mo has a homogeneous bainitic microstructure, while DC and DH exhibit a microstructure consisting of bainite and martensite. The amounts of bainite and martensite are similar for both the materials at both the carbon levels examined. Additionally, the presence of Ni-rich regions was identified for the DC material. SEM and EDX analysis showed that the regions where the diffusion bonded alloying elements were absent resulted in a bainitic microstructure. In the regions where the diffusion alloyed elements were present, copper was more uniformly distributed in comparison with nickel. From the micro-hardness results, it is shown that the hardness values obtained for DC, even at lower carbon level, are similar to those at the higher carbon

level and, it is hypothesized, due to the carbon repelling effect due to nickel. This is hypothesized to compensate for the non-uniform distribution of nickel and the similarity in microstructure to that of DH. For the carbon levels and the conditions employed in this study, tensile strength, yield strength and elongation were slightly better for DH compared to DC, and an increase in the carbon content does not seem to be effective for improving the tensile properties.

Acknowledgements

The authors are grateful to the financial support provided by VINNOVA (The Swedish Governmental Agency for Innovation Systems) and support and technical cooperation provided by Höganäs AB.

REFERENCES

- [1] Handbook, "Designing for P/M- Processing,"
- [2] Bocchini, GF.: *Int. J. Mater. Prod. Technol.*, vol. 15, 2000, p. 172
- [3] Engstrom, U. In: *Powder Metallurgy and Particulate Materials*, 2000
- [4] Maroli, B., Berg, S., Lewenhagen, J. In: *EURO PM2001*, 2001
- [5] Wu, M., Hwang, K.: *Mater. Sci. Eng. A*, 2010
- [6] Castro, F., Sainz, S.: Influence of Ni on hardenability of PM steels, mpiforg
- [7] Sainz, S., Garcia, W., Karuppanagounder, A., Castro, F.: *Powder Metall. Prog.*, vol. 7, 2007, no. 3,
- [8] Hatami, S., Malakizadi, A., Nyborg, L., Wallin, D.: *J. Mater. Process. Technol.*, vol. 210, 2010, no. 9, p. 1180
- [9] Wu, M., Hwang, K., Chuang, K.: *Powder Metall.*, 2008
- [10] Wu, M., Hwang, K., Huang, H.: *Metall. Mater. Trans.*, 2007