Effect of Repeated Quenching Heat Treatment on Microstructure and Dry Sliding Wear Behaviour of Low Carbon PM Steel

Ahmet Güral[1,a], Süleyman Tekeli[1,b], Dursun Özyürek[2,c] and Metin Gürü[3,d]

[1]Gazi University, Tech. Edu. Fac., Mater. Div., 06500, Teknikokullar-Ankara/Turkey
[2]Zonguldak Karaelmas Univ., Karabuk Tech. Edu. Fac. Cast. Div., 78100 Karabuk/Turkey
[3]Gazi University, Eng.&Arch. Fac., Chem. Eng. Dept., 06570, Maltepe-Ankara/Turkey
[a]agural@gazi.edu.tr, [b]stekeli@gazi.edu.tr, [c]dursunozyurek@karaelmas.edu.tr, [d]mguru@gazi.edu.tr

Abstract

The mixed atomized iron powders with 0.3 % graphite and 1 % Ni powders were cold pressed and sintered at 1200 °C for 30 min under pure Ar gas atmosphere. Some of the sintered specimens were intercritically annealed at 760 °C and quenched in water (single quenching). The other sintered specimens were first fully austenized at 890 °C and water quenched. These specimens were then intercritically annealed at 760 °C and re-quenched in water. The experimental results showed that the wear coefficient effectively decreased in the double quenched specimen.

Keywords: Dry sliding wear, low carbon PM steels, intercritical annealing, quenching

1. Introduction

Ceschini et al[1] obtained high wear resistance in the Fe-C-Mo steels with higher Mo content, sintered under conditions giving rise to bainitic microstructure. Simchi et al[2] found that wear mechanisms of PM parts were similar to those of wrought materials although the microstructural features impact the wear rate. Sudhakar et al[3] concluded that 2 Ni iron based PM steels exhibited the maximum wear resistance when hardened and tempered. Previous investigations showed that the wear properties were also affected by martensite content and particle size in dual phase ingot steels[4,5]. 0.2 C Dual phase steel and found that the wear resistance of this steel is higher than that of 0.4 C steel.

In the present study, repeated quenching heat treatments in the intercritical annealing region were applied to the sintered PM steels prepared from Ancorsteel 1000 iron powder with 0.3 % graphite and 1% Ni additions. The effect of single and double quenching heat treatments on the microstructure and dry sliding wear behaviour was investigated.

2. Experimental and Results

For microstructural investigation and wear tests, PM specimens were prepared using atomized iron (Ancorsteel 1000), 0.3 % natural graphite (Alfa Aesar, Germany) and 1 wt.% Ni (Alfa Aesar, Germany) powders (in wt.). Mixed powders were compacted at 700 MPa and sintered at 1200 °C for 30 min under pure argon atmosphere. Some of the sintered specimens were intercritically annealed at 760 °C and rapidly water quenched (SQ). The other sintered specimens were first fully austenized at 890 °C and rapidly water quenched and then intercritically annealed at 760 °C and re-quenched in water (DQ). Hardness measurements were made by Vickers tester using 30 kg load. The dry sliding wear tests were carried out on a standard pin on disc machine. The wear coefficient was calculated by dividing the volume of the worn out material by the sliding distance and applied load.

The Ni-rich martensite islands in the ferrite matrix were obtained in the final microstructure of the SQ specimen after intercritical annealing and quenching (Fig.1a). After intercritical annealing, the Ni-rich martensite islands in the ferrite matrix were seen in the final microstructure of the DQ specimen. However, in the DQ specimen, the size of the martensite islands was not as fine as it was expected (Fig.1b). Through similar heat treatments, martensite islands in ingot steels usually become very fine[6]. The reason for not having finer martensite particle size is thought to be caused by partially dissolved coarse Ni particles. During intercritical annealing, the austenites (martensite after quenching) preferentially formed around Ni rich areas. Thus, the martensitic transformation occurred when the austenite with surrounding Ni-rich areas (austenite with less Ni concentration compared to Ni concentration in the centre) was quenched from the intercritical annealing temperatures.

The martensite volume fraction obtained in the DQ specimen was higher than that of the SQ specimen. This probably occurred because of the fact that some more Ni-rich areas in the austenite iron might have been dissolved when the DQ specimen was re-heated at 890 °C to produce fully martensitic structure before intercritical annealing. Thus, the austenites formed in wider areas during intercritical annealing. The hardness values of 214 and 223

HV were obtained in the SQ and DQ specimens, respectively. Higher hardness value in the DQ specimen was due to slight increase in the martensite volume fraction.

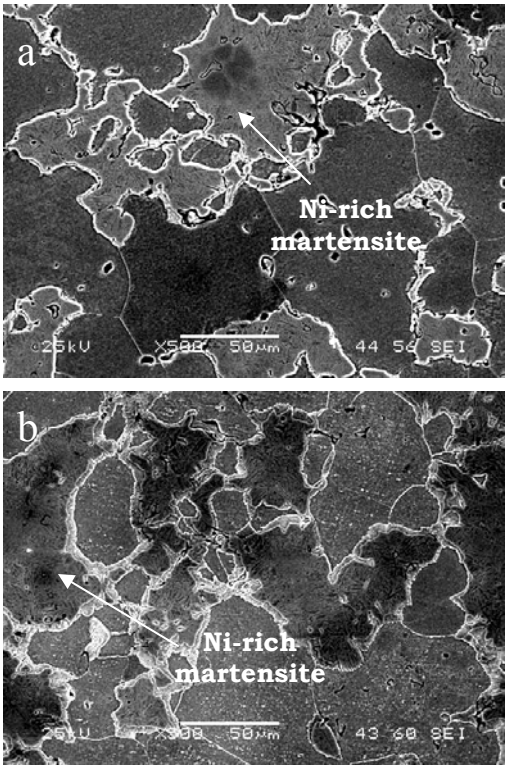

Fig. 1. SEM micrographs of specimens of SQ (a) and DQ (b).

The wear coefficient values of the SQ and DQ specimens are given in Fig. 2. The wear coefficient of the DQ specimens is lower than that of the SQ specimen at all sliding distances. In the DQ specimen, although the martensite islands were not refined, the martensite volume fraction and hardness increased, which explains why the wear coefficient was low in this specimen. In the SQ specimen, the wear coefficient decreased with increasing sliding distance. This is probably due to the local deformation of the surface of the SQ specimen during wear test. This may cause the austenitic areas in the centre of martensite islands to be transformed to martensite, consequently causing the wear coefficient to decrease.

Overall, the wear coefficient was stable in the DQ specimen at all sliding distances. This is assumed to be due to the homogen dissolution of Ni in this specimen. Whang et al. [7] observed a martensite cracking on the worn surfaces of fully martensitic PM specimen. However, in the present study, such martensite cracking was not observed. As the hard martensite islands were embedded in soft ferrite phase, the risk of martensite cracking decreased, which in turn contributed to the decrease of wear coefficient.

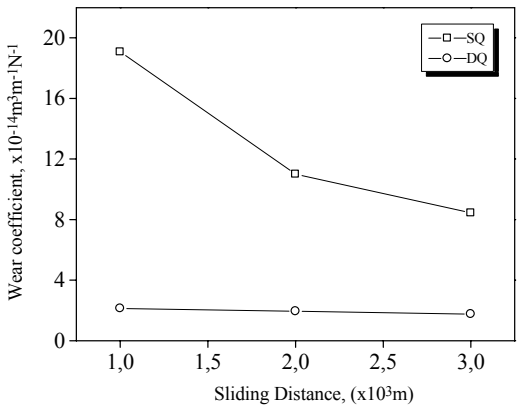

Fig. 2. Wear coefficients of the SQ and DQ specimens.

3. Summary

The martensite volume fraction and hardness in the DQ specimen were higher than that of the SQ specimen. Thus the wear coefficient of the DQ specimen was found to be lower. In the SQ specimen, the wear coefficient decreased with the increase in the sliding distance. However, at all sliding distances, the wear coefficient was stable in the DQ specimen.

Acknowledgements

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4. References

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