Influence of Silicon on Pitting Life of Carburized Steel Treated by Shot Peening

Junya Miyauchi, Ryohei Ishikura, Keisuke Inoue, Kohei Yamaguchi
Daido Steel Co., Ltd., Japan, j-miyauchi@ac.daido.co.jp

Keywords: Residual stress, Shot peening, Carburized materials, Pitting strength, Wear rate

Introduction
With environmental issues as the background, there is a high need for more compact and lightweight gear parts for automobiles and industrial machinery, in order to improve fuel efficiency. Increasing the strength of individual parts is thus required, and a widely used means to achieve higher strength is shot peening (SP), with the results of examinations for the effects of SP for bending fatigue resistance of the gear tooth root and pitting resistance of the tooth surface having been reported.1) 2) Reducing the size and weight of the actual parts could be achieved by increasing strength on both the driving and driven sides, so performing SP on both sides is presumed. Regarding the pitting life of specimens with SP performed on both the driving and driven sides, the effects of the surface hardness, residual stress and surface roughness before testing on the pitting life were reported by Suzuki et al.3) It was also reported that the addition of silicon (Si) that results in an increased hardness after tempering at 573 K and in the keeping of compressive residual stress in a high temperature environment is effective for improving pitting life.4) However, there has never been any report of a study about the effects of alloying elements on the pitting life of steel components with SP performed on both the driving and driven sides.

Objectives
Using steels with different Si addition amounts, specimens with the same surface hardness, residual stress, and surface roughness before testing were prepared, and an evaluation was conducted for pitting life, with the purpose of clarifying what the effects of Si additions would be on the pitting life of the specimens with shot peening (SP) performed on both the driving and driven sides.

Methodology
Specimens
Table 1 shows the chemical composition of the steel specimens used for this study. The steels were prepared in a 150 kg vacuum induction furnace, after which they were hot forged into steel bars with φ32 mm. The steel bars were normalized at 1223 K, and machined into roller pitting testing specimens (drive rollers) of φ26.3 mm × 130 mm through machining. Then after vacuum carburizing at 1223 K, oil quenching and tempering (403 K × 8.4 ks) were performed, followed by the shot peening process as described below. To make surface roughness uniform prior to shot peening, grinding was performed for 150 μm on either side of the test part (the raceway)
after carburizing, resulting in the finish for each specimen of a diameter of φ26.0 mm and a roughness of Ra ≒ 0.25 μm for the rolling surface. Fig.1 shows schematic diagram of roller pitting testing specimen.

The SP conditions are shown in Table 2. Used for the shot materials were the Conditioned Cut Wire (SP1) with a hardness of approximately 800 HV that is commonly used for carburized gears, and the highly hard particles that are made of an amorphous metal (SP2) having a hardness of 900 HV, higher than that of the carburized layers of the specimens. For the processing conditions, a two-stage shot peening was performed, with SP1 particles shot in the first stage with a projection air pressure of 0.5 MPa and coverage of 400%, and SP2 particles shot in the second stage with a projection air pressure of 0.5 MPa and coverage of 1200%.

**Pitting fatigue test method**

For pitting life evaluation, a roller pitting (RP) tester of Komatsu Ltd. (Komatsu Engineering) was used. The test conditions were set at a load of 3540 kgf, slip ratio of 60 % and rotational speed of 1500 rpm. For lubricating oil, commercially available transmission oil was used, with the test conducted at an oil temperature of 323 K and a flow rate of 3 L/min.

For the driven roller used in the test for JIS-SCR420H, in order to achieve the same tempering hardness as the drive roller, hardening and tempering (HT) was carried out on the bearing steel SUJ2, followed by grinding, and then shot peening (SP) under the same conditions as for the drive roller. As for steel A, after vacuum carburizing, quenching and tempering (VCQT), surface grinding was performed, and SP was applied under the same conditions as for the drive roller. The driven roller had a shape with a diameter of 130 mm and a width of 18 mm, and the radius of curvature was set at 1000 mm with the aim of restraining the bearing pressure decreasing due to the wearing during the test that is described below.

For the specimens before and after testing, observations were conducted for the shape of the drive roller, the hardness distribution, residual stress distribution, and cracks in the cross section. The same evaluation method as for the specimens before testing was used. Also for a more detailed grasp of the failure process, some of the tests were interrupted midway, to then conduct another test for inspecting the state of the damage at that point.

**Results and discussions**

**Pitting life and the morphology of fatigue fracture**

Regarding the quality of drive rollers in each specimen, Table 3 shows the surface roughness, surface hardness and residual stress of drive rollers before the testing. The table also includes information of the drive rollers after tempering at 573 K for 10.8 ks. For the drive rollers, the initial hardness, residual stress and surface roughness were the same, while there were differences in hardness and residual stress after tempering. This shows that the quality of drive rollers are matching the purpose of the current study.

Table 4 shows the pitting life of SCr420H and steel A, and the wear depth of the non-pitting areas after the test, revealing that steel A had a pitting life about 3 times as long as that of SCr420H. Delamination damage was observed for both of these specimens, with the surface-initiated pitting...
seen. For both specimens as well, wear was observed in the areas in contact with the driven roller, with a wear depth of 108 μm for SCr420H and 157 μm for steel A. Therefore, the fatigue fracture morphology observed in the current test was pitting accompanied by wear, and it is considered that pitting life is affected by the two factors of wear rate and the wear depth at which the pitting occurs.

### Wear rate

Fig. 2 shows the wear depth of specimens with number of cycles for SCr420H and steel A. The increase in wear depth is mostly linear for both specimens, but differences are observed for the wear rate. Compared to SCr420H, the wear rate is slower for steel A: the number of cycles until the wear depth reaches around 50 μm is $1 \times 10^6$ and $2 \times 10^6$, respectively for these specimens. From this results, it is found that not only the wear depth leading to pitting occurrence but also the wear rate is different between SCr420H and steel A, with a lower wear rate and a deeper wear depth at the time of pitting occurrence for steel A with a longer pitting life. The effects of surface hardness on the wear have been indicated, and it is considered that the difference in surface hardness during the test, which is affected by sliding heat, influenced the wear rate. Fig. 3 shows the surface hardness of specimens with number of cycles for SCr420H and steel A changes, with their wear depth after testing also shown in the figure. For each number of cycles, surface hardness is higher for steel A than SCr420H, with a similar tendency observed for a wear depth of around 50 μm. Therefore, it is considered that the wear rate is slower for steel A with the higher tempering hardness, due to its higher surface hardness during the test.
Wear depth at the time of pitting occurrence

The wear depth at the time of pitting occurrence is discussed. The process leading to pitting occurrence can be considered as comprising the 2 stages of crack initiation and crack growth, and because cracks are observed for both specifications from the initial cycle stage of around $10^5$, it is considered believed that the difference in the wear depth at the time of pitting occurrence is due to the difference in crack growth. Kanisawa et al. reported that, based on the investigations of crack growth behavior for pitting occurrence of SP-performed specimens, compressive residual stress has the effect of suppressing crack growth. Therefore, an examination of the residual stress and crack growth behaviors during the test was made in this study. Fig.4 shows the change in crack depth and residual stress with number of cycles for SCR420H and steel A. It is found that for both SCR420H and steel A, as the compressive residual stress disappeared, the crack depth rapidly increased, resulting in fractures. Compared to SCR420H, the decline in residual stress was small for any number of cycles regarding steel A. It is therefore considered that the difference in wear depth at which pitting occurs between the two specimens is due to the difference how much the compressive residual stress declines.

Since it is considered that the difference in wear depth at pitting occurrence between SCR420H and steel A is due to from the difference in the declining behavior for compressive residual stress, the changes in residual stress during testing were examined. Fig.5 shows the residual stress distribution before and after testing or 523K tempering, which is assumed to be the approximate temperature near the surface layers during testing. For the figures after testing, the wear depth was taken into consideration, and the depth from the initial surface was calculated. For both SCR420H and steel A, the rise in the number of cycles run was accompanied by a decline in compressive residual stress inside and in the surface layers, but there was a difference in the number of cycles at which the compressive residual stress that had been provided through shot peening disappeared between SCR420H and steel A. Whereas for SCR420H, compressive residual stress disappeared at $1 \times 10^6$ cycles, it was retained even at $2 \times 10^6$ cycles for steel A.

It is assumed that residual stress is affected by the wear, the existence of cracks and the sliding heat. While there was a decline in residual stress at any depth after the test compared to the initial state, it is considered that for the topmost surface especially where the decline was substantial, the
main cause of the decline was the existence of cracks. A comparison of the residual stress distribution for steel A after testing for $5 \times 10^5$ cycles and after testing for $1 \times 10^6$ cycles shows that the residual stress distribution changed at a similar pace in the regions that are deeper than the regions where the decline considered to be caused by cracks occurred, with only thermal effects likely accounting for the decline in these deeper regions. For these regions, a comparison was made with the residual stress distribution after tempering the specimens with shot peening performed, showing that the residual stress distribution was equivalent to that of the specimen tempered at 523 K (Fig. 5). This tendency were similar for the SCr420H specimen after testing for $5 \times 10^5$ cycles. It is true that for the SCr420H specimen after $1 \times 10^6$ cycles, the compressive residual stress is higher than that of the specimen after tempering at 523 K, but this is considered to be because it takes a certain time until the heat-affected decline in the residual stress reaches saturation, similar to the results of the study conducted by Ishikura et al. Thus it is considered that the temperature around the surface of a specimen during testing is approximately 523 K, and that as a result of the repeated tests until reaching $5 \times 10^5$ cycles, the residual stress distribution of the specimen declines to the distribution corresponding to this temperature. As for the testing beyond $5 \times 10^5$ cycles, the residual stress distribution depth affected by heat, determines the depth at which the effects of suppressing crack growth disappear, and compared to SCr420H, the decline in compressive residual stress due to thermal effects is smaller for steel A, with the compressive residual stress retained in relatively deeper regions, leading to the result that the depth at which the effects of suppressing crack growth disappear is deeper. This is considered to account for the deeper wear depth for steel A at the time of pitting occurrence.

![Residual stress distribution before and after testing or 523K tempering.](image)

**Model for the changes in damage for the pitting accompanying wear**

Fig. 6 shows the schematic figure for pitting behavior of specimen. For damage in the initial stages, progress in the wear and a decline in residual stress due to sliding heat occur. At about $5 \times 10^5$ cycles, the compressive residual stress declines to a level corresponding to the heating temperature, then the compressive residual stress disappears, and finally, pitting occurs.
Conclusion

In the current study, specimens with the same surface roughness and hardness and residual stress before testing were prepared for 2 steels with different Si addition amount, and an examination was conducted on the effects of the addition of Si on the pitting life of the specimens with shot peening performed on both the driving and driven sides. Stated below are the results of the examination.

1. For the both specimens, the morphology of fatigue fracture comprised the pitting with wear, but the specimen with high Si addition amount showed a longer pitting life.
2. Regarding pitting life with the pitting with wear, the slower the wear rate is and the deeper the wear depth for pitting occurrence is, the pitting life is longer.
3. The wear rate is affected by surface hardness during the test, and with the addition of Si increasing the tempering hardness, which contributes to extending pitting life.
4. The wear depth at which pitting occurs depends on the depth at which the compressive residual stress that has the effects of suppressing crack growth is retained. The addition of Si results in curbing the decline in compressive residual stress in a high temperature environment, which contributes to extending pitting life.

References

2) Watanabe, Yoichi; Narita, Naoki; Umegaki, Shunzo; Mishima, Yoshinao: Tetsu-to-Hagane, 84 (1998), 902.
3) Suzuki, Tomohiro; Ogawa, Kazuyoshi; Hotta, Shoji: Transactions of The Japan Society of Mechanical Engineers (Volume C), 65 (1999), 3765.
4) Ishikura, Ryohei; Inoue, Keisuke; Kobayashi, Yuji; Tsuji, Shunya: Denki-Seiko (Electric Furnace Steel), 81 (2010), 99.
5) Fukagawa, Munemitsu; Iida, Tadashi; Sasaki, Toshimi; Takeuchi Eiichi; Osozawa, Koichiro; Fukui, Shoichi; Otomo, Kiyomitsu: Heat treatment guidebook (Advanced), Taiga Publishing Co., Ltd., 1984, 108.