Experimental Characterisation of Tool Hardness Evolution Under Consideration of Process Relevant Cyclic Thermal and Mechanical Loading During Industrial Forging

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Abstract. The near-surface layer of forging tools is repeatedly exposed to high thermal and mechanical loading during industrial use. For the assessment of wear resistance of tool steels, in previous work thermal cyclic loading tests were carried out to investigate changes in hardness. However, actual results of time-temperature-austenitisation (TTA) tests with mechanical stress superposition demonstrated a distinct reduction of the austenitisation start temperature indicating a change in the occurence of tempering and martensitc re-hardening effects during forging. Therefore, the superposition of a mechanical compression stress to the thermal cyclic loading experiments is of high interest. Tests are carried out in this study to analyse hardness evolution of the tool steel H11 (1.2343) under consideration of forging process conditions. The results show that the application of compression stresses on the specimen during the temperature cycles is able to restrict tempering effects while increasing the amount of martensitic re-hardening.

Keywords: Forging · Tool hardness · Phase transformation · Wear estimation · Martensitic re-hardening · Tempering

1 Introduction

High workpiece temperatures of up to 1250 °C during forging steel result in excessive heating of the surface layer of the forging tools [1]. Numerous investigations show that high surface temperatures in combination with strong cooling due to spray cooling lead to a structural change in the tool surface layer [2]. By this means, microstructural changes are caused leading to tool hardness changes depending mainly on the tool alloy, the maximum tool temperature and the cooling conditions [3] increasing the risk of tool failure or tool deformations [4]. However, recent studies have also proven, that mechanical stress strongly influences the austenitisation-behaviour of

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hot work steels enabling the occurrence of martensitic re-hardening [5]. In general, hardness-changes have a decisive influence on the wear behaviour and thus on the tool life [6]. In each forging cycle the tools are exposed to a combination of thermal and mechanical stresses [7]. Statistical investigations on forging dies show that the main cause of failure of forging tools is due to approx. 70% abrasive wear and approx. 25% mechanical cracking [8]. In industrial practice, the type of damage is strongly dependent on the existing stress collectives. For example, increased wear appears due to thermally induced micro-cracks and abrasion in the tool surface [9]. The growth of tool wear also leads to geometric deviations and a reduction in component quality, which contradicts the demand for near-net-shape production and consistent product quality. In case of a significant wear progress or a tool breakage, high setup costs are incurred in addition to the costly production of new tools. Therefore, reliable information about the expected tool life is necessary for economical process control and the scheduling of set-up times. Moreover, for the design of wear-optimised tools a realistic prediction of the expected tool wear as a function of the forging cycles is required. In addition to the work of Klassen et al. [5], time-temperature-austenitisation (TTA) tests with mechanical stress superposition were carried out by Behrens et al. [10]. By varying the compression load between 30% and 80% of the elastic limit k_{f0} determined at 900 °C of hardened H11 tool steel, a distinct reduction of the Ac_1 temperature of approx. 40 °C was detected as shown in Fig. 1-A for every heating rate tested. The Ac_1 temperature of a tool steel is of high interest for its wear behaviour because at this temperature the phase transformation to austenite starts, which is retransformed to even harder martensite during the tool cooling [11]. In the context of this study, this effect is referred to as (martensitic) re-hardening. An exemplary application of the data from Behrens et al. [10] via an UPSTNO subroutine in the finite element software Simufact.forming 16 is presented in Fig. 1-B, indicating the area of a forging tool where the Ac_1 temperature is exceeded during the forging process. Including the consideration of mechanical stress on the Ac_1 temperature, the area, where martensitic re-hardening effects are expected ($T_{process} > Ac_1$), is significantly increased resulting in a different expected wear behaviour.

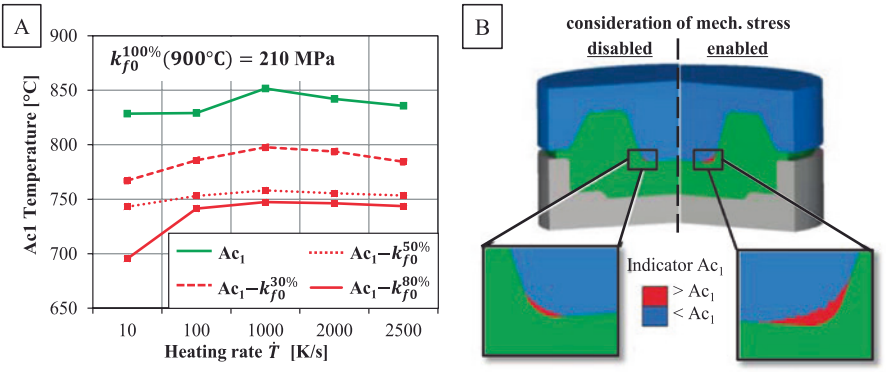

Fig. 1. Results of TTA tests with mechanical stress superposition on H11 tool steel [10] (A)/ Exemplary influence on the size of the of the martensitic re-hardening zone (red) with and without consideration of mechanical stress (B)

As a consequence, further tests are carried out in this study presenting the results of a cyclic thermo-mechanical loading to H11 tool steel. Peak temperatures are varied in regard to Ac_1 and mechanical loading in regard to the elastic limit of the material to test the influence of the parameters on tempering effects and martensitic re-hardening.

2 Methodology

For carrying out cyclic loading tests a forming dilatometer DIL805D by TA Instruments is used equipped with SiO_2 deformation punches (Fig. 2-A). Since investigations on wear-related topics are fundamentally about saving costs, the necessary testing procedure is also strongly connected to an evaluation of testing costs. Therefore, hollow samples are of high interest in order to not only to be able to achieve process relevant heating and cooling rates but also to minimise the amount of nitrogen cooling gas. This sample type is characterised by an increased specimen surface in comparison to the conventional specimen made of bulk material. Because of this decision, two main circumstances using the DIL805D had to be addressed:

1. The primary use case of the deformation unit for the DIL805 is the evaluation of mechanical properties using cylindrical bulk samples (Ø 5 mm x 10 mm). The default hydraulic force control parameters are therefore optimised for this sample type. Using hollow samples with these parameters results in high force deviations as shown in Fig. 2-B, especially during fast heating or quenching segments, where the sample length is rapidly changing. As a consequence the controller sensitivity has to be significantly increased by reducing the proportional value to xp = 0.007. This system value is the most influential parameter for the determination of the control strength in proportional relation to the systems inherent control power after a control deviation is measured. In this case, less power of the hydraulic pressure pump for the regulation of the punch force has to be engaged to accommodate for the reduced specimen cross section. This change reduces force deviations during fast temperature changes to less than 10% of the specified value while allowing heating rates of up to 600 K/s.

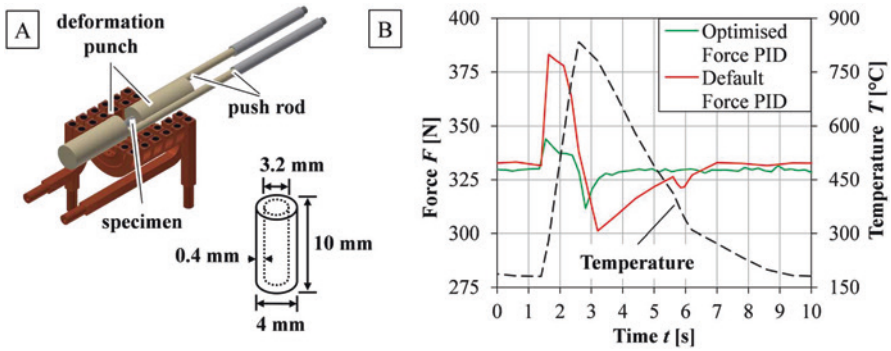

Fig. 2. Dilatometer DIL805D test apparatus and hollow specimen geometry (A)/Optimisation of force PID control parameters for hollow samples (B)

2. The DIL805D default programming capabilities are limited to a certain amount of test segments. Therefore, the DIL control software was extended by TA Instruments with a custom cycle generator module enabling the application of continuous thermal cycles with a constant mechanical stress superposition.

With this test-setup prepared, two types of tests were carried out in regard to the respective test matrices presented in Tables 1 and 2. At first, an extended cyclic re-hardening test is performed by applying sets of 25 thermo-mechanical load cycles with peak temperatures from 800 °C to 900 °C. Mechanical stress is superimposed with three levels in regard to the elastic limit k_{f0} of H11 tool steel determined at 900 °C. The aim of the test is to identify the lowest peak temperature where re-hardening effects can be observed by an increase in hardness. Also, this test is also used to investigate the relationship between the austenitisation behaviour characterised by TTA tests and the wear-relevant hardness. While the dilatometric TTA test used by Behrens et al. [10] is based on tactile measurements to identify phase transformation on a micrometre scale, the hardness evaluation features an optical measurement of indents for the determination of the hardness value. Therefore it must be assumed, that the detection resolution with this procedure is reduced, leading to the assumption that the measurable minimum temperature at which re-hardening occurs is higher compared to the TTA tests.

Table 1. Test matrix for the re-hardening study

Stress superposition [% k_{f0}]	Temperature range [°C]	Temperature Increment [°C]	Cycles	Repetitions
0	800–900	20	25	3
30				
50				
80				

Afterwards, thermo-mechanical loading tests with high cycle counts up to 2000 are carried out to estimate the effects during industrial use. Regarding peak temperatures 600 °C, 750 °C and 900 °C are used to ensure the formation of re-hardening as well as tempering effects. The thermal cycle profile using a heating rate of 500 K/s and the application of the mechanical stress superposition is identical in both parts of this study. Keeping a thermal cycle time of about eight seconds in mind, the repetition number had to be reduced to two because of the high testing time of over two hours per 1000 cycles.

Table 2. Test matrix for the high cycle loading tests

Peak temperatures [°C]	Stress superposition [%k_{f0}]	Investigated cycle numbers	Repetitions
600	0, 30, 50, 80	1, 10, 50, 100, 500, 1000, 2000	2
750			
900			

All tested samples are metallographically prepared for micro-hardness measurement at nine measuring points across the sample length as shown in Fig. 3-A. For this purpose, the samples are first cast in epoxy resin and subsequently wet grinded in several steps with SiC grinding paper ranging from a 220 to a 1200 grid. Afterwards the samples are polished three times using diamond suspension with an abrasive grain diameter of 6, 3 and 1 µm. Both operations are carried out on a Tegramin-30 sample preparation device by Struers. The embedded specimens are then etched with 5% nital acid for light microscopical images of the microstructure. For the micro-hardness measurement the standardised measuring method according to Vickers with a test load of 1.961 N (corresponds to HV0.2) is used.

3 Results and Discussion

3.1 Pretesting

In Fig. 3-B an exemplary overview of the microstructure after 10 loading cycles at 900 °C with 80% k_{f0} is presented showing a characteristic transition from the outer area of the specimen to the center. Because of heat transfer between the sample and the deformation punch leading to lower peak temperatures, the outer area is dominated by tempered ferrite with remaining martensite plates featuring a reduced hardness of 380 HV. The centre of the specimen, where the testing temperature is ensured, only consists of a fine-grained structure, which can be referred to as re-hardened martensite. Hardness in this area is significantly increased to 650 HV compared to the base hardness of 450 HV.

In this study hardness was only evaluated in the middle area of the specimen close to the central welding location of the thermocouple placed at evaluation point 5. This is achieved by statistically averaging the hardness values of the measuring points from position 4 to 6 while also calculating the standard deviation in this area to assess possible fluctuations in hardness.

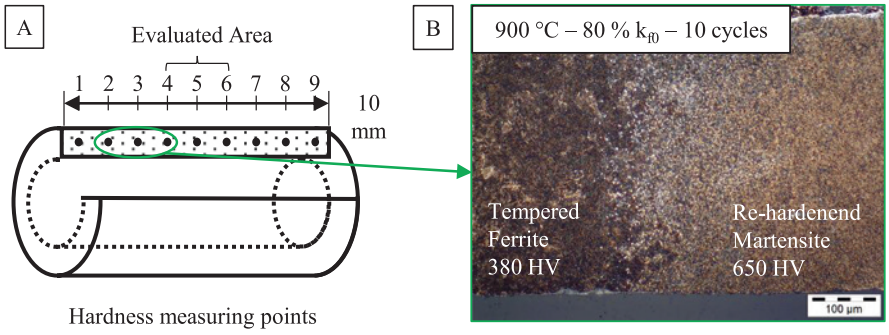

Fig. 3. Hardness measuring locations (A) and microstructural image of the transition area representing the hardness measuring locations from 2 to 4 showing tempered ferrite and re-hardenend martensite after thermo-mechanical loading (B)

3.2 Re-hardening Study

As assumed in Sect. 2, the results of the re-hardening study, plotted in Fig. 4, does not show a clear conformity to the measured Ac_1 temperatures via TTA testing. While without stress superposition (0% k_{f0}) re-hardening can initially be observed at temperatures over 880 °C, the superposition with 80% of the elastic limit k_{f0} will activate this effect already at temperatures of about 850 °C. In agreement with the results of Fig. 1-A the magnitude of the impact due to the mechanical stress level decreases with increased loading. While the measured data agrees with the finding that higher mechanical load leads to lower re-hardening start temperatures in theory, the difference between all results of the mechanical stress superposition tests are relatively minor in practice. The calculated standard deviation of ±15 HV are explained with the inherent measurement inaccuracy of the optical Vickers method. An exception is found in the hardness values of the 50% stress superposition series where the standard deviation values are significantly increased (approx. ±45 HV). The reason for this was found in the evaluation area defined in Fig. 3-A. While in the other test series at this zone either re-hardened or tempered microstructure was found exclusively, in the prominent test series a transitional microstructure comparable to Fig. 3-B was found. A possible explanation for this are slight offsets on the thermocouple welding location or a slight non-concentric placement of the sample in regard to the deformation punches of the dilatometer leading to deviations of the temperature field applied. Keeping this in mind, it must be concluded that the superposition with mechanical stress levels over 30% of k_{f0} leads to no distinct difference on the occurrence of re-hardening effects compared to each other.

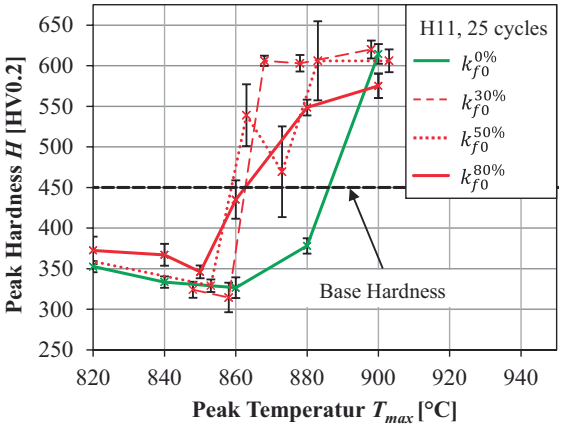

Fig. 4. Hardness over peak temperature for H11 after 25 thermo-mechanical loading cycles, heating rate: 500 K/s

3.3 High Cycle Loading

Because of the findings of the re-hardening study, only the results with mechanical loading of 80% k_{f0} and without additional loading are shown in Fig. 5 for the evaluation of the thermo-mechanical loading test at higher cycle counts and for better comprehension. In general, the results of measurements at 30% and 50% k_{f0}, which are not shown in this picture, are nearly identical to the results depicted below for 80% k_{f0}.

Regarding the three tested peak temperatures, three individual findings can be identified: After the loading at a peak temperature of 600 °C no measurable change in hardness could be observed under consideration of a minor standard deviation of less than 5 HV both with mechanical stress superposition and without. This finding indicates that the superposition applied has no influence on the material specific activation temperature for the occurrence of tempering effects.

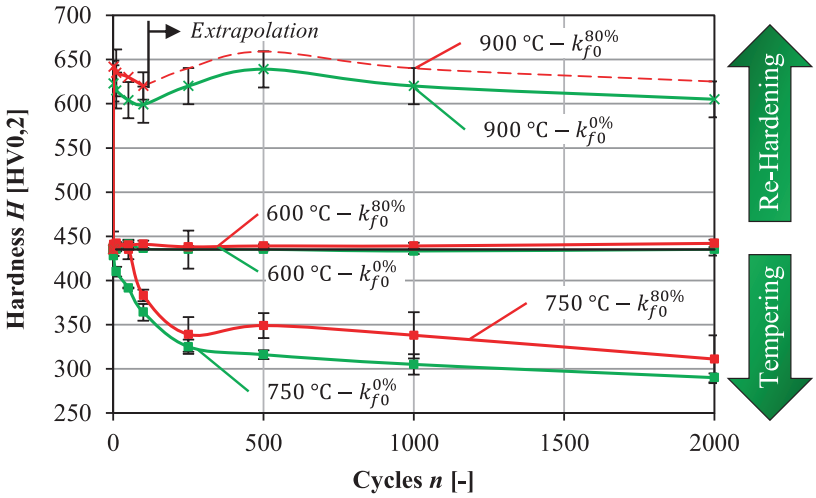

Fig. 5. Hardness results of the thermo-mechanical loading tests on H11 with cycle counts on a process relevant scale

At a peak temperature of 750 °C and up to 250 testing cycles no significant differences between all mechanical loading scenarios can be observed either. During all tests the hardness is reduced from 450 HV down to about 340 HV indicating tempering effects. However, after 500 cycles the reduction of hardness is slowed down by the application of mechanical stress superposition leading to a remaining hardness delta of approximately 30 HV over the full testing cycle spectrum. This finding is mainly explained by the diffusion properties of forcefully resolved carbon in the ordered martensitic matrix. At first, because of the high concentration gradient between matrix and microstructure, carbon can be transferred regardless of the overlaying mechanical stress leading to the identical drop in hardness. However, after the concentration compensation reached a critical point, the superposition with mechanical stress leads

to additional restraint on the martensitic structure slowing down the ongoing diffusion process. The standard deviation of the hardness values for this test series were also found to be in an acceptable range of less than ±10 HV indicating a homogenous microstructural distribution over the evaluation area of all related samples.

At a peak temperature of 900 °C, an immediate increase of the base hardness from 450 HV to over 600 HV is measured after all loading scenarios. During all tests at 900 °C with mechanical stress superposition a steady decrease of specimen length (about 4 µm per cycle) was also observed leading to severe deformation as shown in Fig. 6-A. To prevent a collapse of the specimen, which was found to be happen after a length decrease of about 500 µm, all tests with mechanical stress superposition had to be stopped after 100 loading cycles. Since the microstructure in the deformed area presented in Fig. 6-B is dominated by re-hardened martensite, as can be derived from Fig. 3-B, transformation-induced plasticity is determined as the reason for the sample deformation. This effect describes the occurrence of plastic deformations as a result of elastic mechanical stress during a phase transformation of the microstructure from austenite to martensite [12]. In the case of this study the transformation from martensite to the smaller austenitic structure during heating leads to a reduction of the sample length which is amplified by the mechanical load. During cooling the size increase in length direction due to the retransformation to martensite is also blocked by the deformation punches. Both effects combined cause an incremental reduction of the specimen length during each loading cycle. Still, up to a cycle count of 100, the measured hardness values were approx. 30 HV higher than the measurements with no external mechanical stress applied. However, because of the slightly increased standard deviation of both testing series of about ±15 HV, the influence of the superposed mech. loading is found to be minor in regard to the absolute achievable peak hardness. Nevertheless, the results of the test series with superposed mech. stress were extrapolated in accordance to the unloaded test results by adding a constant offset value of 30 HV to obtain a full data set for upcoming numerical material modelling.

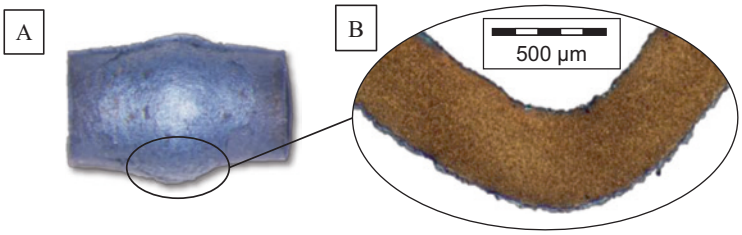

Fig. 6. Deformed specimen after 50 thermo-mechanical loading cycles at 900°C and 80% k_{f0} superposition (A) with martensitic microstructure formed by re-hardening (B)

4 Summary and Outlook

In the present study, the influence of mechanical stress superposition applied to thermal-cyclic experiments to reproduce the tool load in the surface layer while forging was investigated. In previous work, the decrease of the material characteristic Ac_1

temperature was already confirmed by dilatometric TTA tests. Now the occurrence of the associated martensitic re-hardening effects at reduced temperatures could be shown at the example of H11 tool steel. Regarding the two test series carried out in this study two main findings could be identified:

- While the TTA tests indicated a clearly measurable influence of the mech. stress level on the reduction of Ac_1, the results of the re-hardening study were only dependent on the amount of mech. stress to a lesser degree. In summary, it was found that as long as any mechanical load was applied, a significant reduction of about 20 °C to the minimum temperature necessary for re-hardening was observed.
- As a result of the high cycle loading tests, it was shown that tempering effects are influenced by an external mech. stress superposition, resulting in slower reduction in hardness. However, the maximum amount of hardness achievable due to re-hardening was found to be only marginally influenced by the application of mech. stress superposition.

These observations indicate for future work that for more precise wear estimations based on calculated process variables, the normal mechanical contact stress on the surface layer must also be taken into account. In the next step, a user subroutine for Simufact.forming 16 will be created to visualise the material data gathered in this study. Also, widely used nitride tool layers for additional wear resistance are in the focus of upcoming investigations. Since these layers represent a significant chemical modification of the surface layer, the austenitisation and the behaviour of the hardness evolution under thermo-mechanical load will be tested analogously to this study. Finally, laboratory forging tests are planned to validate the results of the material characterization and simulations.

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References

1. Jeong, D.J., Kim, D.J., Kim, J.H., Kim, B.M., Dean, T.A.: Effects of surface treatments and lubricants for warm forging die life. J. Mater. Process. Technol. **113**, 544–550 (2001)
2. Yu, Z., Kuznietsov, K., Mozgova, I., Böhm, V., Gretzki, T., Nürnberger, F., Schaper, M., Reimche, W.: Modeling the relationship between hardness and spray cooling parameters for pinion shafts using a neuro-fuzzy model strategy. J. Heat Treat. Mater. **67**(1), 39–47 (2012)
3. Caliskanoglu, D., Siller, I., Leitner, H., Jeglitsch, F., Waldhauser, W.: Thermal fatigue and softening behaviour of hot work tool steels. In: ICT Conference, Karlstad, 10.–13.09.2002, Issue 1, pp. 707–719 (2002)
4. Marumo, Y., Saiki, H., Minami, A., Sonoi, T.: Effect of the surface structure on the resistance to plastic deformation of a hot forging tool. J. Mater. Process. Technol. **113**(1–3), 22–27 (2001)
5. Klassen, A., Bouguecha, A., Behrens, B.-A.: Wear prediction for hot forging dies under consideration of structure modification in the surface layer. Adv. Mater. Res. **1018**, 341–348 (2014)

6. Kim, D.H., Lee, H.C., Kim, B.M., Kim, K.H.: Estimation of die service life against plastic deformation and wear during hot forging processes. J. Mater. Process. Technol. **166**, 372–380 (2005)
7. Bernhart, G., Brucelle, O.: Methodology for service life increase of hot forging tools. J. Mater. Process. Technol. **87**(1–3), 237–246 (1999)
8. Kannappan, A.: Wear in forging dies - A review of world experience. Metal Form. **36**(12), 335–343 (1969)
9. Persson, A., Hogmark, S., Bergström, J.: Temperature profiles and conditions for thermal fatigue cracking in brass die casting dies. J. Mater. Process. Technol. **152**, 228–236 (2004)
10. Malik, I.Y., Lorenz, U., Chugreev, A., Behrens, B.-A.: Microstructure and wear behaviour of high alloyed hot-work tool steels 1.2343 and 1.2367 under thermo-mechanical loading. In: Materials Science and Engineering, vol. 629 (2019)
11. Behrens, B.-A., Puppa, J., Acar, S., Gerstein, G., Nürnberger, F., Lorenz, U.: Development of an intelligent hot-working steel to increase the tool wear resistance. In: Tooling 2019 Conference & Exhibition, 13.05.–16.05.2019, p. 64 (2019)
12. Behrens, B.-A., Chugreev, A., Kock, C.: Macroscopic FE-Simulation of residual stresses in thermos-mechanically processed steels considering phase transformation effects. In: XIV International Conference on Computational Plasticity, Fundamentals and Applications COMPLAS (2019)