Direct or indirect: Influence of type of retained austenite decomposition during tempering on the toughness of a hot-work tool steel

Christoph Lerchbacher [a],*, Silvia Zinner [b], Harald Leitner [c]

[a] Christian Doppler Laboratory of Early Stages of Precipitation, University of Leoben, Franz-Josef-Straße 18, A-8700 Leoben, Austria
[b] Böhler Edelstahl GmbH & Co AG, Mariazellerstraße 25, A-8605 Kapfenberg, Austria
[c] Department of Physical Metallurgy and Materials Testing, University of Leoben, Franz-Josef-Straße 18, A-8700 Leoben, Austria

ARTICLE INFO

ABSTRACT

A heat treatment slightly differing from the classical 2×2 h tempering treatment has been established in order to eliminate the direct retained austenite decomposition during tempering of the common hot-work tool steel X38CrMoV5-1 . Instead of the direct decomposition into ferrite and cementite during the first tempering step, transformation into martensite has been forced. A quenching dilatometer has been used for heat treatment and for the determination of transformation reactions. The two heat treatments have been compared with respect to toughness behaviour by conducting Charpy-impact tests. The investigations have been performed on samples hardened with quenching rates 5 K/s, 0.5 K/s, and 0.25 K/s, providing different amounts of retained austenite within the as-quenched microstructure. The heat treatment modification does not show improvement regarding the toughness behaviour in case of low cooling rates where the specimen failure is dominated by interfaces, hence, the former interlath retained austenite films. Therefore, the indirect retained austenite decomposition has no positive effect compared to the direct decomposition. In case of the highest cooling rate the failure is dominated by the matrix and the impact toughness could be improved by a factor of 12% at the same hardness level and a dwell time reduction of 15%. For tools of small dimensions where these cooling rates during hardening can be achieved this heat treatment modification should be considered.

1. Introduction

The classical heat treatment of tool steels consists of two procedures called hardening and tempering. Hardening means austenitisation for a specific time followed by rapid quenching, which leads to a martensitic microstructure. Subsequent multi-step tempering leads to the relaxation of the stressed martensite, the formation of secondary carbides and the decomposition of retained austenite [1]. The more-step strategy enables the relaxation of untempered martensite newly formed from untransformed retained austenite during cooling after the first tempering step [2]. In order to gain required material properties, predominantly hardness and toughness, hardening as well as tempering parameters have to be chosen from a narrow range. This is hardly achievable when tool dimensions become large and results in insufficient toughness behaviour.

When no grain-boundary carbides are formed during quenching from austenitisation temperature, as it is the case for the investigated hot-work tool steel, austenite retained after quenching has been found to be responsible for undesired toughness behaviour [3]. Ultra-thin carbon enriched retained austenite films between martensitic laths lead to carbide formation at lath boundaries during tempering [4] which reduces toughness. The amount of retained austenite after quenching strongly depends on the quenching rate and goes hand in hand with a thickness increase of the interlath films [5], hence, the potential for interfacial carbides increases. Regarding large tool dimensions, the amount of retained austenite after quenching is not adjustable in any order by the manufacturer for a given alloy composition. Therefore, to manipulate toughness properties, retained austenite decomposition during tempering could be manipulated.

Decomposition of retained austenite in steels is generally proposed to be a transformation into ferrite and cementite [6] or the formation of bainite [7]. Van Genderen et al. [8] divided the retained austenite decomposition of a FeC model alloy into two successive ways. A preceding ferrite formation occurs before the final transformation into ferrite and cementite takes place. Saha Podder and Bhadeshia [9] showed by means of dilatometer tests that during heating at 450 °C of a bainitic steel ferrite formation does not occur within 1 h of tempering. Previously published work on the investigated hot-work tool steel X38CrMoV5-1 shows that heating to 610 °C already leads to alloy carbide formation from

the retained austenite though no reaction is visible from the dilatometer curves [4]. This carbide formation progresses during tempering. Similar to that, Kulmburg et al. [2] proposed for a high speed steel that carbides form during tempering from the retained austenite which transforms into martensite during cooling. No transformations take place during heating to tempering temperature. All these findings demonstrate that the behaviour of retained austenite during tempering significantly varies for different steel grades and lead to the assumption that individual heat treatments for individual steel grades might improve material properties.

However, traditional multi-step tempering is somehow manifested in steel industry. The intention of the present work is the evaluation of different heat treatment strategies with special regard laid on the retained austenite decomposition. Therefore, the standard tempering treatment has been compared with respect to toughness to a treatment where the retained austenite has been transformed into martensite instead of a direct decomposition into ferrite and cementite. Additionally, the effect of retained austenite decomposition manipulation has been determined with respect to different cooling rates during hardening.

2. Experimental

Dilatometric experiments have been performed in a dilatometer Dil 805A from Bähr Thermoanalyse GmbH. Therefore, samples of 15 mm in length and 5 mm in diameter have been produced. The material has been austenitised at 1020 °C with a dwell time of 30 min. Subsequent quenching has been conducted with quenching rates $\lambda = 0.6$, 6 and 12, which correspond to linear quenching rates of 5 K/s, 0.5 K/s and 0.25 K/s, respectively. Then, the samples have been tempered at 610 °C two times for 2 h simulating the standard heat treatment. The relative length change (relative to the origin length of the sample) of the samples during tempering has been recorded. From these results, modified tempering parameters have been chosen.

Instrumented impact tests conducted on a 300 J impact pendulum from Zwick were performed in order to evaluate the impact toughness of the samples. The mean absorbed energy was measured using 3 specimens per sample state. For the impact tests Charpy U-notched specimens with dimensions of $55 \times 10 \times 7.5$ mm^3 were used. The samples have been heat treated using similar parameters as mentioned above conducting a special instrumentation in the quenching dilatometer in order to simulate real quenching conditions and to provide comparability with the samples heat treated for the length change measurements. The impact tests have been performed at sample temperatures of 200 °C, a common preheating temperature for tools of this material in use. Fracture surface analysis was carried out by scanning electron microscopy using a Zeiss EVO50.

3. Material

The nominal composition of the alloy investigated is given in Table 1. The initial microstructure after hardening consists of a martensitic matrix with nanometric interlath retained austenite films [5]. The volume fraction of retained austenite is $< 3\%$ [3],

Table 1
Nominal composition of the investigated alloy X38CrMoV5-1.

	C	Si	Mn	Cr	Mo	V	Fe
wt%	0.37	1.11	0.41	4.90	1.23	0.33	Bal.

10% [5] and 17% [3] corresponding to the quenching rates $\lambda = 0.6$, 6 and 12, respectively.

4. Results

4.1. Dilatometer

Tempering at 610 °C of the differently quenched samples leads to a dilatation as given in Fig. 1a. The diagram shows the dilatation corresponding to the section from reaching 610 °C to the end of 2 h dwell time. The relative expansion during tempering qualitatively shows similar behaviour for the samples with quenching rates $\lambda = 0.6$, 6 and 12. In order to compare the behaviour of the different samples, the corresponding curves have been translated in a way that at tempering start point the curves have the same level, therefore, no numerical scaling on the vertical axis is given. First, a volume decrease occurs until a minimum is reached after approximately 1600 s of tempering. This volume decrease corresponds to alloy carbide formation within the matrix and within the retained austenite as stated in earlier studies [4,9], hence, carbon enriched retained austenite and matrix become depleted in carbon. Further tempering leads to a volume increase approaching a maximum in volume, corresponding to retained austenite decomposition [9]. The minimum corresponding to each curve has been chosen as base for the calculation of relative changes during tempering. The curves behave quite similar for the range from tempering start to the volume minimum, showing a volume decrease of 0.76%. The extent of the following volume increase beyond the minimum depends on the quenching rate during hardening. The relative volume change with respect to the minimum is 0.5%, 1.27% and 1.62% after tempering for 2 h corresponding to the samples quenched with $\lambda = 0.6$, 6 and 12, respectively. This cooling rate dependent extent of volume increase is most likely related to the amount of retained austenite within the microstructure. In order to prove this, cooling curves after selected times of tempering corresponding to the sample with the most pronounced volume change ($\lambda = 12$) have been recorded. Fig. 1b shows the corresponding cooling curves after tempering for 600 s (10 min), 1600 s (25 min), 3600 s (1 h) and 7200 s (2 h). Again, the curves have been vertically translated in a way to provide an optimum in comparability. The curve corresponding to the sample cooled after 10 min tempering at 610 °C depicts a pronounced reaction showing a volume increase with a start point at approximately 350 °C, indicating the martensitic transformation of the retained austenite. The sample cooled from the minimum (after 25 min) again shows this reaction, but more pronounced and shifted to approximately 370 °C. Cooling after 1 h tempering shifts the reaction to approximately 390 °C and the reaction is not well distinctive. During cooling after 2 h of tempering at 610 °C no reaction occurs. The continuous increase of the martensite start temperature indicates that carbon depletion within the retained austenite develops at least up to 1 h of tempering. The well distinctive reaction during cooling after tempering for 25 min shows that the majority of the retained austenite has a carbon level which allows for a martensitic transformation.

From these results a modified tempering treatment has been chosen. The dwell time of the first tempering step is set to 1600 s (25 min) which corresponds to the volume minimum in the dilatation curve, in order to avoid the direct decomposition of the retained austenite into ferrite and cementite. The dwell time for the second tempering step is set to 3 h, a value which provides similar hardness after heat treatment as in the case for traditional 2×2 h treatment. The different treatments are depicted schematically in Fig. 2. After the first tempering step, the samples are oil-