

# Relationship between processing parameters, alloy atom diffusion distance and surface hardness in laser hardening of tool steel

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## Abstract

Laser transformation hardening of tool steel containing 0.64 wt% carbon and 4.48 wt% chromium was studied. The delivery condition of the steel was annealed, with a microstructure of  $M_7C_3$  type carbides in a matrix of ferrite. Hardening was performed with a direct diode laser. Surface temperature during hardening was measured with a dual wavelength pyrometer and the traverse speed was varied between 1.1 and 33.3 mm s<sup>-1</sup>. A set of samples was quenched in liquid nitrogen to reveal the effect of retained austenite on the resulting hardness.

The surface hardness was greatly influenced by the dissolution of carbides during austenitization. The dissolution of chromium carbides required a high temperature and was assumed to be limited by chromium diffusion. Highest hardness values were achieved with slow traverse speeds and high surface temperatures for both normal and sub-zero hardened samples. Highest measured hardness in samples, in which the surface was cooled by self-quenching, was 822 HV and in samples quenched in liquid nitrogen 926 HV. A good agreement was found between the characteristic diffusion distance of chromium and surface hardness, and the results of this study can be used as a basis for mathematical modelling of laser transformation hardening.

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## 1. Introduction

Laser transformation hardening is a process in which phase transformations in steels are induced by irradiating the surface of a component with a laser beam. Surface regions are heated to temperatures above the  $Ac_1$  transformation temperature, resulting in austenitization. The surrounding material acts as an efficient heat sink. Heat is transported away from the surface by thermal conduction, inducing rapid cooling. Austenitized material, provided with sufficient quantities of carbon, forms martensite upon quenching, producing a hard and wear resistant surface. The properties of the hardened layer are controlled by the energy input, which depends on the beam power density and interaction time. Long interaction times enable more heat to be conducted into the material before the melting temperature is attained at the surface, thus producing a deeper hardened layer. Higher traverse rate and power density result in a faster cooling rate, at the expense of the depth of the hardened layer.

The geometry of the hardened layer depends on the processing parameters and beam-matter interaction, and their effect on the thermal cycle induced by laser radiation. Consequent phase transformations, which determine the hardness of the resulting microstructure, are affected by material composition, initial microstructure and processing parameters.

### 1.1. Phase transformations during laser hardening

The aim of laser transformation hardening is to produce a hard, wear resistant surface in discrete areas of a component. In this aspect martensite  $\alpha'$  is a favoured microstructure, since it is the phase which produces the highest hardness and strength in steels. It is produced from austenite during rapid cooling, and the properties of the formed martensite depend greatly on the properties of austenite prior to quenching.

Austenitization occurs when the temperature of the material exceeds its characteristic austenitization temperature. The temperature at which austenite starts to form in equilibrium conditions is the  $Ae_1$  temperature. In higher temperatures the volume fraction of austenite increases at the expense of other phases, until the material is fully austenitic. The temperature at which the

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transformation to austenite is complete is the  $A_{e3}$  temperature. If heating conditions apply, the transformation temperatures are raised. Then designations  $A_{c1}$  and  $A_{c3}$  are used to represent the critical temperatures [1].  $A_{c1}$  temperature increases slightly with increasing heating rate. This movement is quite insensitive to carbide distribution and carbon content. In contrast to the elevation of the  $A_{c1}$  temperature, the  $A_{c3}$  temperature is structure sensitive and varies considerably with the heating rate [2].

Austenitization of a microstructure consisting of spherical carbides in a matrix of ferrite,  $\alpha$ , begins by nucleation of austenite,  $\gamma$ , about carbides in  $\alpha/\alpha$  grain boundaries [3]. Austenite film grows until it completely envelopes the carbide. It has been assumed that if the austenitization temperature of the surrounding ferrite is not exceeded, further growth of austenite occurs only by carbon diffusing through the austenite envelope from the dissolving carbide to the advancing  $\alpha/\gamma$  boundary. Carbon diffusion through the austenite nodule is the rate controlling process [4,5].

At sufficiently high temperatures ferrite transforms into austenite without carbon diffusion. In such a case, the properties of the formed austenite depend mainly on the extent of carbide dissolution during austenitization. It has been presented that in the case of alloy carbides, the dissolution rate is determined by the diffusivity of the metallic alloy elements [6,7]. Therefore, the austenitization temperature has a pronounced effect on dissolution due to the temperature dependence of diffusion of carbide forming alloying elements [8]. Alloy carbides are thermodynamically more stable in high temperatures and may be unaffected by the thermal cycle [9], or complete dissolving may require long austenitization times and high temperatures.

Since diffusion is the rate controlling process in carbide dissolution, the extent of change depends on the number of diffusive jumps which take place during the thermal cycle [10]. The number of atoms diffusing in unit time across unit area through a unit concentration gradient is known as the diffusivity or diffusion coefficient. It depends on the temperature and concentration of the alloy [11]. The variation of diffusion coefficient  $D$  ( $\text{m}^2 \text{s}^{-1}$ ) with temperature is given by

$$D = D_0 \exp \left\{ - \left( \frac{Q}{RT} \right) \right\}, \quad (1)$$

where  $D_0$  is the frequency factor ( $\text{m}^2 \text{s}^{-1}$ ),  $Q$  the activation energy ( $\text{J mol}^{-1}$ ) and  $R$  is the gas constant ( $\text{J mol}^{-1} \text{K}^{-1}$ ). Under isothermal conditions the characteristic diffusion distance  $d$  (m), assumed to be predominantly controlled by volume diffusion in austenite, is given by [12]

$$d^2 = 2Dt. \quad (2)$$

For a thermal cycle  $T(t)$  the characteristic diffusion distance is given by [12]

$$d^2 = 2D_0 \int_{t_2}^{t_1} \exp \left\{ - \left( \frac{Q}{RT(t)} \right) \right\} dt. \quad (3)$$

Carbon may also diffuse before the  $\alpha/\gamma$  transformation [13], but in such low temperature the diffusivity is sluggish and the effect caused by diffusion in the  $\alpha$ -phase is insignificant.

Dissolution rate of carbides is relevant since the carbides act as sources of carbon to the austenite. Sufficient amount of carbon in solution in austenite is required in order to induce martensite reaction. Carbon in solution in austenite also has the greatest effect on the hardness of the martensite formed during rapid cooling [14]. Therefore, it can be assumed that the extent of carbide dissolution has a pronounced effect on the hardness of formed martensite.

In high-alloy steels, a substantial portion of carbides must remain undissolved to avoid excessive amounts of retained austenite in the product microstructure. Such rapid austenitizing, followed by quenching may develop a unique microstructure comprised of small undissolved carbides in a matrix of ultrafine-grained martensite [15].

Martensitic transformation is diffusionless and occurs upon cooling at rates rapid enough to suppress the diffusion-controlled transformations [1]. Martensite reaction begins at the martensitic start temperature  $M_s$  which can vary from as high as 773 K to well below room temperature. This variation depends on the concentration of alloy elements. The reaction ceases at the  $M_f$  temperature, or the martensitic finish temperature. At this temperature all the austenite should have transformed to martensite. A small portion of austenite may still remain in the structure. This is called retained austenite  $\gamma_{\text{ret}}$ . If the  $M_f$  temperature is below the quenching temperature, more untransformed austenite is retained in the structure. The fraction of retained austenite can then be decreased by sub-zero quenching.

Precipitation and carbide coarsening may occur even at high cooling rates experienced during laser transformation hardening [9,16,17]. This might have an effect on the hardness and properties of the quenched microstructure.

## 1.2. Objective

The objective of this study was to examine the effect of processing parameters on the hardness of laser transformation hardened tool steel. Self-quenched and sub-zero quenched samples were processed with various parameters and the effect of processing parameters was established by hardness testing and metallography.

## 2. Experimental procedure

A high power diode laser was used for hardening. The laser comprised stacks of separate diodes delivering radiation at wavelengths of  $800 \pm 10$  and  $940 \pm 10$  nm. For each power level, half of the delivered laser power comprised 800 nm radiation and half 940 nm radiation. The maximum nominal power of the laser was 3000 W. The time to attain full power was 10 ms, and the time from full power to power-off was also 10 ms. Focal length of the optic was 500 mm and the measured spot size, calculated from the second moment, was 5.3 mm in the traverse direction and 12.3 mm across the traverse direction. Fig. 1 shows the measured intensity profile of the beam.

The laser was mounted on a six-axis industrial articulated robot. The angle between the laser's optical axis and the surface was set to  $85^\circ$  to reduce the amount of power reflected back to the laser. The experimental set-up is illustrated in Fig. 2.

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