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# Thermal diffusivity of traditional and innovative sheet steels

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## Abstract.

The low carbon steels, used for the production of car bodies by deep drawing, are gradually substituted by high strength steels for vehicle weight reduction. The drawn car body components are joined by welding and the welded points undergo a reduction of the local tensile strength.

In developing an accurate welding process model, able to optimized process parameters and to predict the final local microstructure, a significant improvement can be given by the knowledge of the welded steels thermal diffusivity at different temperatures.

The laser-flash method has been used to compare the thermal diffusivity of two traditional deep drawing steels, two high strength steels already in common usage, i.e. a Dual Phase (DP) steel and a TRansformation Induced Plasticity (TRIP) steel, and one experimental high-Mn austenitic TWIP (Twinning Induced Plasticity) steel.

The low carbon steels, at low temperatures, have a thermal diffusivity that is 4-5 times larger than the TWIP steel. Their thermal diffusivity decreases by increasing temperature while the TWIP steel shows an opposite behaviour, albeit with a lesser slope, so that above 700 °C the TWIP thermal diffusivity is larger. The different behaviour of the TWIP steel in respect to the ferritic deep drawing steels arises from its non ferro-magnetic austenitic structure. The DP and TRIP steels show intermediate values, their diffusivity being lower than that of the traditional deep drawing steels; this latter fact probably arises from their higher alloy content and more complex microstructure.

## Introduction

Innovative high-strength steels are nowadays often used in the automotive industries, instead than traditional steels such as the FeP05 grade (EN 10130 standard), both to reduce the vehicles' weight (and hence to save fuel and reduce the air pollution) and to improve the vehicle occupant safety in case of car crash. The main usage requirements of these steels are fracture strength and energy absorption. These steels are produced by using thermo-mechanical processes, in which the rolling temperatures and deformations and the ensuing continuous annealing cycles are strictly controlled. The fast cooling stages of the last process has allowed to obtain a larger variety of microstructures than in the static (batch) annealing process, which was the norm until a few years ago, and in which sheets are annealed in a furnace after being coiled. With this method, several different high-strength steels for automotive uses can be fabricated: TRAsformation Induced Plasticity (TRIP) steels, TWinning Induced Plasticity (TWIP), Dual-Phase (DP).

TRIP steels contain a fraction of retained austenite, which is transformed to martensite during a plastic deformation at room temperature [1]

The TWIP (TWinning Induced Plasticity) steels are alloyed with 15 - 25 wt.% Mn, to obtain a fully austenitic microstructure after a thermomechanical treatment [2].

DP steels consist of ferrite and martensite and are obtained from a final treatment at an intercritical temperature, followed by a fast cooling; the martensite fraction depends from the final holding temperature, the chemical composition and the austenitic grain size [3].

Interstitial Free (IF) steels are alloyed with elements such as Nb and Ti that remove the interstitial elements C and N from solid solution, to favor deep drawing [4].

Car bodies are built by assembling several deep-drawn parts by spot welding. The mechanical properties of the welded joints depend on the microstructures obtained in the melted zone and in the heat-affected zones, which in turn depend upon the thermo-physical properties and the phase-transformation kinetics of the steels that are joined (steels which are often dissimilar) [5].

The thermal diffusivity of the steels is one of the most important parameters not only in the heat treatments [6-9] and in the welding [10] but also during the use of the components, since it influences the wear resistance [11-12]. Nevertheless scarce information are found in literature on the thermal diffusivity of high strength steels. Therefore in this research such property has been evaluated at different temperatures, in order to compare the behavior of the different high strength steels with that of the traditional low carbon deep drawing steels.

The thermal diffusivity was evaluated by the laser flash method, first developed by Parker et al. [13]: a disk specimen is subjected to a pulse of radiant energy at one surface (front) and the thermal diffusivity is computed from the temperature transient on its opposite (rear) surface.

## Experimental

The following steels are investigated: two traditional deep drawing steels, namely a FeP05 and a IF steels, three high strength steels, i.e. two DP steels having 600 and 1000 MPA tensile strength respectively and a TRIP steel, and of one experimental TWIP steel. The chemical composition of the steels is reported in Table I. The examined steel specimens were disks of diameter 10 mm, which were cut from the cold rolled and heat treated steel sheets with full thickness (about 1 mm).

	<b>C</b>	<b>Si</b>	<b>Mn</b>	<b>Cr</b>	<b>Mo</b>	<b>Ni</b>	<b>Ti</b>	<b>V</b>	<b>P</b>	<b>S</b>	<b>Fe</b>
<b>TRIP</b>	0.320	1.600	1.950	0.035	<0.001	0.023	0.005	<0.001	0.017	0.002	Bal.
<b>DP600</b>	0.130	0.050	1.275	0.036	0.19	0.027	0.004	0.002	0.016	<0.001	Bal.
<b>DP1000</b>	0.200	0.220	1.900	0.53	0.14	0.051	0.003	0.009	0.013	0.001	Bal.
<b>TWIP</b>	0.58	0.20	21.650	0.19	0.03	0.070	---	0.250	0.022	<0.001	Bal.

<b>FeP05</b>	0.011	<0.004	0.063	0.012	0.002	0.006	0.029	<0.001	0.010	0.005	Bal.
<b>IF</b>	0.017	0.071	0.633	0.023	<0.001	0.014	0.057	0.003	0.037	0.012	Bal.

Table 1: Chemical composition of the examined steels, (wt. %)

The thermal diffusivity measurements were performed in a water cooled vacuum chamber equipped with an alumina specimen holder and a PG-PBN (Pyrolytic Graphite and Pyrolytic Boron Nitride) furnace. Each specimen was heated continuously at 5 °C/min from room temperature to 1000 °C and the measurements were performed every about 20 °C. The specimen were then held at 1000 °C for a few minutes and some other measurements were carried out.

The energy pulse was provided through a quartz window by a Nd-Yag laser (1.064μm) and the temperature transient was measured with a welded K-type thermocouple. A modified mathematical model [6] was adopted to overcome the requirement of initial thermal equilibrium between the sample and the furnace, which is assumed by the original Parker et. al. model [13].

The samples, both in as received condition and after the heat treatments induced by the thermal diffusivity tests, were examined by X-ray diffraction (XRD) with Co radiation, by optical metallography after Nital etching, and by microhardness tests. Moreover, TRIP and TWIP samples were heated at 250, 500, 750 and 1000 °C, in the same conditions in order to evaluated the change in the microstructure at the intermediate temperatures, and were similarly examined.

## Results

The X-Ray diffraction peaks of the  $\alpha$ -ferrite appear in all examined samples, except the as-fabricated TWIP steel. In most samples, as shown in the metallographic analyses below, other phases are also present, i.e. carbides of iron and alloying elements, martensite and austenite. Owing to the reduce amount of carbon, the quantity of carbides present or formed during the heating and cooling is small in such a manner that their presence was quasi undetectable by XRD, but in the DP steels, after heating at 1000 °C and in the TRIP steel heated at 500-750 °C very low intensity peaks, referred to cementite are also present. In the as received DP 600 and in the TRIP steel up to 500 °C (Fig. 1a), a little amount of austenite was also XRD detected. The deformation of the cubic into tetragonal cell of the iron due to the formation of martensite was not XRD detected, hence either the quantity of martensite is scarce, or the martensite carbon content is limited, such that the martensite peaks are weak and are overcome by the tails of the  $\alpha$ -ferrite ones. In the TWIP steel (Fig. 1b) the austenite is the prevalent phase; however in the sample heated at 1000°C there are also peaks referred to  $\epsilon$  martensite and  $\alpha$ -ferrite. These last phases are also detected in smaller amount in the samples heated at 500 °C. Furthermore, the 1000 °C heating eliminates the preferential orientation induced by the fabrication process.

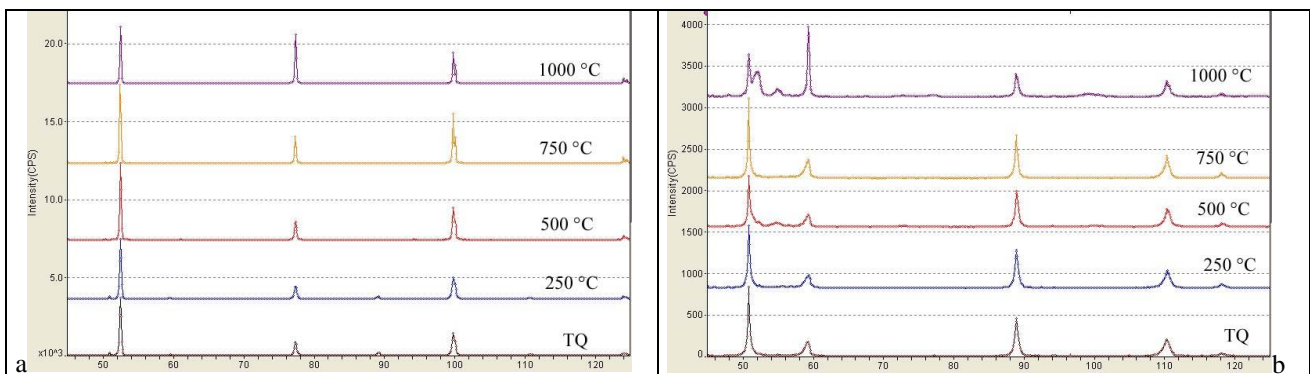


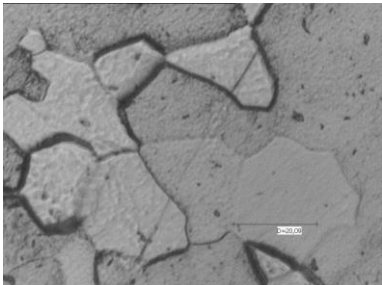
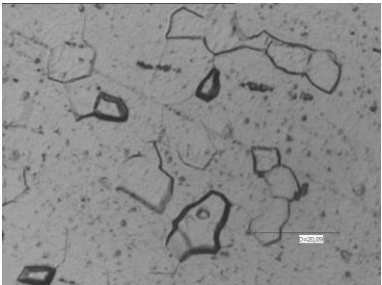
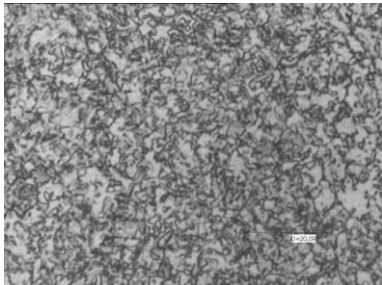
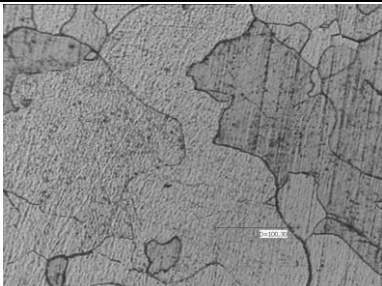
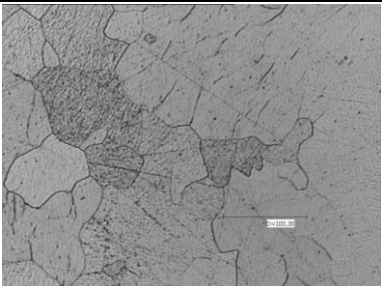
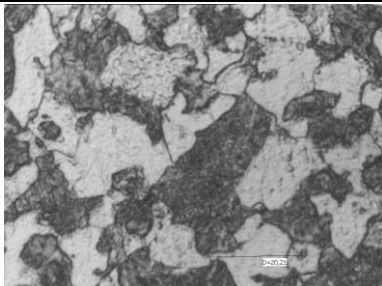
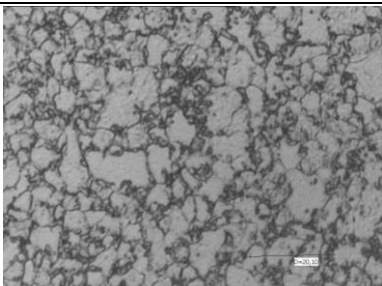
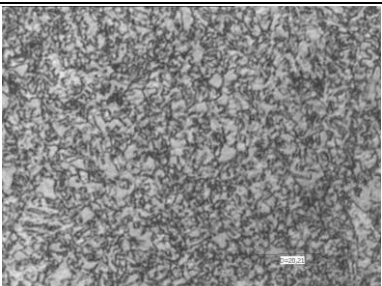
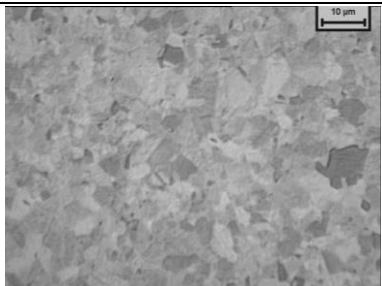
Fig. 1: XRD patterns of TRIP (a) and TWIP (b): as-received (TQ) and after different heating cycles.

The metallographic observations effected after the different heating conditions (Fig. 2), show that

the very slow cooling, after the heating at the highest temperature, produced large ferrite grains, and perlite/bainite structures. The main effect of the heat treatment is to enlarge the grain size, mainly in the deep drawing steels. In fact in these materials carbides are almost absent and consequently the grain growth may not be controlled by the presence of second phases like the precipitates. This statement is confirmed by the different behaviour of the two DP steels: the DP600 (with lower carbon and alloying elements content) always exhibit a larger grain size than the DP1000 (Fig. 2). These grain growth and second phase coarsening effects are confirmed by the decrease of the hardness after the heat treatment, Fig. 2, which is more pronounced in the less alloyed steels, in which the grain growth is not hindered. The second effect of the progressive increase of the temperature is the transformation of the retained austenite, as shown by XRD, into ferrite and carbides, and the coalescence of the precipitates, as visible in Fig. 2. A different behaviour has been observed for the TWIP steel: after heating, large austenitic grains were formed in which twins and  $\epsilon$  martensite platelets are presents [14] (Fig. 2); in this steel after the 1000 °C heating the hardness is 170 HV for the austenite and 192 HV for regions rich of  $\epsilon$  martensite.

The thermal diffusivity results (Fig.3) show almost the same behaviour for the two traditional steels (FeP05 and IF) and for the DP and the TRIP steels. For all of these steels the thermal diffusivity decreases by increasing temperature up to approximately 750 °C. Above 750 °C and up to 1000 °C the thermal diffusivity for the FeP05 becomes nearly independent of temperature while for the other three steels it slowly increases. The TWIP alloy shows an opposite behaviour and its thermal diffusivity slowly increases by increasing temperature.

The different behaviour in the temperature - thermal diffusivity curves for the examined steels is probably due to the dependence of the thermal diffusivity on the crystal structure, the microstructure and the magnetic properties.

As received	 FeP05, 92x68 μm, 95 HV	 I.F., 92x68 μm, 127 HV	 TRIP, 92x68 μm, 265 HV
After heating at 1000°C	 FeP05, 459x344 μm, 92 HV	 I.F., 459x344 μm, 94 HV	 TRIP, 92x68 μm, 205 HV
As received	 DP600, 92x68 μm, 193 HV	 DP1000, 92x68 μm, 357 HV	 TWIP, 90x64 μm, 322 HV

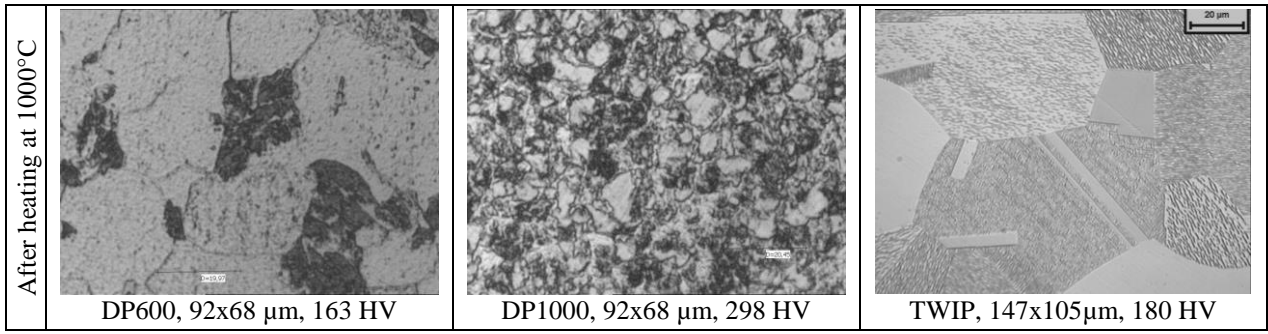


Fig. 2: Microstructures and HV<sub>0.1</sub> microhardness of the steels, as received and after 1000 °C heating.

For all the examined steels, except the TWIP, the thermal diffusivity decreases with increasing temperature where the crystal structure is bcc (body centred cubic). The Armco iron thermal diffusivity values taken from literature show a similar trend [15]. For these steels a thermal diffusivity minimum value is reached around the Curie point (770 °C). Beyond this point the thermal diffusivity starts to increase. This agrees with the literature which states that if a ferromagnetic transition exists in a metal, thermal diffusivity first decreases from room temperature to the Curie temperature, and then it increases [16]. Moreover the decrease in thermal diffusivity may be attributed to the density of crystal vacancies, which increase with the temperature. It wasn't noted a sensible variation of the thermal diffusivity due to the phase transition from bcc to fcc (face centred cubic), which it is expected to occur at temperatures between 840 °C and 910 °C in all the examined steels, except the TWIP steel. However, for pure Iron the described thermal diffusivity variation results is less than 0.5 mm<sup>2</sup>/s [15], hence it may be hindered by the experimental uncertainty, which in the present experiments increases by increasing the temperature.

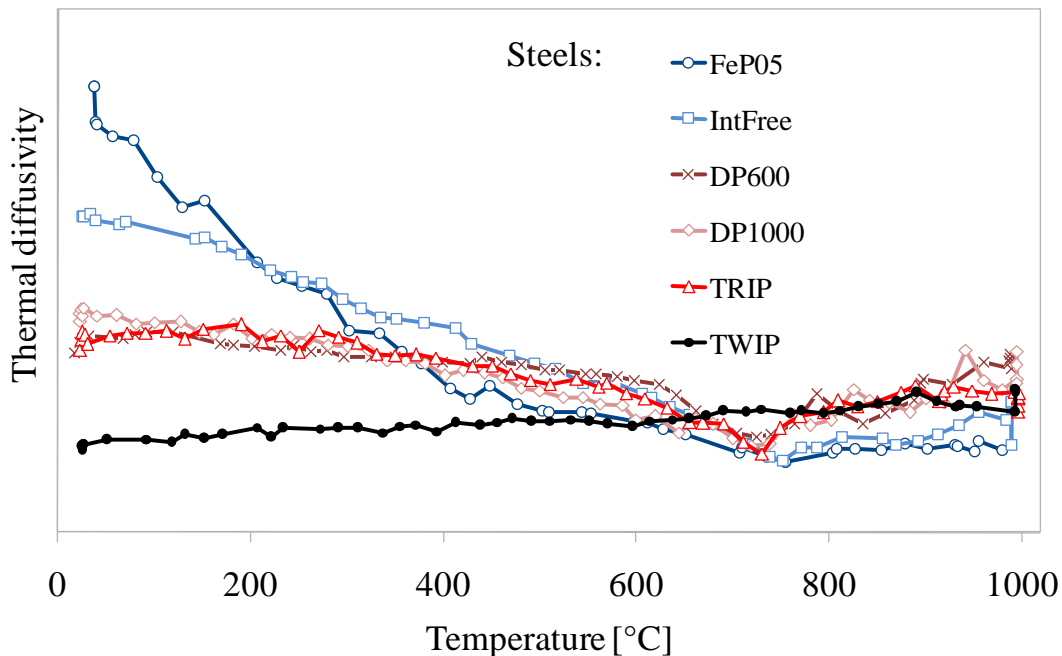


Fig. 4: Thermal diffusivity of the examined steels vs. temperature

By comparing the five examined ferritic steels (i.e. all excepted the TWIP steel) at temperatures below the Curie temperature it is noted that thermal diffusivity is generally higher for those that have a lower amount of Carbon and other alloy elements. Particularly the FeP05 shows the higher values at low temperatures, followed by the IF steel, whereas the values of the three most alloyed ferritic steels (DP600, DP100, TRIP) are roughly one half lower. However, the decreasing in

thermal diffusivity with temperature is steeper for the FEP05 than for the other ferritic steels.

On the contrary, above the Curie temperature the values for the different steels get closer, even if the thermal diffusivity is somewhat higher for the most alloyed ferritic steels.

The TWIP steel, which has a fcc structure and it is not magnetic, shows only a moderate increase of the thermal diffusivity in the examined temperature range. Its thermal diffusivity values are the lowest ones from room temperature up to the Curie temperature of the ferritic steels, while after this point they result to be closer to the values measured for the other steels.

The increase of the thermal diffusivity values with temperature observed in the TWIP steel was previously observed also in some other high-alloy austenitic steels [17-19] and in low-alloy metastable austenite [6,20].

## **Conclusions**

The laser-flash method has been used to compare the thermal diffusivity of two traditional deep drawing steels, two high strength steels, Dual Phase (DP) steel and a TRansformation Induced Plasticity (TRIP) steel, and one experimental high-Mn austenitic TWIP (Twinning Induced Plasticity) steel.

The low carbon steels, at low temperatures, have a thermal diffusivity that is 4-5 times larger than the TWIP steel. Their thermal diffusivity decreases by increasing temperature while the TWIP steel shows an opposite behaviour, albeit with a lesser slope, so that above 700 °C the TWIP thermal diffusivity is larger. The different behaviour of the TWIP steel in respect to the ferritic deep drawing steels arises from its non ferro-magnetic austenitic structure. The DP and TRIP steels show intermediate values, their diffusivity being lower than that of the traditional deep drawing steels; this latter fact probably arises from their higher alloy content and more complex microstructure.

The effect of the heating on the microstructure is to enlarge the grain size and favour the second phases coarsening and the transformation of the retained austenite into ferrite and carbides. In the case of TWIP steel the 1000 °C heating favours the formation, after cooling, of the  $\epsilon$  martensite.

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