CONTRIBUTION TO THE FRACTURE MICROMECHANISM AND MORPHOLOGY OF Cr-Mo-V TOOL STEELS

JOŽE RODIČ,* ALENKA RODIČ* AND FRANC VODOPIVEC**

Complex research of cold work tool steels has been carried out. Toughness mainly depends on the effect of ferrite hardening alloying elements. Toughness statistics are more strongly dependant on energy necessary for formation of the initial crack than on energy necessary for tearing the metal matrix at crack propagation. The differences in fracture micromorphology are smaller as it could be expected according to differences in work consumed on fracturing or according to differences in microstructure. Toughness is lower when a less finer martensite is formed and when inner tensions are present.

INTRODUCTION

A great serial statistically planned research of 17 hard tool steel grades was carried out in Železarna Ravne according to an internal standarized testing method (1). The "ŽR method" also includes the instrumented toughness test by measuring fracture force P (kN) and fracture time τ (μs) as well as work consumption W (J). Due to the nature of these steels the scatter of the measurements at test-pieces fracturing is considerable. In mass testing the complex testing methodics enabled a systematic selection of the test-pieces for metallography and microphractography and the study of the fracture mechanisms and morphology (2,3).

When testing the toughness of hard steel grades according to the classic method we have to do with abnormal fracturing, sometimes even with crumbling of the specimen (Fig. 1a). Because of this the fracture surface increases excessively and an increased amount of work is needed for the fracture. The result can hardly be compared with another specimen with a fracture surface corresponding to a normal test-piece cross-section. The "ŽR method" (Fig. 1b) has shown a sufficient selectivity and reproductibility so that it has been therefore possible to explain the great differences in measurements even among test-pieces of the same statistical population and a different fracture surface morphology (Fig. 2) by peculiarities of the steel microstructure.

*- SLOVENSKE ŽELEZARNE, ŽELEZARNA RAVNE, YU-62390 RAVNE NA KOR.
**- SLOVENSKE ŽELEZARNE, METALURŠKI INŠTITUT, YU-61000 LJUBLJANA

Fig. 3 shows the average toughness data (force P, time τ, work W) and Rockwell hardnesses for one typical representative of the investigated steels in dependence on hardening temperature and on tempering in a very broad variation range from underheated to overheated and from low to high tempered steels. The degrees of determination in correlation among toughness statistics are very large ($R^2 > 0,9$). This is an important observation as it can serve as a basis for conclusions about the micromechanism of steel fracture.

EXPERIMENTAL FINDINGS

With respect to technologically and statistically logical limitations, we have for the whole family of investigated steels found out uniformly valid regression equations for the toughness data and chemical composition. It is clear that these regressions can only be valid for the processing conditions, being a part of our technology. An example for maximum fracture force is shown in nomogram in Fig. 4. This statistical research has shown a somewhat surprising result that technologically permissive variations in the quantity of some primary alloying elements, for example chromium and tungsten, do not have an important influence on toughness. Therefore only four elements with the predominant influence (carbon, molybdenum, vanadium and manganese) have been considered in the nomogram. A discussion on the reasons for this difference is beyond the task of our work and will be therefore omitted.

In Fig. 5 eleven steel grades are classified into characteristic groups. The connection between the content of carbide forming alloying elements divided with carbon content versus the toughness is obvious in spite of deviations at some steel grades. This dependency may be surprising. It shows in fact a positive influence of the alloying elements exceeding the quantity fixed in carbides. Toughness therefore mainly depends on the part of alloying elements in solution in ferrite. The dependency in Fig. 5 shows smaller deviations in steels tempered at 500 to 550^{o} C because after this tempering the microstructure of steel obtained at hardening turns a less important factor. The influence of some hardening data is presented schematically on an example in Fig. 6. It shows some typical recorded fracture time-force curves, with corresponding work consumption (W), hardness (HRC), amount of residual austenite (A) and martensite tetragonality (TM). When the hardened steel is tempered at 250^{o} C, the quantity of retained austenite doesnt change appreciatively, the martensite tetragonality and inner tensions, on the other hand, become lower. Because of this hardness decreases, fracture force increases to twice, while fracture time shows a much lesser increase, and work consumption an even lesser one. This fact together with the form of the recorded dependence of fracture time-force shows that for toughness measured in hard steels the work needed for the fracture propagation is less important than the work needed for elastic-plastic steel deformation before the initial crack appears. Tempering at a higher temperature results in a minor fracture force and a lower work consumption. The most probable explanation for this is that during tempering at 500^{o}C coherent carbides are formed, causing new inner tensions (4). On fracturing these tensions are added to outer tensions. At higher tempering temperatures carbide precipitations increase as well, coherence and inner tensions become smaller or are even suppressed and an increased fracture force and work consumption result. After tempering the diameter of carbide grains are of the order magnitude of 0.1 to 1 μm, hardness is strongly lowered and the propagation of fracture along the test-piece cross-section becomes important for the toughness data.

Uncontrolled deviations in the heating and deformation during the warm-working occur sometimes in steel processing. This causes microstructural peculiarities in steel which are sometimes connected with catastrophic consequences for toughness. Such peculiarities are relatively coarse martensite lamellas and carbide enrichment along some microstructural details, like twin boundaries (Fig. 7). The simplest explanation for this would be that it is much easier for a macro crack to propagate along such details because less energy is needed for the propagation. However this explanation is not correct! For it has been already mentioned that the fracture force and work consumption depend mainly on processes of crack inititiation. The microstructural characteristics are harmful mainly because they facilitate the crack initiation.

When equal hardness values of steel are obtained with different heat-treating the microstructure differs often in the size of martensite lamellas (Fig. 8) and also in the size and amount of carbide grains due to different austenizing temperature. When hardness shows the same value, all toughness statistics are better in the steel with finer microstructure. The fracture is more articulated in steel with lower toughness. The difference in toughness does not result from the difference of the energy consumed to create a new surface. SEM observations have shown relatively fine morphological peculiarities of the fracture, so a stronger magnification is usually used in microphractography. Figs. 9, 10, 11 in 12 show fractures, hardnesses and toughness statistics for the same steel grade having been heat-treated in different ways, yet having practically equal hardness. The steel with higher toughness (Fig. 9) reveals a fine microstructure of tiny carbide grains dispersion in tempered martensite. The fracture surface is transcrystalline with small dimples, it is therefore a ductile fracture. The diameter of dimples corresponds to 2 - 3 times the size of carbide grains. It indicates that the steel cold deformation immediately before decohesion is limited to a very narrow steel band not more than 1 μm from each crack lip. Very similar is the fracture of the steel which was hardened at lower temperature and had a somewhat finer microstructure and more undissoluted carbides as well as much lower toughness data. (Fig. 10) Very seldom small "quasi cleavage" areas have been observed on the surface of this fracture. By comparing this to the previous sample we can state that the ductile transgranular fracture can by no means be sign of a high toughness. It confirms the already mentioned belief that toughness is primarily a result of the energy consumed for crack initiation.

The third sample in Fig. 11 shows a som ewhat less fine microstructure due to a higher hardening temperature. The rupture work equals the work consumed on the previous sample, however, the fracture is of a mixed type, an intimate mixture of the "quasi cleavage" brittle and ductile intracrystalline surface. It is interesting and perhaps also typical that the brittle "quasi cleavage" surface in microscopic scale is more articulated than the ductile surface. Whenever possible,the crack follows martensite lamellas lying near the surface of the macro crack. If no conveniently situated martensite lamellas are disponible, the crack propagates with cold deformation and ductile decohesion. The fourth sample in Fig. 12 shows lower toughness data and a fracture where the "quasi cleavage" surface is only occasionally interrupted by croocked dimples or deformation edges where a smaller deformation took place before the fracture. The fracture surface is at microscopic scale very articulated. After hardening from the proper temperature ($1040^{o}C$), the steel was tempered at 550^{o} C. The toughness decreased strongly and the fracture was like the one in Fig. 10. This is another confirmation that a lower toughness and a change in fracture morphology result from the precipitation of coherent carbides and from inner tensions related to such precipitation. The example in Fig. 12 indicates that the lowest toughness and a practically completely brittle fracture are obtained when coarse martensite is combined with such precipitation. Differences between steels with the same composition, for instance: the fracture is practically equal in spite of more than 6 times greater work consumption (samples in Figs. 9 and 10), the fracture differs strongly although, the toughness is very similar (samples in Figs. 10 and 11) can be explained with the help of the already proposed explanation that the toughness data depend mainly on work necessary for the crack initiation.

CONCLUSION

Complex investigation of cold work tool steels has been carried out in order to establish facts specially important for the toughness. For toughness determination an original impact method was applied. Results indicate that toughness mainly depends on the degree of ferrite hardening and more from the energy for crack initiation than from the energy for crack propagation. The influence of microstructure on toughness should be evaluated from the propensity of steel to crack initiation. That is the reason for differences in fracture micromorphology which are smaller than it could be expected according to differences in fracturing data or according to differences in microstructure.

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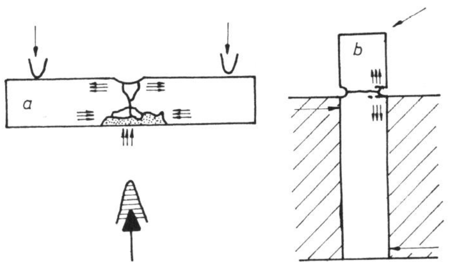


a - classic method of impact testing - abnormal fracturing with crumbling in the case of hard tool steels

b - impact method, developed by Železarna Ravne for complex investigation of hard tool steels.

Fig. 1: Sketch of the breaking mechanism

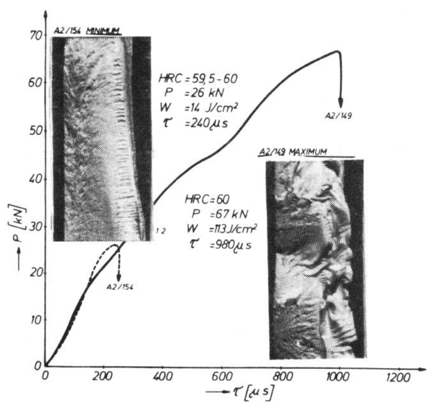
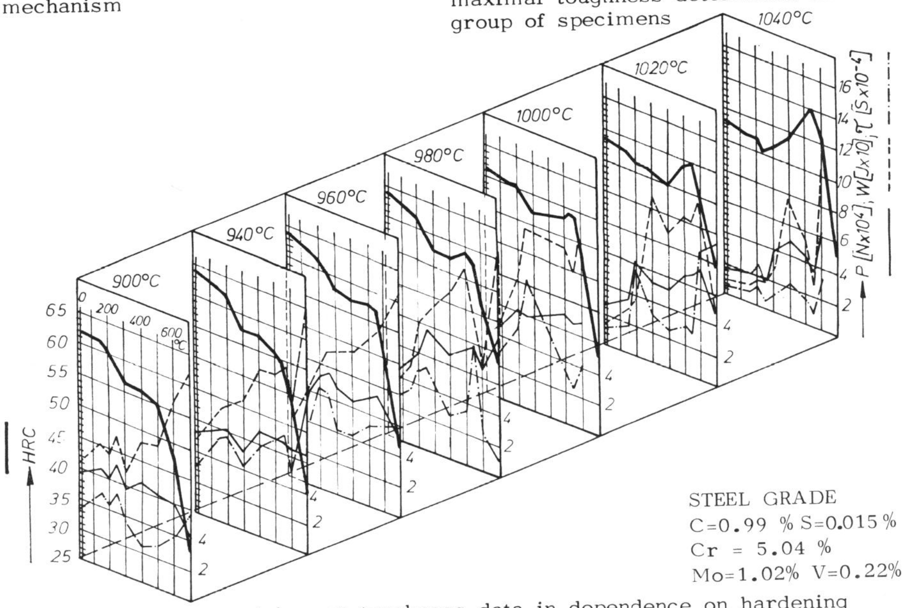


Fig. 2: Comparison of minimal and maximal toughness determined in one group of specimens



STEEL GRADE
C=0.99 % S=0.015 %
Cr = 5.04 %
Mo=1.02% V=0.22%

Fig. 3: Hardness and impact toughness data in dependence on hardening and tempering temperature

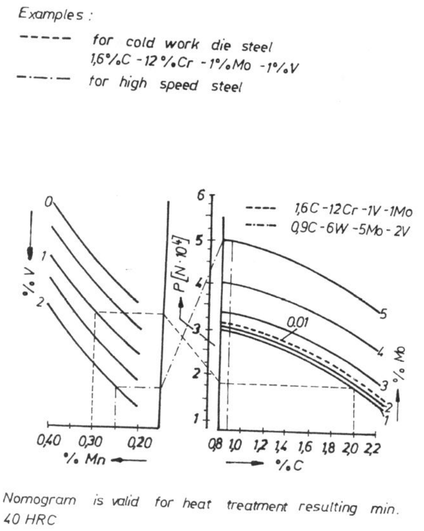


Fig.4: Influence of chemical composition on maximal fracture force

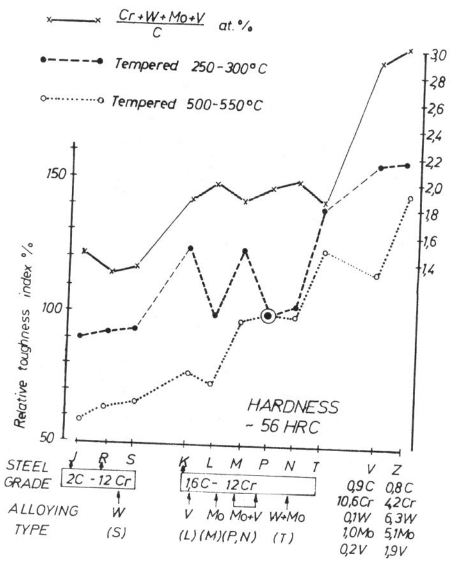
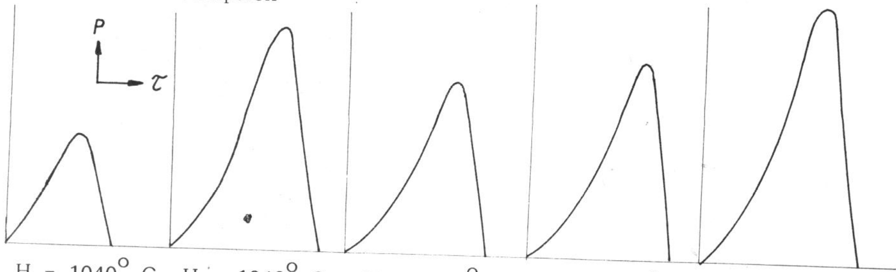


Fig.5: Comparison of the toughness and the carbide-forming elements/carbon ratio

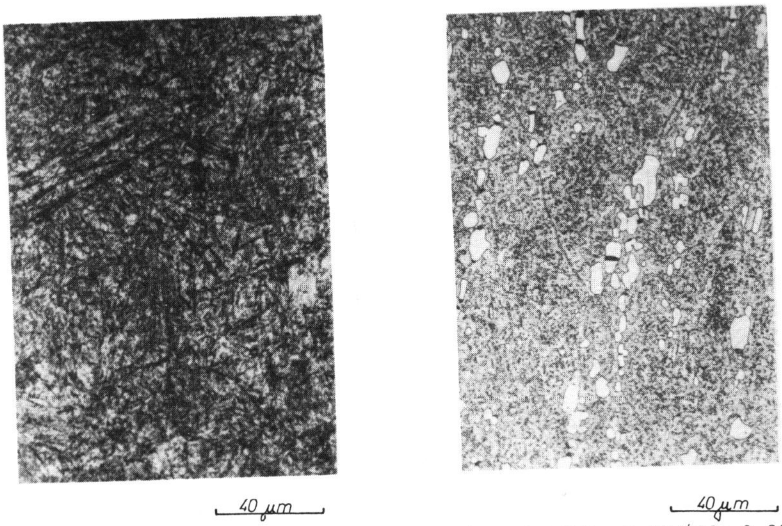
H = Hardening temperature
T = Tempering temperature
HRC = Rockwell hardness
P = Fracture force
W = Work consumption

A = Retained austenite
TM = Martensite tetragonality
STEEL GRADE:
1.52 % C - 11.4 % Cr - 0.82 % Mo - 0.92%V



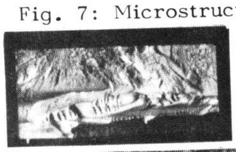
H = 1040° C	H = 1040° C	H = 1040° C	H = 1040° C	H = 1040° C
T = -	T = 250° C	T = 500° C	T = 550° C	T = 600° C
HRC = 64.5	HRC = 59	HRC = 60	HRC = 58	HRC = 50.5
P = 15 kN	P = 30 kN	P = 23 kN	P = 28 kN	P = 34 kN
W = 18 J	W = 21 J	W = 10 J	W = 17 J	W = 21 J
A = 14.3 %	A = 14 %	A = not measured	A = 1.4 %	A $<$ 1 %
TM = 0.007	TM $<$ 0.001	TM = not measured		

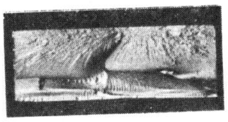
Fig. 6: Influence of tempering effects on impact parameters - registered curves P - τ



0.99 % C-5.04 % Cr-1.02 % Mo-0.22 % V   1.52 % C-11.4 % Cr-0.82 % Mo-0.92 % V

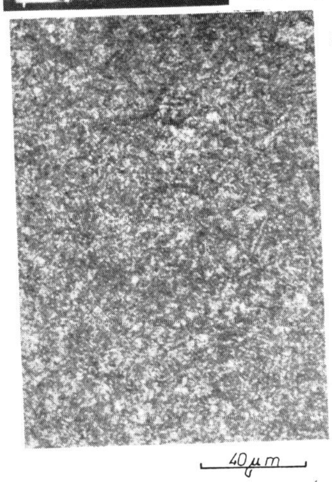
Fig. 7: Microstructures resulting bad toughness (magn. 500 x)





Steel grade OA2 composition
0.99 % C-5.04 % Cr-1.02 % Mo-0.22 % V
Hardened/Tempered

980°C/500°C		1040°C/550°C
	$\overline{\mathrm{HRC}}$	
58.5		58.5
	$\overline{P}$ [kN]	
48.4		36.2
	τ [µs]	
526		382
	$\overline{W}$ [J]	
62.8		35.7



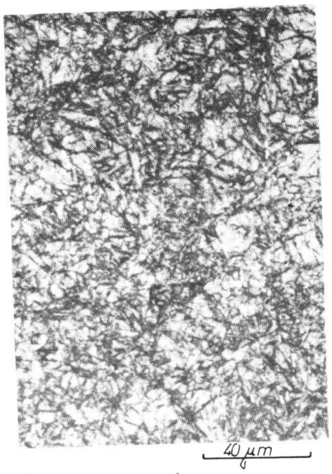


Fig. 8: Microstructure (magn. 500 x), fracture (magn. 0.6 x), heat treatment and toughness data

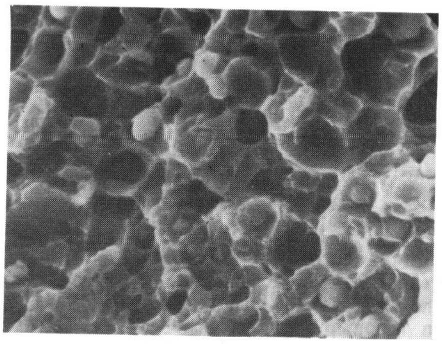


Fig. 9 OA 2 SEM 7200 x
H=980°C T = 250°C HRC = 59
P=65 kN τ = 760 μs W = 103 J

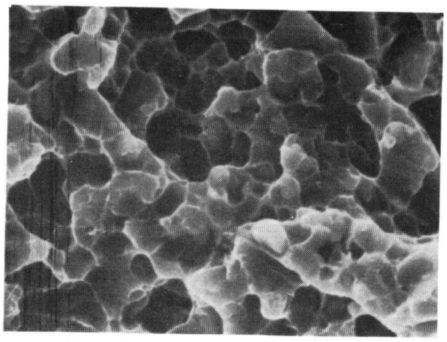


Fig. 10 OA 2 SEM 7200 x
H=920°C T = 200°C HRC = 59
P=35 kN τ = 240 μs W = 17 J

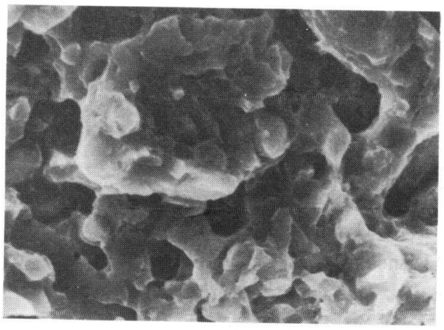


Fig. 11 OA 2 SEM 7200 x
H = 1040°C T = 550°C HRC = 59
P = 34 kN τ = 200 μs W = 18 J

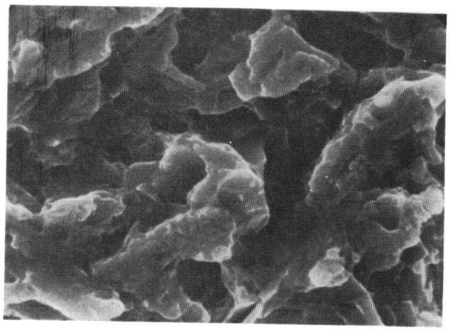


Fig. 12 OA 2 SEM 7200 x
H = 1040°C T = 550°C HRC = 58.5
P = 22 kN τ = 140 μs W = 12 J

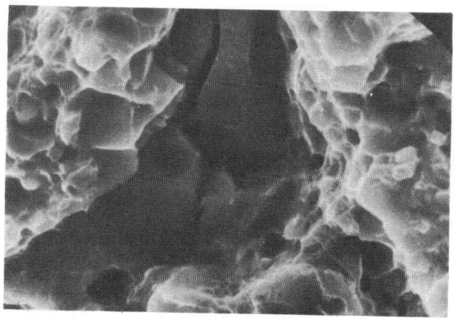


Fig. 13: One detail from the sample in fig. 9

Fig. 9 - 13:

Steel grade OA 2 composition
0.99%C-5.04%Cr-1.02%Mo-0.22%V
H = Hardened T = Tempered
HRC = Rockwell hardness
P = Fracture force kN
τ = Fracture time μs
W = Work consumption J