Fracture toughness of plastic-mold steels: dependence upon heat treatment and microstructure.

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The fracture toughness of low-alloy steels significantly depends upon the microstructure, i.e. upon metallographic constituents amounts and distribution and other parameters like carbides distribution, grain size, etc. Almost all the available studies and experimental data concern single constituent microstructures, whereas specific industrial applications use steels containing variable fractions of tempered martensite and bainite, as well as ferrite-pearlite, due to the production cycle and especially to the heat treatment used. In these cases it is particularly difficult to estimate the fracture toughness on the basis of the available experimental data; thus, it isn't possible in practice to evaluate (and minimize) the crack propagation risk.
Molds for plastic automotive components are a peculiar example, being produced from pre-hardened large blocks. The fracture toughness of the core mixed microstructures, that can occupy parts of the mold face, being sought, K_{Ic} tests have been performed as a function of the sample position in a 1.2738 steel block, and results interpreted at the light of fracture morphology and local microstructure.

Review of fracture toughness data of mixed-microstructure low-alloy steels

Mixed microstructures, such as those usually encountered in slack quenching and tempering of high strength low alloy steels, have not received any attention in terms of fracture toughness (either K_{Ic} or J_{Ic}) characterization. The only measured properties related to this kind of heat-treatment structures date to more than 50 years ago, when notch strength and notch ductility were assessed by Sachs et al. [1] for slack quenched and tempered AISI 1340, 2340, and 5140 steels. Prior slack quenched structures showed appreciable reductions of these properties in respect to prior fully quenched ones; up to 40% in notch strength and 80% in notch ductility in the high tensile strength (1500-2400 MPa) field for the Cr alloyed one (5140).
Only very recently Knott and co-workers [2] have undertaken a campaign to assess the statistical variation of the fracture toughness of mixed microstructures consisting of bainite and auto-tempered martensite (B/M) and obtained by air cooling or salt bath quenching samples of A533B steel austenitized at 1250 or 950°C. Cleavage fractures obtained by testing at –80°C yielded K_{Ic} values of about 55 MPa√m for 45/55% B/M structures with small (8 µm) prior austenitic grain size, and of about 60 MPa√m for 30/70% B/M structures with large (120 µm) austenitic grain size. The fracture toughness (K_{Ic}) of simple upper bainite structures resulted as being 32 and 46 MPa√m, respectively for large and small prior austenitic grain size, whereas the toughness corresponding to simple auto-tempered martensite structures was 89 and 92 MPa√m, respectively.
The above data, while giving a clue to the detrimental effect of bainite on steel fracture toughness, do not allow to single out fracture toughness values to be applied for defect allowance calculations as regards large blocks of low alloy quenched and tempered steels. In fact, the fracture toughness of mixed microstructures consisting of tempered martensite-bainite and pearlite-ferrite, that can occur at the core of the blocks, has not been previously investigated.

Large plastic molds for the automotive industry

Applications, steel grades and critical properties

Large forged steel blooms (typically 1x1 m section and more than 1 m length) are used to fabricate plastic molds, in turn employed to form automotive components, such as bumpers and dashboards, made of thermoplastics, usually polypropylene or reinforced ABS.
Mold design is usually based on previous experience; seldom a complete stress analysis is performed, because the nominal stress level in service is believed to be quite low in respect to the steel tensile strength. No fracture mechanics defect allowance procedure is normal in the mold industry. Yet, some producers reported mold failures during service, due to macroscopically brittle fractures development.
The stress pattern applied to the molds in service arises from the polymer's injection pressure and from the thermal gradients, and could be enhanced by notch effects, due to the geometry of the plastic components that are to be produced, and by the presence of microscopic defects of various origin (particularly weld bed depositions, effected without proper heat treatment). Stresses may be significantly raised during abnormal operations due to incomplete extraction of already molded objects.

Each mold is expected to produce a few millions of pieces in its life, corresponding to the production run of one car model, thus fatigue effects should also be considered.
Wear induced by the reinforced resins flow may be very severe and may be an additional cause for crack nucleation, with the flowing resin infiltrating cracks and acting as a wedge.
Standard compositions of the steels used are given in Table 1. Both 1.2738 (40CrMnNiMo8-6-4, UNI and DIN standards) and 1.2311 (40CrMnMo7) steels are heat treatable, low-alloy steels containing 0,4% C, 2% Cr, 1,5% Mn, 0,2% Mo ca.; the 1.2738 grade has an enhanced hardenability, due to a 1% Ni addition, and is thus preferred in large sections. Ni also promotes toughness. The present work concentrates on the 1.2738 steel.
Despite the high hardenability, in the large steel blocks, various microstructures are present at increasing depths after quench: martensite decreasing and bainite increasing with the surface distance and, at the core, mixed microstructures with fine pearlite and some ferrite appearing initially at prior austenite grain boundaries and increasing thereafter. Tempering, which may last up to 40 h, affects all the microstructures.
In order to reduce cost and manufacture time and to improve logistic management, the traditional cycle (rough machining, heat treating and finishing) has been abandoned by the plastic mold industry in favor of the use of blooms, forged from ingots and quenched and tempered in the steel mill.
Due to the production cycle, any of the above microstructures could be found at the mold face, where notch effects and welding depositions defects are often present.
Thus the relevant steel properties, particularly concerning fracture toughness, fatigue and wear resistance, should be studied as a point-wise function of the microstructure, that in turn is a function of the position in respect to the outer surfaces during the heat treatment. First results of such an undertaking are here presented.

Production cycle

The production cycle of the molds is, usually, the following:

- in the steel mill: casting of steel ingot, forging in order to obtain 1x1 m (or larger) sections, dehydrogenization, oil quenching and tempering, eventually followed by stress relief;
- in the commercial warehouse: removal of decarburized surfaces and sawing to requested dimensions;
- in the mold-machining shop: chip-removal and/or electrical-discharge machining to the mold shape, grinding with or without polishing in selected areas, local surface treatments; upon request, shape corrections using weld depositions.

ESR refining of the steel is usually not possible due to dimensional limits of ESR plants.
The section of the bloom is usually comparable to the section of the original ingot (because larger ingots are not feasible), thus only a slight reduction ratio can be obtained. This problem is partially circumvented by applying a few successive forging steps, each consisting of elongation followed by compression in the same direction, the axis of symmetry of the ingot. Each step can achieve only a limited reduction ratio (1.5 is a possible value), with the total deformation being much less than that obtained in rolling and not comparable in the effects.
Depending on the size, the duration of the hydrogen removal treatment can be of a few days and the duration of austenitizing, tempering and stress relieving can be in excess of 1-2 days for each of these operation. The austenitizing temperature is in the range 840-880°C and the quenching medium is usually oil (use of salt baths has been reported [3]). Tempering is performed in the 550-600°C range to a final 330-300 HB hardness. Heating is usually executed in air.
A rough shape being produced in forging with deep decarburations occurring during heating stages, final material removal from the surface may be up to 20 mm (not including scale). Furthermore, blooms are sometimes sawn to requested size (often asymmetrically); in the case of bumper molds, the bloom is usually sawn to yield a U shape.

Experimental

Specimen preparation

Samples were cut from a commercial bloom used for the production of a bumper mold, taking advantage from the large residual of the U-shaped sawing of the bloom. The detailed thermal and mechanical history of the bloom is not precisely known, nevertheless it is believed to be consistent with the usual and previously described production cycle.
The bloom was delivered to the mold shop showing mechanically worked surfaces; nevertheless, on the basis of the as-delivered dimensions, of the usual production cycle and of some metallographic results, it has been possible to estimate the relative position of the original forged and heat treated surfaces, that are defined *nominal surfaces*.
The chemical analysis, as provided by the steel mill, is reported in Table 1 and satisfies the 1.2738 standard.
The original bloom was a 2420(L) x 1140(T) x 1000(S) mm parallelepiped (nominal

dimensions). The L direction defines the long ingot forging axis in the usual manner, while the S and T directions are believed to be almost indifferent in respect to the casting and forging procedures and define only a conventional reference system.
Four oversize blanks were cut from the positions shown in Fig. 1. The *depth* of a point is defined as its distance from the nearest nominal surface. O and O1 blanks were kept in the conditions determined from the heat-treatment of the original bloom. NT and QT blanks, WxB=41x25 mm, were re-austenitized at 870°C; NT was then normalized, QT oil-quenched; both were thereafter tempered at 560°C for 4h.
A K_{Ic} specimen and a metallographic specimen were machined from each blank; further metallographic specimens as well as a tension specimen were machined from the broken halves of each K_{Ic} tested specimen. Cylindrical 7 mm diameter tension specimens were tested. Metallographic surfaces were perpendicular to the T direction. The orientation and the depth of some relevant points in these mechanical specimens are specified in Table 2.

Results

Optical metallography

In re-heat-treated specimens NT and QT, Fig. 2, the microstructure was tempered martensite (troostite). This microstructure was found to be homogeneous over the whole 41x25 mm section.
The as-received O blank microstructure was observed in regions at different depth, corresponding to the two ends and to the centre of the blank in respect to its larger dimension (the central region was about 3 mm from the K_{Ic} fracture surface). The microstructure was found to be sensibly constant, showing tempered bainite and martensite over the entire specimen. Yet, increasing the depth from 31 up to 217 mm, more regions of tempered bainite were found, with some prior austenite grain boundary constituents appearing at the larger distance.
The as-received O1 specimen, on the contrary, showed quite a large dependence of the microstructure upon the depth, with the end closest to the nominal surface (149 mm distance, Fig. 3a) showing a microstructure similar to that of the O blank mid-point. In the middle (280 mm) and at the furthest end (393 mm) of the same O1 blank, prior austenite grain boundaries constituents increase in volume (Fig.3b). Detailed observations at increasing distances from the nominal surface at 15 mm ca. intervals indicate that pearlite-ferrite was observed first in isolated spots at 197 mm depth and as a continuous constituent beginning from a depth of 212 mm. The K_{Ic} fracture surface was in this mixed-microstructure region.

Mechanical testing

The results of all the mechanical tests (performed at room temperature) are presented in Table 2.
Hardness resulted in the usual range for steel 1.2738 in the commercial pre-hardened condition; a moderately higher hardness was obtained in the re-heat-treated samples, because the tempering temperature was in the lower end (560°C) of the usual range (550-600°C), the tempering time was much shorter (4 hours instead of 1-2 days) and the as-quenched microstructure was probably harder (fully martensitic instead of martensite and bainite).
A remarkable difference was observed in tensile results between the as-received (average YS and UTS: 693 and 904 MPa; hardening exponent, n: 0.1 ca.) and the re-heat-treated specimens (970, 1111 MPa and 0.06).
Fracture toughness tests were performed using actuator displacement control.
K_{Ic} tests [4] were performed upon both O and O1 specimens using a 5 µm/s cross-rate, corresponding to a 0.28 MPa√m/s stress intensity rate and a 4.5 min total test time for the O1 specimen and to 0.44 MPa√m/s and 2 min for the O specimen.
K_{Ic}/J_{Ic} tests were performed on QT and NT specimens using the single-specimen unloading-compliance technique [5], with a rate in the range 3-6 µm/s, corresponding to 0.38-0.53 MPa√m/s.
The only valid fracture toughness data pertains to the O1 specimen (K_{Ic} = 76 MPa√m), with the O specimen yielding not valid results (K_q = 50 MPa√m), owing to K_{max}/K_q being larger than 1.1, and the re-heat-treated specimens having insufficient dimensions (see notes on table 2). For the latter ones calculated critical J values are close to reality, although not perfectly valid; assuming them as valid J_{Ic}, computed K_{IJ} appear to be very close to the respective K_q values. Thus, a K_{Ic} value of the order of 120 MPa√m ca. can be safely hypothesized for both NT and QT conditions. It can be concluded, even with some uncertainty, that the fracture toughness of specimens cut from the original bloom is much lower than that pertaining to specimens of the same compositions characterized by a fully tempered martensite microstructure.

SEM observations of fracture surfaces

NT and QT tension specimens showed fracture surfaces traveled by radial ridges with shear slip lateral faces and very limited external cone zones. O and O1 tension fracture surfaces had a well developed cup-and-cone morphology, with central regions characterized by ductile fracture with void sheet coalescence between primary voids.

The fracture surfaces of the K_{Ic} specimens yielded the following results:

- *O1 specimen* (fracture surface depth 272 mm): mixed, intergranular (typical intergranular facet about 100 µm wide) with quasi-cleavage areas suggesting fracture through pearlite colonies;
- *O specimen* (fracture surface depth 117 mm, Figs. 4a, 4b): prevalently intergranular (typical intergranular facet about 100 µm wide);
- *NT and QT specimens* (re-heat-treated, Figs. 4c, 4d): prevalently intergranular (typical intergranular facet about 10µm wide) with limited ductile fracture areas.

Although the fracture morphologies of O specimen and NT/QT specimens rupture surfaces appear to be intergranular in both cases, a remarkable difference exists in the facet dimensions (100 vs 10 µm) and in the facet appearance; in the O specimen facets are curved with rounded edges and show some segregation effects, whereas facets in NT/QT specimens are more planar and delimited by sharp edges without any segregation.

Close observations of the fracture initiation regions at the end of fatigue pre-cracking has shown that a large blunting has occurred in QT/NT specimens with the formation of a well defined fan of slip lines [6] and multiple shear slip mode II fractures initiating at the blunted notch root. Blunting and shear slip fractures were much less evident in the O1 specimens, and very limited in the case of the O specimen.

Discussion and conclusions

Metallography

Based upon the metallographic observations of the present bloom (not fully reported above) and of another bloom and upon the comparison between CCT diagrams and expected cooling rates, the metallographic constituents at increasing depths in a large plastic mold steel bloom can be estimated as follows:

1. tempered martensite (surface or edges);
2. tempered martensite and bainite;
3. tempered martensite/bainite with an increasing fraction of pearlite-ferrite at the previous austenite grain boundaries, inhomogeneous on a 0.1 mm length scale;
4. tempered bainite and pearlite-ferrite;
5. full pearlite-ferrite (core).

Fracture

The dominant mechanism of fracture appears to be brittle intergranular in the surface portion pertaining to crack propagation. The analysis of one in-service failure occurred in a bumper mold confirmed this result. The dfference between O and O1 specimens on one side and QT and NT specimens on the other side resides in the dimensions of the intergranular facets, the ones pertaining to O/O1 specimens being much larger and less straight than the ones originating in the QT/NT specimens. Such a difference of facet size can be related to different dimensions of the austenitic grain size. Indeed, very long austenitizing times, such as those experienced by the material in the large block from where O/O1 specimens were cut, can be seen as the cause of austenitic grain growth. The intergranular rupture can be in both cases related to embrittling elements migrating to the prior austenite grain borders: both one-step embrittlement and two-step embrittlement (if stress-relieving is applied after tempering) are conceivable [7].

The difference in grain size on the fracture surface cannot be taken as a justification of the large K_c difference between the two groups of results (50/76 vs.119/123). Moreover, previous results on as quenched AISI 4340 steel by Firrao et al. [8] and by Ritchie et al. [9] show that intergranular rupture fracture toughness increases by increasing the austenitic grain size.

Thus, the reported differences in fracture toughness levels among the various microstructures appear to be due to the energy consumption associated with the slip lines deformation and multiple fracture along them.

The difference in fracture toughness between the as-received O/O1 and the re-heat-treated QT/NT microstructures confirms the detrimental effect of uncompleted quenching; this fact and the difference in fracture toughness between the two specimens taken from the as-received bloom also confirms the dependence of the toughness upon the position inside the block.

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Tables and figures

Table 1 - Composition limits of steels used in the mold industry and composition of the 1.2738 heat examined in the present study (wt. %).

	C	Cr	Mn	Ni	Mo	Si	S	P
1.2311 - 40CrMnMo7	0.35 0.45	1.8 2.1	1.3 1.6	-	0.15 0.25	0.2 0.4	<0.035	<0.035
1.2738 - 40CrMnNiMo8-6-4	0.35 0.45	1.8 2.1	1.3 1.6	0.9 1.2	0.15 0.25	0.2 0.4	<0.03	<0.03
1.2738 - Examined Bloom	0.36	2.0	1.5	1.1	0.25	0.25	0.001	0.008

Table 2 - Mechanical properties of the examined bloom (partially from [10]).

Specimens	**Hardness**				**Pre-cracked 3 point bend**							**Uniaxial tension** (T orientation)			
	HRC after quench		HB	HV	orientat.	fracture surface depth	W	B	K_q	B,a min	J_q	depth of reduced section ends	YS	UTS	n
	-		-	-	-	mm	mm	mm	MPa√m	mm	kJ/m²	mm	MPa	MPa	-
O	-		287	307	TS	117	41	25	50$^{(1)}$	13	-	154 - 189	696	888	0.10
O1	-		287	290	TL	272	58	38	76	30	-	319 - 354	690	920	0.11
NT	52	53	330	373	TS	-	41	25	123$^{(1,2)}$	40	83$^{(3)}$	-	966	1110	0.06
QT	54		330	377	TS	-	41	25	119$^{(2)}$	38	72$^{(3)}$	-	973	1112	0.06

Notes: *(1) not valid, K_{max}/K_q>1.1; (2) not valid, 2,5 $(K_q/s_y)^2$ > a or B; (3) not valid, insufficient stable crack propagation.*

Fig. 1 - Positions of O1, O, NT and QT examined blanks in respect to the bloom's nominal surfaces.

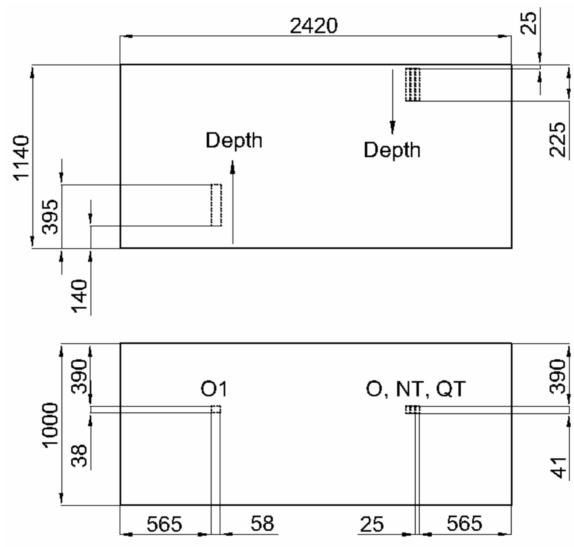


Fig. 2 – Microstructure of Quenched and Tempered (QT) blank. Tempered martensite. 1% Nital etch.

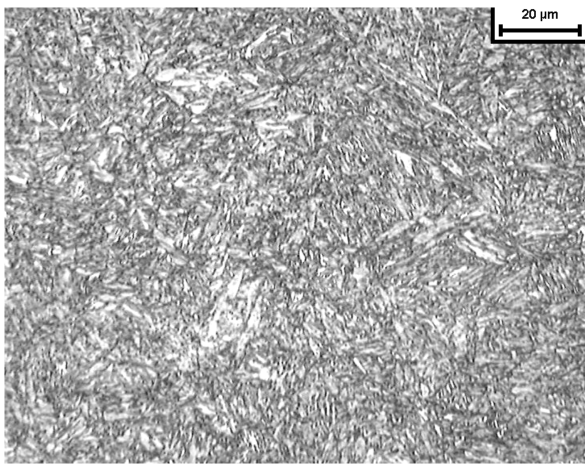


Fig. 3 – O1 blank microstructures at different distances from the bloom surface: (a) 149 mm, tempered martensite and bainite; (b) 280 mm, tempered martensite and bainite with pearlite and some ferrite appearing at prior austenite grain boundaries. 1% Nital etch.

(a) (b)

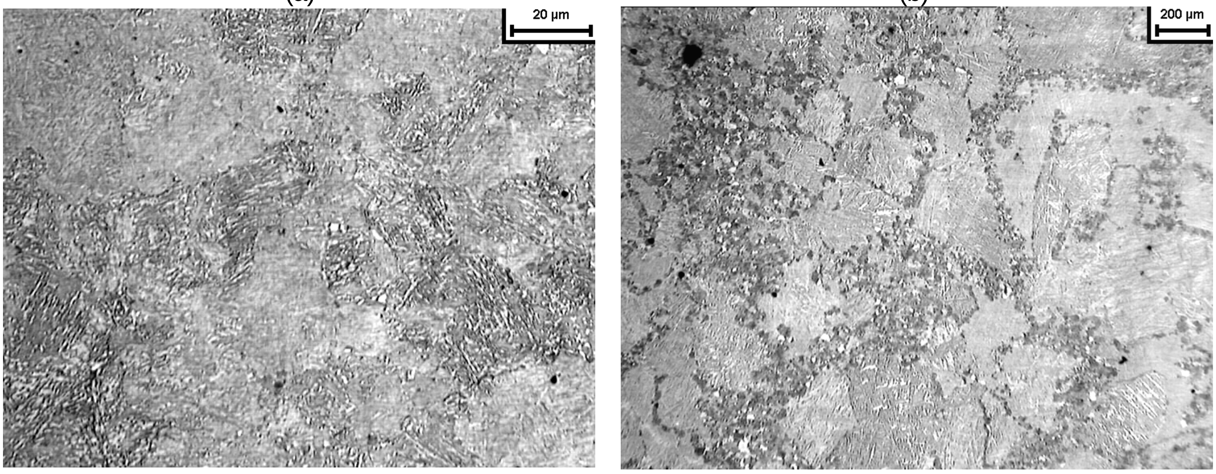
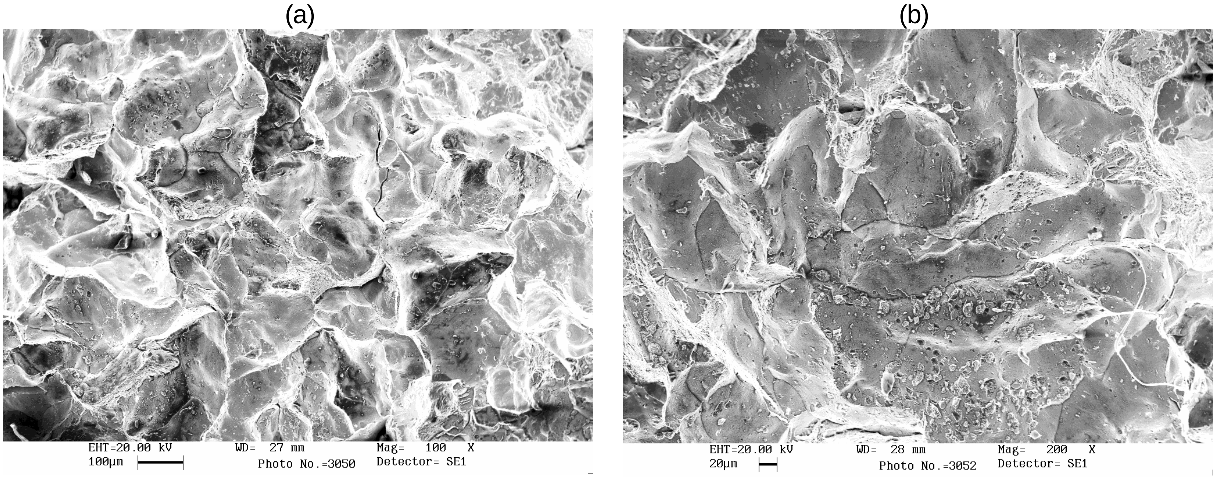


Fig. 4 - Fracture surfaces of O (a,b) and QT (c,d) specimens; intergranular rupture with different facet size and facet morphology.

(a) (b)



(c) (d)